Mechanical Characterisation of High Strength Alloys Produced by Laser Assisted Direct Metal Deposition

By

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THESIS ABSTRACT

High strength alloys particularly steel and titanium based, offer significant advantages of high modulus, stiffness, toughness and good weight to strength ratio if used as porous or lattice structures. These are good candidates for many engineering applications particularly those involving high surface wear rates, large strains and high velocity impact loads. Such applications often demand novel shapes for accommodating special operational and economic requirements, hard surfaces that can withstand severe wear and materials that can sustain dynamic loading and remain strong under demanding conditions. Laser assisted direct metal deposition (DMD) process provides a realistic opportunity to create solid, porous and composite structures from high strength metallic alloys that can be used as coatings, foams and sandwiched structures and as highly stressed components. DMD system especially offers significant advantages for creating novel structures from materials like 316L stainless steel and H13 tool steel by optimizing an array of laser parameters which mainly include the laser power, laser scan speed and powder feed rate. The computer aided control of laser “tool path” provides the advantage of developing novel shapes which also include structures with macro pores, custom shaped cavities and mixed materials.

This research presents an in depth investigation on the utilisation of the specific capabilities of DMD process to develop parts from high strength steel alloys and enhance the range of application of DMD beyond repair and coating. The work includes extensive study of the mechanical behaviour of solid or porous parts processed by DMD under static, dynamic, severe wear and fatigue loading conditions through thorough investigation of DMD process parameters and controlling the quality of built parts. The aim is to develop optimized recipe of process parameters that can help in producing parts which are fully dense, have minimal micro-porosity and good dimensional accuracy.

The following major outcomes have been achieved from this research focussed on developing solid, porous and composite structures from high strength stainless and tool steel alloys. The outcomes fulfil several research gaps in the characterisation of high strength steel parts produced by laser based DMD process under high strain rate loading, sliding wear and fatigue loading conditions. The developed components have been proven good for industrial applications and can be implemented as stable and strong coatings:

- Thorough investigation of DMD parameters for controlling the quality of built parts.
- Development of strong and optimally bonded rectangular and cylindrical coatings of superior quality steel alloys on non-expensive and easily available mild steel specimens as substrate.
• Determining the dynamic characteristics of DMD generated steels against high strain rate compressive loadings and a thorough comparison including mathematical modelling with the corresponding properties of wrought steel alloys.

• In depth evaluation of DMD generated steels against severe sliding wear. It focusses on finding the co-efficient of friction and specific wear rate in comparison to the published behaviour of rolled and wrought steels under similar conditions.

• Investigation of high-cycle fatigue characteristics of composite structures developed from thick coating of alloy steels on mild steel pins using DMD process.

• Study of residual stresses through neutron diffraction techniques to evaluate the effects of laser cladding process on the stress strain behaviour of developed parts

The findings of this research fill a very important gap in implementing the laser generated parts for direct industrial applications, particularly in the development of porous and lattice structures and with high magnitude compressive and impact loading conditions. The obtained results also encourage the enhanced application of DMD machine in producing customized parts apart from coating and repair applications. This research thoroughly establishes the capabilities and limitations of DMD process and is an attempt to maximally utilize the novel features of lase assisted additive manufacturing.
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PhD research has been the longest and perhaps the most tortuous path that I have traversed in my academic life. It comes with lots of learning opportunities, though only few of these I could grab successfully. Its ebb and flow carries with it some new acquaintances, few of which became influential enough that you cannot help but acknowledge their contribution and assistance in achieving the ultimate goal.

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DEVELOPMENT

I hereby declare that I am the sole author of this PhD dissertation and to the best sources of my knowledge it does not contain any content exclusively that has been published by other researchers previously and elsewhere, but in every instance of reporting or referring being provided with appropriate citations.

This is the true and only copy of the thesis that has been submitted in fulfilment of the requirement of PhD degree. This thesis is not submitted anywhere else for the purpose of seeking some other degree or qualification.

I do not have any objection if my dissertation is made available to fellow researchers either in written or electronically.

Syed Haider Riza

August, 2015
DEDICATION

I shall like to dedicate this thesis to my parents who with dedication and without intention of any reward continuously worked hard to make me a better person

And

To my wife who during all the years of our companionship stood beside me in trials and testing periods of life.
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CHAPTER 1
INTRODUCTION

1.1 Background
For the last two decades laser assisted additive manufacturing (LAAM) have been employed in many fields for producing customized parts and performing tedious repairing and coating jobs for expensive and difficult to handle objects. With the passage of time, these technologies have become more viable and provide a sizable niche for complementing, if not replacing the conventional manufacturing technologies of casting, forming and moulding. Main advantage of these systems is non-rerequirement of tooling, but it comes at the cost of reduced dimensional accuracy and lesser strength of produced parts. Therefore, on one hand LAAM techniques provide an exciting opportunity to produce parts from novel materials with properties and configurations that are almost impossible to produce through conventional means; but on the other hand functional use of laser generated parts demands thorough material and mechanical characterization. The main reason behind this necessity of mechanical characterization of LAAM parts or coatings is the non-traditional manufacturing method which does not involve shaping or compacting of the molten metal. The mechanical properties of metallic LAAM specimen cannot simply be determined from the properties of wrought or cast material of the same grade and composition. Melting of very small amount of metallic powder at a particular location followed by its solidification, accompanied by the forward motion of laser beam while completing a particular section, raises serious questions about strength of formed section and inherent micro-porosity. In addition the part is built layer by layer with one section deposited over another which brings forth the issues of inter-laminar bonding strength and induction of residual stresses due to irregular heating and cooling within the part matrix. Therefore, a thorough assessment of important properties like hardness, microstructure, dynamic behaviour and wear characteristics will result in mapping the range of application and should strongly introduce the laser based systems for manufacturing parts that can be used in actual machine and industrial applications and not merely limited to the applications related to repair and coating.

1.2 Laser Assisted Additive Manufacturing of Metallic Parts
Machining or material removal from a raw piece of metal has been a dominant manufacturing process for many centuries. But since the last 30 years, a new domain of manufacturing has emerged that works on a contrary principle, which involves adding material in a sequential manner generally as a series of thin layers upon layers to produce a part. This process has been given many names over the years like rapid prototyping, rapid manufacturing (RM), and solid freeform fabrication (SFF) [1]. But
differences in machine setups, processing techniques, and processed materials cannot diminish one common factor that brings all of these technologies under one umbrella, namely the ability to produce the final part or prototype through material addition (no material removal). The components begin from a single bead of molten powder but develop gradually in a layer by layer fashion with every pass of high power laser or electron beam operating under direct control of a computer-generated solid model. Thus the most appropriate name covering the basic theme of this process is Additive Manufacturing (AM). AM is now the official industry standard term (ASTM F2792) for all applications of the technology. It is defined as “the process of joining materials to make objects from 3D model data, usually layer upon layer, as opposed to subtractive manufacturing methodologies. Synonyms are additive fabrication, additive processes, additive techniques, additive layer manufacturing, layer manufacturing, and freeform fabrication.” [2]

Real-time opportunity offered by laser assisted AM technologies to create complex and near net shapes and profiles while eliminating expensive and time-consuming tooling under direct computer control has generated immense inspiration and motivation for coating, repair and development of parts with un-conventional geometries and dissimilar materials [3, 4]. The application industries include aerospace, electronics, bio-medical and tissue engineering. The continued evolution of AM is expected to offer original equipment manufacturer companies (OEMs) the opportunity to change their supply chain configurations in the future by introducing distributed production [5].

Among the most popular applications of AM technologies is to produce functional parts with minimum requirements of subsequent machining or heat treatment. Producing parts for direct end-use is more challenging than building models and prototypes, so in general AM technologies will take time to fully develop and need to overcome the issues of quality and repeatability [6]. But laser assisted additive manufacturing (LAAM) have already found applications in many crucial and critical areas like development of biomedical scaffolds and implants. It provides a cost-effective and efficient method for producing scaffolds that match the complex anatomical geometry of craniofacial or periodontal structures [7, 8]. It is also opening a niche in repairing of defence and aerospace related components that are cast and machined from costly wrought material with special properties like those of titanium and aluminium alloys [9]. An important aspect of LAAM is that it can be applied as a coating as well as for production of complete parts. Thus these technologies also emerge as a very suitable option for repairing the expensive worn or damaged parts, dies and moulds [10] and for increasing the quality of an ordinary surface by applying a thin coating of property enhancing material [11]. This capability allows this process to be used together with other processes and previously built parts, for example it is possible to build turbine blades directly onto a machined shaft as well as remanufacture or repair worn out parts [12].

[2]
1.3 Mechanical Characterization of Laser Generated Parts

This research is based on a particular LAAM process known as direct metal deposition (DMD). DMD is vastly different from the conventional processes of casting and forming in that the material is processed in powder form instead of handled in bulk form as either solid or molten. A flowing stream of metallic powder is continuously melted by a focused laser beam and the small molten mass is deposited on the compatible substrate already present on the machine table. This localized melting and deposition carries on with the moving laser beam; and once a single layer of cross-section is completely deposited the cycle re-starts for the next layer and so on until the desired height is achieved. DMD process has been extensively used in coating and repair of difficult to process and expensive metallic parts. A wide range of metallic powders including bio-compatible titanium alloy and a variety of steel alloy powders can be used for creating solid and porous structures using DMD cladding.

With this brief introduction to DMD process, it should be evident that in DMD the part is actually “created” sequentially and gradually from the metallic powder without the aid of any die or tooling that can guide the shaping of material. In addition there is no arrangement for applying compacting pressure on the deposited metal like in powder metallurgy or high pressure die casting. Therefore, apart from efficiency of CAD/CAM system the outcome of DMD process mainly depends on a number of machine related parameters, variation of which can result in generation of materials with vastly different mechanical and material properties.

There are several problems and shortcomings that have been reported in relation to material and mechanical characteristics of laser cladded parts and will be discussed in detail in chapters 2 and 3 of this thesis. These may be categorized as follows:

1. Micro-porosity
2. Low Modulus of Elasticity
3. Less dimensional accuracy
4. Non-homogeneity of micro-structure
5. High surface roughness
6. High residual stresses

In addition, there are several characteristics and behaviours that are scarcely investigated and reported for laser generated high strength metallic alloy structures. These mainly include:

A. Compressive dynamic behaviour at high strain rates
B. Wear behaviour under severe and dry conditions
C. Fatigue characteristics for composite (substrate +cladding) structures
There are also some issues and considerations which are particular for the DMD process. For instance, there is a question mark on its ability to produce structures with fine macro-pores of cross-sectional area less than 1 mm² that can be used as bio-medical scaffolds.

1.4 Aims and Objectives of this Research

Since to-date the focus of major interest in DMD applications lies upon coating and repairing applications for complex parts and difficult to process materials. But this research intends to enhance and extend the DMD system towards producing parts that can be directly used in industrial applications owing to its large bed size of 1m x 2m and ability to apply large variety of high strength metallic alloys to produce customized shapes. There are three main motivational objectives for this research. First one is related to a complete characterization of DMD process in view of its capabilities and limitations for producing laser generated parts and not only coatings. Since industrial applications involve solid and porous parts both, and in fact there is growing interest in porous parts for use as scaffolds, fluid flow channels with increased heat transfer and energy absorption structures. Therefore, the initial phase of investigation aims to raise the questions and find answers related to capabilities of DMD in preparing solid and porous structures for industrial applications like producing energy absorbing and impact bearing parts. In case of porous parts main concern relates to minimum size achievable and uniformity of cross-section of (macro) pores. For preparing solid structures on DMD that can be used in intended applications the issues to be resolved include the process imposed limitations on size shape, and height of DMD cladded parts without compromising the dimensional accuracy in diameter. Another important aspect is to observe the repeatability of DMD process.

Apart from investigation of DMD parameters, another major objective of this research is to investigate surface and bulk behaviour of DMD generated metallic structures against normal, shear and fatigue loading. This involves determining the mechanical properties of DMD generated structures to ascertain their quality in comparison to cast and wrought alloys commercially available and find out their usability in certain customized applications. Important mechanical properties that can reveal this behaviour include response of DMD parts under quasi-static and dynamic loading, friction and wear characteristics and fatigue strength under cyclic loading. Therefore, this research investigated the properties of DMD generated high strength steel alloy parts and ascertain their characteristics in respect of certain major aspects like strength, hardness, elastic-plastic behaviour, fatigue and severe sliding wear. The investigation on dynamic compressive behaviour specifically included porous cladded parts in addition to solid ones and numerical modelling of the behaviour in addition to experimental analysis.
The mechanical characterization cannot be complete without material related information which pertains to the investigation of micro-structure, distribution of micro-hardness and material composition as expressed by EDS and XRD analyses. This is the third major objective which aims to identify the composition and micro-structure of cladding, bonding strength between clad and substrate and variation of hardness and strength within the composite structure. The two main research streams along with sub-branches of pursued research work is illustrated in the form of a flow chart presented in Fig 1.1.

Figure 1.1: Flowchart illustrating major streams and sub-branches of the research work

1.5 Research Gaps and Contribution:

In this research the mechanical properties are mainly investigated for laser generated 316L stainless steel (SS) and H13 tool steel (TS) parts which are produced through direct metal deposition process. Mechanical properties of these and similar high strength alloy steels produced from conventional processes in solid, porous and lattice configurations have been thoroughly investigated under different type of loadings by many researchers covering diversified applications [13-18]. But there is a significant gap in the research on properties of the same materials developed by laser cladding. Most of the publications on characterization of structures generated by powder or wire based metallic deposition are focussed on the study of metallurgical aspects particularly evolution and configuration of microstructure [19-21]. And the investigations into mechanical properties are generally limited to evaluating the variation of micro-hardness within the cladding or at cladding-substrate interface and determination of yield and ultimate strengths [22, 23]. There is no doubt that this is not only the beginning rather basic element of characterization, but for full-fledged industrial applications of laser
generated parts, there exists an imminent gap in thorough characterization with respect to dynamic mechanical behaviour which can be related to surface and bulk behaviour under tensile, compressive and shearing loads. This research aims to fulfil this gap while focusing on mechanical properties like high strain rate stress-strain behaviour over elastic and plastic regions, sliding wear characteristics under dry conditions and determination of fatigue strength for composite cladded structures. 

Before highlighting the significant contributions of this research, it is preferable to describe many considerations that form the basic approach and background of this research work. These may be illustrated as follows:

1. All the parts and specimens are tested in as-clad condition with minimal surface finish and no heat treatment.
2. Materials investigated are titanium alloy Ti6Al4V and steel alloys 41C, 316L stainless steel and H13 tool steel. Initial part of the research was dedicated to investigating the possibilities of making metallic scaffolds on DMD. But as this possibility faded away and the focus shifted on developing DMD generated parts for industrial applications, stainless and tool steels remained the materials of interest due to their lesser cost, better manufacturability as compared to titanium alloys and strong bonding and intermixing qualities with other steel alloys particularly mild steel.
3. When creating the parts, emphasis was on economy of process which means using the least amount of powder possible during cladding and operating with the least value of laser power that can generate a near fully dense structure.
4. All the cladding was done by a continuous wave CO2 laser that can provide power up to 5 KW and without using the feedback camera system.
5. The whole set of DMD process parameters, necessary to perform a cladding operation, which is affectionately given the name as “DMD recipe” was finalised after a series of iterative experiments keeping in view the properties of a given material. A detailed account of DMD parametric investigation is provided in chapter 4.

Major contributions arising out of this research are presented as follows:

- The work provides a clear and useful insight not only into the material and metallurgical qualities of DMD generated steel specimen but also into the capabilities and limitations of DMD process vis-à-vis development of high strength metallic alloy parts for biomedical and industrial applications.
- The work also helps in diagnosing the abnormalities in the stress-strain relationship of laser generated parts which is an expected phenomenon for a DMD type method of production.
- Another important contribution is the comparison of dynamic characteristics of fully solid parts with those having cavities or macro-pores. Reason for this approach is that if porous parts
developed on DMD prove themselves to be amply strong and tough particularly under impact loads then a whole new vista of opportunities for developing honeycombs and sandwiched structures of customized shapes and configurations will be opened that can be used in energy absorption applications.

- Extending the dynamic investigation into mathematical modelling of deformation characteristics of laser generated steels in the light of established models already derived for commercially available stainless steels. Subsequent comparison with experimental results indicates close similarities between theoretical and experimental curves in the yielding region but significant deviation in the plastic region.

- In a DMD process, the cladding is done on a compatible metallic substrate, which may be shaved off or retained post cladding. This research investigates into the performance of “composite cylindrical laser cladded structures” for compressive strength, wear properties and fatigue strength. The composite form gives a significant advantage in terms of cost and DMD time saving because it contains a much cheaper mild steel core carrying only 2-3 mm of cladding done from an expensive stainless steel and tool steel alloy. Investigation shows that the performance gap is not huge and composite parts can exhibit comparable behaviour as the completely cladded part which opens the door for huge savings and benefits not only in terms of powder material but in laser beam time and machine expenditure as well.

- Ascertaining the friction and wear behaviour of DMD generated steel alloy specimen covering a large number of cases and possibilities. The experiments on pin-on-disc wear tester involved testing of fully cladded and composite structures under severe sliding wear and dry or unlubricated conditions. This is an important piece of research because there exists an apprehension for LAAM parts that wearing under high loads in dry conditions can remove the cladding at such rapid rates that the whole part may be erased in few minutes.

- Developing a fatigue “profile” of laser cladded composite structures under high cycle fatigue. In comparison to the rolled stainless steel bars, the fatigue performance of DMD generated specimen proved to be much inferior and only slightly improved with heat treatment.

1.6 Organisation of Thesis

The chapters in this thesis are organised in the following manner:

Chapter 1 provides an introduction to the background of research and the approach and contributing aspects of this research. It is neither a detailed literature review nor an in-depth discussion about research investigations.
Chapter 2 is a detailed literature review of laser assisted additive manufacturing systems and takes into account mainstream technologies, nature and quality of parts developed on LAAM and the research gaps that need to be looked into.

Chapter 3 is also a literature review of the metallurgical and mechanical properties of laser generated parts. This chapter also takes into consideration the commercial grade steels in view of the particular mechanical properties investigated in this research for the sake of comparison with the laser manufactured steel alloy parts.

Chapter 4 looks in detail into the DMD operation and influence of parameters that actually drive the whole process. Discussion of strengths, constraints and certain limitations of DMD process are highlighted in view of experiments conducted for part development from titanium and steel alloys.

Chapter 5 deals with the material and metallurgical characteristics of DMD generated steel parts. The analysis encompasses microstructure of cladded parts, XRD and EDS analysis for phase and composition analysis and determination of microhardness within the substrate, at the interface and at various locations in the cladding. This chapter also includes the measurement and analysis of residual stresses generated in composite 316L and H13 specimen of approximate overall thickness of 10 mm with 3 mm cladding and rest the substrate which was mild steel.

Chapter 6 presents the detailed investigation into high strain rate dynamic behaviour of solid, porous and composite cladded parts made from 316L and H13 steel alloys under compressive loading. The research also delves into the micrographs of surfaces after impact and any possible post-impact microstructural changes.

Chapter 7 is also a detailed analysis of sliding wear behaviour of DMD generated solid and composite parts. The wear characteristics are defined in terms of co-efficient of friction and specific wear rate. Micrographic analysis is performed on the worn parts to check the extent of degradation of surfaces. An analysis of wear debris is also presented to determine the extent of oxidation that took place of the cladded materials.

Chapter 8 looks into the high cycle fatigue behaviour of composite cylindrical specimen developed through cladding of 316L and H13 steels on a mild steel pin. The observed fatigue behaviour is compared with the same type of behaviour measured for 316L commercial grade rolled SS.

Chapter 9 is the final chapter that includes concluding remarks and some proposals for future work and the dimensions in which the current state of research can be extended.
CHAPTER 2
LASER ASSISTED AM TECHNOLOGIES – REVIEW OF PROCESS AND APPLICATIONS

2.1 Introduction
High power and highly focused laser beams have been used for cutting and machining applications for nearly half a century, but in additive manufacturing (AM) technologies their focused power has been used for part building through fusion of powders of different materials. Laser based manufacturing systems have become advanced enough to produce functional prototypes and modify the surface properties of existing parts for improved mechanical properties [24]. Additive manufacturing has led to Rapid Tooling (RT) which involves using AM for production of dies and moulds and Rapid Manufacturing (RM) that pertains to employing AM for developing near-net shaped parts suitable for end use with negligible post processing and machining [25]. Laser assisted additive manufacturing (LAAM) technologies have found their niche in those industries and applications where mass production is not feasible or needed and the parts to be made or repaired are too expensive and highly customized. Best examples are knee, hip, dental and craniofacial implants, repairing of gas turbine blades and manufacture of defence related parts.

In all the LAAM processes part is built up by programming the laser movement according to geometry extracted from CAD model. The part model is sliced by the CAM software into a series of cross-sectional layers inside the computer’s memory. Hardware setup consists of a powder bed or a substrate on which powder is sprinkled. Each cross-sectional layer is formed by laser or bed movement in the horizontal plane and simultaneous melting of polymeric, ceramic or metallic powder. Due to high focus and small size of laser beam, heating of powder remains localized and solidification occurs rapidly in small, localized volumes, resulting in fine, as-deposited microstructures [26]. The part is gradually built up as one cross-sectional layer formed upon the previous layer by the laser beam.

A flow chart shown in Fig 2.1 illustrates the basic configuration of LAAM system. Set of inputs shown at top level, that is CAD model, CAM control and laser beam are common in every system. This commonality in nature of inputs is the main strength of LAAM technologies which imparts it an un paralleled flexibility and opportunity to experiment and innovate “during” manufacturing. Direct fabrication of metallic components by laser melting of metallic powders can be performed in one of the two ways (i) powder-in-bed (ii) powder injection through nozzles [27]. Powder-in- bed systems have the very important benefit of supporting the part and any overhang during layer by layer build-up. Selective laser sintering (SLS) and selective laser melting (SLM) cover this domain. Powder injection systems utilise a coaxial nozzle to feed powder into the laser beam focusing on a compatible substrate mounted on the machine table to form the “melt pool”. The profile is generated by either the
controlled movement of laser beam or the machine table. Main technologies in this category are laser engineered net shaping (LENS – moving table) and direct metal deposition (DMD – moving beam). In the DMD process, a different version of the process is also used which involves feeding the material as wire from an off-axis nozzle in addition to powder from co-axial nozzles [28]. The recipient industries are shown as biomedical, aerospace and automotive because advancement in AM technologies has been exemplified in these three key industries due to reduction in lead time, ability to realise highly complex products and the ease with which 3D medical imaging data can be converted into solid objects [29].

Figure 2-1: Flowchart showing configuration of laser assisted additive manufacturing system

Majority of LAAM systems employ CO₂ or Nd:YAG laser. The main difference between the two types lies in their wavelength. The absorptivity of most metals increases by decreasing the wavelength. Tolochko et al [30] investigated the normal spectral absorption of a number of metal, ceramic and polymer powders with two laser wavelengths of 1.06μm and 10.6μm obtained by using each of the two lasers – Nd:YAG and CO₂ respectively. Table 2.1 shows the absorptivity of some common metals used in powder form for laser generated parts and coatings.

Table 2.1: Absorptivity ratio of metallic powders to Nd:YAG and CO₂ lasers [30]

<table>
<thead>
<tr>
<th>Material</th>
<th>Nd:YAG</th>
<th>CO₂</th>
</tr>
</thead>
<tbody>
<tr>
<td>Copper (Cu)</td>
<td>0.59</td>
<td>0.26</td>
</tr>
<tr>
<td>Iron (Fe)</td>
<td>0.64</td>
<td>0.45</td>
</tr>
<tr>
<td>Tin (Sn)</td>
<td>0.66</td>
<td>0.23</td>
</tr>
<tr>
<td>Titanium (Ti)</td>
<td>0.77</td>
<td>0.59</td>
</tr>
<tr>
<td>Lead (Pb)</td>
<td>0.79</td>
<td></td>
</tr>
<tr>
<td>Co-alloy (1% C; 28% Cr; 4% W)</td>
<td>0.58</td>
<td>0.25</td>
</tr>
<tr>
<td>Cu-alloy (10% Al)</td>
<td>0.63</td>
<td>0.32</td>
</tr>
<tr>
<td>Ni-alloy I (13% Cr; 3% B; 4% Si; 0.6% C)</td>
<td>0.64</td>
<td>0.42</td>
</tr>
<tr>
<td>Ni-alloy II (15% Cr; 3.1% Si; 4%; 0.8% C)</td>
<td>0.72</td>
<td>0.51</td>
</tr>
</tbody>
</table>
This chapter provides a detailed review of LAAM in terms of mainstream technologies and the nature and quality of parts developed by these laser based systems. Since its inception major research activity on LAAM has been dedicated to development of biomedical scaffolds and implants, therefore, a significant portion of discussion goes in highlighting the research contributions towards development of solid and porous biomedical structures. But in fact the results of that research is not confined to biomedical domain only but provides an extensive insight into the capabilities, possibilities and limitations of LAAM systems. The discussion also takes into account the commercial development of aerospace and automotive parts using LAAM technologies. Since a large number of avionics research is restricted or classified in nature therefore, the details are not available like wide ranging research publications on biomedical equipment, justifying the reason behind referring to a number of websites for aerospace parts development.

The discussion begins with highlighting the unique features and advantages of LAAM, then moves on to a brief overview of operating methods of different LAAM technologies followed by their merits and de-merits. It proceeds onwards towards an in-depth discussion of different types of parts developed on LAAM and taking into account the role of LAAM systems in surface modification of parts through coating. Final section elucidates the existing gaps and openings available within the research domain of laser assisted metallic parts development.

2.2 Unique Advantages of LAAM
Additive manufacturing methods offer some significant and unique advantages which manifest themselves in the production of functionally graded materials (FGM), porous scaffolds, custom made implants, coating and repair of difficult to process and expensive components [31]. LAAM technologies have gradually discovered a niche in certain applications where other methods are not optimally suitable or lack in crucial capabilities. In the following lines some peculiar advantages of LAAM systems are highlighted that are beneficial to many key industries. These advantages render LAAM as techniques which actually complement rather than eliminate the traditional manufacturing systems and strongly put up their case as technology of future.

2.2.1 Absence of Tooling
In the entire LAAM cycle, the biggest advantage is absence of any tooling that can guide and define the shape of object. This is the benefit that opens the door to a number of novel and high end applications [32]. One of the biggest issues associated with manufacturing processes like forming, casting and forging is the use of tooling like dies and moulds. Developing a given set of tooling is a costly and tedious process but often justified by the fact that vast majority of plastic and many metal parts are produced in huge quantities. However, there are still many specialized and costly parts which are needed in small quantities, for which production of moulds presents a myriad of technical
problems, and hence effort and expense behind tooling becomes prohibitive and non-feasible. One of the best examples of this phenomenon is in the aerospace industry and to a lesser extent in biomedical engineering. That is why the aerospace and implant manufacturers form the biggest markets for additive manufacturing technologies. Atzeni and Salmi [33] performed a breakeven cost analysis for manufacturing an aircraft’s landing gear through high pressure die casting compared with SLS. According to their research SLS not only emerges as more economical but much faster due to elimination of tool and die production time.

In electronics and household industries parts are needed in bulk quantities with an operational life of few years only. But parts for airplanes and implants and prostheses for biomedical applications are made in very limited quantities and must stay in service for decades. Even a commercial airplane is built of millions of parts and most of these need to be strong, lightweight and stable under demanding conditions. When these parts are broken or damaged, there must be a rapid and feasible way of repairing or replacement. In case of re-production or repair using moulds, the process of transporting and setting up the particular mould is time consuming besides the hassle in maintaining a proper inventory of every part. But in a LAAM system entire part detail is preserved in the CAD model, therefore one computer can execute the job of hundreds of moulds. The use of software in place of mould not only reduces the production/repair lead time to a great extent about also eliminates all the costs associated with the handling and transportation of the needed die or mould. All the data related to part geometry is available at one location with an indefinite virtual life and can be instantly used to invoke laser power for reproduction or repair.

2.2.2 Mass Customization

Mass customization is a manufacturing technique that combines the flexibility and personalization of customization with cost effectiveness of mass production [34]. In this respect LAAM systems have been proved extremely successful particularly in the manufacturing of biomedical equipment and parts for aircrafts. In fact LAAM has made mass customization a cost-effective and efficient option for high tech industries. A very good example is fuel-injector produced on SLM and shown in Fig 2.2 [35]. The component design was extremely complex due to combined optimization of airflow and fuel swirling in addition to incorporating the integrated cooling channels. LAAM enabled the whole part to develop as a single piece from cobalt-chrome alloys with 40% reduction of weight and 50% saving in cost.
Certain products like hearing aids, dental implants and prosthetics for limbs require that every unit be custom-made to match the body of the specific end user. Therefore, there is no room for mass production but opportunity exists for mass customization through LAAM. For every new product, traditionally, cost depends on the design, required tooling and the quantity intended. But for LAAM every product produced simply does not involve these cost altering factors because every part is generated in exactly the same manner. There are examples of many specialised prostheses that are developed as custom-fit components using LAAM technologies without incurring additional cost due to change in design or dimensions. SLS has been used for fabrication of many customized prostheses. One good example is the fabrication of polystyrene resin prototype to correct nasal defects [36]. Another remarkable instance is the manufacture of Passive Dynamic (PD) Response ankle–foot Orthoses (PD-AFOs) by SLS [37, 38, 39]. Fig 2.3 illustrates some designs of foot orthoses developed on SLS.

![Figure 2.3: PD-AFOs manufactured by SLS (a) Single strut AFO with topology optimization [37] (b) PD-AFOs with range of bending stiffness values. Stiffest orthosis on the left, most flexible orthosis on the right [38] (c) Novel prosthetic foot designs optimized through CAD [39]](image)

These assistive devices seek to improve walking ability for persons with various neuromuscular disorders. SLS allows the manufacture of PD-AFOs directly from digitised shape information of the patient’s limb, which eliminates the necessity for preliminary moulds, hand lamination and finishing procedures. In this way LAAM technology significantly expands the options for exploring and developing new designs and optimizing their characteristics without development of new tooling.
2.2.3 Complex Geometrical Features

Another unique advantage of additive manufacturing technologies is their ability to create complex geometrical shapes that would be impossible to do with any other method of manufacturing. While machining complex geometrical features, subtractive manufacturing or machining methods often run into the problem of tool clearance and tool accessibility. This is basically due to a finite size of cutting tool which requires some amount of space to reach the machining zone. Tool clearance is usually overcome through dividing the part into modules and subsequent joining by means of adhesives or fasteners. But besides creating structural weaknesses joining is not always possible especially in case of small or fragile parts. For such objects, AM is simply the only manufacturing option available which can create complex geometric features with no cost penalty in manufacturing. Fig 2.4 illustrates some LAAM produced parts that exhibit complex features and intricate structures. Warnke et al [40] produced ultra-light mesh structures with over 450 holes and channels per cm$^3$ in a single scaffold. The porous titanium alloy cube produced by them is shown in Fig 2.4(a). Mullen et al manufactured a hip augment and can be seen in Fig 2.4(b). The structure exhibited a major pore diameter of 500 µm, a fully interconnecting porosity of 65%, and a compressive strength (in the build direction) of 53 MPa [41]. Using DMD setup Mazumder et al [42] developed and characterized a metallic scaffold with complex geometrical features as one piece, from widely used Ti6Al4V alloy mimicking the condylar ramus unit (CRU) of a Yucatan minipig as shown in Fig 2.4(c).

Figure 2-4: Complex shapes produced by LAAM systems (a) Porous titanium alloy cubes of side length 0.5 cm [40] (b) Porous titanium hip augment [41] (c) CAD image and an identical Ti6Al4V scaffold of the CRU of a Yucatan minipig produced by the DMD process [42]
2.3 Brief Review of LAAM technologies

In this section the operating principles of four mainstream LAAM technologies are presented. Within 10 years of emergence of first commercial laser manufacturing system, SLS in 1989, three other major systems also came into being which are SLM, LENS and DMD. Every system has its share of development and contribution in enriching the AM arena and till to-date exhibited a sustained progress curve despite serious challenges from traditional and non-laser based additive manufacturing techniques [43].

2.3.1 Selective Laser Sintering (SLS)

Selective laser sintering involves the selective use of a laser to build up a model layer by layer from a fine powder bed. A typical SLS system is shown in the schematics of Fig 2.5(a) [44]. The fine powder particles adhere and solidify (or sinter) when illuminated by a laser beam. Using a beam deflection system each layer is scanned according to its corresponding cross section as calculated from the CAD model. The deposition of successive powder layers with a typical thickness in the range of 20 to 150µm is realized using a powder deposition system. There are many machines available commercially which differ in the way the powder is deposited (e.g. by roller or scraper), the inert atmosphere (Argon or Nitrogen) and the type of laser used for cladding (CO₂ laser, lamp or diode pumped Nd:YAG laser, disk or fibre laser) [45]. The immediate advantage offered by SLS in comparison to well established stereo-lithography technique was non-requirement of support structures, since the un-sintered powder provides support during the part build-up [46].

SLS inherently offers good opportunity to create controlled porous structures as shown by a Polycaprolactone (PCL) specimen shown in Fig 2.5(b) and thus is very suitable for manufacturing TE scaffolds [47]. In fact, SLS offers good user control over the scaffold’s microstructures by adjusting the main processing parameters, which are laser power, scan speed and powder bed temperature [48].

Figure 2.5: Selective Laser Sintering process and parts (a) Schematics of SLS process [44] (b) SLS processed PCL specimen with 3D orthogonal porous channels [47] (c) A manifold made on SLS from Durable Polyamide (Nylon) [51]
Compared to other non-metallic RP systems such as fused deposition modelling (FDM), three dimensional printing (3DP), stereo-lithography, or inkjet printing, SLS provides many benefits for creating porous structures. Layer-by-layer additive fabrication in SLS allows construction with complex internal and external geometries. Second, virtually any powdered biomaterial that will fuse but not decompose under a laser beam can be used for fabrication. For scaffold manufacturing, SLS does not require the use of organic solvents, can be used to make intricate biphasic scaffold geometries, and does not require the use of a filament (as in FDM). It offers an easy method to incorporate multiple materials; it is fast and cost effective, thus making it a well-adapted technology for the fabrication of tissue engineering scaffolds [47, 49].

SLS is a sintering (not melting) process, in which the powder particles are adhered to each other at high temperatures below their melting point. The powder binding mechanism in an SLS system has been adequately described by Kruth et al [50] and can be classified into solid state sintering, chemically induced binding, liquid phase sintering with partial melting and full melting. SLS got itself well established rapidly for processing of polymer powders, as can be viewed in Fig. 2.5(c) from the top manifold assembly produced from nylon powder [51]. But due to sintering "constraint" SLS could not prove equally efficient for metallic powders. During early stages of development, materials that could be fabricated in SLS included: polycarbonate (PC), nylon, nylon/glass composite, wax, ceramics, elastomeric and metal-polymer powders, and later on biocompatible powders such as Hydroxyapatite and poly (L-Lactide). Initial work on metals was reported around year 2002 for high-strength powder mixtures, such as Fe-Cu, WC-Co, TiC-Ni/Co/Mo, TiCN-Ni, TiB2-Ni, ZrB2-Cu and Fe3C-Fe. But the most important aspect is that each mixture contained two metal powders and only the powder having the lower melting point was fused to act as the binder for high melting and high strength powder [52, 53].

2.3.2 Selective Laser Melting (SLM)
Selective laser melting is also known as Direct Metal Laser Melting (DMLS) although the former name is currently more popular. Some of the problems apparent in SLS systems were resolved by using a high-energy laser beam to eliminate the use of binders and directly fuse the high melting point metallic powder in layers successively deposited one over other as ultra-thin two-dimensional cross-sections. The main goal was to produce metallic parts with 100% density [54]. The typical SLM system is shown in Fig 2.6(a) which uses a very precise laser beam of only 0.03 mm diameter that builds, along Z axis in steps of only 0.05 mm; metal parts of any complexity [55]. A wiper coats a metallic powder upon a metal plate. The laser beam melts the powder on the focusing level of the powder in welding beads. The melt process is similar to the deposit welding. The powder is melted and fused with the metal plate. After the treatment in the focusing level the metal plate is lowered and
for next layer deposition the wipers lay again metallic powders. Thus second layer of metallic powder in the focusing level of the laser beam is now present. The laser beam works on this level and merges the metallic powder together with the previous melted and solidified weld beads. In this manner the part is built and composed as ultra-thin layer upon layer. The process achieves a building speed of 5 cm$^3$/h in fully dense steel.

When we speak of fully dense objects created by SLM and other laser melting technologies, it is extremely important to note that compactness and density of created part is directly proportional to the quality and size of powder used. Shellabear and Nyrhila [56] reported the gradual development in powder quality till the production of a tool steel material called DirectSteel H20 which achieves a density of almost 100%, an ultimate tensile strength of up to 1,100 MPa and a hardness of up to 42 Rockwell C directly from the DMLS process. Therefore, mechanical properties of finished components are comparable to those of the elementary powder metal materials used in the laser building process. Currently parts produced by SLM require no post-processing; neither infiltration with other materials nor post-heat treatment. These parts are produced completely dense and homogenous with negligible microscopic pores or voids. Many metallic alloys can be processed such as low melting point alloys, zinc, bronze, stainless steel, tool steel, titanium and cobalt-chrome alloys [57].

Figure 2.6: SLM process and illustration of produced parts (a) Schematics of SLM process [55] (b) Cooling element from aluminium AlSi10Mg alloy [58] (c) Die model for bevel gear forming made of nickel-based alloy powder [59]

SLM has shown great potential to produce fully functional parts that can be employed as customised biomedical implants for medical applications, metallic alloy parts for industrial applications and lightweight products for aerospace. Citim GmbH of Germany has developed many functional prototypes from metallic alloys on SLM with very good mechanical properties. Fig 2.6(b) shows a cooling element from aluminium alloy AlSi10Mg which possessed a 0.2% yield strength of 240 MPa and maximum tensile strength of 410 MPa; values which are better than cast aluminium
alloys [58]. Osakada and Shiomi [59] developed a die model for bevel gear from nickel-based alloy powder of size 75 µm with the composition as 83%Ni, 9.4%Cr, 1.8%B, 2.8%Si, 2.0%Fe and 0.4%C. The finished part is shown in Fig 2.6(c) with size of the model as 45mm x 35mm x 10.6mm high. The relative density of the model was measured as 88% and the Vickers hardness as 740.

2.3.3 Laser Engineered Net Shaping (LENS)
LENS was developed in Sandia National Laboratories, USA, to fabricate metal components directly from CAD solid models and reduce the lead times for metal part fabrication [60]. The system as illustrated in Fig 2.7(a) consists of an Nd:YAG laser, an argon filled glovebox, a 3-axis computer controlled positioning system, and a powder delivery or feed unit. The beam is brought into the glovebox through a window mounted on the top of the glovebox and directed to the deposition region using a plano-convex lens. The powder delivery nozzle is designed to inject the powder stream directly into the focused laser beam and the lens and powder nozzle move as an integral unit. Tool path patterns to build each layer are obtained by electronically slicing the previously built CAD solid model into a sequence of layers. Physically each layer is fabricated by first generating an outline of the key component features and then filling the cross-section using a rastering technique. The substrate on which the part is to be built is positioned on an X–Y table which is computer controlled to create the master pattern required to build the structure. The build height of the structure is controlled by movement in the Z direction of the laser and powder delivery system. The build is created by subsequently depositing one layer on top of one another until the assembly is completed [61, 62].

![LENS system](image)

**Figure 2.7:** LENS system (a) Schematic representation of the LENS process [61] (b) Single line, single wall specimen produced by LENS [62] (c) Repaired Ti6Al4V bearing housing of a gas turbine engine [64]

LENS has the unique capability of producing parts with thin walls and high depth-to-diameter aspect ratios. Parts have been fabricated with 0.356mm (.014 in) diameter holes having a depth-to-diameter
aspect ratio of more than 70:1. Fig 2.7(b) illustrates the ability of LENS system to build appreciably high structures with consistency in wall thickness. Another unique processing feature of LENS is the capability of selectively applying metal to existing parts for repairing worn or broken parts while maintaining the integrity of the parent material [63]. Fig 2.7(c) shows the damaged region in a housing of a titanium alloy bearing, the inner side of which was repaired using a LENS system [64].

2.3.4. Direct Metal Deposition
The concept of Direct Metal Deposition (DMD) was conceived at University of Michigan and developed by Precision Optical Manufacturing (POM) in collaboration with TRUMPF Germany [65]. Direct Metal Deposition is an additive manufacturing (AM) technique in which metallic structures can be built layer by layer through laser melting of metallic powders and subsequent cladding on a substrate material [66]. Since its inception in 1999, this AM technology has matured enough, enabling and encouraging researchers to experiment with the solid and porous specimen having controlled and variable porosity [67]. Therefore, DMD offers building up of complex structures from metallic powders of steels, titanium, nickel and cobalt based alloys. The structures developed are reported to possess good material, micro-structural and mechanical properties which are generally better than cast parts; directly through CAD models, without additional tooling [68].

![Figure 2.8: Illustration of DMD setup (a) Schematics of DMD process [84] (b) Configuration of POM-DMD-505 machine [69]](image_url)

Operationally DMD is similar to LENS and essentially different from powder-bed based systems like SLM. The schematics of process is shown in Fig 2.8(a). The picture of commercially available POM-DMD is presented in Fig 2.8(b) [69]. The machine operates through a CNC controlled CO\textsubscript{2} laser focusing on a substrate for first layer deposition or already cladded material's melt pool for subsequent layers; while at the same time injecting, melting and depositing small amount of metal powder (flowing through coaxial nozzles) to build the part as sequence of thin layers, one upon another. The
laser scans the substrate under the instructions of a program that is driven by the features and parameters of the CAD model, imported from a commercial solid modeller. Thus in this way the object can be directly manufactured without any complex tooling. A feedback control system in DMD machine provides a closed loop control to maintain uniform deposition thickness, thus significantly reducing post-processing time [70].

2.4 Merits and de-merits of LAAM processes
Before a detailed review of the research is done on each of these technologies, a comparative table is presented for quick introduction to the processing features and respective merits and demerits of each of these four systems. The information presented in Table 2.2 illustrates an overview of the advantages and disadvantages with respect to materials and applications.

<table>
<thead>
<tr>
<th>Process</th>
<th>Merits</th>
<th>De-merits</th>
<th>Materials Processed</th>
<th>References</th>
</tr>
</thead>
<tbody>
<tr>
<td>Selective Laser Sintering (SLS)</td>
<td>• Widest range of materials processed.</td>
<td>• Difficult to produce pure metallic parts.</td>
<td>Calcium Phosphates, Polycaprolactone (PCL) Carbon-fibre composites, hydroxyl-apatite (HAp) and poly (vinyl alcohol) (PVA), poly(L-lactide) (PLLA)</td>
<td>44 – 53</td>
</tr>
<tr>
<td></td>
<td>• No extra support needed other than the powder bed.</td>
<td>• Rough, grainy and porous surface finish.</td>
<td></td>
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<tr>
<td></td>
<td>• Good for Intricate and complex shapes.</td>
<td>• The large shrink rates increase the tendency for the prototype to warp, bow or curl.</td>
<td></td>
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<tr>
<td></td>
<td>• Fast build times.</td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td></td>
<td>• Good part stability</td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td>Direct Laser Metal Sintering OR Selective Laser Melting (SLM)</td>
<td>• Produces high density metallic scaffolds and implants.</td>
<td>• Not suitable for large lots or large parts, because of slowness of process.</td>
<td>Ti6Al4V, Stainless Steel 316L, Pure Titanium Grade -I &amp; II, Co-Cr-Mo alloys, Ti6Al7Nb alloy</td>
<td>40, 55, 57, 76-80, 89</td>
</tr>
<tr>
<td></td>
<td>• Negligible post processing is needed.</td>
<td>• Laser beam size is 0.03 mm diameter which builds, along Z axis in steps of only 0.05 mm.</td>
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<tr>
<td></td>
<td>• Highly complex geometries can be easily produced.</td>
<td>• The process is relatively expensive.</td>
<td></td>
<td></td>
</tr>
<tr>
<td></td>
<td>• Part well-supported by powder bed during build-up</td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td>Laser Engineered Net Shaping (LENS)</td>
<td>Fully dense parts with negligible micro-porosity.</td>
<td>Finishing procedures required post cladding.</td>
<td>Commercially pure Titanium, TiO₂, Co-Cr-Mo alloys, Tantalum*</td>
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<tr>
<td>• Wide range of metal powders can be processed.</td>
<td>• Residual stresses produced in the generated prototype.</td>
<td>*The last two materials are used for coating on titanium implants.</td>
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<tr>
<td>• Functionally graded scaffolds can be produced with controlled porosity and material properties.</td>
<td>• Limitation of slope in producing inclined surfaces.</td>
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<tr>
<td>• Low dimensional accuracy.</td>
<td>• Finishing procedures required post cladding.</td>
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<tr>
<th>Direct Metal Deposition (DMD)</th>
<th>Very good for coating and repairing applications for damaged expensive and large size parts.</th>
<th>The built part needs separation from the substrate.</th>
<th>Ti6Al4V, Stainless steel alloys 41C &amp; 316L, Tool steel H13</th>
</tr>
</thead>
<tbody>
<tr>
<td>• Fully dense prototypes can be produced</td>
<td>• Surface finish is rough and demands finishing procedures.</td>
<td>• Residual stresses after cladding are induced.</td>
<td>42, 65-73, 95, 96, 101</td>
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<tr>
<td>• Wide range of metallic powders can be processed.</td>
<td>• Low dimensional accuracy.</td>
<td>• Problem of oxidation for some metals particularly with titanium.</td>
<td></td>
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<tr>
<td>• Composite and functionally graded parts can be easily produced.</td>
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2.5 Nature and Quality of parts produced by LAAM

LAAM technologies have matured enough to present themselves as a viable manufacturing option particularly for customised applications. Laser generated parts differ in their nature of applications as well as in their properties after cladding or built-up from powder bed. In this section, an analysis is conducted to assess the nature and quality of parts produced on various LAAM systems which include characteristics like dimensional accuracy, surface roughness, micro-porosity and control of geometric features. It must be kept in view that this discussion is not related to material and mechanical characterization of LAAM generated parts; a discussion that will be taken up in the next chapter. Another important consideration is that only metallic parts are considered for analysis although there exists a huge volume of published literature on LAAM produced polymers and ceramics.
The laser assisted part building process is defined by a large number of process and material based parameters which have an effect on the properties of parts produced. For SLM and SLS these include the processing parameters such as laser power, scanning speed, scan line spacing (hatch distance), size of powder particle, thickness of powder layer, scanning strategy and speed, working atmosphere and temperature of powder bed. In LENS and DMD systems spatial distribution (mode) of the laser beam, shielding gas flow, powder delivery gas flow and powder delivery method are additionally important parameters [72]. Material-based input parameters are powder granulo-morphometry, chemical composition, thermal, optical, metallurgical, mechanical and rheological characteristics. Physical behaviour of “laser radiation–powder–substrate” system includes absorption, reflection, radiation and heat transfer, phase transformations, a moving interface between solid phase and liquid phase, fluid flow caused by surface tension gradient and mass transportation within the molten pool, and chemical reactions. Fig 2.9 illustrates a number of variable parameters that are associated with the formation of melt pool and generally affect solidification process and powder utilization efficiency [73].

The microstructural features and mechanical properties of the as-deposited part are significantly affected by the melt pool size, the rate at which cooling and subsequent solidification occurs at the solid–liquid interface of the molten pool, and by the thermal cycles that may occur during the deposition process [74]. Changing the process parameters can generate a non-linear response, for example changing laser power or scanning speed can lead to various undesirable effects such as irregularity and balling effect. Therefore, it is important to establish the relationships between the principal LAAM parameters, surface morphology and bulk properties [75, 76].

The combination of parameters significantly affect the surface smoothness and morphology of laser deposited material. Smoothness or roughness of the surfaces of as built laser cladded parts is an important consideration in assessing the quality of a particular additive manufacturing process. Yadroitsev and Sumarov [76] used SLM to examine surface morphology of one-pass thin wall from
SS 904 powder which is produced by gradually increasing layer thickness from 40 μm up to 80 μm in steps of 10 μm. The surface roughness and morphology varies with the scan speeds (V) and hatch distance. When hatch distance is less than track width, reduction of the powder consolidation zone leads to the situation when the laser beam directly interacts with the powder resulting in a rougher surface. At lower scan speeds i.e. \( V = 0.04 - 0.06 \) m/s for all the range of layer thickness, the surface of the wall was rough. Surface imperfections resulted from irregularity and distortions of single vectors due to powder excess. For \( V = 0.14 \) m/s and layer thickness \( h = 70 \) μm and greater, small irregular pores appeared. With increasing scan speed upto 0.2 m/s, pores became regular, enlarged, elongated and perpendicular to the sintering direction even at smaller layer thickness (\( h = 60 \) μm). For smaller layer thickness (\( h = 40-50 \) μm) the surface of the wall is smoother as a result of fine structure of the sintered tracks (the height of the track is smaller and the remelted depth into underlying track is bigger).

Dadbaksh et al [77] have discussed the effects of temperature gradient over the SLM powder bed that can generate deformation and delamination within the cladding. The non-homogenous temperature gradient exists due to cool argon gas entering from one side of the powder bed and resulted in a biased cooling direction. Their research also revealed the presence of micro-porosity within the steel parts produced parallel and perpendicular to gas flow and can be seen to be around 18 per cent. This indicates that the layout of parts has a negligible influence on porosity formation and powder consolidation conditions. This suggests that the temperature gradients can be controlled through proper part layout and design of scan strategy but for controlling porosity, other parameters like laser power and scan speed needs to be adjusted.

SLM offers the significant advantage of producing very fine and porous metallic structures while at the same time accommodating variety of shapes that are not limited to only the prismatic ones. This makes it the preferred technology for producing porous structures like metallic scaffolds and implants. For bone implants, extremely fine micromesh structure, the implant will be reduced in weight by about 90% while still retaining enormous strength and stiffness within the material. The cancellous bone can be "grown" inside thin mesh structures of prefabricated bio-compatible titanium alloy Ti6Al4V. This facilitates for highly defined, patient-specific external scaffold structure, and controlled internal structural characteristics. The result should be a stable bone–titanium alloy composite, which would have an application as a bone replacement in tumour or reconstructive surgery.

Vandenbroucke and Kruth [78] have reported development of dental prostheses on SLM employing two biocompatible metal alloys, Ti6Al4V and Co-Cr-Mo. They used M3 Linear machine from the German company Concept Laser GmbH which uses a diode-pumped Nd:YAG laser with beam spot size of 200 μm and maximum power of 95W on the building platform. Noticeable fact is the
requirement of more energy density (195 J/mm\(^3\)) with less build-up rate (1.8 cm\(^3\)/h) for titanium alloy when compared with corresponding values of 85 J/mm\(^3\) and 4.0 cm\(^3\)/h for Co-Cr-Mo alloy. It may be noted that energy density determines the energy supplied by the laser beam to a unit volume of metallic powder being processed. Ying Lin et al [79] employed SLM to develop lumbar spine inter-body fusion cages using biocompatible Ti6Al4V powder. The chosen dimension of the cage was 24.5 mm (L) by 14 mm (W) by 14 mm (H), which is composed of 3.2 mm by 3.2 mm by 3.2 mm microstructures with the minimal feature size of 500 µm. Uklejewski et al [80] employed selective laser melting (SLM) to develop a prototype of a minimally invasive resurfacing hip arthroplasty (RHA) endoprosthesis with the original multi-spiked connecting scaffold (MSC-Scaffold) from Ti6Al7Nb alloy powder. This implant offers cement-less fixation into the peri-articular trabecular bone in addition to adequate opportunity of bone tissue ingrowth into the inter-spiked space of the MSC-Scaffold. The process parameters applied during the SLM manufacturing of the prototype were: layer thickness 50 µm, scan speed 125 mm/s, built rate 4 cm\(^3\)/h.

LENS has also been used to create tissue engineering scaffolds and biomedical implants but did not prove so successful as SLS and SLM. Bandyopadhyay et al [81, 82] in a series of papers reported the fabrication of porous Ti implants with mechanical properties matching those of natural bone. Commercially pure (CP) titanium powder was used to develop porous samples on a substrate of rolled Ti plates 3 mm thick. The system employs a 500 Watts Nd:YAG laser in a glove box containing an argon atmosphere. Laser powers of 250 and 300 W were chosen to partially melt metal powders during deposition process to create the desired porous structures. Scan speeds of 5, 10, 15 and 18 mm/s were used to fabricate structures with varying porosity. Similarly, powder feed rates of 18, 23, 28 and 38 g/min were used to vary the porosity in the samples. Also, the distance between two successive laser scans was varied between 0.76 and 9.52 mm to tailor the pore size and distribution. Initially the microstructural study revealed irregularity in shape of the pores for all samples processed under various processing parameters. The size and connectivity of pores were also low. After changing many process parameters highly interconnected porosity coupled with low density was observed in the samples processed at low laser power, high hatch distance, high powder feed rate and high scan speed. Figure 2.10 shows hip stems that were directly fabricated using LENS with various amounts of macro-porosities to reduce bulk density and effective stiffness.
It is possible in a DMD process to generate structures with controlled properties at micro and macro levels without significant heat treatment since the laser creates a very small heat affected zone. Laser assisted DMDs can be used on almost any metallic surface and possess the ability to mix metals in more than one ways to create graded structures [74]. Developing bone implants and tissue engineering scaffolds on a DMD machine is a challenging task. This is due to the fact that originally DMD was conceived to produce only solid structures with very little porosity [75]. The developed scaffold was tested and characterized thoroughly from the viewpoint of micro-structure as well as mechanical properties. Most important outcomes were the lacking of mechanical and surface criteria recommended for TE scaffolds. This requires post-fabrication operations like annealing and sandblasting.

Apart from biomedical scaffolds and implants, there have been other attempts to create LAAM based porous structures for industrial purposes. Porous and cellular metallic structures and their special sub-class called “metallic foams” are known to have many interesting combinations of physical and mechanical properties such as low specific weight, high gas permeability, and high thermal conductivity. For this reason, there exists a great interest in their practical applications in terms of filtration and separation, heat exchangers and cooling machines, storage and transfer of liquids, and fluid flow control [84 - 86]. Lattice structures have been enjoying a sustainable development curve while finding numerous applications in engineering and medicine, some of which include lightweight and stiff structures, energy absorption devices at predictable and uniform stress level and heat transfer applications [87, 88]. Yadroitsev et al [89] used ultra-fine resolution of SLM process to manufacture geometrically complex filters with customized shapes and orientation in order to decrease flow resistance (back-pressure) and improve filtration performance. They also experimented with a heterogeneously structured filter in which the first five layers (layer thickness 200 µm) were synthesized from a nickel alloy + polycarbonate MP mixture with the composition ratio 10:1 and in each of the following five layers the proportion of nickel alloy was increased from 12:1 to 22:1 in steps of 2. Fig 2.11(a) illustrates a customized orientation filter made from 316L stainless steel. Paul et al [90] used continuous wave CO$_2$ laser to develop a number of porous structures from Inconel 625 with porosities ranging from approximately 2 – 11%. Fig 2.11 (b) displays one of many
porous blocks which is conspicuous by the non-uniformity of edges and the top surface, an effect that is typical of laser generated structures when the height of build increases beyond 12-15 mm.

Figure 2.11: LAAM generated porous structures (a) Filter from stainless steel grade 316L with micro-sized square channels. Size of the channel is 150 µm x 150 µm, the wall thickness is 120 µm [89] (b) Laser generated porous structure from Inconel 625 [90]

Statistical analyses performed to correlate porosity with process parameters indicate that the powder fed per unit traverse length was the important independent process parameter that governed the porosity. Transverse traverse index came next in significance as an independent parameter, while laser energy per unit traverse length showed little effect on the porosity as an independent parameter. But in combination when powder fed per unit traverse length and transverse traverse index were increased, the porosity also increased significantly.

Since in LAAM the part is developed at a very small scale which is comparable to the beam spot size and due to controllability of beam’s contour, many attempts have been made to produce functionally graded materials (FGM). In an FGM, physical, chemical, biochemical, and mechanical properties change with position. The manufacturing process of an FGM can usually be divided in building the spatially inhomogeneous structure (“gradation”) and transformation of this structure into a bulk material (“consolidation”) [91]. Using LAAM the main interest in FGM is to produce parts with varying degree of composition from compatible metals. DMD is particularly helpful due to possibility of simultaneous powder flow from different feed hoppers that can contain separate materials. In all graded structures the problem of cracking and material blending pose the major challenge. Which composition will provide the best results is a matter of continuing research and intensive mathematical modelling. While experimenting with creation of graded structures from TiC and CP Ti powders on Ti6Al4V substrate using LENS system, Liu and DuPont [92] postulated that these frequently occurring cracks resulted from a high level of combined process-generated thermal stresses and interfacial stress mismatch. In brittle materials, relief of thermal stresses can generally
only occur in the form of cracking, whereas in ductile materials the stress relief can be accomplished by plastic deformation.

Muller et al [93] described the preparation of a wall with 15 mm height, 90 mm length and 0.8 mm thickness – with a material gradient perpendicular to the substrate (z-axis) from alternate deposition of 316L SS and Stellite 6 powders. They used a direct metal deposition system with multiple powder feeders. Powders were deposited using two separate hoppers, with the first contained elemental 316L powder (A), and the second contained elemental Stellite 6 powder (B). The composition of the deposited powder was achieved by controlling the powder flow rates from each feeders A and B. For modelling, the operation simulation techniques and control theory were extensively used to correlate powder flow rate, powder flow tinning, bead formation and material blend behaviour. The proposed and cladded specimen are shown in Fig 2.12 (a) & (b) respectively. Good correlation was found between the simulated and the deposited material distributions. Although the sample produced did not possess visible pores or cracks, the composition was a bit different from the intended value and the gradient was non-linear.

Domac and Baughman [94] investigated the improvement in properties of metallic thermal protection system in an aircraft by replacing an Inconel 718 bracket with a graded component produced by AM methods from combining Inconel 718 and Ti6Al4V alloys. They used LENS system to produce constant composition blends of the two materials by building walls 25 mm wide x 12.5 mm tall and columns 9.5x 9.5 x 12.5 mm [0.375 x 0.375 x 0.5 in] tall. Cracking was observed in the walls when the composition blend was 50 percent each and in the columns when the blend was 40 percent Ti6Al4V and 60 percent Inconel 718. The graded composition deposits developed macroscopic cracks before the full transition from Ti6Al4V to Inconel 718 was achieved, as shown in Fig 2.13(b). The 2.5 in. wide wall developed a large crack when the composition reached a target mix of approximately 58 percent Ti6Al4V and 42 percent Inconel 718, based on powder feed control, with the crack propagating to complete fracture of the build during shutdown and removal, as shown in Fig 2.13(a).
There are some dedicated efforts to produce bi-metallic tooling from materials like H13 tool steel and copper for improved die performances. LAAM is an attractive option for creating complex shapes and thus manufacturing of conformal cooling channels seems a very probable solution with respect to increasing the heat transfer rate from a plastic injection mould. Pogson et al [95] experimented to explore the possibility of tool steel moulds with copper cooling channels. They produced several 10mm x 10mm block from 20/80 mixture of copper and H13 tool steel using laser re-melting process. Process parameters include 69-83 W of laser power and 150-200 mm/s of scan speed while allowing a beam overlap of 25%. The results were not encouraging due to high degree of micro-porosity, non-uniform mixing of two powders and appearance of cracks during cooling.

Imran et al [96] compared the performance of tool steel core pin against a DMD cladded copper alloy core pin in a high pressure die casting system. They prepared the composite pin by first depositing a 0.5 mm thick intermediate layer of 316L stainless steel and then cladding H13 tool steel upon the compatible steel base. Due to the presence of internal copper core and external hard and tough tool steel, the bi-metallic core pins outperformed the solid tool steel pins in terms of material build up and eventual failure of core pin due to sticking with the casting and restricting proper die opening.

Sahasrabudhe et al [97] investigated the fabrication of bimetallic structures using LENS in two set of experiments. In the first one, direct fabrication of compositionally graded structures of SS316L and Ti6Al4V alloys was attempted, while in the second approach, a NiCr intermediate bond layer was used for fabricating a bimetallic structure of SS410 and Ti6Al4V. In case of direct deposition, cracks were observed at the interface without delamination. The stainless steel and titanium alloys proved to be immiscible and upon deposition of more layers of Ti6Al4V, delamination occurred due to increase in thermal stress. While using NiCr alloy as the intermediate layer better bonding and more consistent set of properties like microhardness was achieved. EDS images showed clear delineation among the deposited layers. Some porosity was detected in the intermediate layer but it did not cause delamination.
2.6 Laser Assisted Surface Modification of Metallic parts

Laser assisted surface modification has been perhaps the most practiced form of LAAM. Since the high power laser beam is used to melt and deposit a given metallic powder, therefore, opportunity is ripe to apply an expensive but hard and tough alloy upon an inferior quality substrate for improved properties particularly wear and impact resistance. There are hundreds of publications on different types and techniques of laser aided surface engineering. In a way all the operations performed on LENS and DMD can be regarded as some form of surface engineering because it alters the surface of substrate. But the type of surface modification in which we are interested in this section pertains to the use of any major LAAM system for coating and repair operations prevalent in engineering and biomedical applications as an attempt to improve the quality and functionality of a given part.

Lallemand et al. [98] have used 1kW Nd:YaG laser to deposit Cobalt based powder on austenitic steel wires. Well bonded coating for 316L wires increased the hardness of surface from 380 HV (original wire) to around 550 HV (with coating). An interesting effect noted was the restoration or reduction in mechanical properties of cold drawn wires in the heat affected zone due to laser heating. The influence of laser cladding depends mainly on the temperature increase in the wire during the process, as well as on the cooling rate. The width of heat affected zone is directly linked to the laser power, the wire diameter and the laser scanning speed.

Krishna et al. [99] used LENS for fabrication of functionally graded Co–Cr–Mo coated Ti6Al4V alloy implants. The main purpose was to create a metallurgically sound interface between the two alloys, increase the surface hardness and improve wear resistance. Due to non-compatibility of cobalt and titanium and to prevent detrimental effects of mismatch in elastic moduli, thermal expansion coefficient and hardness between the two materials, functionally graded coating was applied having an intermediate layer with a gradual compositional variation in between the top wear resistant exterior and the interior substrate. Due to rapid cooling rates in laser processing, gradient structures did not show any intermetallic compounds of Ti and Co in the transition region or on the top surface. Elimination of intermetallic compounds is beneficial in terms of better wear and biocompatibility of these graded structures. Gradient coatings with 86% Co–Cr–Mo in the top surface showed an increase of 184% in surface hardness. The finer grain size, uniform microstructure and high hardness of laser-processed gradient coating is capable of providing better wear resistance compared with conventionally cast alloy of this type.

Balla et al. [100] have used laser engineered net shaping technique to apply compositionally and structurally graded coating of TiO₂ on Commercially Pure Ti substrates. The graded structures with varying concentrations of TiO₂ on the top surface were found to be non-toxic and biocompatible. These unitized structures with open porosity on substrate side and hard, low friction surface on the top surface can eliminate the need for multiple parts with different compositions for load-bearing
implants such as total hip prostheses. Gradient coatings with only 50% TiO$_2$ in top surface showed an increase of 4.74 times in the value of surface hardness. The finer grain size, uniform microstructure and high hardness of this laser-processed gradient coating can potentially provide excellent wear resistance.

Cu–30Ni alloy was successfully laser deposited on a rolled C71500 plate substrate by Direct Metal Deposition technology [101]. Deposition of 15 cross-hatched layers on the substrate was done according to a designed experimental combination which optimizes the combination of three major process parameters namely laser power, laser scanning speed and powder feed rate. Microhardness of the cladding was reduced with respect to substrate but exhibited a consistent value in the range 115 – 130 HVN.

Sun et al [102] used laser surface alloying technique to form wear resistant layers on 70MnV cast steel rolls with NiCr–Cr$_3$C$_2$ powders. Cast steel rolls were used as substrate on which NiCr–Cr$_3$C$_2$ powder was preplaced and deposited with the help of 4 kW CO$_2$ laser. For improvement in microhardness and sliding wear, best results were obtained for relatively high laser scanning speed and lesser thickness of deposited layer. In comparison to the substrate, microhardness increased from 326 HV to around 850 HV while wear resistance improved by a factor of 8.8.

Metallic implants, despite possessing biological and mechanical compatibility, need surface engineering or modification to improve their biological performance in order to moderate and control the response of surrounding living tissue. The surface modification can happen in two ways: (i) changing the surface morphology to increase the surface roughness which in turn enhances the surface area of implants for more cell attachment and induce a better mechanical interlocking of cells and implant [103] (ii) modification of chemical characteristics of implant surface coming in contact with living cells through “coating”. Coating is the most logical option to combine the best of metals and bio-ceramics. Thus by coating metallic implants with materials like calcium hydroxyapatite, the rate of implant fixation can be significantly improved with prolonged longevity [104].

Hydroxyapatites (HAp) are ceramics that have chemical similarities to the inorganic component of bone and tooth [105]. Laser assisted coating offers much improved adherence of coating to substrate in comparison to all other coating methods [106]. HAp cannot be directly melted and deposited on a substrate through a laser beam because it is transparent in the infra-red region. Cheng et al [107] developed a methodology for laser assisted HAp coating on titanium substrate using Nd:Yag laser which involves placing a powder precursor containing HAp and titanium on the substrate and using acetone as the temporary ‘adhesive’. Laser penetrates the HAp powders and heats the metal powders and substrate. Since HAp does not absorb laser, HAp powders will remain at low temperature before they are entrapped into metallic layer and form a strong metallurgical bonding with metallic substrate.
So the HAp is not directly coated but in an indirect way made to bond with the metallic substrate and entrapped within the metal matrix. Therefore, the coating microstructure consists of three layers, as shown in Fig 2.14, and includes - a metal layer, a molten zone (a dense composite coating), and a less dense HAp coating near surface, as shown in the SEM micrograph below. Authors reported good bonding strength and good mechanical properties for this coating.

![SEM micrograph of HAp coated Ti6Al4V substrate using an Nd:YAG laser with power of 100W and scanning velocity of 1 mm/s](image)

Figure 2.14: Cross-sectional SEM micrograph of HAp coated Ti6Al4V substrate using an Nd:YAG laser with power of 100W and scanning velocity of 1 mm/s [107]

Coating of bio-compatible metals one over other has also been reported to improve upon some deficiency of the base metal. Titanium is preferred for implant manufacturing but it has wear problems. Wear problems in turn lead to osteolysis and asceptic loosening. Krishna et al [108] demonstrated the functionally graded coating of Co–Cr–Mo upon Ti6Al4V alloy implants using Laser Engineering Net Shaping (LENS). The coating seeks to ensure a metallurgically sound interface between the two alloys with an increment in overall surface hardness, which should minimize the likelihood of localized Hertzian failure during implant service. Balla et al [109] reported coating of tantalum on titanium to increase the implant potential for enhanced/early biological fixation. System used was LENS equipped with 500 W Nd:YAG laser to deposit tantalum coatings of 1.5–2.0 mm thickness and 10 mm diameter on a substrate of 3 mm thick rolled, commercially pure Ti plates. With tantalum, complete melting was achieved at 450 W of laser power and the coatings exhibited good bonding between individual layers without any cracks or fusion defects.

### 2.7 Research Gaps, Problems and Challenges

The concept of direct part development from computer generated model using a combination of laser power and CAM control is indeed fascinating but not without problems and limitations. All the systems described in the previous section have a number of inherent constraints that restrict the biomedical and industrial applications to enjoy the luxury of "printing" accurately replicated organs and prototypes exactly replicating the CAD model. While discussing the existing research gaps and ensuing problems in commercializing the LAAM technologies, it must be appreciated that these systems have yet to reach their maximum potential and continuous improvement is reported with every passing year. That’s why we are careful in analysing the research gaps and have tried to base
In early works employing SLS for medical applications, the problems were reported to be of maintaining high spatial resolution and that of faceting and segmentation due to limitations in accuracy of manufacturing setup itself [82]. One extremely important research study to be carried out particularly for laser systems working with powder beds will be to look out for the energy absorption and penetration into the powder bed. Wang et al [53] reported that for a bed of WC-Co powder mixture the accumulated energy is absorbed in the powder bed surface up to a certain depth. A high concentration of the energy can be observed near the powder bed surface, and almost all the absorbed energy (~96%) is concentrated within a depth of only 0.4 mm. This high concentration leads to a large temperature gradient in depth and thus theoretically to a strong curling/warping tendency. If the process is not well controlled, cracks and delamination may occur. This is more probable for the first layer of powder; and the created pattern may fail to stick to the baseplate. There is also a significant need for research into improvement in the pore size and inter-connectivity. This in turn, depends on the size and shape of metal powder as well as the relationship between melting depths or track heights and laser parameters particularly peak power and power density [110].

Another important research problem to solve in laser cladded parts is high surface roughness usually found with bio-compatible metals like titanium and its alloys. The surface roughness depends on many factors: type of material, powder particle size, layer thickness, laser and scan parameters, scan strategy and surface post-treatment. Because of the stair effect due to the layer-wise production, surface roughness of a sloping plane depends on the sloping angle. The stair effect can be reduced by decreasing the layer thickness or by increasing the sloping angle [75]. Yasa et al [55] also cited low surface quality of SLM produced parts as one of the major drawbacks in addition to 1-2% micro-porosity introduced during laser processing. Meier and Haberland [111] while developing a porous filigree and solid impeller from austenitic steel powders on SLM noticed that only a moderate scanning speed (around 150 mm/s) leads to smooth surface structures whereas each laser track is noticeable clearly. An increase of the scanning speed causes a fragmentation of the scanning tracks and thus the surface is not cohesive. It seems that the fragmentation is initiated with higher scanning speeds on the vertical walls than in case of the horizontal surfaces. But it must be kept in view that the specific values are process dependent and cannot be applied overboard on every LAAM system. Like in DMD a scanning speed of 5.8 mm/s was considered optimum when depositing Inconel 718 powder on the substrate of same material [71]. The main difference is of course the laser power used which was 90W in case of SLM while 650 W for DMD cladding.
As laser assisted processes use thermal energy to melt the metallic powder and form the desired shapes, therefore, the specimen produced cannot be entirely independent of thermal defects and distortions. But for LAAM systems, encouraging factor is the presence of highly focused laser beam at lowest possible powers and presence of a very small heat affected zone (HAZ) [76]. In a study on configuration of HAZ for carbon steel AISI 1010 under a laser beam of intensity 130 J/mm, Knupfer and Moore [112] demonstrated the existence of two distinct regions, one close to the heat source upto a depth of approximately 0.25 mm and other inside the bulk material. First region shows carbon dissolution and phase transformation into martensite and bainite while second region exhibited plastic deformation and grain refinement due to re-crystallization.

Gu and Shen [113] conducted a well-researched investigation on the “balling” phenomena associated with the melting and fusion of two different types of 316L stainless steel powders under a focussed laser beam on DMLS system. The balling phenomena occurs when the molten material does not properly melt the underlying substrate due to the surface tension, which increases viscosity at solid-liquid interface and tends to spheroidise the liquid. This results in a rough and bead-shaped surface, obstructing a smooth layer deposition and decreasing the density of the produced part [114]. A schematic illustration of this very important potential defect in laser based part generation is presented in Fig. 2.15 that can affect surface quality, strength of cladding and bonding strength with the substrate. Their study revealed that if laser power is low enough to completely melt the metallic powder then coarse hardened balls are formed which make the structure inherently weak. Also if scanning speed is higher than optimum, then improper melting gives rise to micro-size balls. Their conclusion recommends increasing the input energy density, which can be realized by increasing laser power, lowering scan speed, or decreasing powder layer thickness, for successfully alleviating the balling phenomena.

For LENS and DMD systems which involve powder deposition through nozzles coaxial to the laser nozzle; it is a common problem that all the powder coming into the fusion zone does not necessarily contribute to the specimen growth. Thus there is a significantly large deposition of unmelted and
partially melted powder at unwanted locations. Occlusion of pores occurs due to this effect in porous structures resulting in need of post processing operations. By comparing the energy available in the melt pool and the energy required for the fusion of the whole 'powder feeding' arriving near the melt pool at a certain average temperature; one can estimate the proportion of efficient powder contributing to layers build up [115]. In coaxial powder injection laser cladding, the melt pool, under the irradiation of a laser beam, is free to deform due to surface tension and other forces. The growth and evolution of the liquid/gas interface, i.e. free surface of melt pool determines the quality of solidified geometry and surface roughness of the clad layer. A comprehensive research investigation is required to capture the interactive relations of the free surface geometry and the process physics, and to develop a precise method of free surface tracking [72]. In a laser-generated melt pool, large gradients of temperature and surface tension exist, and this generates waviness in the top surface of cladded parts as they gain in vertical height. This problem is especially severe in those technologies which do not work on a powder bed such as in LENS and DMD.

Another important research area in need of exploration is to understand the true nature of interaction between the number of variables involved in laser melting of powder, cladding and subsequent part building through layer deposition of one over the other. What is the composition of melt pool, how far its thermal effects propagate, what is the mass flow rate of melt pool and what is the nature of heat transfer among the layers and with the substrate? Experimental investigations alone may not be of much use in understanding these processes and the most difficult part is to take measurements in situ due to localised heating of laser beam [116]. Thus the only viable option is the numerical, mathematical and finite element modelling for a deeper understanding and adequate description of the system and the process. This area has been explored scarcely and can be considered as an area of research with huge potential for further development.
3.1 Introduction
As this research is based on mechanical characterization of high strength alloys particularly stainless and tool steels produced on DMD system, therefore, the review presented in this chapter forms the core discussion regarding published literature and ongoing research in the relevant area. Rapid manufacturing and laser assisted additive manufacturing (LAAM) techniques for producing metallic parts and components have been instrumental in revolutionizing and simultaneously challenging the conventional manufacturing strategies and applications [117]. Due to several constraints, these technologies are not seriously competitive with manufacturing techniques like forming, rolling, cold working and forging at mass production level; but at customized and low volume production level LAAM technologies offer distinct advantages of creating intricate shapes without tooling and with better control over material properties augmented by the availability of wide range of materials-to-process in powder form. Out of different LAAM systems, Direct Metal Deposition (DMD) possesses very good capability to work on high strength and tough austenitic stainless steels, H13 tool steel and biocompatible metals like titanium alloys and tantalum [67]. DMD provides the facility to produce complex geometrical features and functionally graded structures through CAD control. In a closed-loop DMD system a whole new class of optimally designed materials can be produced by depositing multiple materials at different parts of a single component with high precision [118]. It also provides the unique opportunity to manufacture large size parts that can be used as prototypes in numerous applications owing to its large bed size of 2m x 1m area [119]. These parts can be used in applications such as highly stressed machine components, plates under large compressive loads and barriers for restricting impact and shock loads due to physical projectiles or explosive forces.

The parts developed on DMD are actually made in a manner that involves creation of material and part simultaneously. Parts with complete shape and form are generated as a result of sintering of powdered material and subsequent layer by layer deposition over a “compatible” substrate through the assistance of high power and highly focussed laser beam. Unlike powder metallurgy, there are no compacting forces involved but there are process associated heat transfers which can result in generation of residual stresses and non-uniformities like micro-porosity and non-homogenous microstructures, irrespective of the materials cladded. Rapid heating and cooling cycles during laser scanning of the surface can cause severe microstructural changes such as phase transformation or dynamic recrystallization in the melt pool heat affected zone (HAZ). These effects tend to accumulate
when multiple layers are cladded one over the other. Thus the most essential need for any laser generated part is to ensure the mechanical and metallurgical character of structures generated in addition to consistency of mechanical properties vital for employment in engineering applications like those mentioned above [120,121].

In this chapter the published research on investigation of metallurgical and mechanical properties of laser generated high strength metallic alloys are analysed in detail particularly the evolution of micro-structure, micro-hardness across the cladded material and substrate, stress-strain behaviour, wear properties, fatigue strength and existence of residual stresses. More emphasis is laid on those investigations which involve steel alloys either as clad material or substrate, but cobalt based and nickel based alloys which have been extensively used for creating better surfaces through laser cladding are also considered. Since this research is based on DMD process, therefore, out of all the LAAM processes only those investigations are considered which are based on DMD or similar processes with some variations. Only a couple of works are mentioned in the whole chapter about SLM for the sake of comparison. High strain rate dynamic testing and dry sliding wear of DMD generated parts are the two major topics of this research, since these are the avenues which are still to be thoroughly investigated. Therefore, it was considered necessary that in the sections related to mechanical properties and wear characteristics of laser generated metallic structures a brief but broad based study taking into consideration conventional metallic alloys should be included to not only highlight the importance and potential applications of DMD cladded metallic structures but also to create a proper linkage with the research that has already been done for high strength alloys.

The discussion on micro-hardness of laser generated metallic structures is a ubiquitous topic that is sometimes linked up with micro-structure and sometimes with surface properties like wear. Therefore, the references to micro-hardness are provided in both the sections on mechanical properties and wear characteristics, but in our opinion hardness is more related to wear, that’s why more detailed discussion on hardness citing some hardness profiles within the laser cladded specimen will be found in the review of wear behaviour of laser generated alloys.

3.2 Metallurgical Characteristics and Mechanical Properties

For mechanical characterization of laser generated metallic structures, generated as claddings or complete parts, it is essential to find the values of strength and hardness. But this data is in turn related to the development and formation of microstructure, particularly within the cladding and at clad-substrate interface. For comparison with commercially available rolled and wrought high strength stainless steel alloys, it is important to have a brief review of their characteristics and range of applications that can be compatible with high strength alloy parts produced by DMD or similar laser based process.
Commercially available high strength steel alloys demonstrate some peculiar characteristics that forbid a straightforward transition towards ascertaining their design performance based on the characteristics and behaviour of carbon steels. The noticeable deviating instances are non-linear stress strain relationship in the elastic region, no sharply defined yield point and substantial strain hardening in the plastic region of deformation [122]. Although there is a serious shortage of research on the mechanical behaviour, particularly dynamic and high strain rate behaviour of laser cladded high strength steel alloys, yet many attempts have been made to develop the mechanical profile of stainless steels particularly austenitic type manufactured through some sintering process. Chawla and Deng [123] mixed blended and binder-treated powders of Fe–0.85Mo, 2 wt.% Ni, and 0.6 wt.% graphite using a proprietary process developed by Hoeganaes Corporation. At a density of 7.4 g/cm³ the Fe-Mo-Ni sintered steels exhibited a Young’s modulus of 172 GPa and an ultimate tensile strength of 750 MPa. Dewidar [124] sintered 316L stainless steel powder to form compacts at pressure levels of 150, 250 and 350 MPa. Best values for yield strength in compression was measured to be approximately 130 MPa for samples obtained at sintering temperature and pressure of 1300 °C and 350 MPa respectively in nitrogen atmosphere. The compressive stress-strain curve for sintered 316L in Argon atmosphere revealed three distinct regions of deformation typical for high strength steel alloys, namely linear elastic, plastic collapse, and densification due to work hardening.

High strength stainless steel alloys in the form of sandwiched structures are considered a powerful candidate for bearing impact loads without fracture and efficiently absorbing the energy of explosions and blasts [125, 126]. These sandwiched plate structures have been proven as good as solid plates for resisting impact forces of blasting, and may be developed from repetition of core geometries like pyramidal truss, square honeycomb and folded (or corrugated) plate [127]. Rathbun et al [128] tested the dynamic performance of solid and sandwiched beams made from 304 SS under impact velocities of 140-470 m/s. Their results, later supported by the investigations of Radford et al [129], concluded that wherever the loading is dominated by bending instead of stretching like impulses representative of those expected from nearby shocks, the metallic sandwich structures outperform monolithic solids of equivalent weight. And the best core topologies that provide simultaneous crushing and stretching resistance include square honeycombs.

Microstructural and strength related properties of a material are closely related as described and elucidated by Ashby [130, 131]. There are numerous publications that take into account both these characteristics for laser generated high strength metallic alloys. DeLima and Sankare [132] used mono-mode laser to deposit 316L on steel substrate using coaxial nozzles in an argon environment. The cladded structures were found to be fully austenitic by XRD analysis. The phase growth in the region between clad and substrate was epitaxial fine dendritic. In the areas where substrate re-melted,
epitaxy was lost and equi-axed grains were observed in the micro-structure. As each clad layer was deposited over the previous one, nucleation of cracks at the pores between layers was observed that could lead to crack growth and micro-porosity inside the cladding. Uniaxial tensile strength was also measured for different cladded samples. Values differ with change in cladding parameters. Best values obtained were 261 MPa for yield strength, 539 MPa for ultimate strength and only 3.7 GPa for Young’s modulus. The value of Young’s modulus was very low in comparison to 180-200 GPa reported for commercial grades of 316L stainless steel [133].

Majumder et al [134] used diode laser to clad multiple layers of 316L SS on mild steel substrate. The process parameters used, which they considered optimum, were laser power density as 0.073 kW/mm², scan speed 5 mm/s and powder rate to be 203 mg/s. The microstructure of laser cladded steel was found to be predominantly cellular with an average grain size of 10 µm and appears different from the usual austenitic structure. Microhardness values vary for different values of power density, and authors concluded that average microhardness of the cladded layers decreases with increase in applied power density. This effect can be attributed to coarsening of grains within the final microstructure.

Zhang et al [71] used DMD to deposit Inconel 718 superalloy on a heat treated substrate also of Inconel 718. Laser parameters selected were 650 W beam power with 6.7 mm/s scan speed. SEM micrographs illustrated in Fig 3.1 indicate some interesting features. In Figs 3.1(b) and (c) the columnar dendrites formed at crystallographic orientations <111> and <200> exhibited epitaxial growth from the substrate. Fig 3.1(b) shows that there is no heat affected zone within the cladding and at interface due to a positive temperature gradient from top to bottom. Fig 3.1(d) highlights fine equiaxial grains formed at the top layer, which may be due to the effect of reduction in temperature gradient of liquid as compared to the solidification rate of melt pool.
Tabernero et al [135] also investigated the mechanical and metallurgical properties of Inconel 718 upon Inconel 718 substrate using a high power diode laser with the laser parameters as: laser power 1100 W, scan speed 700 mm/min and powder feed rate as 5.2 g/min. Two different cladding strategies were adopted as illustrated in Fig 3.2.

Results of tensile tests reveal significantly lesser values of ultimate strength as compared to wrought material. Another noticeable fact is the reduction of tensile strength for spiral cladding in comparison to zigzag cladding strategy as shown in Fig 3.3. Stress strain curves of Fig 3.3(a) indicate some early failure which mainly occurred in laser deposited material. Fig 3.3 (b) illustrates a 55% reduction in strength for spirally cladded specimen when tested in a state without any bonded substrate. Precipitation hardening of the specimen induced improvement in the mechanical properties because of homogenisation of the microstructure of the deposited material. The tests on treated specimens showed higher tensile strength and ductility increments as much as 100%.
Bhattacharya et al [136] investigated the laser cladding of cupro-nickel alloys which have found good use in marine engineering due to their good corrosion resistance and anti-fouling properties in seawater. A DMD system with CO$_2$ laser was used to perform the cladding on rolled C71500 plate substrates. It was observed that layered Cu–38Ni alloy specimen microstructures primarily consisted of columnar dendrites, growing almost perpendicular to the interfaces, with subsequent transition from columnar to equiaxed structure at distances away from interface toward top surface of the layers. This evolution of microstructure is very similar to what is described by Zhang [71] and previously illustrated in Fig. 3.1. Electron Back Scatter Diffraction (EBSD) analysis indicates that in every cladded layer, grains have certain orientations with respect to layer boundaries, as shown in Fig 3.4(a). It is also observed that the laser formed specimen has a narrow grain size distribution (5–140 µm), with 90 % of the grains below 100 µm as illustrated by the histogram in Fig 3.4(c). It can also be observed in Fig 3.4(b) that there is a difference in sizes of grains, which are found at the layer boundaries in comparison to grain sizes inside the cladded layers. The conspicuous fact is that in the region where layers consist of larger grains some of which run across the layers with the result that layer boundaries are distinctly decorated with very small-size grains.
Bhattachraya et al [66] investigated the microstructural evolution and micro-hardness of AISI 4340 medium carbon, high strength low alloy steel with rolled mild steel plate used as substrate. In all, 8 layers were deposited using a DMD system. First layer, generally called the alloy or dilution layer, was deposited with 800 W laser power, 3 g/min powder feed rate and a scan speed of 500 mm/min. Subsequent 7 layers were cladded at 500 W, 5 g/min and 450 mm/min. Under equilibrium processing conditions, the microstructure of AISI 4340 should contain ferrite and pearlite, but very high cooling rates \( (10^5 – 10^7 \text{ K/s}) \) typical of a DMD process pushes the microstructure of DMD cladded steel alloys into martensitic range.

![Figure 3.5: Microstructural evolution in DMD cladded layers of AISI 4340 medium carbon steel (a) Interface layer (b) Alloy layer (c) Top layer [66]](image)

As can be observed in Figs 3.5(a) & (b) that interface and alloy layers have some fine tempered martensite with relatively larger grain size. Fig 3.5(c) clearly illustrates the needle shaped martensitic structures which gets tempered or rounded towards the bottom layers. SEM and TEM investigations reveal mainly the presence of ferrite and martensite phases in the clad with few occurrences of cementite and negligible presence of austenite.

Choi and Chang [137] used a DMD system with feedback control to deposit H13 tool steel on a 6.4 mm thick AISI 1018 steel substrate. The main idea of feedback control is to prevent over-deposition by using an adaptive control capability to quickly reduce laser power as soon as over-deposition is detected. Cladding was done with CO\(_2\) laser at 1.75 kW and powder feed rate of 5-11 g/min. As illustrated in Fig 3.6, the columnar dendrite grows mainly along the deposition direction and their direction is usually perpendicular to the clad boundary, along the direction of higher temperature gradient. The two micrographs also show slightly different morphology. It was observed that average secondary dendrite arm spacing (-2 μm) near the top surface (Fig. 3.6(a)) is shorter than that in the lower part of cladding due to the morphology changing annealing effect owing to cyclic heat flow from the upper layers.
A 2 kW CO\textsubscript{2} laser with a system similar to DMD was used by Zhang et al [138] to deposit 316L SS powder on 10 mm thick carbon steel substrate. The process parameters were 1000W laser beam power, 3 mm/s scan speed and 6 g/min powder feed rate. The thickness of each cladded layer was found to be uniform (about 0.5 mm) with the presence of a thin remelted layer between the layers. Fig. 3.7 shows the microstructure of laser direct deposited 316L with the continuous dendritic structure conspicuous in Fig 3.7 (b). Strong evidence of epitaxial growth off the prior solid interface can be observed within each deposited layer and highlighted in Fig 3.7(a). It should be noted that Fig. 3.7(b) is a magnified view of Fig. 3.7(a), displaying the typical dendritic structure.

Regarding mechanical properties of laser cladded 316L specimen, a slight directional difference was noted in the measurement of tensile strength. In a direction perpendicular to layer orientation, ultimate tensile strength was found to be 626 MPa while in the parallel-to-layer direction the value was
recorded as 694 MPa. The specimen also exhibited appreciable ductility with an average value of 40% elongation before fracture.

Wang et al [139] fabricated a thick plate of size 150mm×75mm×20mm from martensitic 1Cr12Ni2WMoVNb steel using laser melting deposition (LMD) process which is similar to LENS. The substrate used was a 15 mm thick A3 steel which is grounded and degreased before laser deposition. Due to the rapid cooling rate, the newly laser deposited 1Cr12Ni2WMoVNb steel exhibited a microstructure consisting of martensite, retained austenite and inter-dendritic phases. Morphologically the laser deposited steel has some banded feature between adjacent layers (about 200µm). The banded feature is referred to as the interlayer heat-affected zone (ILHAZ), and it is associated with the thermal effects during laser melting deposition process. The room-temperature tensile properties of the laser deposited 1Cr12Ni2WMoVNb steel. The ultimate tensile strength reaches 1223 MPa, which is comparable to the wrought bar, but the elongation and the reduction in area are only half of the wrought bar. The evaluated microhardness of the laser deposited 1Cr12Ni2WMoVNb steel was non-uniform. The average microhardness (with a test load of 200 g) in a single deposition layer was 424 HV, which decreased to 401HV in the interlayer heat-affected zone.

Ravi et al [140] used direct laser fabrication or LENS system to deposit SC420 stainless steels on a 12 mm thick stainless steel substrate. To consolidate any internal pores and relieve any residual stresses within the build, the as-deposited SC420 samples were treated using a Hot Isostatic Pressing (HIP) under argon gas at 1100°C and at a pressure of 100 MPa for 2 h. SEM micrographs exhibit a tempered martensite microstructure in both as-deposited and HIPped samples, as illustrated in Fig. 3.8 (a) & (b). Along with the martensite phase, fine carbides and ferrite were also observed along the grain boundaries. HIP resulted in an increase in the amount of ferrite. In the as-deposited samples due to high cooling rates fine carbides were produced in the grains, but high temperature HIPping dissolved the fine carbides into the matrix. But at the same time slow cooling rate during HIPping process led to the precipitation of the carbide particles along the grain boundaries and their coarsening as well. Both conditions were examined using X-ray diffraction, which confirmed the presence of martensite, ferrite and the iron and chromium rich carbide.
Figure 3.8: Microstructure of LENS deposited SC420 stainless steel (a) as-deposited condition (mainly martensitic structure), and (d) After HIP operation (with more extensive carbide formation) [140].

Mechanical properties like yield and tensile strength were also examined for the SC420 martensitic stainless steel specimen in as-deposited and HIPped conditions under two sample configurations. One set of specimen was cladded with a smaller cross-section (parallel to the substrate) of 13 x 13 mm and the other set with a larger cross-section of 13 x 70 mm, as shown in Fig 3.9(c). The main ideas was to observe the effect of possible annealing in the smaller cross-section due to the fact that there is lesser time available to reject the heat and cool down as compared to the specimen in which laser has to traverse a distance of 70 mm before returning for the next pass. But this difference in cladding pattern was only reflected in change of elongation, 12-15% for the smaller section as compared to 2-4% for the larger section, without exhibiting any change in the yield and tensile strength. HIPping resulted in 6 to 9% improvement in elongation and tensile strength while also reducing scatter in the ductility. Figs 3.9 (a) & (b) illustrate the mechanical properties in terms of stress-strain curves for horizontal and vertical cladded specimen in as-deposited and HIPped conditions. Lesser elongation before fracturing is clearly visible in graphs of Fig 3.9(b) for the specimen tested in as-deposited condition.

![Stress-strain curves for SC420 steel specimen prepared by LENS](image)

**Figure 3.9:** Stress strain curves for SC420 steel specimen prepared by LENS (a) Horizontal specimen with less annealing (b) Vertical specimen with more annealing effect (c) Layout of specimens [140]

### 3.3 Wear Behaviour of Laser Cladded Alloys

In many applications metal parts undergo rubbing against each other or harder ceramic surfaces and may experience high rates of wear and rapid surface degradation. Therefore, with the expanding range of applications and more stringent demand on wear performance parameters, at least for the last 50
years, study of friction co-efficient, temperature rise during wear and amount of metal worn out with contact load and sliding speed for metals and particularly steel alloys, has been a continuous research endeavour. Generally accepted Coulomb’s friction formula is sufficiently inadequate from the research point of view and fails to take into account the subtle interactions and variations present in the variables involved in wear. Bayer [141] has highlighted the importance of three major aspects in studying wear behaviour. These are (i) nature of contact, which includes type, condition and mechanical properties of materials in addition to the presence of hard particles at the interface, (ii) type of motion associated with the contact such as rolling or sliding, and (iii) the environment surrounding the contact like dry or lubricated. Archard developed a set of wear relationships [142] and subsequently surface profile based contact model [143], which highlighted the importance of surface topography in understanding wear behaviour involving elastic and plastic contact surface deformations during rubbing and sliding. According to this theory the volume worn ($V$) during sliding wear obeys a relation that involves sliding distance ($S$) and area of contact ($A$) and defined by the eq. (3.1) given as:

$$\frac{V}{S} = \frac{1}{3} KA$$

(3.1)

where $K$ is a constant called wear coefficient. Equation (3.1) can also be written as:

$$V = k S L$$

(3.2)

Note that $k$ is termed as “specific wear rate” and is essentially different from $K$. Here $L$ is the normal load between the contacting surfaces.

The study of sliding wear is complex and must be investigated as a ‘system behaviour’ in which a number of factors like adhesion of asperities, wear debris and sub-surface crack initiation and growth simultaneously play their part in influencing the quantities like wear volume, nature of wear and co-efficient of friction [144]. Lim et al [145] has clearly demonstrated that coefficient of friction for steels exhibits significant variation in stark contrast to the notion of ‘$\mu$’ having a constant value for a given pair of materials as mentioned usually in textbooks. Another noticeable fact in their research was that at sliding velocities below 1 m/s, the value of ‘$\mu$’ was greater than one which substantially reduced at velocities above 1 m/s. The possible reason may be that under dry conditions at low velocities there is direct metal-to-metal contact and asperities create adhering patches, like tiny spot welds, which must be plastically sheared or pulled apart. Yoon et al [146] investigated the variation in frictional characteristics of AISI 52100 steel on itself in response to changes of the dynamic parameters of a pin-on-disk apparatus. The experimental results had generally demonstrated that the coefficient of friction increased with the normal stiffness of evaluating system. It exhibited the highest
value for spring loading and the lowest for pneumatic loading while being intermediate for dead weight loading. Scherge et al [147] demonstrated the strong influence of dynamic variables on friction and wear behaviour of metals. Main reason is the variation in overall energy of the typical system like pin on disc during wear.

Laser cladding has already been used as an effective technique for enhancing the wear properties of metals including steels through coating. In fact due to their unique characteristics laser have been considered and applied as a preferred source to improve the surface qualities of steels in terms of hardness, toughness and wear properties. Chen and Xue [148] used 500 W pulsed Nd:YAG laser to deposit high vanadium tool steels on low cost H13 tool steel substrate. Microhardness was observed to be increasing with the percentage of vanadium in the cladded powder. Fig 3.10 illustrates the difference in as-clad hardness of 9% vanadium and 15% vanadium samples. The values improved significantly after double tempering at 540-550 °C.

Metal-matrix composites (MMC) contain a hard phase like tungsten or chromium carbide in a relatively softer matrix of Inconel or Stellite and are considered as useful materials for improving abrasion resistance. Nurminen et al [149] applied a set of MMCs with different carbide phases on a low carbon steel substrate using Nd:YAG laser operating at a power of 3 kW and a scan speed of 1000 mm/min. For dry abrasion test against a rubber wheel, Inconel 625 in the phase of chromium carbide showed the lowest wear rate followed by M2 tool steel in vanadium carbide. St-Georges [150] compared the wear resistance among laser cladding and hardfacing for metal-matrix composites containing NiCr matrix with WC particles coated on low carbon steel substrates using 4 kW Nd:YaG laser. Their test data showed that Ni-WC laser cladded samples exhibited much better wear resistance in terms of material removal in an abrasion test than hard-faced coatings and D2 tool steel. Janaki Ram [151] reported poor wear resistance of laser cladded coating when performing the similar
abrasion test for laser deposited CoCrMo powder on a wrought CoCrMo substrate. The cladding was performed with a LENS system at a laser power of 285 W. Despite comparable hardness of cladded and deposited materials, cladding showed a material removal of 6.45 g in comparison to 2.7 g for wrought material during abrasion test. Authors attributed the inferior wear resistance of laser generated material to the presence of the carbide phase as irregularly shaped, very thin, long, interconnected particles, as a thin continuous network in the deposit microstructure which is susceptible to be easily removed during the dry sand/rubber wheel wear test.

Chen et al [152] applied Nickel based self-fluxing alloy coatings reinforced with hard cermet phase (WC-12%Co) for optimum wear resistance on 0.45% carbon steel substrates employing CO2 laser operating at 2 kW. The measured microhardness was in the range of 700 - 900 HV in the coating zone, even exceeding 1200 HV in WC region, and in the range of 350 - 400 HV in the substrate. When tested by the ball-on-disc metallic dry sliding wear tester, mould steel substrate experienced a specific wear rate of $12.982 \times 10^{-6}$ mm$^3$/N.m in comparison to $0.054 \times 10^{-6}$ mm$^3$/N.m for the composite laser based coating. Masanta et al [153] studied the possible effects of the choice of substrate material on the tribological performance for TiB$_2$–TiC–Al$_2$O$_3$ composite coating deposited by laser cladding process. The average micro-hardness values of the coatings developed on AISI 1020 steel and AISI 304 stainless steel substrates were measured in the range of 1400–2200 HV and 1600–1900 HV respectively. The dry sliding wear behaviour of the coating was assessed by using ball-on-disc tribometer at various normal loads ranging from 9.8 to 39.2N. Coefficient of friction did not show any significant dependence on the choice of substrate material. Wear rate varied between $1.5 \times 10^{-7}$ g/N.m and $2.6 \times 10^{-7}$ g/N.m while exhibiting significant dependence on the laser scan speed. Difference in the values of thermal conductivity of the two steel substrate materials is considered to have a significant effect on the microstructure and tribological performance of the coatings.

Sun et al [154] have used 4 kW CO2 laser to apply NiCr–Cr$_3$C$_2$ alloy powder on 70MnV cast steel roll that was used as the substrate. The thickness of deposited layer was reported to vary from 0.48 mm – 0.58 mm depending on laser scanning speed. The samples with highest scanning speed and the least thickness exhibited most significant improvement in wear properties as compared to those of the substrate. Navas et al [155] employed laser cladding for depositing cobalt based Tebaloy-800 powder on AISI 304 stainless steel substrate. Cladding was done with a CO$_2$ laser at 1800 W, 240 mm/min and powder feed rate between 10-15 g/min. Sliding wear tests were performed on the cladding using ball-on-disc configuration against AISI 52100 chromium steel balls of 6mm diameter and 850 HV and with block-on-ring arrangement at 15, 30 and 40 N against a hardened AISI 1043 steel ring (480 HV) while maintaining a constant sliding speed of 1.31 m/s.
The microhardness and wear results shown in Fig 3.11 illustrate very significant improvements in every aspect and thus highlighting the practical importance of surface modification through laser cladding in improving the wear characteristics of low cost general purpose metals. But there is one aspect, which indicated degradation and that is the value of co-efficient of friction, which showed a value of 0.447 for the substrate and 0.71 for the cladding after stabilization and reaching the steady state after prolonged sliding.

3.4 Residual Stresses in Metallic Laser Claddings

The induction of residual stresses due to focussed and high power laser beams performing cladding and coating operations has been a matter of concern particularly for high strength metallic alloys. Residual stress fields may superimpose on externally applied stresses and may significantly alter the nature of overall stress pattern [156]. In DMD processes determination of residual stresses has been an important part of characterization due to high power of laser beam (>500 W and may reach up to 3 kW), localized formation of melt pool, thermal mismatching between cladded metal and substrate material and rapid solidification rate of molten metal [70, 66, 136]. Parts produced by similar LENS process are also not immune to the development of residual stress profile as a result of laser cladding. Hofmeister et al [157] employed an empirical scaling approach to correlate the cooling rate with the length of melt pool in a LENS process. A wide range of laser power and scan speeds were taken into account and the logarithm of the cooling rate at the solid-liquid interface was plotted against the logarithm of the length of the melt pool. Fig 3.12 illustrates the plot for 316SS and H13TS and shows that regardless of the power, traverse velocity, and conduction path, the square of length of the molten pool largely determines the cooling rate.

Figure 3.11: Microhardness and wear improvement for laser cladding of Tribaloy-800 powder on AISI304 substrate (a) Microhardness across the substrate and cladded section (b) Comparison of wear rates for clad and substrate (c) Volume loss against sliding distance for the Tribaloy-800 cladding [155]
Rangaswamy et al [158, 159] discussed the residual stress distribution in 316L SS and Inconel 718 parts of simple geometries formed by LENS system. The residual stresses were measured by neutron diffraction and contour methods. It was found that for 316L specimen, the residual stresses were primarily uniaxial i.e. along the (Z) direction which is normal to cladding or along the height of cladding. In Fig 3.13 the results of measurement for residual stresses by contour and neutron diffraction method are shown. The maximum tensile or compressive stresses measured were within the tensile yield stress level for stainless steel i.e. in the vicinity of 445 MPa. The important feature is that stresses are tensile near the edges of cladded pillar but changed to compressive near the centre. Such a state of residual stresses with compressive stresses within and tensile stresses at the boundaries points towards the situation when the edges remained at higher temperatures than the interior, thus constraining the shrinkage of exterior surface by the cooler interior.

Finnie et al [160] used 1.5 kW CO₂ laser at 600 mm/min and a powder feed rate of 6 g/min to deposit Stellite F powder on rectangular AISI 304 stainless steel substrate. The cladding was done without preheating of substrate and also when the substrate was preheated to 800°C through induction

Figure 3.12: The measured cooling rates at the first local minimum behind the melt pool versus the project melt pool length for 316SS and H13 [157]

Figure 3.13: Residual stress measurement for a 316L pillar made by LENS (a) Picture illustrating sizes and measurement axes (b) Plots of residual stresses as measured by neutron diffraction and contour methods along X and Y axes [159]
heating. Residual stresses were then measured for both types of claddings by measuring strain at a selected location while a cut of progressively increasing depth is introduced into the part. The expected error with this method in the measured peak stress is estimated to be about 6% while that in the cladding can be about 10%. Fig 3.14 shows the measured value compared with those predicted by numerical modelling.

The most noticeable aspect is significant reduction in residual stresses due to pre-heating and reversal of residual stresses within the cladding from dominantly tensile to totally compressive as a result of preheating. Suarez et al [161] also used CO₂ laser at 5 mm/s and 2.2 g/min to deposit Stellite B powder on AISI 304 austenitic steel plates. They used ED-XRD stress measurement technique on a Synchrotron setup to measure the residual stresses on samples with and without preheating. The planar residual stresses in the cladding were found to be tensile in the range of 250-300 MPa which gradually changed to compressive at the interface and upto a depth of 5 mm into the substrate. The preheating alleviates the tensile stress along the height of cladding to around 20 MPa but the stress values inside the substrate almost remain unchanged.

3.5 Fatigue Behaviour
Most metallic parts are subjected to repeated or cyclic loading causing fatigue and an early failure. Research data on crack propagation due to cyclic loading has been available for a large number of high strength steel compositions. Barsom et al [162] investigated and developed a crack growth model for high strength construction and maraging steels. According to them the primary factor affecting crack growth rate is the stress intensity factor (Kᵢ) with mechanical properties, composition and specimen geometry exhibiting minor effects. High strength steels exhibit a peculiar behaviour showing fatigue crack initiation at the surface when subjected to fatigue at high-stress amplitude and low cycles, while in very high cycle fatigue region (VHCF) exceeding $10^7$ cycles fatigue failure of some high-strength and case-hardened steels occurs at small internal defects in the subsurface zone
For high strength steels, the fatigue life under very high cycle fatigue loading exists in three phases which include (i) formation of fine-granular-area (FGA) at some subsurface inclusion (ii) enlargement of FGA into a fish-eye (iii) fish eye to the critical crack size. Hong et al [164] have calculated that the crack initiation life \( N_i \) contributed by the formation of FGA is less than 70% when the total fatigue life \( N_f \) is below \( 10^6 \) cycles, but when total fatigue life exceeds beyond \( 10^7 \) cycles, i.e. in VHCF regime, the value of \( N_i/N_f \) is larger than 95%.

Since laser cladded steel parts have been employed in various industrial application, therefore, it is extremely important to thoroughly ascertain the high and low cycle fatigue behaviour of such parts to avoid catastrophic failures. Niederhauser and Karlson [165] studied the fatigue behaviour of laser cladded cobalt-chromium alloy having 25.5 wt.% chromium on a 200 x 100 x 25 mm thick plain carbon steel plate with 0.51% carbon. Two plates were prepared with clad heights of 2 and 1.6 mm and the fatigue tests were carried out with load varying as sawtooth wave shape applied at a constant strain rate of \( 10^{-2} \) s\(^{-1}\). The test strain amplitudes were chosen between 0.3% and 1%. For stress amplitudes of 700 MPa the cycles to failure were merely 500, but increased to around 15000 when the stress amplitude is reduced to 475 MPa. There were two important observations from these fatigue tests. First one relates to remarkably low scatter in fatigue behaviour despite the complex metallurgical compositions produced due to dilution and HAZ in the interface layer between substrate and cladding. The second one pertains to non-zero mean stresses showing up due to residual stresses created during laser cladding and remain present throughout the lifetime for low strain amplitudes. But at strains above 0.6% these residual stresses vanish and, hence, the mean stress falls towards zero.

During mechanical or thermal fatigue, the main concern is stability and integrity of the interface layer in the laser cladded parts. The main reason is the high degree of variability and large number of parameters contributing to the interface layer, strength of which determines the bonding quality of clad to substrate. Ganesh et al [166] used a CO\(_2\) laser with powder delivery system and a CNC table control to deposit layers of pure Stellite 21 powder and Stellite 21 mixed with 316L on a 316L SS substrate. The specimens were tested by rotating bending fatigue (RBF) method while being subjected to identical sinusoidal loading. There was not much difference in the number of cycles to failure for pure and mixed cladding which on the average was observed to be 1 million cycles. Fig 3.15 shows the structure of failed surfaces for the two types of claddings. Most noticeable fact is that in every case, the fatigue failure initiated from the substrate and eventually failure occurred at the interface. In Fig 3.15(a) the arrow “O” indicates the initiation region for fatigue failure under RBF test.
In repeated bending tests on Stellite 21 laser cladded on X5CrNi18-10 austenitic stainless steels, Koehler et al [167] found cycles to failure to be from 0.5 – 2 million for repeated bending tests (R=0) and 2–7 million cycles for alternating stress tests (R=-1), where ‘R’ is stress ratio. For X5CrNi18-10 base material laser cladded with Stellite 21, compressive residual stresses were found in the cladding and at the interface of cladding and base material. The fatigue strength dropped by 2.6% under alternating bending load fatigue test (R=0) and by 9.3% for repeated load investigations (R=-1). The noticeable departure from the investigations of Ganesh et al [166] was that the crack initiated from the surface of cladding which the authors attributed to the negative influence of residual stresses induced during laser cladding process [168].

Spierings et al [169] investigated the fatigue characteristics of SLM built 316L specimen. The process parameters include 103 W laser power, 425 mm/s scan speed and thickness of each deposited layer to be 30 µm. Fatigue test were conducted to ASTM E466 standard testing within the high cycle fatigue (HCF) range (N >=10^4) at R = 0.1 using a 50 Hz sinusoidal load under control to a maximum of 10^7 cycles. The fatigue limit was measured to be 269 MPa which is almost equal to the values found for wrought 316L specimens. Fig 3.16 shows the full range of S-N curves for 316L SLM produced parts with surface finishes ranging from R_a = 5 to 50 µm. For the investigated 316L material, there seems to be no significant effect of surface treatment on fatigue lifetime but a strong influence does exist on the fatigue limit as indicated by a stress amplitude bandwidth of approximately 150 MPa.
Hutasoit et al [170] investigated the relationship of fatigue life and nature of residual stresses by depositing Stellite 6 powder on AISI 4130 steel substrate. Fatigue tests were performed on cantilever type rotary bending test rig, under load of 100 N, 150 N and 200 N with loading ratio $R=-1$. An artificial notch was created on the surface of laser clad region of fatigue test specimen which provides the crack initiation site for further crack propagation during fatigue loading. Therefore, fatigue life investigation was based on crack propagation stage until the specimens fractured. In comparison to the AISI4130 itself the fatigue life of laser cladded Stellite 6 specimen exhibited a 50% reduction, a fact that is attributed to the existence of tensile residual stresses within the cladding. Another important investigated fact implies that if for the same size structure, thickness of cladding is increased then fatigue life will reduce due to increased area of higher tensile residual stresses in axial stress direction.

3.6 Research Gaps

The research aspect of investigating mechanical and material properties of metallic structures is very wide and diversified at the same time. We cannot accept the properties of laser generated high strength alloys to be the same as cast and wrought metallic alloys. Therefore, this whole gap is available as a potentially fruitful research opportunity. The research in this direction is definitely advancing and bringing forth many valuable results but still there are a number of avenues left to be thoroughly investigated. It is quite evident from all the published research that laser cladding is mainly considered as closer to welding, joining and coating applications instead of part manufacturing that could be beneficial in industrial applications. There has been very good development in the fields of biomedical applications and high tech aerospace parts but in general LAAM technologies have been treated at the periphery and not as a mainstream production technology. Keeping in view the limitations discussed earlier, this approach is not totally irrational but there are opportunities to alter this trend, particularly with high powered and large machine bed systems like DMD.
The most apparent gap exists in the dynamic testing of laser generated high strength alloy parts at high strain rates. The reason may be that since its inception, laser generated metallic parts have been mainly used in biomedical applications where such requirements do not arise or these have been mainly applied as coatings. This research is an attempt to open this niche and the investigations have already been acknowledged as the published research on high strain rate compressive dynamic behaviour of DMD cladded 316L SS and H13 TS parts [171].

In metallurgical investigations, major emphasis is on the investigation of microstructure which lies at the core of other mechanical properties like strength and hardness. But there is a serious lack in the amount and quality of research that provide an inter-relationship of observed micro-structure and the measured mechanical properties. The laser based systems actually produce parts from powders and the micro-structure affecting all other properties is so heavily dependent on the combination of laser process parameters, that it becomes a herculean task to determine the type of micro-structure that will be created by a given set of parameters. So this relationship is yet to be fully established. Another problem is the quality of bonding at the interface of clad and substrate. This problem will become more significant when composite specimen is formed by combining substrate of low cost and inferior mechanical properties with high strength cladding. How much the property and material mismatch will occur is yet to be investigated for industrial applications.

The research work on investigating the sliding wear behaviour of laser cladded surfaces created on stronger and tougher metallic substrates also needs to be thoroughly established. The practical situations governing wear applications are so varying that even for commercially available wrought, rolled and cast metallic alloys, a significant gap exists to completely establish tribological behaviour. It is only recently that wear investigations on laser cladded steel alloys are emerging and there is a long way to go. Particularly important avenues to be investigated are the variability in stability of laser cladding against prolonged and severe wear and the bonding with substrate against large tangential frictional forces.

Induction of residual stresses is an unavoidable phenomenon in laser based creation of metallic structures, a problem which is more pronounced in high power LENS and DMD systems. Large number of research papers assume the state of residual stress to be planar throughout (along the thickness or Z-direction), but there should be a better picture of residual stress formation in three principal directions particularly at the clad-substrate interface, where there is always a large discontinuity in values and pattern of residual stresses. There is also an existence of lesser investigation of the state of residual stresses based on the materials and process parameters controlling the outcome of LAAM system.
4.1 Introduction

Direct Metal Deposition is an additive manufacturing (AM) technique in which metallic structures can be built layer by layer through laser melting of metallic powders and subsequent cladding on a substrate material [20]. Therefore, DMD cladding offers building up of complex metallic structures with controlled material, micro-structural and mechanical properties directly through CAD models, without additional tooling [68]. Like all laser based additive manufacturing processes, quality of DMD cladded parts also depend on the crucial parameters like laser power, laser scanning speed and powder feed rate. For acquiring proper insight into the dynamics of complex physical and chemical processes involved and ascertaining the quality of part or coating produced with confidence, it is therefore mandatory to perform a thorough analysis of the influence of laser process parameters and to develop appropriate numerical models of the process. This research attempts to enhance the application area of DMD by optimizing process parameters for creating parts with at least 5 mm height with different geometries and profiles from high strength alloys.

Yadroitsev et al [172] discussed the importance of range of characteristics for three important process variables for SLM but that is very much relevant to DMD process as well. These include (i) Powder in terms of composition, particle size and optical and heat transfer characteristics (ii) Laser in view of power, spot size, beam spatial distribution and scanning speed; and (iii) Manufacturing strategy in relation to orientation and positioning of beam with respect to substrate. In a laser based DMD process it is extremely important to ensure that parts are fully dense with minimum microporosity while at the same time rigidly bonded to the substrate. This demands proper melting of powder stream and appropriate overlapping of consecutive tracks side by side to form a single layer of deposition. When a typical cross-section of a laser deposited track is analysed as shown in Fig 4.1 (a) & (b) then a number of important parameters come under consideration in addition to a number of affected regions [173, 174]. De Oliveira et al [174] derived two main parameters from these sectional values which are important for the track characterization. First one is the “Dilution” parameter which is defined as the molten area or area of melt pool ($A_m$) divided by the total area which comprises the clad area ($A_c$) plus the molten area. Dilution signifies the relative amount of substrate material that has been molten and gets mixed with melted powder during cladding. Although some dilution is necessary for creating a good metallurgical bond between deposited metal and substrate but large dilutions result in degradation of the quality of track in terms of surface roughness and dimensional
consistency. The second parameter is the clad angle ‘α’ which depends on track width and height. Clad angle should be large enough to avoid gaps and eventual porosity between adjacent tracks.

Figure 4.1: Nomenclature and schematics of DMD cladding process (a) cross-section of a typical deposited track (b) Schematics of cladding highlighting dilution and HAZ [174].

Kaplan and Groboth [175] take into account the energy and mass balance during laser cladding process to find out the optimum value of significant parameters. Their numerical and experimental investigations bring forth the importance of processing or scanning speed for controlling the height and width of deposited track. Increasing the processing speed usually resulted in reduction in clad width and height. The clad height was observed to be proportional to the powder feed rate up to a certain limit (around 1.5 g/s) for Stellite 6 on mild steel substrate. Similar limitation was observed on the energy input during cladding of large sectional area in that after reaching a power input of 3000 W, melting of additional powder was stopped and increasing the powder flow rate only enhanced the powder wastage rate.

Pinkerton and Li [176] discussed the wastage of laser power due to reflection from the powder stream and wastage of heated powder which remained uncaptured by the melt pool. They calculated that about 20% of incident power is lost due to these occurrences for a CO₂ laser. This power loss is additional to the losses existent due to inefficiencies inherent in the generation of laser beam. In a later publication, the same authors modelled the geometry of moving melt pool and the cladded track using energy and mass balances [177]. For 316L and H13 steels their model predicts the shape of melt pool, in a plane parallel to the surface of substrate, to be elliptical with aspect ratio tending to increase with the increase in absorbed power.

Ghosh and Choi [178] developed computer simulation based on numerical models for multi-pass and multi-layer DMD cladding of H13 tool steel on mild steel substrate. Their model depicts heat accumulation in the substrate as a result of multiple passes which leads to a slightly higher melt pool depth as compared to a single pass only. The power impinging on the metal-substrate system depends on the material absorptivity and laser power. Heat affected zone is very small but heating rates are
very high of the order of $10^4$ K/s. During the transient state of cooling, cooling rates drop drastically with time due to imbalance in the total amount of heat added and the heat dissipated. With the absorption of initial thermal energy, the specimen undergoes yielding and the surface contracts resulting in high tensile stresses at the surface and compressive stresses in the core.

4.2 Main Features of DMD Setup

In this section the salient features of POM-DMD setup is described in some detail that was employed to prepare numerous specimen and metallic parts for investigation in this research. DMD process made it possible to generate structures with controlled properties at micro and macro levels without significant machining and heat treatment post-cladding, the main reason for which was the small heat affected zone and optimum combination of process parameters. The schematics of DMD process and the picture of inside chamber of POM-DMD machine is presented in Fig 4.2.

![Schematics and picture of POM-DMD setup](image)

Figure 4.2. Schematics [65] and picture of POM-DMD setup

The POM-DMD machine uses a CO$_2$ laser (wavelength 10.6 µm, divergence 1 mrad with maximum power of 5 kW), the powder delivery system including 4 feed hoppers, sensors for feedback loop for dimension control, the inert-gas (argon & helium) environment, and the CAD driven motion control system for the laser beam. A feedback control system in DMD machine provides a closed loop control to maintain uniform deposition thickness, thus significantly reducing post-processing time [19]. The CNC controlled CO$_2$ laser focuses on a compatible substrate or already cladded material's melt pool while at the same time injecting, melting and depositing small amount of metal powder (flowing through coaxial nozzles) to build the part as sequence of thin layers, one upon another [67]. The deposition rate of laser can range from 24 to 160 cm$^3$/hr and the deposition or scanning speed can vary from 50 to 1800 mm/min.
Fig 4.3 (a) illustrates the large bed size of DMD machine (2 m x 1m) inside the cladding chamber which also provides a height of 1 m. This is the largest workspace among all laser assisted AM systems and provide the unique advantage of not only repairing but manufacturing large size metallic parts for industrial applications. In the same figure, a three jaw chuck is also visible which provides the facility for cylindrical cladding. In this operation, the cylindrical workpiece (substrate) is held and slowly rotated in the chuck while laser beam performs cladding on the cylindrical surface. In this research a number of composite specimen with inner core of mild steel and outer cladding of H13 and 316L alloy steels have been produced and tested for dynamic performance, wear characteristics and fatigue responses. Fig 4.3(b) shows four separate powder feeder hoppers in which different metallic powders can be stored and injected simultaneously through the coaxial nozzle to facilitate intermixing of powders and deposition of composite powders on to the substrate.

Laser assisted DMDs can be used on a wide range of metallic surfaces and possess the ability to mix metals in more than one ways to create graded structures [32]. Table 4.1 presents the names of more popular metallic powders that have been successfully used for cladding in DMD along with the range of materials that can be used as substrate. It is obvious that not every material can be used as substrate for any powder used. The mentioned powders have been successfully processed over the last 20 years on different DMD systems and the appropriate references are already provided in chapters 2 & 3 particularly in Table 2.1.
Table 4.1: Metallic powders and compatible substrates that can be processed on DMD

<table>
<thead>
<tr>
<th>DMD Powder</th>
<th>Rockwell Hardness</th>
<th>Tool Steels</th>
<th>Stainless steels</th>
<th>Low carbon steels</th>
<th>Cast Iron</th>
<th>Ni-alloys</th>
<th>Co-alloys</th>
<th>Cu-alloys</th>
<th>Ti, CP Ti</th>
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<tr>
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<td>17-4 PH SS</td>
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<tr>
<td>7</td>
<td>P20</td>
<td>36-44</td>
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<tr>
<td>8</td>
<td>P21</td>
<td>45-49</td>
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<td>Y</td>
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<td>B. Cobalt based</td>
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4.3. Experimental Methods:

For evaluating the influence of DMD process parameters on the developed metallic structures, it is necessary to analyse the effects of variation in a particular parameter upon every other parameter of interest. Here it is appropriate to introduce the terms “controlling parameters” and “controlled variables” which respectively pertain to the values that can be varied on the machine like beam power, powder feed rate and laser scan speed and the variables which are dependent on those machine parameters but cannot be directly controlled like track width and track height. The main criteria in finding the optimized set of controlling parameters were to minimize laser power and reduce powder wastage for a well bonded structure with minimal surface defects and micro porosity. The controlled variables that have been investigated in this research include:
- Track properties like width and height which govern the hatch distance or sideways overlapping movement of laser to form a layer of cross-section

- Dimensional accuracy particularly in cases when specimen height becomes greater than 5 mm and for porous structures where overlapping of tool path cannot be avoided. Height of 5 mm is set as the threshold value because it is well beyond the thickness of coatings generally done with DMD systems and enables the cladding to be used as a part or component.

- Possibility of creating different geometries like completely solid, thin walled, porous and composite structures

- Bonding quality and strength between clad and the substrate

- Surface quality of cladded structure

- Micro-porosity with the help of optical microscopy and SEM in terms of frequency of occurrence of micro pores and their volumetric percentage within the bulk of the cladded parts

The materials investigated during this research include titanium alloy Ti6A-4V and high strength steel alloys which include 41C, 316L and H13. The substrate used for titanium alloy was commercially pure titanium grade 2 plates while for all steel alloy claddings mild steel plate was used as the substrate. The thickness of substrate was in the range of 10-25 mm. Claddings done on plates with thickness below 10 mm at laser power equal to or more than 500 W and particularly in instances where cladding height rises beyond 3 mm, tend to induce a concave bend in the top surface of substrate. The undesirable bending eventually generates non-consistency in z-height and results in accumulation of dimensional errors due to uneven concentration of laser power on the melt pool thus effectuating non-uniform melting and deposition of powder. Chemical composition of all powders used is presented in Table 4.2 according to manufacturer/supplier’s data sheet:

### Table 4.2: Chemical Composition of metallic powders used in Laser Cladding on DMD

| A. 316 L stainless steel powder supplied by SULZER Metco Australia |
|-----------------|---------------|--------------|-----------|------------|----------|-----|-----|
| Composition % wt. | Iron | Chromium | Molybdenum | Nickel | Manganese | Silicon | Carbon | Other |
| 62-72 | 16 -20 | 2 - 4 | 10 - 14 | 1 | 2-3 | 0.03 | <0.5 |

| B. H13 tool steel powder supplied by Alloys International Australasia pty Ltd. |
|-----------------|---------------|--------------|-----------|------------|----------|-----|
| Composition % wt. | Iron | Chromium | Molybdenum | Niobium | Manganese | Silicon | Carbon |
| Balance | 5 - 7 | 1.5 - 2 | 1 - 2 | 0.4 | 1 | 0.35 |

| C. 41C stainless steel powder supplied by SULZER Metco Australia |
|-----------------|---------------|--------------|-----------|------------|----------|-----|
| Composition % wt. | Iron | Chromium | Molybdenum | Nickel | Manganese | Silicon | Carbon |
| 66 | 17 | 2 | 12 | 1 | 2 | - |

| D. Ti6Al4V titanium alloy powder supplied by HAYNES Metal |
|-----------------|---------------|--------------|-----------|------------|----------|
| Composition % wt. | Titanium | Aluminium | Vanadium | Iron |
| 90 | 6 | 4 | 0.3 max. |
Since one of the initial objectives of this research was to investigate the possibility of using DMD system for creating biomedical scaffolds. These structures are mostly porous with preferred size of the pores as less than 1 mm in order to facilitate cell proliferation and tissue development. Therefore, initial experimental work was focussed on producing porous scaffolds from one of the best known metallic alloys for biomedical applications i.e. Ti-6Al-4V. But the immediate problems encountered were (i) strong tendency of titanium to react with oxygen even in the presence of shielding gas and (ii) inability to produce wall thicknesses in the range of 0.2 to 0.5 mm due to large beam spot size needed to generate enough power for melting the titanium alloy powder. Despite these negative results it is considered necessary to mention the experimentation on Ti-alloy powder on DMD because it illustrates the directions and routes of conducted research and provides some useful insights into capabilities and limitations of DMD setup.

In contrast to Ti6Al4V, DMD cladding of high strength steel alloys was highly encouraging. Consequently, the focus shifted on steel alloys for preparing structures for subsequent mechanical characterization. Four different types of structures were produced from high strength alloy steels on the DMD setup which include (i) completely solid (ii) thin walled (iii) porous, and (iv) composite.

Generally, for solid structures there is little restriction on the major controlling parameters which are laser power, powder feed rate and laser scan speed. Cladding can be done at high laser powers in the range of 1.5 – 2.5 kW which can generate a large beam spot size in excess of 2 mm while working with a high powder flow rate greater than 7 gm./min. The main objective is to quickly deposit maximum possible material that gives good bonding with the substrate and produces a cladded structure with minimal micro-porosity. But for producing thin walled structures, it is mandatory to reduce the beam spot size to a dimension that is comparable to intended wall thickness. For maintaining the dimensional accuracy as the height increases, it is also essential to maintain z-height between the coaxial nozzle and the top layer of cladding. The beam spot size depends on a variable called s_focus which in turn depends upon the optical system which produces the laser. The value of s_focus defines the focus point of laser in the space between nozzle and substrate. Zero s_focus maintains the focus in the middle of this space while a negative value moves focus closer to the substrate and results in a smaller beam spot diameter.

Figure 4.4 (a) & (b) present the pictures of completely solid pins of 8 mm diameter cladded up to 7 mm height on 10 mm diameter MS rods used as substrate, in as-cladded condition and after turning on lathe to remove dimensional inaccuracies and variations. The surface roughness after cladding is comparable to cast surfaces and there is a sufficient quantity of unmelted powder stuck to the surface which is apparent from the picture shown in Fig 4.4(a). But the roughness conspicuous on the surface does not go very deep into the bulk and the thickness of layer removed after machining to expose the clean interior is in the range of 0.1 mm.
The convenience of cladding solid parts is lost when we intend to produce geometries like hollow and porous structures. In contrast to solid structures, during preparation of thin walled, porous and composite structures from DMD cladding, specific problems arise which demand alteration and adjustment not only in laser process parameters but also to the cladding scheme. In Fig 4.4(c) a number of thin walled cladded specimen from 316L SS are shown which are produced at different combinations of laser power, scan speed and powder flow rate. All the specimens are finished after cladding of 11 layers. Laser power was varied between 500 to 1000W, the powder flow rate between 3.5 to 6 g/min and scan speed between 80 to 150 mm/min. It is evident by looking at specimens 4 and 5 in Fig 4.4 (c) that when powder flow rate was increased for rapid cladding rate then laser power must increase to utilise the available powder but in compromise wall thickness also increases accordingly. The deteriorating dimensional accuracy is also visible with the increasing height of the walls. Best accuracy is obtained with specimen ‘2’ where cladding was done at 450-500 W and 3.5 g/min at 80 mm/min, but it comes at the cost of more time required for attaining a certain height.

The composite structures have a cylindrical geometry and involve the cladding of 316L and H13 steels on the external cylindrical surface of a 5 mm diameter mild steel rod held in a rotating chuck as shown in Fig 4.3(a). The main problem is the speed at which chuck should be rotated so that molten powder is appropriately deposited and bonded to the MS rod rotating beneath. Good bonding was achieved when the MS rod was rotated at a tangential speed of 12 mm/min for the initial and final two layers and 18 mm/min for the intermediate three layers. Laser power was also varied from 750 W for the initial layer to 600 W for the next two and 500 W for the last three layers. The first layer was deposited at slower speed and higher power to ensure good bonding and settling of the cladded material on the substrate. Since the composite rod when sectioned does not clearly depict the core and cladded cross-sections, therefore, its picture is not presented here but the optical micro-graphs exhibiting the micro-structure are presented in Fig 4.5 in which cladding is clearly differentiated from

![Figure 4.4: Solid DMD cladded structures from H13 and 316L alloy steel powders with MS substrate (a) in as-cladded condition (b) after turning on lathe (c) thin walled structures with different parameters.](image-url)
the core. The boundary separating MS from alloy and tool steel cladding is distinctly visible along with the coarse grains at the interface and presents a good view of excellent bonding between clad and substrate.

![Optical micrographs illustrating DMD cladded composite structures with mild steel rod as substrate (a) 316L SS cladding (b) H13 TS cladding](image)

Figure 4.5: Optical micrographs illustrating DMD cladded composite structures with mild steel rod as substrate (a) 316L SS cladding (b) H13 TS cladding

For porous structures, the main consideration was size of the pores. But the track width imposes a restriction on the size of pores (rectangular shape) which cannot be lesser than the track width; therefore, a large number of iterations were made to settle for the parameters that can provide the smallest beam spot size. For high strength steel alloys like 316L, 41C and H13, the minimum beam spot size that could be achieved was 0.9 mm at a laser power in the range of 450 – 500W and s_focus equal to -17. Fig 4.6 illustrates variety of porous structures prepared with steel alloys.

![As cladded porous structures from 41C stainless steel powder (a) Cylindrical - 20 mm external diameter and 12 mm high (b) 18 x 10 x 7 mm high – pore size 2 x 2 mm (c) 21 x 21 x 12 mm (high) – pore size 2 x 2 mm (d) 10 x 10 x 3mm (high) – pore size 1 x1 mm.](image)

Figure 4.6: As cladded porous structures from 41C stainless steel powder (a) Cylindrical - 20 mm external diameter and 12 mm high (b) 18 x 10 x 7 mm high – pore size 2 x 2 mm (c) 21 x 21 x 12 mm (high) – pore size 2 x 2 mm (d) 10 x 10 x 3mm (high) – pore size 1 x1 mm.

The selection of optimum controlling parameters also depends to a great extent on the melting point of metal that is processed in addition to the intended type of structure. The DMD cladding experiments were designed in such a manner that initially all the controlling variables are varied interactively to produce several tracks of varying characteristics. Then based on the output values of track width and height, different structures as described above were created from CAD models employing optimal...
parameters. Table 4.3 displays the track height and width for titanium alloy and Table 4.4 exhibits values for high strength steel alloys. Since there is very little difference in the melting temperatures of austenitic (316L) and martensitic (H13) steels, therefore, the final outcome in terms of track width and height is almost similar for both types of alloys.

Table 4.3: DMD parameters for cladding of Titanium alloy powder (Ti-6Al-4V) on Grade 2 Ti plate

<table>
<thead>
<tr>
<th>Cladding Ref #</th>
<th>Powder Feed (g/min)</th>
<th>Laser Beam Power (Watts)</th>
<th>Laser scanning speed (mm/min)</th>
<th>Track width (mm)</th>
<th>Track Height (mm)</th>
</tr>
</thead>
<tbody>
<tr>
<td>Ti-S-1</td>
<td>2</td>
<td>800</td>
<td>250</td>
<td>2.45</td>
<td>1.15</td>
</tr>
<tr>
<td>Ti-S-2</td>
<td>2</td>
<td>1000</td>
<td>60</td>
<td>2.8/2.5*</td>
<td>1.3</td>
</tr>
<tr>
<td>Ti-S-3</td>
<td>2</td>
<td>1200</td>
<td>100</td>
<td>2.8</td>
<td>1.32</td>
</tr>
<tr>
<td>Ti-S-4</td>
<td>4</td>
<td>1400</td>
<td>80</td>
<td>3.4</td>
<td>1.45</td>
</tr>
<tr>
<td>Ti-S-5</td>
<td>4</td>
<td>1600</td>
<td>60</td>
<td>3.05</td>
<td>1.4</td>
</tr>
</tbody>
</table>

* track width reduced slightly by manipulating the z_height

Table 4.4: DMD parameters for cladding high strength steel alloy powder on mild steel substrate.

<table>
<thead>
<tr>
<th>Cladding Ref #</th>
<th>Powder Feed (g/min)</th>
<th>Laser Beam Power (Watts)</th>
<th>Laser scanning speed (mm/min)</th>
<th>Track width (mm)</th>
<th>Track Height (mm)</th>
</tr>
</thead>
<tbody>
<tr>
<td>ST-S-1</td>
<td>4</td>
<td>800</td>
<td>60</td>
<td>1.25</td>
<td>0.3</td>
</tr>
<tr>
<td>ST-S-2</td>
<td>5</td>
<td>800</td>
<td>100</td>
<td>1.2</td>
<td>0.25</td>
</tr>
<tr>
<td>ST-S-3</td>
<td>6</td>
<td>800</td>
<td>150</td>
<td>1.2</td>
<td>0.2</td>
</tr>
<tr>
<td>ST-S-4</td>
<td>4</td>
<td>1000</td>
<td>60</td>
<td>1.5</td>
<td>0.45</td>
</tr>
<tr>
<td>ST-S-5</td>
<td>5</td>
<td>1000</td>
<td>100</td>
<td>1.4</td>
<td>0.40</td>
</tr>
<tr>
<td>ST-S-6</td>
<td>6</td>
<td>1000</td>
<td>150</td>
<td>1.4</td>
<td>0.35</td>
</tr>
<tr>
<td>ST-S-7</td>
<td>4</td>
<td>1200</td>
<td>60</td>
<td>1.7</td>
<td>0.55</td>
</tr>
<tr>
<td>ST-S-8</td>
<td>5</td>
<td>1200</td>
<td>100</td>
<td>1.6</td>
<td>0.5</td>
</tr>
<tr>
<td>ST-S-9</td>
<td>6</td>
<td>1200</td>
<td>150</td>
<td>1.6</td>
<td>0.45</td>
</tr>
<tr>
<td>ST-P-1</td>
<td>4</td>
<td>500</td>
<td>80</td>
<td>1.0</td>
<td>0.2</td>
</tr>
<tr>
<td>ST-P-2</td>
<td>1.8</td>
<td>400</td>
<td>60</td>
<td>0.75</td>
<td>0.1</td>
</tr>
</tbody>
</table>

4.4. Results and Discussion
The main interest in parametric investigation is focussed on the track dimensions for a given set of controlling parameters. Track dimensions reflect upon the efficiency of utilising the powder, dimensional accuracy of the finished part and the possibility of creating fine porous structures. The track width corresponds to the spot size of laser beam with an error of ± 10% depending on the laser power, s_focus and z_height.

4.4.1. Thermal Cycle in a DMD process
Before deciding upon the selection of optimum laser power, it is essential to understand the heating and thermal cycle of powder particles illuminated by the laser beam. The size of gas atomized powder particles range between 50 – 100 µm. There have been many extensive studies conducted to develop an insight into the mechanism of heating, melting and re-solidification of metallic powder. It must be noted that for powder bed systems like SLS and SLM the mechanism of heat transfer from laser to powder is different from systems like LENS and DMD which involve melting of a flowing powder stream. But there are many similarities once the powder particle becomes heated, melted and re-
solidified and the manner in which heat gradient affects the lower layers during cladding. Qi et al [72] developed a numerical model to study the heat transfer between laser and powder stream flowing through a coaxial nozzle for the DMD system. A schematic of the coaxial nozzle showing laser beam impinging on the substrate and surrounding powder flow is illustrated in Fig 4.7(a). The value of stand-off distance is always maintained to be 20 mm by varying the z_height as the layer thickness increases. Another important consideration in analysing DMD cladding is the attenuation of laser power as it passes through the powder stream due to loss of energy by reflection and absorption on powder particles. Fig 4.7(b) presents the reduction in laser power intensity with respect to the radial distance from the centre of beam. At an outward radial distance of 0.3 mm, more than 5 times reduction in beam intensity can be observed.

![Figure 4.7: Interaction of laser beam and powder stream in a DMD system (a) Schematics of coaxial nozzle in a DMD system, and (b) Plot of laser beam attenuation with radial distance from the beam centre [72]](image)

Thus in a DMD process only a small area of the surface of substrate irradiated by the laser beam gets heated and melted, and eventually forms a molten pool. This brings us to couple of very important consequential effects observed in the DMD process. First one is that the powder particles are not melted in air but after they strike and entrapped by the liquid substance in the melt pool while the second is related to the formation of very small heat affected zone (HAZ) due to concentration of beam power near the centre. In their numerical model, Qi et al [72] assumed immediate melting of powder particles which in turn were assumed to be perfectly spherical, as these fall into the liquefied melt pool. But the melting and re-solidification model similar to that proposed by Konrad et al [179] seems more realistic which explains that the laser-powder grain interaction can be divided into three stages: (1) preheating, (2) melting and re-solidification, and (3) thermalization. The model was described for a SLM process but it can be adequately adjusted for the DMD process. In the co-axial powder stream, preheating occurs before the powder grain enters melt pool. It is the period during which powder particles absorbs sensible heat to raise their surface temperature to near melting point.
and forming a thin skin of liquid, usually a few microns, on the surface of the powder sphere. The powder then becomes part of the melt pool and eventually resolidifies due to large temperature gradients existing between the top and bottom surface of the substrate. The presence of small HAZ indirectly affects the thermal state of clad since almost simultaneously to re-solidification of localized melt pool, the thermalization stage begins in which the powder particle starts attaining a uniform internal temperature.

4.4.2. DMD cladding of structures with cavities and pores
During the initial phases of this research, the objective was to investigate the possibility of producing porous structures from biocompatible metals like titanium and some steel alloys using DMD process that can be used as biomedical scaffolds. Development of such structures poses the unique challenge of optimizing the cladding geometry for creating uniform thin walls, while maintaining the strength and dimensional accuracy of the structure. In a DMD process main requirement is to generate sufficient heat in the laser beam that can adequately melt the powder coming out of co-axial nozzles. If the beam power is not able to efficiently melt the metallic powder, then cladding will be improper besides wastage of powder and sticking of unmelted powder to the base or substrate. But increasing the beam power has a down side as well. If beam power is continually increased for proper powder melting the beam spot size also increases since the value of s_focus must now be increased to handle more power, which means the focus point of beam must be moved in the +Z direction or away from the machine bed. This increment in spot size results in limiting the prospects of producing finer details or macro-pores, because the width of cladding cannot be less than the beam size. Another important outcome is the increased "bead size" due to more heat and more powder melting, thus resulting in a wider and higher track.

During the investigation to find out optimum controlling parameters for developing porous structures from Ti6Al4V alloy, initial results for track width and height are shown in Table 4.3. The values indicate that track width increases to approximately 2.8 mm with the power only around 1200 Watts. In comparison, after the cladding of biocompatible stainless steel powder track width was found to be 1.7 in the same power range. This observation was enigmatic in view of the concept that track width almost exactly corresponds to the beam spot size. The observed anomaly brings forth a very important conclusion for DMD based laser cladding that the track properties are also dependent upon the type of material cladded in addition to laser process parameters. The observed difference is attributed to low thermal conductivity and higher melting temperature (melting point 1670°C) of titanium alloy powder. Thus adequate melting of titanium powder needs more heat concentration which should be accompanied by less scanning speed, and the outcome is wider and higher track. On the other hand, stainless steel powder needs appreciably less “power concentration” and exhibit good bonding to the mild steel substrate between 600 – 800 Watts.
Fig 4.8 shows the results of cladding according to the parameters shown in Table 4.3 for Ti6Al4V powder. While observing the titanium plate, it is clearly visible that first clad (Ref # Ti-S-1 in Table 4.3) did not bond to the base Ti plate and after deposition it simply eroded away from the substrate. This is an evidence of the fact that at low powers (<1000 W for titanium), melt pool is not sufficiently generated and whatever powder is melted, it forms a circular bead on itself instead of bonding with the substrate. This phenomenon makes it impossible to create a form or a shape with “features” to some height. Apart from less power, another reason of poor bonding was the high laser scanning speed that allowed for lesser melting time to the incoming powder. Thus by increasing the beam power and reducing the laser scanning speed, titanium alloy cladding with firm bonding to the substrate was obtained, as shown in adjacent tracks/layers. But the solid and properly bonded clad was obtained with track width consistently in excess of 2.5 mm. Attempt was also made to continuously vary the z-height with every track instead of keeping it constant within a value of 50% of track width. This innovation helped in offsetting the effect of conical beam spot and resulted in reducing the track width and height to a small extent. But for high melting titanium powder, still the track width remained in the vicinity of 2.5 mm.

After this initial investigation, the attempt was made to create a DMD cladded Ti6Al4V model with featured cavities involving curved geometries. CAD model from which the DMD tool path was extracted is presented in Fig 4.9. The controlling parameters recipe was selected to be Ref # Ti-S-1 corresponding to minimum laser power that can provide good bonding and create fully dense structure. Intended dimensions of the laser cladded model were kept flexible and initially the external dimensions were set as 18 x 32 mm.

When cladding was finished, resulting structure is shown in Fig. 4.10 (a), which clearly indicates complete inability to generate any feature and the deposition obtained was in the form of discretised
beads. When the size of model was increased to 24 x 36 mm, some features began to emerge but still blurred and ragged as elucidated in Fig 4.10(b). Most importantly no macro porosity could be produced, as all the designed pores were filled by the molten metal powder.

![Image](a) ![Image](b)

Fig. 4.10 Laser cladding on DMD using Ti6Al4V powder for model SM-1 (a) External dimensions 18 x 32 mm (b) External dimensions 24 x 36 mm.

Finally, when the model size was increased to 40 x 68 mm with each plateau width greater than 2.8 mm, only then the DMD cladding process was able to produce the profile replicating the exact CAD model, as shown in Fig 4.11. These investigations clearly indicate the constraints experienced during deposition of high melting temperature titanium alloy powder.

![Image]

Figure: 4.11 DMD fabricated Ti6Al4V alloy part with full features replicating the CAD model SM-1.

On the contrary, results obtained for stainless steel (41C) cladding were quite different. Table 4.4 illustrates manipulation of controlling parameters for producing the nine tracks (Ref# St-S-1 till St-S-9) whose cross-sectional views are presented in Fig. 4.12 (a), (b) & (c). In this case beam size was controllable within 1.25 mm. The results are more or less uniform in respect of bonding, minimal micro porosity and track properties as shown in the slight necking which was observed at the boundary of substrate and cladding for low power and less powder feed. But overall the bonding is good and narrow tracks are consistently produced with low track heights.

![Image](a) ![Image](b) ![Image](c)

Figure 4.12: Cross sectional views for DMD cladded tracks of 41C Stainless steel powder (a) Tracks for 800 watts (b) Tracks for 1000 watts (c) Tracks for 1200 Watts (Parameters increase from right to left).
Encouraged by the possibility of working with a "finer" beam, attempt was made to reproduce the CAD model coded SM-2 showing many intricate features, as shown in Fig. 4.13 with alloy steel powders.

![Figure 4.13: CAD solid model SM-2 with smaller pore sizes](image)

Initially the model size was chosen as 30 x 30 mm (external wall to wall) with diameter of circle as 18 mm. The resulting prototype is presented in Fig. 4.14, exhibiting good replication of the CAD model with all ribs 1.2 mm thick and all cavities neatly produced.

![Figure 4.14: DMD fabricated featured structure with cavities from 41C SS powder replicating CAD model SM-2](image)

The only deficiency found was the inconsistent height of the ribs. Careful monitoring of the DMD process led to the observation that for a model with changing geometrical features, laser beam has to overlap previously cladded structures and at several instances it needs to switch between the locations while altering the tool path. This becomes a major source for distorted geometry and dimensional inaccuracy post cladding. The effect is conspicuously evident in Fig. 4.14 as non-uniform heights across the model accompanied by some ugly blobs of deposited metal. For obviating this operational hindrance, the cladding strategy was slightly altered as follows:

i. Develop a separate program for every feature. This effectively eliminated the frequent laser switching while moving from one feature to the other.

ii. In every program it was ensured that there is no intersection of cladding while producing different features.

Through this modified strategy, specimen was produced as shown in Fig. 4.15 with external dimensions restricted to only 12 x 12 mm in size while exhibiting distinct pores. The variation in rib heights was also controlled within a range of 0.25 – 0.5 mm.
4.4.3. Relationship of DMD Process Parameters and Track dimensions

There are approximately 12–14 variables which in some way influence the characteristics of the cladded part. These variables include the following: actual laser power, beam spot diameter, spatial distribution (mode) of the laser beam, shielding gas flow, powder delivery gas flow, travel speed, powder feed rate, material properties (absorptivity, melting point, thermal conductivity, etc.), powder characteristics (particle size distribution and particle shape), powder delivery method (side injection or concentric injection), height increment per layer, percentage overlap between tracks, and tool-path patterns [72].

Sometimes a term “Specific Energy” is used to illustrate the cumulative effect of most important process parameters like beam power and laser scan speed. It can be defined as:

\[ E_s = \frac{P_b}{d_b v_s} \]  

(4.1)

where \( P_b \) = laser beam power, \( d_b \) = beam spot diameter and \( v_s \) = processing or laser scan speed. Specific energy is directly proportional to the height of deposited layer but has little effect on its width. The value of specific energy may vary from 50 J/mm² to 150 J/mm² [19].

Finding out the interacting influences of every parameter on the other is a lengthy and tedious task, but fortunately there are some parameters that are more significant and important in view of controlling the desired material properties of deposited structures. These can be described as laser beam power \( (P_b) \), laser spot diameter associated with \( s\_focus \) value, laser scanning speed \( (v_s) \) and (metallic) powder feed rate \( (f_p) \) [42, 180]. These parameters can be regarded as "controlling parameters" of the DMD system. Through their optimization, track width \( (W) \), track height \( (H) \) and micro/macro-structure can be varied in a determined manner. Thus these three variables can be described as the "controlled variables". Laser scanning speed coupled with powder feed rate probably plays the most important role, since at higher traversing speeds possibility of generating microporosity in the specimen increases due to less powder deposited over greater track lengths. Therefore, the first part of investigation is related to arriving at the optimum combination of \( P_b, f_p \), and \( v_s \), with the outcome estimated in terms of track width and height. Lesser width provides finer control over the features while lesser track height is important for creating a stable structure and controlling the vertical
dimension as the cladding proceeds. In addition to attaining optimum required track width and height, controlling parameters were also studied for good bonding to substrate without shape deterioration as the specimens attain height, particularly at the corners. It must be noted that mechanical properties of cladded parts directly depend upon the quality of controlled variables.

The DMD process parameters are also affected by the materials that are used for cladding. Every metallic powder possesses its own particular set of properties, most important of which are the melting point and absorptivity of laser radiation. In this research materials investigated were titanium and high strength steel alloys, and it was clearly demonstrated that both the materials require a vastly different set of DMD parameters for cladding and the nature of parts produced by these metallic powders also carry their typical constraints.

A number of experimental investigations were conducted that take into account the effects in variation of three principal controlling parameters upon track width and track height. The materials investigated were titanium and high strength steel alloys. But after initial investigations, Ti6Al4V was not considered further due to oxidation problems and the limitations of producing fine porous structures with pore sizes ~1mm and that could be used for biomedical applications, using DMD. Very little deviations were observed in the outcome for 41C, 316L and H13 track dimensions due to negligible differences in their melting temperatures and bonding affinity with the mild steel substrate. The results plotted as line and bar graphs are shown and discussed as follows:

1. Track width and height are plotted against laser scan speed at maximum power of 5 kW and at 60% lesser beam power at 3 kW. The results are shown in Fig 4.16. The powder flow rate was 10 gm./min for 5 kW cladding and 8.8 gm./min for 3 kW deposition.

![Figure 4.16: Variation of track width and height with laser scan speed (a) at laser power of 5 kW (b) at laser power of 3 kW for stainless steels](image-url)

There are two significant observations apparent from these graphs. First one relates to the fact that by changing laser beam power from 5 to 3 kW there is very little effect on the track width which remained
in the vicinity of 2.5 mm. This observation is very important for cladded porous or hollow structures where wall thickness imposes a restriction on track width. Second observation pertains to the continuous reduction in track height from 1.2 to 0.23 mm at 3 kW with laser scan speed $v_s$ increasing from 185 to 1000 mm/min; while on the contrary exhibiting more or less constant value of 1 mm for $P_b$ equal to 5 kW. This clearly demonstrates that in graph of Fig 4.16 (b) the existing laser power could not keep pace with the rapidly moving beam for melting and depositing the available powder in the flowing stream. Since the beam spot size was same due to unchanged $s_{focus}$ and the beam remained focussed on the melt pool, therefore, there was not much difference in track width but track height kept on reducing with increase in $v_s$ due to lesser quantity of powder deposited in every track. Therefore, there is significant impact on track height due to variation in beam power and the scanning speed.

2. Variation of cladded track dimensions were observed with respect to powder feed rate for two different $P_b$ values of 4 and 3 kW. The laser scan speed was held constant at 500 mm/min. The measured values of track width and height are plotted in Fig 4.17 (a) & (b) for two different values of beam power.

![Graph 1](image1.png)

![Graph 2](image2.png)

Figure 4.17: Variation in track dimensions with the powder feed rate (a) For cladding at 4 kW (b) For cladding at 3 kW for stainless steels

Fig 4.17(a) & (b) shows that at 4 kW when $f_p$ increases beyond 9 g/min, the track width was measured as 2.9 mm while its value rises to 2.8 mm in the same range of $f_p$ for 3 kW cladding. When $f_p$ was increased beyond that value then track width decreased slightly instead of increasing which shows that the width of melt pool cannot be extended beyond the beam spot size and there is no need of increasing powder flow rate once the critical balance between beam power and powder melted is achieved, which in this case is approximately 9 gm./min. The track height also follows the same pattern increasing to the maximum value of 0.45 mm for $f_p$ around 9 gm./min and then showing no further increment with additional powder input.
3. Fig 4.18 shows the measured track width and height with variation in beam power at constant powder flow rate of 5.1 gm./min and laser scan speed of 500 mm/min. Since the values of $P_b$ can vary in the range from 500 W to 5 kW, therefore, the graphs are plotted for two different zones as illustrated in Figs 4.18 (a) & (b). We designate 2 to 5 kW as the higher range while 0.5 to 2 kW was considered as the lower power range. The graphs of low power range were extended to 2.5 kW to ensure consistency of values among the two separately conducted experiments. The most important result obtained from these plots is the jump in track width between 1000 and 1500 W. This behaviour shows that if laser power is maintained below 1000W then the track width can be maintained near to 1 mm which is good for making thin walled and porous structures, while for $P_b$ greater than 1000 W the deposited track width experienced a steep rise to 2 mm. The main limitation is imposed by the need to increase $s_{\text{focus}}$ which results in an increase in aperture of final mirror to accommodate higher laser power. Thus the cladding regime with power greater than 1000 W in conjunction with higher $f_p$ and $v_s$ values can be suitable for cladding of solid structures but impractical for preparing fine porous structures like biomedical scaffolds from biocompatible materials.

Cumulative observation of graphs in Figs 4.17 and 4.18 makes it evident that the track height is more dependent on powder flow rate rather than laser power. Increasing $f_p$ from 5.1 to 9.2 g/min in Fig 4.17 shows an increase in track height from 0.25 to 0.46 mm at constant $P_b$, but in the whole range from 0.5 to 5 kW, as shown in Fig 4.18, the increment in track height is only from 0.1 to 0.26 mm when the value of $f_p$ was maintained constant at 5.1 g/min.

Figure 4.18: Variation of DMD cladded track dimensions with laser power (a) High power range from 2 to 5 kW (b) Low power range from 0.5 to 2 kW for stainless steels.
4.4.4. Morphology of Cladded Surfaces

SEM micrographs of as cladded surfaces of layers formed in a DMD process also help in comprehending a number of important characteristics of the DMD cladding process and parameters. For a clad that is composed of 5 tracks deposited one upon the other, SEM micrographs with relatively lower magnification (around 500) like those shown in Fig 4.19 (a) & (b) show a ragged surface with high roughness, nature of which vastly differ depending on the values of parameters used during cladding. The 41C micrograph in Fig 4.19(a) was cladded with the controlling parameters set as 1000 W, 5 g/min and 100 mm/min while surface visible in Fig 4.19(b) was prepared at 1000W, 6 g/min and 150 mm/min. It was also observed carefully that at every power there is a critical set of values for f_p and v_s which if exceeded not only results in wastage of powder but also degrades the quality of cladded bulk and the surface.

![Figure 4.19: SEM micrographs of single track layers cladded from 41C powder (a) Cladding at 1000 W, 5 g/min and 100 mm/min (x481) (b) Cladding at 1000W, 6 g/min and 150 mm/min (x696)](image)

Another important observation obtained from the study of morphology of as-cladded surfaces is the presence of burnt and oxidised powder particles at laser power levels higher than the optimally required values. These particles do not form part of cladding and exist in addition to partially melted particles that stick to the side surfaces of cladded part and substrate and get accumulated inside the pores and cavities of non-solid structures.
Figure 4.20: SEM micrographs highlighting burnt powder particles sticking to the cladded surface (a) 41C cladding at 1200 W (x1.36k) (b) 316L cladding at 1500 W (x417).

The micrograph in Fig 4.20(a) illustrates the fine burnt debris accumulated on the surface of a 41C specimen’s surface which was cladded at 1200 W, while in Fig 4.20(b) a large number of burnt particles are visible on the surface of 316L rectangular padding produced at 1500W and at high powder flow rate of 6 g/min.

Figure 4.21: SEM micrographs showing individual tracks for steel alloy surfaces cladded at 1000W, 4 g/min and 100 mm/min (a) 316 L surface at magnification (x212) (b) H13 surface at magnification (x116)

Additional conspicuous effect that can be observed on the cladded surfaces of solid parts done with principal controlling parameters set to higher than optimized values for reducing the clad time is the wider tracks and highly rough surfaces with molten blobs of powder particles sticking to the surfaces. As described in the previous section that once the laser power increases beyond a certain value then the optical requirements of beam demand an increase in $s_{\text{focus}}$ which results in an increase in beam spot diameter causing a larger track width to be deposited. That effect is clearly evident in the micrographs shown in Fig 4.21 (a) for 316L and in Fig 4.21(b) for H13 DMD cladded
surfaces. If the cladding power is maintained up to 600 W for steel alloys with s_focus equal to -15 then the beam spot size and hence track width can be easily maintained between 0.9 to 1mm, but according to graphs of Fig 4.18 the track width rapidly increases beyond 1000 W, as more power is available for melting available powder in the flowing stream.

### 4.4.5. Optimizing Controlling Parameters

Based upon all the mentioned observations, impact graphs can be plotted that would illustrate the cumulative effects of principal controlling parameters for the investigated high strength alloys which are high strength steels and Ti6Al4V in this investigation. Fig 4.22 presents the variations in track width and height with different combinations of principal controlling parameters. For high strength steel alloys, most appropriate parameter values for cladding strategies that can result in good bonding with substrate, minimum micro porosity and powder wastage, are recommended to lie within two highlighted zones ‘A’ and ‘B’ as shown in Fig 4.22. For solid specimen cladding, controlling parameters should be selected within the zone ‘A’ while for cladding of thin walled and porous structures, zone ‘B’ will fetch accurate results. Increasing laser power beyond 600 W and powder feed rate over 3 gm./ min results in degradation of quality of porous structures. On the other hand selecting $P_b$ above 1200 W and $f_p$ more than 5 gm. /min causes the burnt and partially melted powder particles to stick and accumulate on the cladded surface and on the substrate. Higher values of controlling parameters also result in more powder wastage.

![Figure 4.22: Variation in track width and height against the cumulative effect of principal controlling parameters for DMD cladding of high strength steel alloys](image-url)
4.5. Summary

Quality of any part developed through DMD cladding process not only depends on the optimized combination of laser and machine parameters but on the nature of geometry created and type of material processed. In this research a number of specimen were produced from titanium (Ti6Al4V) and high strength steel alloys (41C, 316L and H13) using DMD cladding. Initial objective was to develop biomedical scaffolds which are porous structures with pore size preferably less than 50 microns. Cladding with Ti6Al4V showed two prominent drawbacks which were track width greater than 2 mm and high affinity to oxygen during cladding. In view of these problems titanium alloy was not further considered for DMD part development.

High strength steel alloys proved to be more suitable for generation of solid, porous and composite parts through DMD cladding. The parametric investigation was done in detail with particular emphasis on the cladded track dimensions. Out of all the effective parameters in a DMD process, laser beam power, powder flow rate and laser scan speed were found to be the most influential. The specimen with different geometries mentioned have been successfully created from steel alloys and the DMD parameters were adequately controlled and manipulated for the desirable shapes. The investigation highlights not only the importance of certain parameters but to take the whole cladding process into account which includes cladding scheme and materials involved. The study although fetches satisfactory results in terms of DMD cladding from steel alloys but needs more thorough investigation into cladding of titanium alloys from the point of view of eliminating oxidation of generated parts and possibility of creating delicate features and small size macro pores. This study also uncovers many deficiencies of the DMD process when considered from the viewpoint of part generation through cladding for industrial usage.
5.1. Introduction

The material and metallurgical analysis is performed on the DMD cladded high strength steels which include 41C, 316L and H13. Former two traditionally exhibit the tendency to retain austenite at room temperature while H13 tool steel contains martensite as the major phase in its microstructure. Repeated investigations proved the unsuitability of DMD system for preparation of biomedical scaffolds and implants from Ti6Al4V claddings; therefore, metallurgical and material research for titanium alloys was not pursued beyond initial stages as already described in chapter 4. The mechanical properties of alloy steel parts produced by DMD cladding are primarily related to the evolution of micro-structure, cladding associated micro-porosity, micro-hardness along and across the cladding and secondarily to the residual stresses generated during laser deposition of metallic powder on the substrate. Therefore, in this chapter all these aspects are analysed and discussed before commencing a detailed analysis of dynamic and wear properties particularly high strain rate compressive strength, subsequently presented in chapters 6, 7 and 8.

Inevitable presence of residual stresses in laser generated structures can have profound effects upon the behaviour exhibited under dynamic loads particularly fatigue strength. Therefore, discussion on residual stress is included in this chapter to complete the investigation on material behaviour. In addition to residual stresses, fatigue resistance or endurance is also an important material behaviour that needs to be looked into and compared with the same materials available in wrought and rolled forms, but that study is taken up separately in chapter 8 to assign due focus to fatigue investigation. For this study, the samples consist of clad and substrate both so that the nature of residual stresses can be elucidated at the interface in addition to the cladding. Experiments were conducted to evaluate planar residual stresses generated during cladding across the specimen thickness from substrate to the top surface of cladding. The technique used is the well-established “Neutron diffraction” method which has the unique capability of finding out strains within the bulk of stressed material and not limited to surface measurements only. The study was sponsored by Australian Nuclear Sciences and Technology Organisation (ANSTO) located at Lucas Heights, NSW Australia. These studies provide a very useful insight into prospects of development of composite structures from DMD cladded steels, particularly in view of their intended use in industrial applications and in situations with unusual shape and loading requirements such as high load bearing lattice structures.
5.2. Microstructural Investigation

Mechanical properties like compressive strength under impact loading and wear resistance against sliding friction is dependent on the homogeneity of microstructures developed during the production or cladding process. Processing of tool and alloy steels at high temperatures, such as casting and hot working gives rise to coarse-grain microstructure and carbide segregation along the grain boundaries and inside cells due to presence of high alloy contents like chromium, tungsten and molybdenum [182]. The coarse-grained, dendritic and segregated microstructure causes a reduction in strength and toughness. Hot working processes like hot rolling can also produce refined and homogenous microstructure independent of high temperature grain residuals like retained austenite, provided solidification is rapid and processing temperatures are prevented from falling to room temperature and re-rise. Depending on cooling rate, the matrix microstructure of a rapidly solidified tool and stainless steel part generally varies from ferrite to martensite. [183]

DMD process is essentially comparable to similar hot working processes involving fusion and re-solidification at a high rate but with some specific advantages: (a) DMD setup exercises great control over processing parameters, an optimized combination of which provides a realistic opportunity to produce refined and homogeneous microstructure, (b) melting and re-solidification is localized to a very small area defined by the diameter of laser beam, which results in a rapid solidification rate, (c) layer by layer deposition of the cladded metal does not allow the specimen to cool down to room temperature and then re-heated. Therefore, high strength combined with good toughness is achievable through development of fine microstructures as illustrated by the results derived during this research.

5.2.1 Experimental Methods:

The prime focus of this research is upon developing functional parts in solid and porous forms from high strength steel alloys using DMD. The microstructural investigations and evaluation of microhardness were mainly carried out across the bulk of produced solid and porous specimen, since the major intended applications need parts to be cut off and used independently from the substrate. The specimens were produced at different laser powers and relevant process parameters in order to study the accompanying changes in microstructure and microhardness. Significant amount of research has already been published which illustrate the variation in alloy steel cladded microstructures from the substrate to top layers and within the heat affected (HAZ) and re-melted zones, but in almost all these studies, the shape of investigated sample is rectangular prismatic. This study, in contrast, concentrates on investigating the patterns of grain formation at the interface and on adjacent cladded layers for composite cylindrical specimen. Composite cylindrical specimen chosen for analysing the condition of grains at the interface between mild steel substrate and the alloy steel clad were prepared by cladding 316L and H13 on the outside surface of a slowly rotating 5 mm diameter mild steel pin over a length of 35 mm approximately. This structure is more challenging than the rectangular and
prismatic formation normally studied because in annular cladding the problem of poor bonding may emerge in a more pronounced manner as a result of gravitational and centrifugal effects on the just deposited annular tracks due to continuous rotation of the substrate during cladding.

In all, twelve solid and composite cylindrical parts and six porous rectangular prismatic parts have been generated through laser cladding of 316L and H13 alloy steels on mild steel substrate using the DMD setup. Based on the detailed parametric investigations of the quality of DMD generated parts, all the porous specimens were generated at 500 watts laser power and specific reasons were already discussed in chapter 4. The lengths of solid cylindrical parts were 5-6 mm with a diameter of 5 mm, depths of rectangular specimens were ranging from 5 to 10 mm with a cladded area of 20 x 10 mm and 10 x 10 mm having 1 mm wall thickness while composite cylindrical parts were produced by 1.5 mm thick cladding on the outer surface of Φ 5 x 35 mm long mild steel bars. Rectangular and solid cylindrical specimens were cut off from the substrate while the composite cylindrical specimens were slitted at arbitrary cross sectional locations to produce test samples of lengths 5 mm so that the substrate pin and annular cladding is adequately revealed. Fig 5.1 presents a sample of porous, solid and composite specimen after etching with the appropriate solutions.

![Etched samples for study of microstructure](image)

**Figure 5.1: Etched samples for study of microstructure (a) Porous (b) Solid (c) Composite**

The samples were then fine polished up to 3-micron level for rendering the surface scratch free to a large extent and then tested for Vickers microhardness on Bueheler hardness tester. Later on the same samples were mirror polished to 1-micron level in a sequential manner for microstructural investigation. The austenitic 316L stainless steel specimens were etched using Carpenter 300 series etchant which consisted of 8.5 grams ferric chloride, 2.4 grams cupric chloride, 122 ml alcohol, 122 ml hydrochloric acid and 6 ml nitric acid. Ferritic H13 tool steel was etched with 2% Nital (196 ml ethanol and 4 ml nitric acid) for examination of microstructure under Leica optical microscope that can provide magnification up to 1000x. For all the prepared specimens the microstructure is studied near the top layer and at the interface and the optical micrographs are presented for 316L and H13 solid, porous and composite parts in Fig 5.2.
Figure 5.2: Evolution of microstructure with laser power and energy density (a) 316L 500 Watts (x500) (b) H13 500 Watts (x500) (c) 316L 1000 Watts (x500) (d) H13 1000 Watts (x500) (e) 316L 1500 Watts (x500) (f) H13 1500 Watts (x500)

The samples were produced on DMD setup for a combination of laser power, scan speed and powder feed rate, the combination of these three controlling parameters being defined as energy density. The values of energy density are shown for every micrograph of Fig 5.2 in Table 5.1. A relatively higher value of specific energy or energy density (> 300 J/mm²) is chosen after the observation that it greatly
improves the problems of micro-porosity and results in fully dense structures with consistently superior mechanical properties and very good bonding to the substrate.

Table 5.1: Laser Power and Energy Density for 316L SS and H13 TS cladding

<table>
<thead>
<tr>
<th>Reference in Fig 5.2</th>
<th>Cladded Material</th>
<th>Laser Power (Watts)</th>
<th>Energy Density (J/mm²)</th>
</tr>
</thead>
<tbody>
<tr>
<td>5.2 (a)</td>
<td>316L SS</td>
<td>500</td>
<td>300</td>
</tr>
<tr>
<td>5.2 (b)</td>
<td>H13 TS</td>
<td>500</td>
<td>375</td>
</tr>
<tr>
<td>5.2 (c)</td>
<td>316L SS</td>
<td>1000</td>
<td>455</td>
</tr>
<tr>
<td>5.2 (d)</td>
<td>H13 TS</td>
<td>1000</td>
<td>495</td>
</tr>
<tr>
<td>5.2 (e)</td>
<td>316L SS</td>
<td>1500</td>
<td>480</td>
</tr>
<tr>
<td>5.2 (f)</td>
<td>H13 TS</td>
<td>1500</td>
<td>515</td>
</tr>
</tbody>
</table>

5.2.2 Results and Discussion

Out of a large number of optical micrographs obtained for the prepared set of specimens, six are presented in Fig 5.2 with Fig 5.2(a) & (b) showing microstructures of porous 316L and H13 rectangular specimen generated at 500 watts of laser power. Fig 5.2 (c) to (f) present microstructure obtained from solid cylindrical specimen of the two materials cladded at 1000 and 1500 watts. The microstructure of steel alloy structures developed on DMD has been well studied. Some of the important aspects that have been well established are that the laser and process parameters play a significant role in the evolution of microstructure and the microstructure exhibits varying phase and grain configurations from the HAZ to top cladded layers [20, 21]. The majority of microstructure observed in the cladding of H13 tool steel near the substrate and adjacent layers above HAZ resembles tempered martensite with some retained austenite [181]. Martensite in steels is a super-saturated solid solution of carbon in ferritic iron. The tempering effect occurs due to frequent heating and cooling cycles during the deposition of layers one upon the other. On the other hand, uppermost layers which did not undergo the pseudo-heat treatment exhibit untempered martensite.

A detailed discussion on the evolution of microstructure for both types of cladded steels is presented with reference to Fig 5.2, but before that discussion the acquired XRD images for H13 and 316L cladded at 1000 and 1500 watts is also presented in Fig 5.3 to ascertain the type and extent of phases present within the microstructure.
The microstructural configuration is strongly dependent on the type and extent of phases present in the cladding structure. XRD graphs shown in Fig 5.3 help in determining the phase configuration of the H13 and 316L DMD cladded specimen. Niu and Chang [183] discussed in detail about the XRD based phase analysis of laser cladded M2 High Speed Steel on mild steel substrate. In this study the XRD investigation of laser cladded tool steels leads to very similar results. The dominant phases found to be present in the H13 microstructure include bcc lattice structure mainly composed of ferrite and martensite while in 316L a large presence of bcc and fcc structures were found that may consists of γ-austenite and metallic carbides due to high alloy content within the stainless steel powder. The diffracting planes associated with the peaks for bcc structured ferrite and martensite are <110>, <200>, <211> and <220> while those exhibiting the presence of fcc austenite phase have the orientation <111>, <200>, <220> and <311>.

In both the H13 specimen highest peaks as shown in the Figs 5.3 (a) & (b) correspond to the bcc crystal structure with <hkl> value as <110> and the values of d₀ as 2.03923 and 2.05 nm, respectively. The <110> peak intensity in Fig 5.3(a) is approximately 35% higher than in Fig 5.3 (b) and probably
illustrates more dissolution carbides into the matrix at higher powers. The two XRD plots obtained for 316L cladding as shown in Figs 5.3 (c) & (d) exhibit a strong presence of fcc crystal structure at the diffracting plane orientation <111> with corresponding 2θ equal to 43.507° and d₀ = 2.07846 nm. Adjacent to this peak another higher peak exists having bcc- structure and which corresponds to the diffracting plane <110> at 2θ equal to 44.485° with d₀ = 2.035 nm. Therefore, it can be concluded that the microstructure developed by 316L laser cladding has strong presence of α and γ – iron along with metallic carbides (MₓC) particularly those of chromium.

Figs 5.2 (a) & (b) reveal the fact that for both 316L and H13 specimen produced at 500 watts, the microstructure is not fully evolved. From the iron-carbon diagram it can be easily found out that for hypo-eutectoid steels the final formation should consist of ferrite and martensite. Also by looking at the cooling curves of steel, it is evident that rapid solidification as experienced in laser cladding will result in creation of martensite as the major phase in the final microstructure. In Fig 5.2 (a) for 316L specimen most of the grains are lamellar dendritic and not oriented in a single direction. As it can be observed that the grains are substantially long and mostly oriented in radial direction, which proves that the columnar grains are essentially aligned with the direction of heat transfer i.e. radially outwards. In fig 5.2(b) for the H13 tool steel, grains show interspersed ferritic grains in a matrix of lath martensite. But the needle like martensitic formation has not yet begun and the martensitic plates are not aligned in the same direction. This microstructure is typical of ferritic steels in which martensitic plates do not exist with parallel sides; instead these show as lenticular shapes due to constraints in the matrix which oppose the shape change resulting from transformation.

At higher laser powers and more energy focussed per mm² the microstructure appears to grow more homogenous and bit more equi-axed. For 316L samples produced at 1000W as illustrated in Fig 5.2(c), the microstructure shows the beginning of formation of cellular grains that are typical of austenitic steels. On the other hand the grains for 316L cladded at 1500 W as shown in Fig 5.2(e), are finer and exhibit well defined boundaries which are a result of the concentrated laser power, small melt pool size and rapid solidification [184]. The orientation of all the grains is not in a single direction which is due to uneven heating and cooling or existence of temperature gradients within the structure during cladding. Fig 5.2 (d) exhibits formation of martensitic laths in which ferritic grains are interspersed at regular intervals. But in Fig 5.2 (f) the microstructure becomes more homogenised and the ferritic structure is dissolved into martensitic phase to a great extent.

Very high magnification SEM micrographs are also shown in Fig 5.4 for H13 (magnification factor = 11.23 k) and 316L cladding (magnification factor =7.72 k) done at 1000 Watts. The H13 cladding shown in Fig 5.4(a) exhibits needle shape fully developed martensitic microstructure with regions of different grain orientations. In a DMD cladded structure, the grain directions are expected to show different orientations due to laser scanning strategy with every layer oriented at 90° with respect to
the previously cladded layer. The microstructure is consistent with a hint of micro-porosity and presence of few unmelted or partially melted powder particles. Although 316L is identified by a cellular structure with well-defined grain boundaries but due to very high cooling rates, there is appreciable presence of ferrite, interspersed dendritic structure and traces of martensite in the final microstructure of 316L, a fact also corroborated by optical micrograph in Fig 5.2(c).

Figure 5.4: High magnification SEM micrographs of DMD generated structures cladded at 1000 Watts (a) H13 cladding (x 11.23k) (b) 316L cladding (x 7.72 k)

From all these microstructures it is evident that the energy density in which laser power is the most influential component has a significant effect on the microstructural evolution of DMD generated high
strength steel parts. Since at low laser powers the energy dissipation occurs more rapidly from the metallic specimen, there is lesser time available to develop the settled granular or martensitic microstructure for the two types of steels. But at higher power, the powder melting is better and the tempering of substrate and cladded layers take place frequently that results in development of more uniform and cellular microstructure. The observed patterns of microstructure for DMD generated 316L and H13 parts reveal the reason for lesser modulus of elasticity when tested under high strain rate compressive loads, according to the results presented in Chapter 6. But the impact absorption and wear resistance performance were fairly exceptional even for as-cladded samples without any post cladding heat treatment.

When observing the microstructure and hardness across the interface between substrate and cladding, three distinct regions with different microstructures can be observed [32]. These three regions are formed with each cladding pass and can be distinguished as interface (including HAZ), columnar grain and the equi-axed grain regions. In two micrographs of Fig 5.5 developed for H13 and 316L annular cladding over mild steel bar, MS substrate, interface and cladded regions are easily distinguishable. The boundary between clad and substrate is clearly visible with large grain sizes at the interface as shown in Fig 5.5 (a) & (b). At the interface between clad and substrate heat treating of the existing grains takes place which allows the grain size to grow larger than the substrate. In comparing the micrographs of Fig 5.5 (a) and (b), the HAZ for H13 cladding is smaller with grains at the interface approximately 2-3 times larger in size while in Fig 5.5(b), the interface grains for 316L cladding are 10-12 times larger in size. This effect is due to larger thermal conductivity of H13 tool steel (24.3 W/m.k) in comparison to 316L stainless steel (14.6 W/m.k), which allows for quicker heat dissipation from the H13 grains and heat retention in 316L structure for larger time.

Figure 5.5: Optical micrographs of interface layers and grain formation in cladded layers for composite specimen (a) H 13 cladding on mild steel pin (x 200) (b) 316L cladding on mild steel pin (x 200).

Another conspicuous observation is that the substrate (mild steel pin) grains are more elongated, larger in size and oriented in the radial direction beneath the interfacial layer for H13 cladding in
comparison to 316L. This also proves that grains become columnar in the direction of heat transfer which is the case for H13 specimen as shown in Fig 5.5 (a). On the other hand solid dissolution of grain occurs at consistently higher temperatures in 316L cladded structure to form exceptionally large size grains at the interface. As the heat flows in a direction from clad to the substrate, just above the interface, grains or primary dendrites become long, slender and perpendicular to the interface. Finally in the regions where temperature gradients exist in several directions and not particularly oriented in a single direction, equi-axed grains can be observed as in the core of substrate material.

The effect of re-shaping and re-orienting the interfacial grains due to directional heat transfer and differential thermal properties at the interface is also illustrated in the SEM micrographs of 316L composite specimen shown in Fig 5.6. In Fig 5.6(a) the interface boundary is shown (magnification factor = 16.15 k) which not only illustrates the radial orientation of microstructure but also the presence of fine unmelted powder particles and voids at microscopic levels. Fig 5.6(b) shows the grain sizes and clearly identifiable grain boundaries for the same specimen at the interface and within the region between cladding and substrate (magnification factor = 3.47 k).

![SEM micrographs of interface for composite 316L specimen](image1)

**Figure 5.6:** SEM micrographs of interface for composite 316L specimen (a) Microscopic discontinuities (b) Granular structure at the interface

### 5.3. Micro-hardness

In a DMD cladded structure, the value of micro hardness is not constant at every point with the main difference existing between the substrate and the cladding. This inconsistency may be attributed to the existence of HAZ and fused and remelted zones within the interface between clad and substrate. Thus the region consists of microstructure which is not entirely consistent and homogenous. Another possible reason may be micro-cracks and micro-porosity which Dadbashk et al [77] has clearly mentioned for stainless steel parts produced by selective laser melting. Yasa and Kruth [55] also reported a presence of 1-2% micro-porosity in laser generated 316L SS parts. In this research the use of high energy density owing to least possible beam spot size and slow scan speeds, has greatly minimized the problem of micro-porosity in the DMD cladded 316L and H13 parts. The image in Fig
5.7(a) shows an optical micrograph of a polished surface of DMD specimen which still shows some pores while the SEM micrograph shown in Fig. 5.7(b) illustrates the gouges present in as-cladded surface of steel parts produced on DMD which shows that laser cladded parts are not entirely free from the presence of surface inconsistencies and pores at the microscopic level.

Figure 5.7: Presence of microscopic pores and voids in DMD cladding (a) Optical microscope image of 316L cladding (x500) (b) SEM micrograph of 41C SS cladding (x3.02k)

Within the DMD cladded structure, three distinct zones exist which can be inferred and depicted from the study of microstructure. It is interesting to note that the value of micro hardness is different for every region. Grum and Zanidrsic [185] defined these three zones as follows:

- At the top surface of the cladded layer
- Within the bulk of cladding if thickness is greater than 3 mm
- In the transition zone between the cladded and un-remelted modified layer, where only diffusion will occur.

Fig 5.8 presents the microhardness values for the cylindrical composite specimen over a radial distance of 5 mm with the spacing between the consecutive readings as 0.5mm. The interface between clad and substrate is approximately 1 mm wide and exhibits a significant drop in the hardness value when viewed from the clad side. The drop is more conspicuous for H13 specimen while for 316L parts substrate and clad hardness values are close and comparable. The values of hardness obtained are comparable to the hardness of wrought stainless and tool steel alloys, which lie in ranges of 460 HV to 580 HV for H13 and between 160 to 240 HV for 316L steels; both without heat treatment. Within laser cladding, hardness values representing a strong core can be attributed to the tempering effect during layer by layer deposition of metal.
Figure 5.8: Variation of microhardness from substrate to cladding for DMD generated 316L and H13 samples.

The location and width of transition region at the interface encompassing HAZ and re-melted zones should be of particular concern for applications involving fatigue or large magnitude loadings. In this region, the solid diffusion of grains is appreciable leading to highly non-linear variation. It should be noted that the transition curves shown in Fig 5.8 are only approximations subjected to a high degree of inaccuracy, but in conjunction with Fig 5.5 it can be seen that the interface for H13 is much smaller than that for 316L deposition. As has been noted in Fig 5.5 that the grain sizes at the interface are different in size from both the substrate and clad, therefore, this non-homogeneity can lead to higher residual stresses and initiation of crack or dislocation in this particular region especially under variable loads.

Figure 5.9 shows the values of Vickers microhardness for DMD generated 316L and H13 parts across the top layer after cutting-off from the substrate with readings taken at a spacing of 0.2 mm. So these values do not take into account the effects of substrate and intermediate heat affected and re-melting zones; which are already considered in substantial detail in the published literature. The statistical data showing the range, mean value and standard deviation for the microhardness values shown in the graphs of Fig 5.9 is presented in Table 5.2. The noticeable fact is that the values measured for H13 parts show a higher standard deviation as compared to the 316L parts, which illustrates larger hardness variation in H13 among the clad and substrate.

Table 5.2: Statistical Parameters for ‘HV’ values for 316L and H13 solid parts

<table>
<thead>
<tr>
<th>Statistical Parameters</th>
<th>316L Solid 1000 Watts</th>
<th>316L Solid 1500 Watts</th>
<th>H-13 Solid 1000 Watts</th>
<th>H-13 Solid 1500 Watts</th>
</tr>
</thead>
<tbody>
<tr>
<td>Maximum</td>
<td>217.2</td>
<td>265.4</td>
<td>538.3</td>
<td>485.7</td>
</tr>
<tr>
<td>Minimum</td>
<td>189.7</td>
<td>236.5</td>
<td>468.9</td>
<td>437.9</td>
</tr>
<tr>
<td>Mean (95% CI)</td>
<td>199.8</td>
<td>251.5</td>
<td>512.1</td>
<td>457.27</td>
</tr>
<tr>
<td>Std. Dev.</td>
<td>8.33</td>
<td>9.38</td>
<td>18.02</td>
<td>11.65</td>
</tr>
</tbody>
</table>
Figure 5.9: Microhardness distribution within the cladding of solid cylindrical parts with respect to radial distance in top layers for (a) 316L stainless steel specimen, and (b) H13 tool steel specimen.

The statistical analysis of microhardness data shown in Fig 5.9 reveals a very interesting behaviour for DMD cladded 316L and H13 specimen. There is a 20% increase in the hardness values for 316L specimen but on the other hand a 10% reduction in the microhardness for H13 specimen, based on the mean values, with the increase of laser power used in cladding. As expected that during the material deposition and subsequent re-solidification effect, top layers of the clad will cool at a faster rate and therefore, experience more tempering effect. With the increase in laser power, heat input increases resulting in greater tempering effect, which shows itself as increased hardness for 316L. But for H13 specimen with a martensitic microstructure and also due to the need for higher energy density for
proper melting, the microhardness may slightly reduce due to lesser heat dissipation and creation of finer and more interspersed microstructure.

5.4 Residual Stresses

5.4.1. Impact of Residual stresses on Laser cladded structures

The undesirable existence of residual stresses is unavoidable whenever there exists an unbalanced heat transfer during the process. The formation of residual stresses occurs primarily during the production process that is during laser cladding in a DMD process, and is generally associated with differential cooling and non-uniform plastic deformation. The presence of residual stresses in produced stainless and tool steel parts may result in premature yielding and can lead to a loss of stiffness accompanied by a reduction in load-carrying capacity [186]. It should be realized that laser assisted DMD process involves localized, time varying and uneven heating of the substrate metal resulting in appreciable temperature gradients within its bulk. Furthermore, the layer upon layer deposition of cladded metal generates temperature gradient along the vertical section of cladding. The contours of temperature gradients thus generated would certainly introduce appreciable amount of residual stresses in both clad and substrate with a significant probability of residual stresses existing in appreciable magnitudes along the thickness of cladding. The amount of residual stress variation thus needs to be checked from heat affected zone to the top layer of cladded metal.

For laser cladding process, in general, the major causes of process-induced residual stresses in a clad layer are ascribed to two effects: (1) thermal mismatch among the clad, the heat-affected zone (HAZ) and the unaffected cold substrate when a clad cools after it re-solidifies, and/or (2) solid-state phase-transformation induced volumetric change in the clad and the underneath HAZ of the substrate when the clad cools after it re-solidifies [187].

Ghosh and Choi [178] described the induction of thermal related stresses in laser cladded parts as inevitable because melting and solidification occur at very high rates and the rate of heat transfer is extremely fast due to small dimensions of the laser processing system. In addition to micro scale heat transfer, the macro scale mass transfer and non-equilibrium phase-change kinetics are also the contributing factors resulting in conditions conducive to the generation of residual stresses within the cladding system. Bruckner et al [188] highlighted the importance of controlling laser process parameters for controlling the magnitudes of induced residual stresses and hence limiting the related danger of cracking to appear in the cladding. Their one dimensional mathematical model attempted to qualitatively explain influence of plastic flow, phase transformations and differences of heat expansion coefficients on the residual stresses. The model predicts reduction in residual stresses due to increased heating time or lower scan speeds during cladding, preheating of the substrate and lower differences in thermal expansion coefficients of cladding material and the substrate.
Since in many cases, the DMD cladding is used as a coating to repair damaged objects, therefore, any presence of residual stress gradient pose a serious threat to the stability of coating. In most cases, tensile residual stresses may arise in the clad, which can become a cause of cracking within the clad and de-bonding from the substrate, and eventually bring upon an adverse effect on the ultimate mechanical performance parameters such as fatigue strength or resistance to stress corrosion, whereas the existence of compressive stresses at the surface of the clad material could improve its service life [189]. In their concluding remarks regarding measurement of residual stresses in cobalt based Stellite 6 coating on a low pressure turbine blade, Bendeich et al [190] mentioned about presence of significant tensile and compressive stresses in the repaired region of the blade both in the cladded layer and in the parent metal. In both the Stellite 6 and parent metal, the tensile stresses were found to be located near the surface which would initiate crack formation in these regions under fatigue conditions.

The detrimental effects of residual stresses could be lessened or expected to be resolved by various approaches proposed by researchers particularly in case of laser cladding. Pei and Hosson [191] suggested optimizing the coating thickness or to introduce a compliant interlayer for the reduction of the thermal stress as a commonly applied strategy. The intermediate layer should minimize the thermal mismatch between the clad and the substrate. In case of compliant films melting at low temperatures, a thick multilayer of essentially discrete composition should be deposited rather than a layer with gradual changes in composition. Residual stresses are a major factor in crack generation in the intermediate layer of bimetallic structures. Sahasrabudhe et al [192] investigated the residual stresses induced cracking at the interface between 316L and Ti6Al4V laser generated bimetallic structures, which is associated with the formation of brittle intermetallics. By introducing an intermediate layer of NiCr alloy with thickness around 750 µm, the delamination and formation of cracks at the interface was greatly minimized.

Improvements can also be induced by optimizing laser processing parameters, preheating the substrate during cladding, or performing post-cladding stress-relieving treatments. The main idea is to reduce the thermal strain difference between coating and substrate and to facilitate heat equalization at relatively high temperatures at which most materials undergo plastic deformation rather than brittle fracture. Furthermore, substrate hardening and the resulting reaction stress in the coating may be avoided in this way [192,193].

5.4.2. Measurement of Residual Stresses

The residual stresses, can only be deduced from the measured strains and subsequently combining these values with material constants. Moreover extracting the strain tensor information from within a particular cladded layer is complex and requires innovative experimental approaches. In order to circumvent the experimental difficulties, numerical modelling and finite element based simulation have been considered as considerably accurate and easily accessible approaches for predicting residual
stresses in the laser cladded specimen [194] as well as in welded high strength steel parts [195]. But numerical modelling not only requires comprehensive database of physical, thermal and mechanical properties of the materials being modelled but at some stage of investigation, experiments become inevitable for validating the numerical predictions. As described in detail by Withers and Bhadeshia [196], the residual stresses may vary in materials along macroscopic distances e.g. in metals, and also on the scale of grain and atomic size e.g. in most composite materials. Furthermore, the solutions of differential equations used to model the thermo-mechanical evolution of residual stresses as a function of temperature, possess an inherent degree of approximation which is further accentuated by different assumptions employed for model simplification. That’s why accurate prediction of the state of residual stresses is very difficult and thus experimental evaluation of laser generated residual stresses is persistently adopted as the most reliable method of investigation.

Residual stress measurement techniques can be unambiguously divided into destructive and non-destructive methods. Destructive methods involve methods like slitting, deep-hole drilling and contour methods while non-destructive methods mainly revolve around diffraction methods using X-rays and high energy neutron beam [197]. Withers et al [198] have discussed and compared in detail the merits and demerits of both types of techniques. Hole-drilling and slitting provide depth profiles while contour method enables observation of maps of residual stresses over an area. On the other hand non-destructive techniques focus on measurement of some parameter which is directly related to stress. For example, X-ray and neutron diffraction methods are based on the measurement of lattice strains by studying the variations in crystal lattice parameters of a polycrystalline material. The first method measures the residual strain on the near-surface of the material; and the second one measures the residual strain within a volume of the sample. X-ray diffraction is limited to a depth of approximately 5 µm from the surface for steels due to poor levels of X-ray penetration while neutron beam can probe up to a depth of 50 mm into steel samples [198]. Therefore, for obtaining successive values of residual stresses across the clad and substrate, neutron beam is the preferred non-destructive method. However, this method requires a special neutron source which is not only extremely expensive and hazardous but requires a nuclear reactor and a complex operational setup that must also ensure health and safety of personnel and environment.

For this study Australian Nuclear Sciences and Technology Organisation (ANSTO) located at Lucas Heights, NSW has provided the facility to perform the experiments for residual stress measurement using neutron diffraction method on the equipment named KOWARI.
5.4.3. Sample Preparation

The aim of this work is to study the development and variation in the magnitudes of residual stresses generated as a result of cladding high strength steel alloys on mild steel substrates. Since the properties of manufactured parts are dictated by the DMD process parameters, the research intends to investigate how the pattern and magnitude of residual stresses induced in the parts will vary by changing laser power during cladding. Another important objective is to corroborate the experimental results with the existing numerical modelling and FEA investigations.

For this investigation a set of 6 samples (forming a matrix 3x2) were produced with 2 different steel alloys, 316L SS and H13 TS, and cladded at 3 different laser beam powers of 1000 W, 1500 W and 2000 W. The main idea is to investigate the variation in magnitudes and nature (tensile and compressive) of residual stresses with laser power and associated process parameters. All samples have approximately the same dimensions which consists of a 3 – 3.25 mm thick clad on an 8±0.5 mm thick substrate plate of mild steel of size 55 x 55 mm. The finished samples prepared for testing on the neutron diffractometer are shown in Fig 5.10.

![Figure 5.10: Samples cut by SiC blades and prepared for Neutron diffraction measurements](image)

Due to completely solid nature of cladding the cladding was done at higher powers and with greater energy density. The laser parameters used for producing these samples are presented in Table 5.3:

<table>
<thead>
<tr>
<th>Laser Power (Watts)</th>
<th>Powder feed rate (g/min)</th>
<th>Laser scan speed (mm/min)</th>
<th>S_Focus (degrees)</th>
<th>Track width (mm)</th>
<th>Track height (mm)</th>
</tr>
</thead>
<tbody>
<tr>
<td>1000</td>
<td>4.2</td>
<td>100</td>
<td>-15</td>
<td>1.0</td>
<td>0.75</td>
</tr>
<tr>
<td>1500</td>
<td>6.1</td>
<td>300</td>
<td>5</td>
<td>1.2</td>
<td>1.0</td>
</tr>
<tr>
<td>2000</td>
<td>8.5</td>
<td>300</td>
<td>5</td>
<td>1.5</td>
<td>1.8</td>
</tr>
</tbody>
</table>
5.4.4. Experimental Method and Setup

Residual stresses were measured using KOWARI stress diffractometer in the three principal directions (Transverse (T), vertical (V) and Normal (N)) as illustrated in Fig 5.1(a) with respect to the mounted samples. The residual stresses are evaluated through the whole thickness of the substrate and cladding system. Strains in both kind of clads, 316L SS and H13 TS, were measured using either α-Fe(211) or γ-Fe(311) reflections in 90 degrees geometry, with a wavelength adjusted accordingly. Because of large expected stress gradients and in order to avoid surface effects, a spatial resolution of 0.5 mm in through-thickness direction was maintained. The planar sample geometry allows a vertically large gauge volume of 0.5x0.5x20 mm³ to be used for all directions (normal and two in-plane directions), making measurements relatively fast. The measurement grid comprises of multiple through-thickness positions with a step size of 0.5 mm. For the overall sample thickness of ‘t’ mm, the total number of measurement points were estimated to be 2*t. The experimental setup with the mounted specimen is shown in Fig 5.11.

Owing to the planar sample geometry the zero plane stress condition is applicable (also experimentally verified in similar studies) and it will be used to determine d₀ profiles and calculate in-plane stress components. The aimed accuracy of strain measurements is 5×10⁻⁵ µstrain, achieved in ~8 min beam time, which eventually resulted in stress determination with an estimated accuracy of 15 MPa. The experiments were split into two runs. Totally, 6 samples were measured in as-built conditions. The total measurement time will be estimated as follows: No. of samples × 3 direction × No. of measurement points × 8 min. Since accurate sample positioning of 0.05 mm was required for 0.5 mm resolution, some additional time should be allocated for sample alignment and positioning.
The KOWARI diffractometer at ANSTO uses neutron diffraction for measuring residual stresses. The diffracted radiation is collected by a PSD detector of area 30 x 30 cm$^2$. A bent silicon monochromators was set at a take-off angle of 75.9° to generate a wavelength of approximately
1.6739 °A. Primary and secondary slits were used to generate a gauge volume of 0.5 x 0.5 x20 mm$^3$ for measuring all the principal directions, which include the direction of cladding or parallel to laser scan (T), perpendicular to cladding direction in the plane of cladding (V) and in a direction normal to the plane of cladding (N). The data measurements were conducted at two different scattering angles for two different types of steel specimen. For the austenitic 316L, the value of scattering angle was set as $2\Theta \approx 101.6^\circ$ while for the ferritic H13 tool steel the set value was $2\Theta \approx 90.4^\circ$. For calculating the residual stresses in relevant directions, following formula as given by Kohler et al [199] and shown in equation 5.1 has been used:

$$\sigma_i = \frac{E_{hkl}(1-\nu_{hkl})}{(1+v_{hkl})(1-2v_{hkl})} \varepsilon_i + \frac{v_{hkl}E_{hkl}}{(1-2v_{hkl})(1+v_{hkl})} \sum_j \varepsilon_j, \ j = 1,2,3 \text{ and } j \neq i \quad (5.1)$$

where $\sigma_i$ and $\varepsilon_i$ are the principal stress and strains respectively and $E_{hkl}$ and $v_{hkl}$ are the Young’s modulus and Poisson’s ratio in a direction perpendicular to the diffraction planes $\{hkl\}$. For the austenitic steel $\{hkl\}$ index was $\{311\}$ while for ferritic steel the analysed diffraction plane was $\{211\}$. The values of E and v for these planes are given as:

$E_{211} = 220 \text{ GPa}, \ v_{211} = 0.28, \ E_{311} = 175 \text{ GPa}, \text{ and } v_{311} = 0.31.$

5.4.5. Results and Discussion

The residual stresses as found by neutron diffraction method show a gradual change not only in the magnitude but also in the type of stress. The behaviour shown is different for the two types of steel alloys investigated. In addition, the pattern and nature of residual stresses also exhibit some variation with the laser power employed during cladding. The measurements were made for three dimensional state of strain, but since the stress in the normal direction is negligible therefore, at every point a plane state of stress is evaluated.

From the bottom surface of substrate to the interfacial region, the stresses show a concave pattern. This means that in the substrate region, stresses are generally tensile at the bottom but rapidly change to compressive in the middle region. Near HAZ and re-melted zone residual stresses are generally found to be less than 100 MPa, due to tempering effects. The plots of plane state of residual stresses as measured by the neutron diffractometer for three different laser powers along the thickness of substrate and cladding are shown in Fig 5.12 for H13 specimen and in Fig 5.13 for 316L samples. In Fig 5.12(a) a vertical line is drawn to indicate the separate regions of substrate and DMD cladding. It must be appreciated that the values of residual stresses in this region cannot be estimated by joining the points ‘S’ and ‘C’ with a straight line, where ‘S’ is the last recorded point on substrate side and
‘C’ is the first recorded point on the cladding side. This construction applies to all the other plots as well.

Figure 5.12: Plots showing state of planar residual stresses across thickness of DMD cladded specimen (a) H13 - 1000 Watts (b) H13 – 1500 watts (c) H13 – 2000 watts
Figure 5.13: Plots showing state of planar residual stresses across thickness of DMD cladded specimen (a) 316L – 1000 watts (b) 316L – 1500 watts (c) 316L – 2000 watts
The important observations as enunciated from Figs 5.12 and 5.13 are presented below:

- The stress magnitudes and their nature (tensile or compressive) depends on the thermal expansion of material, size of the dilation component of the transformation and the elastic modulus of the material [200].

- State of stress in the top layer is dominantly compressive with the highest value of -427.5 MPa recorded for H13 cladded at 2000 watts. The existence of compressive residual stresses provides a better integrity of the overall structure and resistance to cracking and better hardness. This observation is proved during the impact testing and wear investigations with results presented in chapters 6 and 7 respectively.

- The distribution of residual stress is not symmetric across the thickness of clad and substrate. In every sample, the stresses at the bottom of substrate are tensile and at the top of cladding are compressive. The value of compressive stress shows a steady increase with the laser power and values are noted to be -427 MPa for H13 and -103 MPa for 316L cladded at 2000 watts. Inside the substrate near its middle region, stresses are invariably compressive except for H13 2000 watts.

- Compressive stresses at the top layers can be attributed to low martensitic transformation temperatures particularly in H13. Since steels with low martensitic transformation temperature produce compressive residual stresses in their martensitic region, therefore, the values of compressive residual stresses are higher for H13 tool steel specimen. The presence of fully developed martensitic microstructure in the top layers of H13 cladding can be verified by visualizing the micrographs in Fig 5.2.

- In all 316L specimen residual stress peaks in the vicinity of +200 MPa (tensile) are noted. As has been mentioned previously that in steel alloys, there are pronounced effects of phase transformation on dilation within the cladding, particularly the formation of martensite (α’) helps in generation of compressive stresses. Since 316L is an austenitic type of steel and due to rapid solidification of the laser melt pool, it is impossible for the high temperature austenite to undergo complete transformation into martensite and consequently, a certain quantity of retained austenite (γ’) would remain in the clad at the room temperature after cladding. The presence of the retained austenite which has less specific volume as compared to the martensite, however, adversely affects any efforts to reduce the magnitude of tensile residual stresses induced by thermal shrinkage in the clad.

- In Fig 5.12(a) the points S and C indicate the last recorded point in substrate and first recorded point in the clad respectively. The most noticeable aspect is that with the increase in laser power, for both the alloys, difference of residual stresses between S and C keeps on increasing. Table 5.4 illustrates the difference of residual stress values with the laser power.
Table 5.4: Difference in residual stresses between adjacent substrate and clad points across the interface

<table>
<thead>
<tr>
<th>Laser Power (Watts)</th>
<th>H13 tool steel</th>
<th>316L Stainless steel</th>
</tr>
</thead>
<tbody>
<tr>
<td>Residual stress at ‘S’ (MPa)</td>
<td>Difference in Residual Stress (MPa)</td>
<td>Residual stress at ‘C’ (MPa)</td>
</tr>
<tr>
<td>Direction</td>
<td>T</td>
<td>V</td>
</tr>
<tr>
<td>1000</td>
<td>5.63</td>
<td>30.72</td>
</tr>
<tr>
<td>1500</td>
<td>57.9</td>
<td>-79</td>
</tr>
</tbody>
</table>

5.5 Summary
The metallurgical investigations on the DMD cladded 316L and H13 high strength steel alloys were carried out in view of finding out the microstructure, micro-hardness and residual stresses induced during laser cladding. Since the main emphasis of this research is upon evaluating the capabilities of DMD setup to produce fully functional parts other than coatings, therefore, porous, solid and composite parts were developed and appropriately tested for metallurgical properties. The main purpose behind choosing porous and cylindrical geometry was that these configurations have not been reported as such for detailed material investigations and the same parts were later on used for compressive dynamic and wear testing.

The microstructure was largely heterogeneous and seen to be significantly dependent upon the energy density used during cladding. According to this investigation laser power has been the most important parameter influencing the material properties; therefore, the graphs are labelled according to the beam power for reference. The values of micro-hardness exhibit the range comparable to the hardness of wrought alloys, a fact that can be explained by appreciating the number of pseudo heat treatment cycles which the specimen undergo during numerous passes of laser beam during cladding process. The residual stresses calculated by the neutron diffraction method also exhibits encouraging sign with compressive stresses at the top cladded layer and in the middle portion of substrate. The magnitudes of tensile residual stresses were also appreciably less than the yield strength of material which provides for a compact structure with high hardness and good strength. These properties are evaluated in as-cladded condition without any post cladding heat treatment and therefore, these properties can be further improved by applying adequate annealing and tempering.

[101]
CHAPTER 6
COMPRESSIVE STRESS-STRAIN BEHAVIOUR UNDER STATIC AND
HIGH STRAIN RATE IMPACT LOADING

6.1 Introduction and Background

The objective of this chapter is to investigate the mechanical properties of DMD generated high strength steel alloy parts and ascertain their characteristics in the yielding zone through quasi static testing and under high strain rate compressive loading using a Split Hopkinson Pressure Bar (SHPB) apparatus. High strength stainless steel alloys demonstrate some peculiar characteristics that forbid a straightforward transition towards ascertaining their design performance based on the characteristics and behaviour of carbon steels. The noticeable deviating instances are non-linear stress strain relationship in the elastic region, no sharply defined yield point and substantial strain hardening in the plastic region of deformation [122]. That’s why the ensuing research on stainless steel alloys encompasses every possible investigation aspect which includes numerical modelling, finite element prediction and formulation of constitutive stress strain relationships [16].

The parts developed on DMD are actually made in a manner that involves creation of material and part simultaneously. Parts with complete shape and form are generated as a result of sintering of powdered material and subsequent layer by layer deposition over a “compatible” substrate through the assistance of high power and highly focussed laser beam. Unlike powder metallurgy, there are no compacting forces involved and there are process associated heat transfers which can result in generation of residual stresses and non-uniformities like micro-porosity and non-homogenous microstructures, irrespective of the materials cladded. Thus the most essential need for any laser generated part is to ensure the character of structures generated and the consistency of mechanical properties vital for employment in engineering applications like those mentioned above [5, 6].

This effort provides a useful insight into the quality of DMD generated specimen particularly under high strain rate compressive loads and also helps in diagnosing the abnormalities in their stress strain relationship which is expected for a novel method of production. The materials investigated are 316L stainless steel (SS) and H13 tool steel (TS). Mechanical properties of austenitic stainless steels from commercially available grades 304 and 316L have been investigated by many researchers covering diversified applications [17-18]. But there is a significant gap in the research on properties of these materials developed by laser cladding. Mechanical properties under focus are stress strain behaviour over elastic and plastic regions and determination of yield strengths and moduli of elasticity.
Another important aspect of this investigation is the comparative study of different configuration of parts produced on DMD. These include fully solid parts, those having cavities or macro-pores or “porous specimen” and the ones which comprise of two different materials and called as “composite parts”. Reason for this approach is to look into more functionality of laser generated parts and finding out the probable means of cost savings for similar nature of usage. The proved strength and toughness of porous parts developed on DMD particularly under impact loads can open a vista of opportunities for developing honeycombs and sandwiched structures with customized shapes and configurations for a number of wide ranging engineering applications. The composite parts are prepared with low cost and easily available mild steel at the core and with DMD cladding of stainless and tool steel on the outer surface in a 2:3 ratio by volume respectively. If their stress-strain characteristics under high impact loading prove themselves to be 80% of the strength of the fully cladded or solid laser generated specimen then it will be a significant economic achievement.

This research also extends the investigation into mathematical modelling of deformation characteristics of laser generated parts in the light of established models already derived for commercially available stainless steels. Subsequent comparison with experimental results indicates close similarities between theoretical and experimental curves in the yielding region but significant deviation in the plastic region.

6.2. Experimental Methods:
6.2.1. Sample Preparation:
All the specimens were produced on the POM-DMD machine operating with a CO₂ laser that can generate beam power up to 5kW. Like in all LAAM processes the composition and micro-structure of any part or coating created on DMD machine depends on a set of machine and process parameters, which are controllable during operation and the final outcome depends on their optimum combination. The most important parameters are laser power (in watts), laser beam diameter (in millimetres), laser scanning speed (in mm/sec) and powder feed rate (in g/min). Sometimes a term “Specific Energy” is defined as: laser power/ (beam diameter x processing speed) [19]. The value of specific energy for the deposition varies from 375 to 1000 Joules/mm². Riza et. al [20] have earlier described how the combination of DMD process parameters affect the prospects of developing different types of structures particularly when achieving the accuracy of dimensions comparable to the minimum possible width of laser beam and maximum possible height of deposited layer. Table 6.1 shows the optimized DMD process parameters for creating 316L SS and H13 TS solid and porous specimen.

Table 6.1: DMD Process parameters
<table>
<thead>
<tr>
<th>Parameter</th>
<th>Unit</th>
<th>Value</th>
</tr>
</thead>
<tbody>
<tr>
<td>Laser energy</td>
<td>Watts</td>
<td>500 - 700</td>
</tr>
<tr>
<td>(Beam) S- Focus*</td>
<td>Degrees</td>
<td>-15</td>
</tr>
<tr>
<td>Scan speed</td>
<td>mm/min</td>
<td>30-80</td>
</tr>
<tr>
<td>Powder feed rate</td>
<td>g/min</td>
<td>3–4</td>
</tr>
<tr>
<td>Cover gas flow (Argon)</td>
<td>Litres/min</td>
<td>2</td>
</tr>
<tr>
<td>Carrier gas flow (Argon)</td>
<td>Litres/min</td>
<td>7</td>
</tr>
<tr>
<td>Carrier gas flow (Helium)</td>
<td>Litres/min</td>
<td>2</td>
</tr>
<tr>
<td>Shielding gas flow (Argon)</td>
<td>Litres/min</td>
<td>18</td>
</tr>
<tr>
<td>Shielding gas flow (Helium)</td>
<td>Litres/min</td>
<td>5</td>
</tr>
<tr>
<td>Shaping gas flow (Argon)**</td>
<td>Litres/min</td>
<td>10</td>
</tr>
</tbody>
</table>

*S-focus is an optical machine parameter that controls the beam spot size. For the value -17, beam spot size is from 0.9-1 mm.

**Shaping gas is also called nozzle gas that shapes the powder stream coming out of the nozzle into the focus of laser beam and subsequently deposited on the substrate after melting.

It is noted from the material properties literature [202] that 316L SS and H13 TS have significant differences in their mechanical properties which are not entirely independent of their chemical composition. The chemical composition of the two powders obtained from supplier’s data is presented in Table 6.2.

**Table 6.2: Chemical Composition of metallic powders used in Laser Cladding on DMD**

A. 316 L stainless steel powder supplied by SULZER Metco Australia

<table>
<thead>
<tr>
<th>Element</th>
<th>Iron</th>
<th>Chromium</th>
<th>Molybdenum</th>
<th>Nickel</th>
<th>Manganese</th>
<th>Silicon</th>
<th>Carbon</th>
<th>Other</th>
</tr>
</thead>
<tbody>
<tr>
<td>Composition % wt.</td>
<td>62-72</td>
<td>16 -20</td>
<td>2 - 4</td>
<td>10 - 14</td>
<td>1</td>
<td>2-3</td>
<td>0.03</td>
<td>&lt;0.5</td>
</tr>
</tbody>
</table>

B. H13 tool steel powder supplied by Alloys International Australasia pty Ltd.

<table>
<thead>
<tr>
<th>Element</th>
<th>Iron</th>
<th>Chromium</th>
<th>Molybdenum</th>
<th>Niobium</th>
<th>Manganese</th>
<th>Silicon</th>
<th>Carbon</th>
<th>Other</th>
</tr>
</thead>
<tbody>
<tr>
<td>Composition % wt.</td>
<td>Balance</td>
<td>5 – 7</td>
<td>1.5 - 2</td>
<td>1 - 2</td>
<td>0.4</td>
<td>1</td>
<td>0.35</td>
<td></td>
</tr>
</tbody>
</table>

In total 30 solid and porous specimens were produced by the laser assisted DMD process, which ascribes 3 specimens per material per type for each test. Composite parts are produced by cladding of 316L and H13 powders on the outer cylindrical surface of a 5 mm diameter mild steel bar and were used only for high strain rate impact testing. The final outer diameter of composite parts was
maintained as 8 ± 0.25 mm. Specimen of desired lengths were then cut to accurate required dimensions by a SiC blade on high speed Secotom cutting machine. For the porous parts, shape of cavities is designed to be semicircular for quasi-static testing and quarter circular for dynamic testing. Subsequent to cladding, all the solid and porous specimen were turned to the required size and then parted off from the mild steel substrate. Fig. 6.1(a) & (b) shows the specimens each for porous H13 and 316L materials for dynamic testing along with the corresponding CAD model. Fig. 6.1(c) & (d) shows machined specimen and corresponding CAD model to be used for quasi-static testing.

Three identical specimens were tested for each case to get average results. Table 6.3 presents the necessary data including length and diameter for all the finished parts that are used in this experimental investigation.

<table>
<thead>
<tr>
<th>Material</th>
<th>Type</th>
<th>Quasi Static Testing</th>
<th>% of Solid Volume</th>
<th>High Strain rate testing</th>
<th>% of Solid*/cladded** Volume</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td></td>
<td>Outer Dia.</td>
<td>Length</td>
<td></td>
<td>Outer Dia.</td>
</tr>
<tr>
<td>H13</td>
<td>Solid</td>
<td>4.06</td>
<td>3.89</td>
<td>-</td>
<td>8.96</td>
</tr>
<tr>
<td>H13</td>
<td>Porous</td>
<td>8.1</td>
<td>4.53</td>
<td>73.43</td>
<td>9.02</td>
</tr>
<tr>
<td>H13</td>
<td>Composite</td>
<td>-</td>
<td>-</td>
<td>-</td>
<td>8.08</td>
</tr>
<tr>
<td>316L</td>
<td>Solid</td>
<td>4.03</td>
<td>3.71</td>
<td>-</td>
<td>8.97</td>
</tr>
<tr>
<td>316L</td>
<td>Porous</td>
<td>8.06</td>
<td>3.91</td>
<td>73.16</td>
<td>8.95</td>
</tr>
<tr>
<td>316L</td>
<td>Composite</td>
<td>-</td>
<td>-</td>
<td>-</td>
<td>8.1</td>
</tr>
</tbody>
</table>

**6.2.2 Testing Procedures:**
Out of the total, 18 finished specimens were then tested for high strain rate compressive loading on a Split Hopkinson pressure bar (SHPB). This equipment provides the opportunity to test materials at strain rates from 50 s⁻¹ to around 10⁴ s⁻¹, which is sufficient to simulate ballistic impacts, explosion scenarios and high speed crashes [203]. Figure 6.2 illustrates the mode of operation of SHPB and the dimensions of actual setup on which the experiment was performed [204].
The Split Hopkinson pressure bar or Kolsky compression bar is a combination of two long symmetrical bars of uniform cross-section and usually made of the same material. The general rule for selecting the bar material is that the yield strength of bar always remains higher than the maximum compressive strength of specimen for the corresponding strain rate [40]. A bearing and alignment fixture allows the bars to move freely while retaining precise axial alignment. A high velocity elastic stress wave is generated by impact of the striker (energized by high pressure gas) on the incident bar. Since the tested parts are made from high strength steel alloys, a maximum available gas pressure of 0.25 MPa was used to create high strain rate impacts at more than 2500s\(^{-1}\). When the compression wave in the incident bar propagates to the interface between the incident bar and the specimen, part of it is reflected back into the incident bar while the rest is transmitted into the specimen and compresses the specimen. The interaction of the stress waves in the specimen with the specimen/transmission bar builds the profile of the transmitted signal. By suitable measurement techniques involving strain gauges, oscilloscopes and data acquisition system and by the application of elastic wave theory, details of the applied disturbance can be reconstructed. Since the bar remains elastic, it can be used to measure either loads or displacements or both [205, 206].

The signals in the form of incident, transmitted and reflected waves, which eventually transform into stresses and strains are measured by the strain gauges. Strain gauge A measures the incident and reflected waves while transmitted wave signals are measured by strain gauge B. Fig. 6.3(a) shows all these waveforms and their sequential relationship along the time scale, while Fig. 6.3(b) represents the matched reflected and transmitted waves which actually provide the values of stress and strain respectively.
In all, 12 parts were tested on a servo-hydraulic MTS 7973 machine of 250 kN capacity for quasi static testing. The machine is equipped with load cells, extensometers and the data acquisition software that provides the load vs. displacement graph simultaneous to the loading and deformation of the specimen between the moving platens. The strain rate during the quasi-static tests was maintained at 0.001 s⁻¹.

6.3. Results and Discussion

The stress-strain curves are derived from the quasi-static and high strain rate dynamic testing for solid, porous and composite specimen of 316L stainless steel and H13 tool steel. High strain rate dynamic curves provide a good insight into the toughness, dynamic yielding strength and impact resistance of the laser deposited material but contain insufficient information for behaviour in the elastic/static yielding region and on response to continuously applied loading with small strain rates. This gap is filled by the quasi static curves which analyse the behaviour of two grades of high strength DMD generated steel alloys under large compressive loads in (i) the yielding region and (ii) up to 40 % strain, which is considered to be the extreme level of true strain that can be endured at satisfactory performance levels. Due to vastly different types of responses to compressive loading at quasi static strain rates (0.001 s⁻¹ and high dynamic loading strain rates (> 2500 s⁻¹), the results are described in separate sections.

6.3.1 High Strain Rate Dynamic Testing:

Split Hopkinson Pressure Bar (SHPB) is used to develop stress strain curves for 316L and H13 laser cladded specimens of solid and porous types. Main feature of this setup is to check the response of specimen under high strain rate conditions simulating high velocity impacts. The initial dimensions of the solid specimen tested were approximately 9 mm outer diameter and 6.25 mm length. Composite specimens were tested at almost double the strain rate at a specimen length of around 3 mm. The
dimensional variations during sample preparation were kept within ± 0.1 mm. Careful measurement of dimensions after the impact, as shown in Table 6.4, indicate the trends representative of very high material strength and superior stiffness. The volumetric strain for H13 is three times less than 316L for fully solid parts which is in accordance with the values of strength of H13 to be 3 times the strength of 316 L as can be observed from manufacturer’s data [202]. But since composite specimens have a mild steel core, therefore, the values of volumetric strain were found to be in the ratio of 1:5 for H13 against 316L.

Table 6.4. Specimen dimensions as measured after dynamic testing

<table>
<thead>
<tr>
<th>Type of Specimen</th>
<th>Outer Diameter (mm)</th>
<th>Length (mm)</th>
<th>Change in volume (mm³)</th>
<th>% change in volume</th>
</tr>
</thead>
<tbody>
<tr>
<td>H13 TS solid</td>
<td>9.19</td>
<td>5.97</td>
<td>-1.6078</td>
<td>-0.4044</td>
</tr>
<tr>
<td>H13 TS porous</td>
<td>10.31</td>
<td>5.82 (*)</td>
<td>-</td>
<td>-</td>
</tr>
<tr>
<td>H13 TS composite</td>
<td>8.18</td>
<td>2.73</td>
<td>-8.3028</td>
<td>-5.473</td>
</tr>
<tr>
<td>316L SS solid</td>
<td>9.54</td>
<td>5.49</td>
<td>-5.1807</td>
<td>-1.303</td>
</tr>
<tr>
<td>316L SS porous</td>
<td>10.81</td>
<td>5.43 (*)</td>
<td>-</td>
<td>-</td>
</tr>
<tr>
<td>316L SS composite</td>
<td>9.35</td>
<td>1.61</td>
<td>-41.447</td>
<td>-27.279</td>
</tr>
</tbody>
</table>

*For porous parts it is very difficult to calculate accurate changes in volume by measuring exterior dimensions only due to flow of material inside the cavities

In SHPB system, the impact was initiated through a striker bar excited by a gas pressure of 0.25 MPa that transferred its energy to the incident bar, which in turn generated an impact velocity of 21.98 m/s. For a specimen of length 6.25 ± 0.1 mm, this amounts to a nominal strain rate (NSR) in the range of 3461-3573 s⁻¹ while for 2.9 ±0.1 mm composite specimen, the value reaches in the vicinity of 7320 – 7850 s⁻¹. Composite specimens are tested at higher strain rates to thoroughly ascertain the characteristic of interface or bonding layer between the substrate and cladded material. When equilibrium state is achieved, the actual strain rate is a function of wave propagation velocity and the amplitude of reflected wave. Reflected wave inherently carries the information regarding deformation of the specimen struck by the incident bar and thus the plot of strain rate vs. time illustrates the extent of deformation which the specimen undergoes in response to the impact and the mode of dissipation of impact wave. Higher values of actual strain rates (ASR) indicate more deformation and lesser time to zero value denotes better impact absorption capability or higher toughness.
Figure 6.4: Graphs of true stress, true strain and actual strain rates with respect to time for (a) Solid H13 (b) Porous H13 (c) Solid 316L (d) Porous 316L (e) Composite 316L, and (f) Composite H13 specimens under high strain rate compressive dynamic loading

Fig. 6.4 shows the values of strain rate, true stress and true strain as calculated and plotted against time for all the six types of specimens. Strain values are scaled up by a factor of 1000 while strain rates are scaled down by a factor of 5 to make possible the meaningful presentation of three quantities on the same graph. A close study of all the graphs presented in Fig. 6.4 leads to the following results:
a) As expected according to material properties, DMD generated H13 TS is stronger and stiffer than 316L SS. The solid H13 TS shows a maximum ASR of 2228 s\(^{-1}\) while 316L goes up to 3391 s\(^{-1}\) under same values of impact velocity.

b) For solid H13 specimen, the time taken by strain rate to drop to zero value is just 20 μs as compared to 100 μs for solid 316L. This indicates demise of reflected wave and clearly proves high impact energy absorption capability of laser DMD generated H13 TS specimen.

c) For porous specimen of both material types, actual strain rate (ASR) values are higher than the solid parts and in fact reach 3800 s\(^{-1}\) which is greater than the applied strain rate. This may be due to the resonating effects in the reflected wave and is expected due to presence of cavities in the porous parts.

d) The composite specimen with smaller lengths show much higher values of ASR for both the material types but the trend exhibited is very similar to the one shown for solid parts. For solid specimen ASR of H13 is approximately 1.5 times that of 316L while for composite parts this ratio drops slightly to 1.4. This may be attributed to the presence of mild steel in the core and only 61% of high strength material by volume in the whole specimen.

e) There is a distinct difference in the manner in which the strain spike dies down among the two materials. In H13 TS, the spike instantly gets back to nearly zero with subsequent negligible vibrations, but in 316L SS, the ASR attains zero value gradually after a delay of 0.001 s. It clearly proves the superior impact resistance and energy absorption capability of laser cladded H13 TS.

f) When compared to the variation of stress vs. time, the values of strain in all cases is lagging by an interval of 20μs. This is also accompanied by an initial interval of zero strain. This behaviour corroborates with the behaviour in quasi static testing and strongly suggests the presence of micro-porosity in the laser generated specimen.

g) The peak value of stress is in the range 1250-1300 MPa for solid specimen and in the range 925-975 MPa for porous parts. The difference in stress values is accompanied by an increase in strain from 0.0591(solid) to 0.0969 (porous) for H13 and from 0.161(solid) to 0.194 (porous) for 316L parts. These values prove the capability of porous specimen to absorb the impact energy by deforming more at lower stress values. This is an indicator that highlights the advantageous use of porous structures in preparing blast and explosion proof lightweight structures without sacrificing overall strength.

h) For composite specimen the stress rises to 1156 MPa for 316L while it reaches a peak value around 1385 MPa for H13. The accompanying values of strain are 0.3 and 0.09 for the two materials respectively. This is a very peculiar behaviour which illustrates H13 composites showing a behaviour similar to H13 porous specimen but proving to be a softer material in case of 316L composites. In comparison the strain is 1.5 times higher for 316L composite than
evaluated for porous cladded specimen of the same material. This brings us to an important conclusion that when the difference in strength among core and clad materials is vastly tilted in favour of cladding then the properties of cladded material, H13 in this case, will determine the overall behaviour. Thus composite H13 specimen mostly behave as the H13 parts with core treated as pseudo-cavity, but in 316L specimen since the strength of core and clad is comparable, therefore, the softer core made of MS demonstrated its significance through greater overall deformation than the case when parts were porous but completely developed from 316L cladding.

i) Another noticeable aspect is the time at which stress values become maximum. For H13 TS this time is synchronous with the peak strain rate but in all other cases, this peak is achieved at the end of second region or before initiation of stress drop down. This means that for both the porous and 316L solid specimen, deformation and stress builds up simultaneously until the impact is absorbed, while for H13 solid part, major portion of impact was absorbed as soon as the impact was received.

Fig. 6.5 shows the graphs of true stress and true strain, which are calculated using elastic wave theory and plotted for all the six types of specimens. There are three distinct and noticeable regions of the stress strain curves explained as follows:

i. The first region in dynamic stress-strain curves in every graph of Fig. 6.5 exhibits a rapid rise of stress which is observable in all configurations and corresponds to the almost instantaneous peak surge in strain rate at 0.003 s due to impact of incident bar. In this initial region true strain has a range of 0 to 0.02 during which H13 TS shows stress rise up to 1250 MPa while the maximum value for 316L SS is 800 MPa. Stresses created by the impact are within the elastic range of H13 TS but much beyond the yielding stress of 316L SS. Therefore, we see the peculiar manner in which the stress rises to its peak value in 316L solid, composite and porous specimen. As previously shown in Fig 6.4 that the stress rise for 316L parts is accompanied by large strains i.e. greater than 10%, which indicates existence of plastic deformation. For instance in the stress strain curve for 316L solid part, it can be observed that the stress rises at constant slope up to a value of 545 MPa which corresponds to the ultimate strength of material and then the graph shows an increase in stress up to 801 MPa but strain also increases to 2.14%. This is the initial region of work hardening and plastic deformation.
Figure 6.5: True Stress vs. True Strain curves under high strain rate dynamic loading for H13 TS and 316L SS solid, porous and composite specimen.

ii. Second region visible as the relatively flat portion in every plot of Fig. 6.5 is the stress plateau after impact and is accompanied by increasing strain. It encompasses the zones of elastic and plastic deformations both, as the incident pressure wave is transmitted through the specimen. Another noticeable factor in this region is that apart from H13 TS solid specimen, the value of
stress is increased with respect to the value in first region. This is the zone of energy absorption and plastic deformation.

iii. Third region which is more conspicuous in Figs. 6.5 (c) & (d) occurs when stress begins to drop towards zero value after the termination of impact related effects and later accompanied by recovery of elastic strain. For H13 solid specimen, the maximum strain is 0.0591, subsequently recovered to 0.0464, which remains as the final plastic strain. On the other hand 316L SS is compressed to a strain of 0.161, from where it recovers to the final value of 0.1478. As the ultimate strength of 316L is 3 times less than that of H13, therefore, the values of final plastic strain are corresponding to the ratio of strengths like the volumetric strain ratio given earlier.

iv. The main difference between the stress-strain curves of solid and porous specimen is in the initial slope encompassing elastic and yielding region. For both H13 and 316L solid parts, yield stress is achieved at nearly 0.2% strain, after which H13 part deforms upto 5% strain and 316L part experienced upto 14% strain in the plastic region. But in contrast, porous H13 was deformed by 4% of its original length at a true stress value of around 850 MPa (much below the yield strength value of 1200 MPa) while 316L porous part went through a strain of 4% at the stress value in the vicinity of 775 MPa. Afterwards, for porous parts of both the materials, a steady increase with a much lesser slope was observed in true stress value upto 15% strain. Since the porous specimen did not break after undergoing such large value of strain and the strain energy during an impact is the product of stress and strain, therefore, the observed stress-strain behaviour shown in Fig 6.5 is very encouraging for absorption of shock energy as it undergoes large amounts of strain without impending fracture.

v. For the composite parts, the stress strain curves under dynamic loading show very different behaviours for 316L and H13 cladded specimen. For H13 part, the initial peak stress goes up to 1284 MPa with a true strain of 0.48 %. But unlike solid part, the stress increases to 1385 MPa accompanied by a true strain of 8% just at the beginning of third region. It can be observed that true stress is increased by about 9% but the true strain value is incremented by about 40% up to the onset of third region where stress begins to drop. This is due to the presence of mild steel core but the behaviour is mainly governed by the stronger material which is H13 and not the softer MS core. But for 316L composite, the stress level is more or less same for the three specimen types, but in composite parts the strain is very large even when compared to the porous parts which shows the strong influence of MS at the core, and corroborates the derived postulate that in composite parts, if the strengths are comparable then the stress-strain properties are not entirely
governed by higher strength cladding but softer core too has strong influence on the eventual values.

Solids, porous and composite specimens show different behaviours related to absorption of impact. But particularly noticing the behaviour of porous parts, important conclusions can be drawn regarding their impact absorption capabilities. As illustrated in the curves of Fig 6.5 the initial region for porous specimen indicates a higher strain which is 4% for H13 and 3.5% for 316L. In the beginning high strain is due to the presence of cavities, but when the pores close down and the walls of cavities come closer, value of stress continues to rise. In between first and second region the stress varies from 815 to 1072 MPa for H13 and from 748 to 893 for 316L porous specimen. The values of maximum strain as the stress approaches zero are 17.78% for H13 and 22.5% for 316L respectively. Strain recovery is 1.2% for both H13 and 316L porous parts. This demonstrated behaviour of very little stress variation during absorption of impact and extended region of strain hardening without rupture or crushing is suitable for energy absorption and dissipation applications.

6.3.2. **Quasi Static Stress Strain curves**:

The dynamic curves provide very useful information regarding stress strain behaviour at high strain rate dynamic conditions but are rarely useful for understanding the elastic behaviour and static yielding of the parts. Therefore, solid and porous specimens made of two steel types were examined under uniaxial and continuously applied compressive loads at strain rate of 0.001 s\(^{-1}\). Since it has been observed that the behaviours of composite specimen are governed by either of H13 and 316L solid parts, therefore, these are not investigated separately. The stress strain relationship covers elastic and plastic regions both. A noticeable observation during testing was that the solid pieces were almost flattened (crushed) by the compressive forces without breaking or fracture, but porous pieces showed lesser length reduction due to internal space available for lateral expansion within the cavities. Fig. 6.6 shows the true stress versus true strain curves plotted for all four types of specimen. Some important observations inferred from these curves which are not available from the dynamic curves are presented as follows:

1. All the curves are non-linear in elastic and plastic regions both as expected from the stress strain behaviour of high strength steel alloys [207]. The four curves, although differing with each other in numerical values of yield strength and modulus of elasticity, display a common trend of dual modulus of elasticity in the region before yielding. The exhibition of dual modulus in the stress strain curve can be explained in terms of micro-porosity that is almost always present in parts which are produced by sintering or melting of powder. The inherent micro-porosity may act as a “notch” and results in abrupt change in slope of stress-strain curve within the yielding region [208].

[114]
2. The presence of dual slope within the yielding region generates two values of Young’s modulus ‘$E_1$’ and ‘$E_2$’ which are evaluated and presented in Table 6.5. The $E_1$ exists up to 0.32 % strain while $E_2$ covers the range from 0.35 to 3% strain. Values of $E_1$ for both the materials are approximately half as compared to $E_2$. It means appreciable deformation in the beginning without significant values of stress. The resulting slope of both the curves confirms that the value of compression modulus of the produced materials is quite low in comparison to the values found in manufacturer’s catalogues and handbooks for the commercially available grades.

3. There is not much information available on the exhibition of low modulus of elasticity for laser cladded steels but very recently de Lima and Sankare [132] have reported the exactly same behaviour for specimen produced by laser based additive manufacturing of 316L powder on the substrate of same material. Although the values reported in this chapter are not that low but in comparison to those for rolled sheets are still low enough as is clearly evident from Table 6.5.

Figure 6.6: True Stress strain curves for H13 TS and 316L SS solid and porous specimen under quasi static loading

(a) (b) (c) (d)
Table 6.5: Young’s Moduli in the Yielding elastic region

<table>
<thead>
<tr>
<th>Material</th>
<th>Zone designation</th>
<th>Range in % strain</th>
<th>Young’s Modulus (Experimental) (GPa)</th>
<th>Young’s Modulus - commercial grade alloys (GPa) [202]</th>
</tr>
</thead>
<tbody>
<tr>
<td>Stainless steel</td>
<td>Primary Yielding</td>
<td>0 to 0.32</td>
<td>(E₁) 17</td>
<td>193 – 200 (Annealed)</td>
</tr>
<tr>
<td>316L</td>
<td>Secondary Yielding</td>
<td>0.35 to 2.7</td>
<td>(E₂) 38</td>
<td></td>
</tr>
<tr>
<td>Tool steel H13</td>
<td>Primary Yielding</td>
<td>0 to 0.3</td>
<td>(E₁) 26</td>
<td>210 (Hot Worked)</td>
</tr>
<tr>
<td></td>
<td>Secondary Yielding</td>
<td>0.35 to 2.2</td>
<td>(E₂) 45</td>
<td></td>
</tr>
</tbody>
</table>

4. Since high strength steel alloys do not exhibit a distinct yield point, this prompts the use of 0.2% proof stress as a measure of yield strength for such materials [209]. The values of 0.2% proof stress for 316L SS is in the range 250 - 325 MPa and for H13 TS, 0.2% proof stress is evaluated to cover the range 1225 - 1300 MPa which is 30% lesser than the yield strength of quenched H13 tool steel. And since these values show good repeatability, it proves that the steels “produced” by laser assisted DMD process are competitive enough in strength and ductility values to commercially available high strength steel alloys.

An important characteristic of “strengthening under dynamic loading” is also evident when static and dynamic curves are viewed together. Wu and Jiang [210] reported 74% increase in crush strength of honeycomb structures under dynamic loading as compared to quasi-static loading conditions. Deshpande and Fleck [211] mentioned the fact, based on substantial research evidence, that compressive strength of metallic foams remains insensitive to applied strain rates of 1000 s⁻¹. In our investigation, this behaviour is significantly evident for both 316L and H13 solid and porous parts. In dynamic curves, the stress of 800 MPa soon after impact is accompanied by a strain of 2.2%, while under quasi-static loading, this value of stress is achieved after 9.7% strain. For H13 this dynamic hardening effect is also very prominent as stress level reaches up to 1192 MPa at 0.1% strain while under quasi-static loading equivalent values are obtained at strain of 4.7%. This effect may be due to the sudden action of impact of dynamic loading, which brings the part beyond elastic limit into the regions of strain hardening almost instantaneously. This causes the expanded regions of static curve to squeeze and generate a curve with much steeper slope and thus more strength and hardening of the loaded material.
6.4. Constitutive Relationships

Due to significantly different behaviour of high strength steel alloys in comparison to carbon steels, there is a continuous research for developing mathematical models that can constitute the stress-strain characteristics of these metals. Rasmussen [209] in his detailed analysis developed relationships based on the classic Ramberg-Osgood relationships [212], which make use of the Ramberg-Osgood parameter (n) to extend the linear elastic region to include plastic deformation. Rasmussen’s equations can model the stress strain behaviour of stainless steels under tension as well as in compression in a piecewise manner while employing only three parameters, which include initial Young’s modulus, 0.2% proof stress and 0.01% proof stress.

The full range stress strain relationships proposed by Rasmussen for stainless steels are given in the following two equations:

\[
\varepsilon = \begin{cases} 
\frac{\sigma}{E_0} + 0.002 \left( \frac{\sigma}{\sigma_{0.2}} \right)^n \text{ for } \sigma \leq \sigma_{0.2} \\
\sigma - \sigma_{0.2} \frac{\sigma - \sigma_{0.2}}{E_{0.2}} + \varepsilon_0 \left( \frac{\sigma - \sigma_{0.2}}{\sigma_u - \sigma_{0.2}} \right)^m + \varepsilon_{0.2} \text{ for } \sigma > \sigma_{0.2}
\end{cases}
\]

In these equations the exponent ‘m’ is given as: \( m = 1 + 3.5 \frac{\sigma_{0.2}}{\sigma_u} \). The modulus to the curve at 0.2% proof stress is estimated as: \( E_{0.2} = \frac{E_0}{1 + 0.002 \left( \frac{n}{7} \right)} \) where \( n = \frac{\ln(20)}{\ln(\sigma_{0.2}/\sigma_{0.01})} \) and \( e = \frac{\sigma_{0.2}}{E_0} \).

Ashraf et al [213] also made use of the Ramberg-Osgood relation to present equations that define strain as a function of stress in two similar steps. In every case the initial range of strain exists from 0 to 0.2%, while the second equation covers every strain value beyond 0.2%. The equations derived by Ashraf et al [213] are more suitable for modelling stainless steel specimen under compression, because these take into account two very important features of high strength steel’s behaviour under compressive loading, which are (i) absence of ultimate stress and strain (\( \sigma_u \) and \( \varepsilon_u \)); instead these values are substituted with 1% proof stress (\( \sigma_{1.0} \)) and a correction factor, and (ii) inclusion of strain hardening co-efficient \( n'_{0.2,1.0} \). Later Quach et al [214] combined the analyses of Rasmussen and Ashraf et al to develop stress strain equations for stainless steels spread over three regions. First zone is the same but second zone is from 0.25 to 2% and the additional third region extends from 2% to all greater values. The full range of equations presented by Quach et al is presented below:
Based on the models proposed by Rasmussen and Quach et al, the stress-strain curves are plotted for 316L SS specimen and compared with the experimental true stress vs. true strain curve with the plots as shown in Fig. 6.9. Since in Rasmussen equations, the expressions for $\sigma_u$ and $\varepsilon_u$ are developed from the test data for tensile coupons, therefore, for compressive strains, these are adjusted according to the equations given by Quach et al and thus the ultimate compressive stress and strain is used according to the relationships as follows:

$$\sigma_{u,com} = \sigma_{u,ten} (1 + \varepsilon_{u,ten})^2 \quad \text{and} \quad \varepsilon_{u,com} = 1 - \frac{1}{1 + \varepsilon_{u,ten}}$$

For H13 TS, the theoretical and experimental curves are totally out of conformity and this mismatching is particularly due to huge difference between 0.2% proof stress and $E_0$. The parameters obtained from the experimental data and used in plotting the theoretical curves are $E_0 = 38.4$ GPa, $\sigma_{0.2} = 325$ MPa, and $\sigma_{0.01} = 51$ MPa. The values of Ramberg-Osgood parameter and the strain hardening coefficient are evaluated as 1.6863 and 7.926 respectively. Comparing to the values available in [209] and [212], it is obvious that ‘n’ is too small and $n'_{0.2,1.0}$ is exceedingly high; a behaviour that is also evident from the graphical comparison.

![Figure 6.9](image-url)

(a) (b)

Figure 6.9. Comparison of Mathematically modelled and Experimental stress strain curves for laser generated 316L SS (a) Normal view (b) Magnified view of the yielding region
It is important to note that in mathematical modelling of the behaviour of laser cladded 316L SS, the effect of primary yielding region is kept to a minimum in calculation of main parameters. As stated earlier, the primary yielding signifies the presence of micro-porosity, an effect that is not present in stainless steels produced by forming and forging. The theoretical and experimental curves are in close proximity up to 2% strain, after which the deviation becomes prominent. It can be observed in Fig. 6.9 (b) that the deviation is mainly due to high non-linearity and low compression modulus in the primary yielding zone for experimental curve, but as the values move beyond 0.5% strain, differences between mathematical model and experimental curve are narrowed to within +10%. Rasmussen curves show very little strain hardening while Quach curves indicate significant strain hardening effects. But the actual strain hardening rate as shown from experimental data is much higher than the values suggested by the mathematical model. This may be due to high hardness of the part (161 HV50) and low modulus of elasticity. This fact can be observed in the stress-strain power law for plastic region of deformation.

6.5. SEM and Microstructural analysis after impact

The DMD cladded H13 and 316L specimens exhibited enough strength to absorb the energy of impact without breaking but in most of the cases they deformed plastically. Particularly encouraging observations include the behaviour of porous and composite specimens, which, although undergoing large strains, retained their structure without disintegration. Such behaviour is much desirable in impact absorption applications. This section attempts to look into the post-impact surface characteristics and micro-structure of all material types and configurations. This will provide a good insight of the influence of impact upon the specimen at micro-granular scale and will also demonstrate the mechanical strength and integrity of laser assisted DMD generated high strength steel alloys. Initially a number of SEM micrographs are presented for the analysis of surface distortion or deformation followed by the presentation and discussion of the microstructures as developed post impact for all the six types of specimens. Major interest is in finding out the characteristics for solid specimen and the composite parts particularly at the interface between MS core and outside cladding of H13 and 316L steel alloys.
Fig 6.10 shows several SEM micrographs that reveal the surface condition of specimens after the high strain rate impact experienced in the Split Hopkinson’s bar. For solid specimen, there was no apparent damage to the surface but the high magnification picture with magnification factor greater than 5000 reveals intensive scratching and some material removal at the surface. Most noticeable effect can be observed in Fig 6.10 (b) which shows opening up of surface to expose the burnt particles deposited during cladding and formation of a continuous crack that obviously is not deep enough to cause disintegration. In Fig 6.10 (c) the condition of outside cladding on the composite specimen is shown after impact. It looks more bruised than the solid part, the reason may be the higher strain rate set during composite parts testing. But the structure remained intact and was never broken in a single instance during three experiments that were conducted. At the interface in low magnification micrograph as shown in Fig 6.10(d), greater distortion can be observed while within the core, few cracks are noticeable. It appears that softer core provided a cushioning effect as it shows more deformation under impact than the surrounding cladding. But overall the structure maintained its configuration and did not show any fracture or macro cracks.
In the micrographs of Fig 6.11 the post impact surface condition of H13 solid and composite specimen is presented. The phenomenon conspicuously visible is less intensity of surface distortion in case of H13 parts. This outcome can be expected after looking at the dynamic curves for both the materials as presented in Fig 6.5. The surface as shown in Fig 6.11 (b) appears to be scratched and bruised but not severely gouged as observed for 316L in Fig 6.10. There is an impact induced long and shallow opening in the surface visible in Fig 6.11(a) that is not deep but it amply exposes the deficiencies inherently associated with the development of parts using a laser cladding process like some degree of micro-porosity and lack of compacting forces. For the H13 composite part, the larger depression of MS core in comparison to the outer H13 cladding demonstrates the difference in strength of two materials. Another noticeable difference is the accumulation of fine burnt particles at the interface. This may be due to the unmelted carbon present in the bulk of cladded material as H13 has significant carbon percentage.

Figure 6.11: SEM micrographs of H13 solid and composite specimen after impact (a) H13 solid (x4.55k) (b) H13 solid (x2.88k) (c) H13 composite - clad (x3.13k) (d) H13 Composite - Interface (x1.25k)
The microstructural study of DMD generated solid and composite specimen from 316L and H13 powders is also carried out to determine the extent of grain dislocation or grain boundary distortion due to high velocity impact which these specimen experienced in Split Hopkinson bar. There were no apparent signs of damage or fracture in the surfaces when inspected visually. For comparison, the microstructures of 316L and H13 parts before impact are presented in Fig 6.12 as determined after etching in the optical microscope. Fig 6.12 (a) shows the typical austenitic structure of 316L stainless steel with grain boundaries and some d-ferrite stringers.

Fig 6.12 (b) shows the fine dendritic microstructure at the interface within the cladding which are not developed enough to exhibit distinct grain boundaries due to low power cladding. The microstructure of H13 solid part shown in Fig 6.12 (c) vividly demonstrates the martensitic structure which may be regarded as similar to the microstructure of parts produced by powder metallurgy. Fig 6.13 illustrates the post impact microstructures of 316L solid and composite parts. The most important and recognizable effect is the re-arrangement of grain structure into round or curved patterns thus exposing
the grain boundaries to a great extent as very clearly illustrated in the low magnification picture of Fig 6.13(b).

Figure 6.13: Microstructures of 316L solid and composite specimen after impact (a) 316L solid (x500) (b) 316L solid (x50) (c) 316L Composite clad and interface (x50) (d) 316L composite - clad (x200)

The detailed microstructural features shown in Fig 6.13(a) at a magnification factor of 500 also reveal re-arrangement of grain boundaries and elongation of austenitic grains in comparison to the same magnification optical micrograph of Fig 6.12 (a) for solid 316L specimen after high strain rate impact. Same pattern can also be observed over a larger area within the 316L cladding in the composite part at a lesser magnification factor of 200 as illustrated in Fig 6.13(d). But for 316L solid part, at a magnification factor of 50, the overall microstructure shows significant distortion enforced upon originally uniform arrangement of grain structure followed by emergence of curved grain boundaries due to high force of impact, as shown in Fig 6.13 (b). The effect of grains overlapping upon one another can also be visualized at the interface of composite 316L specimen in Fig 6.13(c) as the impact causes two different materials to penetrate into each other and redefining the grain structure at the interface.
The microstructural changes that occurred in H13 solid and composite parts can be observed in optical micrographs of Fig 6.14. The grain structure is clearly jumbled up for solid H13 parts shown in Figs 6.14 (a) & (b) at magnification factors of 200 and 500 respectively. In comparison to the microstructure before the impact as shown in Fig 6.12(c), dark ferritic segregation is conspicuous, otherwise there are no significant signs of grain dislocation and widening of grain boundaries as observed for 316L parts. Within the H13 cladding, disorientation of grain structure can be observed with ferritic regions, which shows the rearrangement and dispersion of martensitic structure into fine grains dispersed into ferritic matrix as observed in Fig 6.14(c). The composite H13 specimen shows the elongation of grains at the interface with some solid diffusion of two materials but the interface boundary can be seen as intact with a degree of distortion into curved profile, as shown in Fig 6.14(d). Comparison with the microstructure shown in Figs 6.12(c) & (d) it is evident that in H13 composite parts, the outer cladding of high strength material absorbs the bulk of impact and little effects are transferred to the softer core, which resulted in lesser distortion of the interfacial grain boundary and also lesser degree of solid diffusion at the interface.
6.6 Conclusions

Two high strength steel alloys 316L SS and H13 TS were used in powder form to create cylindrical specimen with three different geometrical configurations i.e. solid, with macro pores and composites of rolled MS bar and DMD cladding. All the six types were tested for strength and ductility under high strain rate dynamic conditions. Solid and porous specimen were also tested under quasi-static condition for determining the elastic and yield behaviour of the cladded material. All the specimen showed very good ability not only to withstand the axial load but also to bear the impact load at high velocity and were found to be close to most of the mechanical properties obtained for these alloys manufactured commercially. A noticeable shortcoming from manufacturing aspect is the lack of any forming and shaping forces involved during production of part that effectuates micro-porosity and fractional non-homogeneity in the microstructure. Due to this and absence of any post-processing operation, some inconsistencies are observed in mechanical properties. But the variations observed are not entirely random causing the specimen from the same material to behave entirely differently under similar testing conditions. Instead, the observed variations during quasi static testing are confined within ±10% of the average values presented in this chapter. Repeatability in dynamic testing is still better and restricted to ± 6% of the average values. The values for 0.2% proof stress and maximum elastic strength are very close to those available for commercial grade steels but the noticeable departure is low value of modulus of elasticity. This can be improved by heat treatment considerably.

The stress strain curves obtained are highly valuable and demonstrate the capability of DMD system to produce parts of high strength and stiffness that can withstand impact loads at high strain rate. The determination of dynamic properties for different configurations adds substantially to the mechanical characterization of DMD cladded high strength alloy steels. Lack of physical disintegration in porous and non-separation or cracking at the interface layer in composite parts presents a very significant advantage in employing DMD cladding for novel industrial applications. The advantage of DMD cladding of high strength steels becomes more prominent when viewed in the backdrop of specimen being tested in as-cladded condition without any heat treatment to alleviate residual stresses or enhance strength and ductility. This investigation opens the door of employing superior mechanical properties of H13 in applications much diversified and not limited to only tool manufacturing. For 316L SS, the behaviour can be regarded as highly satisfactory, because it bears the impact loads without fracture up to the stress values, which are much higher than the ultimate stress quoted in manufacturer’s catalogue.
CHAPTER 7

SLIDING WEAR BEHAVIOUR ON A PIN-ON-DISC APPARATUS

7.1 Introduction

Laser assisted DMD process provides a realistic opportunity to create solid and porous structures from high strength metallic alloys that can be used as coatings, foams and sandwiched structures and as highly stressed components. This chapter describes the investigation which evaluates the quality and capability of DMD generated high strength steel alloy parts to withstand severe sliding wear against mild steel counter face. This mode of wear investigation is chosen because steel rubbing against steel under high speeds and loads without the presence of any lubricant has been considered as one of the situations where co-efficient of friction and wear rate are in the highest range of values among metal on metal friction and wear. For laser cladded steels, which are generated by melting and subsequent deposition in a DMD process, friction and wear investigation is extremely necessary to ascertain the quality of their bulk against dis-integration and their surfaces against rapid distortion while bearing continuously applied tangential forces for extended periods.

Almost all the published wear tests involving laser cladding have investigated coatings, so this research attempts to close a gap through research on wear characteristics of complete parts produced by LAAM instead of laser coated specimen. The difference between part and coating can be understood in that the coating is spread over a wide area with little height or small thickness (up to 1mm); but the specimen tested in this study are made as 6 mm diameter pins with finished height of 5 mm. The experimental apparatus is the well-known pin on disc wear tester and the experimental procedure conforms to ASTM G99 standards. The main idea is to assess the wear behaviour of DMD generated high strength steels and compare their coefficient of friction and wear rate with the commercial grades of same type of materials.

The H13 tool steel and 316L stainless steel specimens were produced on DMD as cylindrical pins and annular paddings on mild steel substrate. The parts for wear testing were cladded as cylindrical pins and tested for sliding wear against a mild steel disc under severe conditions. Since the cylindrical pins produced by DMD process represent a new type of material and demands in-depth mechanical characterization, therefore, a thorough analysis for wear has been performed for several experimental regimes. The experimental data was tabulated and analysed graphically and statistically to ascertain the friction and wear properties of two different types of high strength steel alloys. The study also investigates the relationship of primary and secondary wear factors in terms of wear loss from the DMD specimen and the energy dissipation during sliding wear. The results highlight the strong and consistent behaviour of laser generated specimen under different conditions of abrasive wear and
exhibit very little signs of material degradation and extensive micro-chipping and flaking under high loads. The results were highly encouraging and certainly open a niche for using laser assisted DMD process to create customised parts from high strength metallic alloys for industrial applications. The results also indicate the importance of looking into the wear behaviour of steels from a combined perspective of major influencing factors instead of treating these individually and in a discrete manner.

7.2 Experimental Methods and Materials

7.2.1 Sample Preparation
There are three types of specimen produced by laser assisted direct metal deposition for this investigation. These include:

(i) 316L and H13 solid cylindrical pins with laser deposition on mild steel (MS) pins used as substrate. The cladded pins were turned and faced on lathe to produce a final diameter of 6 mm and a length of 5 mm. The diameter of substrate pin was 10 mm and its length 35 mm so that it can be conveniently held in the arm of wear tester.

(ii) 316L and H13 composite pins. The composite pin is actually a mild steel pin of diameter 5 mm over which a cladding of 1.5 mm thickness was done which resulted in final diameter of pin to become 8 mm. The cylindrical surface was ground and the end fine turned to finish the specimen for wear testing.

(iii) 316L and H13 annular padding on mild steel disc. Width of padding is 6 mm and its height is 3 mm. The mild steel disc was turned to the diameter of 165 mm and faced to create a flat surface before laser cladding. During preparation of ‘disc specimen’ the ongoing sequence of localized heating by laser and rapid cooling due to small heat affected zone resulted in slight warping of the cladded disc surface. Therefore, post cladding the disc was faced again to create a flat bottom surface and remove waviness during rotation.

All the specimens were laser cladded on the POM-DMD machine operating with a CO$_2$ laser that can generate beam power up to 5kW. In a DMD process the dimensional accuracy and surface finish of the part depend on the uniformity and repeatability of the clad height and width. Clad height and width are sensitive to many machine and process parameters, the appropriate combination of which results in producing the cladding with very low dilution, strong bonding with the substrate, good profiles of built section and appearance [138]. For sample preparation the selected DMD parameters for different types of specimen are presented in Table 7.1. The selected recipe is based on the consideration of minimum power consumption and adopting parameters that can produce fully dense structures with minimal micro-porosity.
Table 7.1: DMD Process Parameters used in sample preparation

<table>
<thead>
<tr>
<th>Parameter</th>
<th>Unit</th>
<th>Values</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td></td>
<td>Cylindrical pins</td>
</tr>
<tr>
<td>Laser energy</td>
<td>Watts</td>
<td>500</td>
</tr>
<tr>
<td>(Beam) S-Focus*</td>
<td>Degrees</td>
<td>-17</td>
</tr>
<tr>
<td>Scan speed</td>
<td>mm/min</td>
<td>60-80</td>
</tr>
<tr>
<td>Powder feed rate</td>
<td>g/min</td>
<td>2.5-3</td>
</tr>
</tbody>
</table>

*S-focus is an optical machine parameter that controls the beam spot size. For the value -17, beam spot size is from 0.9-1 mm.

7.2.2 Material Properties:

a) Chemical composition: The mechanical properties of cladded specimen for wear testing depend upon the chemical composition of 316L stainless steel and H13 tool steel powders used for preparing DMD specimen. Percentage by weight of constituting elements is provided in Table 7.2. Composition of H13 is conspicuous with lesser percentage of chromium and absence of nickel as well as high percentage of iron and carbon in comparison to austenitic stainless steels like 316L.

Table 7.2: Chemical Composition of metallic powders used in Laser Cladding on DMD

| A. 316 L stainless steel powder supplied by SULZER Metco Australia |
|------------------------|-----------------|-----------------|-----------------|-----------------|-----------------|-----------------|-----------------|-----------------|
| Element                | Iron | Chromium | Molybdenum | Nickel | Manganese | Silicon | Carbon | Other |
| Composition % wt.      | 62-72| 16-20    | 2-4        | 10-14  | 1         | 2-3     | 0.03   | <0.5 |

| B. H13 tool steel powder supplied by Alloys International Australasia pty Ltd. |
|------------------------|-----------------|-----------------|-----------------|-----------------|-----------------|-----------------|
| Element                | Iron | Chromium | Molybdenum | Niobium | Manganese | Silicon | Carbon |
| Composition % wt.      | Balance | 5-7    | 1.5-2       | 1-2     | 0.4       | 1       | 0.35   |

b) Microstructure: Microstructural characteristics are an important indicator of the material properties of metals. Fig 7.1 illustrates the optical micrographs for 316L solid and composite and H13 solid and composite pins used in wear testing. The microstructure is observed within the cross section in a direction parallel to the laser scanning and shows typical austenitic and martensitic structures in Fig 7.1 (a) and Fig 7.1(b) respectively. The structure is uniform and dense with negligible non-homogeneities. For H13 pins the microstructure is almost exactly similar to H13 annealed rods with hardness 20 HRC [215]. In the composite pins, interface of width 0.4 – 0.5 mm between mild steel core pin and outer cladding is conspicuous by large size grains and a distinct boundary layer.
Figure 7.1: Optical micro-graphs showing microstructures of laser cladded solid and composite 316L and H13 pins. (a) 316 L solid pin (x500) (b) H13 solid pin (x500) (c) 316L composite interface (x200), and (d) H13 composite interface (x200)

c) **Surface Roughness**: ASTM G-99 recommends testing of several properties for the pin and disc prior to wear test so that behaviour of materials undergoing sliding friction can be ascertained thoroughly. The first is the surface roughness of pin and disc. The surface roughness parameters based on three points test for the machined mild steel disc, 316L & H13 DMD generated pins before wear and 316L and H13 cladding on mild steel disc before wear are presented in table 7.3.

Table 7.3: Surface roughness parameters of specimen before wear

<table>
<thead>
<tr>
<th>Amplitude Parameters</th>
<th>Mild Steel Disc (Normal)</th>
<th>Mild Steel Disc (Worn)</th>
<th>H13 &amp; 316L Laser Cladded Pin</th>
<th>H13 &amp; 316L Cladding on disc</th>
</tr>
</thead>
<tbody>
<tr>
<td>Arithmetic mean roughness ($R_a$)</td>
<td>0.89 µm</td>
<td>7.59 µm</td>
<td>1.71 µm</td>
<td>3.85 µm</td>
</tr>
<tr>
<td>Total Profile Height ($R_t$)</td>
<td>5.95 µm</td>
<td>76.8 µm</td>
<td>10.3 µm</td>
<td>28.2 µm</td>
</tr>
<tr>
<td>Surface roughness depth ($R_z$)</td>
<td>5.11 µm</td>
<td>53.46 µm</td>
<td>8.33 µm</td>
<td>22.8 µm</td>
</tr>
<tr>
<td>Root mean square roughness ($R_q$)</td>
<td>1.08 µm</td>
<td>10.63 µm</td>
<td>2.08 µm</td>
<td>4.36 µm</td>
</tr>
</tbody>
</table>

d) **Micro hardness**: Vickers microhardness tests were conducted for mild steel disc under 100gf, 316L cladding under 300 gf and for H13 pins under 500 gf. Overall 5 readings were taken in horizontal and vertical directions which gave an average value of 168 HV for the disc, 213.2 HV for 316L and 489.6 HV for H13 pins. Bressan and Schopf [216] reported a Vickers micro-hardness value of 687 HV for wrought H13 bars available as commercial grade tool steel. In comparison, the hardness of laser cladded H13 is appreciably less, which is understandable because the pins in this research are used in as-cladded condition without any heat treatment and cold or hot working. Hardness variation within composite pins is presented in Table 7.4.

Table 7.4: Vickers micro-hardness variation for 316L and H13 composite pins

<table>
<thead>
<tr>
<th>316L Composite</th>
<th>H13 Composite</th>
</tr>
</thead>
<tbody>
<tr>
<td>MS Core</td>
<td>Interface</td>
</tr>
<tr>
<td>192.6</td>
<td>182.45</td>
</tr>
<tr>
<td>MS Core</td>
<td>Interface</td>
</tr>
<tr>
<td>180.95</td>
<td>167.05</td>
</tr>
</tbody>
</table>
Slight increase in hardness values can be observed in comparison to the cylindrical pins made from single material. Main reason is that composite pins are cladded at higher laser power and thus more heating and slightly rapid cooling rate contributes to the increment in hardness.

### 7.2.3 Experimental Setup

The friction and wear tests were conducted on a Koehler Instrument’s pin on disc wear tester. This equipment has been used with several variations like as mentioned by Saikko [217] and Fujisawa et al [218], but in this research the equipment was used in its standard form which conforms to ASTM G-99 standard. The hardware setup and schematics of the wear testing process are illustrated in Fig 7.2. The setup consists of a cylindrical pin manufactured by the DMD process held rigidly in the arm with its flat face maintained perpendicular to the rotating surface of mild steel disc finished to a diameter of 165 mm. A normal contact load (L) is applied by the pin on the disc by means of a pulley supported hanging weight connected to the pin holding arm. During sliding the rotational speed of disc can vary and the load can also be increased or decreased thus generating a particular frictional force at the sliding interface. As the pin slides on the disc, it forms circular worn tracks with certain amount of materials removed from pin and disc both depending upon their strength, hardness and surface condition.

Equipment is loaded with an adequate array of sensors and data acquisition software to measure the resulting frictional force and depth of wear track formed by the sliding pin on the disc. The frictional force developed during wear test was continuously measured by a beam type load cell while disc wear was monitored using an LVDT position sensor. The sensed data was transferred to the software through a dedicated controller and data acquisition system. The graphs are dynamically plotted between wear vs. time and frictional force vs. time.

![Figure 7.2: Pin on disc wear tester with labelling of important parts (a) Picture of equipment (b) Process schematics](image-url)
7.2.4 Experimental Design

In this investigation, it is important to note that (i) tested specimens tested were manufactured by a method entirely different from casting or metal forming/working or even powder metallurgy because there is no compacting of powder (ii) specimens were tested in as-cladded condition without any heat treatment that can improve their mechanical properties, and (iii) specimens were tested as components and not as coatings. Therefore, the main focus of this research is to ascertain the wear behaviour of laser based DMD generated stainless and tool steels in terms of co-efficient of friction, specific wear rate and the energy dissipated during wearing out of a given volume from the specimen. If these measured variables are found to be significantly deviating from the values for commercial steel grades found in the published literature, then serious concerns would be raised about the quality of laser cladded steels. Another very important concern for this research is to check the possible disintegration of laser cladding or its separation and delamination from the substrate under mechanical force and heat due to friction. Thus in addition to wear, this experiment also checks the integrity of laser cladded steels and their bonding strength with the mild steel substrate.

Keeping in view the novelty of process and material, a series of experiments were conducted to prove or disprove the viability and suitability of the DMD produced specimens considering the following aspects:

1. Consistency of experimentally determined co-efficient of friction ($\mu$) for laser generated steels in comparison to the range of $\mu$ as given for commercial steels
2. Quantity of wear volume ($V$) determined over sliding distance ($S$) greater than 2 km to estimate extent of wear over the expected life cycle. This is also very important in cases where laser cladding is applied as coating or a thin layer for repairing an expensive part.
3. Changes in $\mu$ and $V$ with the variations in sliding speed ($c$), sliding distance and normal load ($L$). Ideally $\mu$ should remain consistent but experimentally that does not happen and thus needs thorough investigation.
4. Extent of variance and dispersion in the experimental wear data
5. Energy dissipated per unit sliding distance and for a given volume removal.

Accordingly the experiments were designed into two sets of configurations with first set divided into two regimes and the second set into four regimes. When designing the experiments, it is essential to categorize the factors that are considered as most influential and the factors which are most important to measure during the experiments. We call the influential factors as “primary” and measured factors as “secondary”. Among primary factors normal contact load, sliding speed and sliding distance are considered as most influential while co-efficient of friction, wear volume and energy dissipated due to friction are considered as significantly important secondary factors. Different designations for experimental design are described in Table 7.5. It can be seen that the regimes ‘A’
and ‘B’ belong to set #1 while regimes ‘C’, ‘D’, ‘E’ and ‘F’ are placed in set #2. Experiments in set #1 attempt to evaluate wear behaviour under changing factors and to determine their correlation. While experiments designated to set #2 take into account special conditions as mentioned before and designed to have a deeper insight into the non-linearity of friction and wear.

Table 7.5: Design of Experimental configurations

<table>
<thead>
<tr>
<th>Experimental Configuration</th>
<th>Description</th>
</tr>
</thead>
<tbody>
<tr>
<td>Set # 1</td>
<td></td>
</tr>
<tr>
<td>A.</td>
<td>Laser cladded pin on MS plate at variable speed and constant load</td>
</tr>
<tr>
<td>B.</td>
<td>Laser cladded pins on MS plate under load variations at constant speed</td>
</tr>
<tr>
<td>Set # 2</td>
<td></td>
</tr>
<tr>
<td>C.</td>
<td>Laser cladded pin on worn MS plate</td>
</tr>
<tr>
<td>D.</td>
<td>Laser cladded pins on MS plate at high speed and high load</td>
</tr>
<tr>
<td>E.</td>
<td>Laser cladded composite pins on MS plate</td>
</tr>
<tr>
<td>F.</td>
<td>MS pin on laser cladding over steel substrate used as disc</td>
</tr>
</tbody>
</table>

In case of dry severe sliding wear involving steels either in contact with itself or moving against any non-metal, not only ambient temperature but temperature rise of the materials rubbing against each other generate a significant effect on the coefficient of friction [219, 220]. In a pin on disc tester, the temperature of pin and disc during wear cannot change at the same rate and therefore, it is a logical conclusion that coefficient of friction and wear rate should change if the laser cladded steels are used as disc instead of pin. This consideration prompted to use regime ‘F’ in which the pin was of mild steel while the contacting surface of rotating disc was a 10 mm wide and 3mm high laser cladded circular padding from each of 316L and H13 powders.

7.3 Results and Discussions

Main research goals behind wear investigations are to engineer and design materials that (i) have lesser wear rate and can last longer, (ii) have reduced friction coefficient - which saves energy, improves efficiency, and generally results in smoother operation, and (iii) can survive extreme environments where conventional fluid lubricants are not effective. Therefore, wear characteristics of laser generated high strength steel alloys were examined under “severe” conditions in six different experimental regimes. Ramalho and Miranda [221] illustrated the important difference between mild and severe wear in terms of the depth of wear tracks formed by sliding rubbing action. In mild wear the surfaces are generally covered by oxide layers generated during rubbing and result in track debris of the order of several nanometres. In contrast during severe wear, the metal to metal contact is firmly
established, plastic deformation of surface occurs and the worn debris consists of particles up to several hundred micrometres.

Large amount of data from twelve series of experiments based on six experimental regimes were collected from the data acquisition system of pin on disc tester. The irrelevant noise was eliminated and the data being averaged, tabulated and graphed for analysing the values of secondary factors, their statistical significance and the influence of primary factors on secondary factors. In the ensuing discussion, the derived tables and graphs are presented according to the sequence of experimental regimes followed by the observed facts.

7.3.1. Average Trends and Correlation of Factors
Regimes A and B help in comprehending the average friction and wear behaviour of laser cladded parts and determining the correlation of $\mu$, specific wear rate and energy dissipated per unit sliding distance ($E/S$) with other influencing factors like sliding speed, normal contact load, sliding distance. The word average is used to emphasize the fact that in the first section of analysis, the transient and dynamic behaviour of quantities are not taken into account.

![Figure 7.3](image)

(a) 316L Variable speed at constant load
(b) H13 Variable speed at constant load
(c) 316L Variable load at constant speed
(d) H 13 Variable load at constant speed

Figure 7.3: Variation of co-efficient of friction ($\mu$) and energy dissipated per unit sliding distance ($E/S$) with normal contact load, sliding speed and sliding distance for H13 and 316L pin specimens.
Figs 7.3 (a) and 7.3(b) represent the variation of $\mu$ and energy dissipated per unit metre ($E/S$) for two pin materials H13 and 316L when contact load is kept constant at 19.62 N and the disc rotating speed is changed from 150 to 300 rpm in three equal steps, while Figs 7.3(c) & (d) illustrate the changes in same quantities when load varies from 19.62 to 49.05 N in increments of 9.8 N (1 kg hanging weight) and the disc rotating speed is held constant at 300 rpm.

From the plots of Fig 7.3 it is evident that the value of co-efficient of friction is strongly dependent on the normal load in comparison to changes in sliding speed and distance. For the range of speed increment, average value of $\mu$ remains more or less the same, but with the increment in load, $\mu$ increases from 0.5 to 0.65 for 316L laser generated pin specimen and from 0.62 to 0.83 for H13 pins. This clearly implies higher impact of normal load upon $\mu$ in comparison to the effect of increment in sliding speed. This behaviour is comprehensible because higher loads result in more penetration of asperities while higher speed at the same load is a cause of shearing off the interlocking protrusions at the interface.

Same trend can be observed for energy dissipated per unit metre of sliding distance ($E/S$) for both the materials. There is no significant change in $E/S$ over the range of speed variation, but with increase of load $E/S$ increases from 9.9 to 31.63 J/m for 316L and from 12.16 to 32.58 J/m for H13 pins. Another important observation is that almost complete transition in $\mu$ with change of speed (regime ‘A’) or load (regime ‘B’) occurs within first 500 m of sliding distance and it does not experience any further significant variation as sliding distance goes beyond 500m. It means that under a particular condition of load, speed and environment, the state of friction remains stable if not constant with respect to sliding distance.

The impact of sliding speed ($c$), sliding distance ($S$) and contact load ($L$) on $\mu$ can further be highlighted through the study of correlation coefficients ($r$) for two different types of materials tested for friction and wear. In conjunction with the values of $\mu$ presented in plots of Fig 7.3, correlation coefficients ($r$-values) provide a useful insight into the influence of primary factors upon $\mu$ for the two DMD generated high strength steel alloys. In terms of performance against sliding wear, the two alloys exhibit vastly different behaviour as indicated by the values of $\mu$ and wear rates. The $r$-values with respect to three primary factors is presented in Table 7.6:

<table>
<thead>
<tr>
<th>Factor</th>
<th>316L</th>
<th>H13</th>
</tr>
</thead>
<tbody>
<tr>
<td>Sliding speed ($c$)</td>
<td>-0.0257</td>
<td>0.7746</td>
</tr>
<tr>
<td>Sliding distance ($S$)</td>
<td>0.2527</td>
<td>0.9811</td>
</tr>
<tr>
<td>Normal Contact load ($L$)</td>
<td>0.7298</td>
<td>0.9766</td>
</tr>
</tbody>
</table>
The values indicate a weaker relation of primary factors with $\mu$ for 316L specimen while higher values of ‘$r$’ can be observed for H13 parts. It means that for 316L specimen $\mu$ decreases very slightly with speed and increases at a low rate with sliding distance. On the other hand, increment in sliding distance and speed both have significant influence in increasing the value of $\mu$ for H13 pins. But for both 316L and H13 parts, $\mu$ has strong dependence on normal contact load at the interface. But it must be kept in view that these values only show the relation of $\mu$ with the mentioned factors and not the rate of increment or decrement over a given range.

In addition to co-efficient of friction the wear volume ($V$) is also an extremely important parameter because it will define the surface quality of material created by the DMD process. Fig 7.4 provides the quantities of wear volume calculated per kilometre of travel ($V/S$) during sliding of 316L and H13 pins on rotating mild steel disc. The data is plotted for each of the two materials and covers both the regimes A and B.

![Figure 7.4: Variation of wear volume per km sliding distance ($V/S$) in relation to sliding speed and normal contact load](image)

<table>
<thead>
<tr>
<th>Variable</th>
<th>Avg. Coeff. of friction</th>
<th>$V/S$ (mm³/km of travel)</th>
</tr>
</thead>
<tbody>
<tr>
<td>316L Variable speed</td>
<td>0.72</td>
<td>0.260</td>
</tr>
<tr>
<td>316L Variable load</td>
<td>0.74</td>
<td>0.912</td>
</tr>
<tr>
<td>H13 Variable speed</td>
<td>0.58</td>
<td>0.618</td>
</tr>
<tr>
<td>H13 Variable load</td>
<td>0.62</td>
<td>4.555</td>
</tr>
</tbody>
</table>

For further elaborating the pattern of material removal during multi-speed and multi-load regimes, Fig 7.5 presents the dot plots of fractional $V/S$ against normal load and the tangential speed with which the surfaces in contact traverse over one another. The plotted values represent similar trend for both 316L and H13 specimen.
Fractional $V/S$ is obtained by dividing the volume removed for every segment of multi-load and multi-speed experiment by the total volume removed during the entire regime. The sum of all the fractional values in every graph is therefore equal to one. The important observations from Figs 7.4 & 7.5 can be illustrated as follows:

1. The co-efficient of friction has no direct impact on the total wear volume. It also appears that variation of speed while keeping normal load constant has lesser significance in increasing the wear volume.

2. With the increment in load the increase in wear volume for 1 km of sliding jumps to 3 times the value for variable speed for H13 and almost 8 times for 316L. This data clearly indicates low wear rate for the hard material. It is interesting to note that the hardness of H13 specimen is more than twice that of 316L specimen. Data presented in Fig 7.4 illustrates that the ratio of wear volume is almost equal to hardness ratio for multi-speed regime and gets doubled of hardness ratio for variable load regime. This is an important observation which underlines the significance of hardness to reduce wear volume.

3. Wear volume per km sliding travel (not cumulative wear volume) is correlated with co-efficient of friction. This implies that for the experimental regime having higher $\mu$, $V/S$ will also be higher. The reason is that both are strongly dependent upon normal contact load, condition of sliding surfaces which mainly include surface roughness and presence of metal to metal contact or oxidised film at the interface and to a lesser extent on sliding speed. But this does not mean that both increase and decrease at similar rates. This fact is evident if we refer to values of $V/S$ for two regimes when examined for 316L pins. At variable speed average $\mu$ is 0.57 which is increased to a value of 0.62 for variable loading. But among these two regimes $V/S$ increased from 0.62 mm$^3$/km to 4.55 mm$^3$/km. Thus higher $\mu$ between “same materials” under a given set of conditions is an indicator of greater wear from the surfaces.
4. There is a threshold or critical load beyond which the wear rate increases at a drastic rate. Fig 7.5(b) illustrates significantly small values of $V/S$ up to 30N which increases to very high values as the load reaches 50N. The graph shows 82% of volume removal at load of 49 N.

5. The value of $V/S$ significantly drops with increasing speed as highlighted in Fig 7.5(a), but there is a slight increase in wear rate with speed before it drops to a value less than one third of the initial value. This phenomenon can be understood in terms of shearing of asperities on the surfaces in contact being broken down at higher speeds. The experimental data suggests that for alloy steels on mild steel, this critical speed is between 1.8 and 2 m/s after the distance traversed becomes 1 km.

7.3.2 Detailed analysis of secondary factors ($\mu$, $V$ and $E/S$)

The plotted data in Figs 7.3 and 7.4 present only the average values for secondary factors $\mu$, $V$ and $E/S$. It is now considered necessary to carefully look into the variations of secondary parameters with sliding time under a wider set of operating conditions to firmly establish the wear quality of DMD cladded steels. The appropriate set of information collected from the data acquisition system of pin on disc tester for experimental regimes C to F as described in Table 7.5, is tabulated and presented in Table 7.7.

<table>
<thead>
<tr>
<th>Experimental Regime</th>
<th>Sliding Distance</th>
<th>Sliding speed</th>
<th>Contact Load</th>
<th>Wear Volume</th>
<th>Co-efficient of Friction</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td>(m)</td>
<td>(m/s)</td>
<td>(N)</td>
<td>(mm$^3$)</td>
<td>$\mu$ $\mu_A$ $\Delta$ $\sigma$</td>
</tr>
<tr>
<td>S</td>
<td>C_316L</td>
<td>5152.87</td>
<td>1.6961</td>
<td>49.05</td>
<td>14.5949</td>
</tr>
<tr>
<td>L</td>
<td>D_316L</td>
<td>4865.96</td>
<td>2.0285</td>
<td>49.05</td>
<td>3.443</td>
</tr>
<tr>
<td>V</td>
<td>E_316L</td>
<td>4711.88</td>
<td>1.9639</td>
<td>49.05</td>
<td>3.7595</td>
</tr>
<tr>
<td>$\mu$M</td>
<td>F_316L</td>
<td>2650.77</td>
<td>1.0993</td>
<td>49.05</td>
<td>1.0695</td>
</tr>
<tr>
<td>$\mu_A$</td>
<td>C_H13</td>
<td>3768.19</td>
<td>1.5705</td>
<td>49.05</td>
<td>4.769</td>
</tr>
<tr>
<td>$\Delta$</td>
<td>D_H13</td>
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<td>2.04165</td>
<td>49.05</td>
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<td>E_H13</td>
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<td>$\sigma$</td>
<td>F_H13</td>
<td>2197.98</td>
<td>1.2249</td>
<td>49.05</td>
<td>0.8219</td>
</tr>
</tbody>
</table>

7.3.2.1 Co-efficient of Friction

When we refer to the published literature then $\mu$ for steel on steel is found to be a highly fluctuating variable particularly during transition from static to kinetic friction and in dynamic sliding. Yang et al [222] have investigated the unstable fluctuations and stick-slip motion for carbon steel pairs in acceleration and deceleration. Another important concern about $\mu$ is the highest value it can attain. In engineering handbooks the range of $\mu$ for mild steel sliding on alloy steel is given from 0.53 – 0.74,
which can become as low as 0.23 for cases when surfaces are hard and have comparable hardness values [223]. But Lim et al [145] have reported $\mu$ around 1.0 for iron on steel at low velocities and rough surfaces. Mozgovoy et al [224] reported $\mu$ as high as 1.4 when investigating sliding friction of pre-hardened tool steel against 22MnB$_5$ boron steels at 40°C. The fluctuations or pulsations clearly noticeable in the graphs of $\mu$ during sliding wear is representative of a highly non-linear behaviour governed by a large set of variables, some of which pertain to material types and some are related to the operating conditions.

For all experimental regimes $\mu$ shows a highly irregular graph when evolved against time. Main reason is that the testing conditions ensure metal to metal contact and high contact load during sliding. In Figs 7.6 & 7.7 the evolution of co-efficient of friction with time is presented for regimes ‘C’ to ‘F’ for laser cladded 316L and H13 specimens respectively. For analysis it is considered more appropriate to present the plots of $\mu$ vs. time instead of showing dot plots exhibiting mean value of $\mu$ for a particular experiment. Observing the variation of $\mu$ over time provides far better insight and in-depth comprehension of friction behaviour. Some important observations are presented below:

1. The initial condition and build-up of graphs is similar to stick-slip friction behaviour because of severe wear conditions and due to the fact that normal contact load was applied prior to initial rotation of the disc. In all the graphs of Figs 7.6 and 7.7 except for 316L pin sliding on worn surface, an initial peak of the highest value can be observed followed by significant fluctuations. This behaviour represents an initial transition from static to kinetic friction and subsequent stick-slip behaviour which results in an initial surge to maximum value of $\mu$ and then exhibiting lower but fluctuating values during motion.

2. In case of sliding over an already worn surface the behaviour shown for 316L SS and H13 TS is markedly different. In case of laser cladded 316L pins as shown in Fig 7.6(a) there is an initial peak but that does not represent the maximum $\mu$, instead there is a sequence of peaks higher than the initial one and the value of $\mu$ becoming stable around 0.58 after sliding of approximately 3 km. This pattern shows the rapid wearing of the surface of pin while sliding against the worn surface, and hence peak surges in co-efficient of friction. The drop in $\mu$ after a given peak represents formation of a thin oxidised layer which quickly breaks away to give way to another surge. Fouvry and Merhej [225] discussed the formation of oxidised third body layer and significance of wear debris trapped under larger area of contact and their cumulative effect in lowering the co-efficient of friction. This effect can be observed in Fig 7.6(a) which shows that as the wear debris settled down in the area of contact, an oxidised third body layer is formed and the value of $\mu$ decreases and becomes stable. In contrast for H13 pin sliding on a worn surface the graph of Fig 7.7(a) depicts behaviour more similar to the general case of initial peak and maximum $\mu$ at the start of sliding. Afterwards the value of $\mu$ oscillates about 0.95 and finally settles to around 0.65 after 1.8 km of
sliding. The harder surface of H13 pins is responsible for higher $\mu$ throughout than its 316L counterpart due to lack of softening of the pin and retention of greater surface roughness at the interface. The graph of Fig 7.7(a) highlights the absence of waviness pattern in contrast to Fig 7.6(a) and thus lesser influence of stick-slip behaviour during friction.

Figure 7.6: Evolution of co-efficient of friction over time for laser generated 316L SS for experimental configurations C to F (a) C_316L (b) D_316L (c) E_316L, and (d) F_316L

Figure 7.7: Evolution of co-efficient of friction over time for laser generated H13 TS tested for experimental regimes C to F (a) C_H13 (b) D_H13 (c) F_H13, and (d) E_H13
3. The lowest values of $\mu$ are observed under the condition of high load and high sliding speed as evident from graphs in Figs 7.6(b) and 7.7(b). The main reason for the lowering of $\mu$ are twofold; (i) at high speeds the asperities at area of contact are sheared off rapidly due to higher energy input at the interface, and (ii) due to higher energy dissipation, formation of oxidised layer at the interface and burnishing of surfaces in contact also occur at a faster rate. For H13 pin, graph shows a slight upward trend up to 2 km which may be due to removal of oxidised layer which is always a possibility under dry and severe conditions. This results in an increased surface roughness and hence higher co-efficient of friction. For 316L pins this oscillation between forming and breaking of oxidised layer is more pronounced and occurs at a faster rate. The difference in two behaviours can be attributed to superior strength and hardness of laser cladded H13 pin.

4. The case of composite pins for the two materials is quite typical and needs special consideration from two aspects. First one pertains to probable de-lamination of laser cladding from mild steel core as a result of high tangential forces acting on the pin due to friction. Second concern deals with differential wearing of the core and clad due to difference in strengths and hardness. The first aspect will be duly highlighted during discussion of SEM micrographs for the worn pins. For the wear and friction behaviour, graph of Fig 7.6(c) presents a uniform behaviour for 316L pins with stabilising of $\mu$ around 0.52, but for H13 pins the fluctuations are much higher as indicated in Fig 7.7(c). These observation lead to the conclusion that since 316L generated samples are comparable in strength to the mild steel core and therefore, the whole pin (including core and clad) wears at more or less the same rate. But for H13 composite pins, bulk of wearing occurs on the outer cladding with intervals when the core comes in contact with the disc. Fluctuations around mean value of 0.6 in Fig 7.7(c) indicate major contact existing between outer annular area occupied by H13 cladding and the mild steel disc.

5. When the experimental arrangement is reversed and the cladded disc takes the place of mild steel disc, higher overall values of co-efficient of friction are observed. For laser generated 316L padding on the rotating disc, $\mu$ is highest at the initial transition stage from static to kinetic friction. But once the sliding continues at constant speed, value of $\mu$ quickly settles down to the range between 0.45 and 0.5. Ideally after sliding more than 1 km, $\mu$ should be much closer to the values observed for high speed and high load as shown in Fig 7.6(b) which did not happen. Thus, the fact should never be ignored that when ascertaining wear behaviour on a pin on disc apparatus, there is always some difference in $\mu$ when the pin material is used as disc and vice versa. The reason for this difference is that the pin is loaded by the hanging weight in a direction normal to the rotating disc, and if pin wears at a higher rate than the disc, it has greater tendency to penetrate into the disc creating more resistance to motion and hence a higher co-efficient of friction.
6. For the mild steel pin sliding over the H13 cladding on the rotating disc, a gradual decrease in $\mu$ is observed but at a slow rate and accompanied with large fluctuations. For similar reason as explained above, here again the net value of $\mu$ is significantly higher than that found for high load and high speed. But larger fluctuations may have following reasons: (i) pin is experiencing make and break contact with H13 cladding due to high wear rate against a much harder material, and (ii) the arm behaves like a cantilever beam and the pin may experience some large amplitude vibrations due to high friction and elasticity of the supporting arm.

7. From the graphs of Figs. 7.6 and 7.7 it is clearly evident that the nature of $\mu$ is stochastic and thus demands at least the basic statistical analysis to ascertain the extent of variance in the data as obtained from data acquisition system of wear tester. Guicciardi et al [226] have demonstrated that wear data for pin on disc sliding wear tests exhibit dispersion in the range of 28–47% for the disc and 32–56% for pin material. The four columns under the heading of ‘co-efficient of friction’ in Table 7.7 provide the maximum, average, range and standard deviation for the complete set of values recorded by the wear tester. In addition $R^2$ values for coefficient of friction and linear trend-lines for its graph are presented in Figs 7.6 and 7.7 for all experimental regimes under consideration. It can be observed that in regimes ‘D’ and ‘E’ value of 3 standard deviations ($3\sigma$) remains under 20% which is the sign of consistency of data. But in regime ‘F’ where surface roughness increases slightly, $3\sigma$ is 27% for 316L and 25% for H13 while for regime ‘C’ with rougher surfaces, $3\sigma$ increases to 43.5% for 316L and 55% for H13 specimen. This clearly indicates rapid increase in dispersion of wear data with increase in surface roughness.

8. The skewness of data is a strong indicator that $\mu$ for any experimental regime does not follow normal distribution. In such cases box plots are a good means for statistically analysing the zones of data concentration. Box and whisker plots are presented in Fig. 7.8 for 316L and H13 laser cladded pin specimens for experimental regimes ‘C’ to ‘F’. The information from these plots provide a good idea of the dynamic and kinetic friction co-efficient for a given experimental regime. The box in the
plot contains half of the data points and hence if size of box is small it means that in this range average value of $\mu$ is present. The maximum value shown by the top whisker shows that static friction coefficient lies in this vicinity. Minimum value represented by the bottom whisker is an indicator of intermittent lowering of co-efficient of friction due to stick-slip behaviour, periodic formation of oxidised layer and debris collection in the interface. For regime ‘C’ the values are spread over a wide range and show a high degree of variance. For other regimes more resembling usual wear conditions, box plots give a range of $\mu$ between 0.35 – 0.55 for 316L and 0.45 to 0.7 for H13 laser cladded pins. In comparison to the values found in published literature as mentioned earlier for steels against steel counterface, $\mu$ shows a consistent value comparable to commercial grades of steel under unlubricated conditions.

7.3.2.2 Specific wear rate and energy expenditure

While studying friction and wear behaviour of laser generated high strength steel alloys, one of the most important aspects to realize is that the wear rate may be a highly inconsistent quantity and independent of the co-efficient of friction. Blau [227] described that mechanical energy during sliding can be converted to heating of pin and disc which ultimately dissipates into atmosphere, or vibrations that creates sound, or to material deformation, or the creation of new surfaces by fracture. Therefore, depending on the distribution of energy between the materials in contact, wear rate can be different even for materials exhibiting same co-efficient of friction. Modi et al [228] investigated the abrasive wear of high carbon steels and discussed in detail the correlation of wear rate with mechanical properties like hardness, strength and fracture toughness in reference to Hornbogen’s equation of wear rate [229]. Thus, in light of these investigations, it is imperative to look into wear rate and the amount of energy spent in overcoming friction during a particular experimental or regime wear independently from co-efficient of friction.

For calculating the volume of material removed during sliding wear, difference of mass before and after the experiment is measured for pins for regimes C to E and for disc for regime F. The specific wear rate ($k$-value) can be evaluated from the experimental data according to eq. 7.1, which provides the value of wear volume per unit load per unit sliding distance according to Archard’s formula [143] and given as:

\[ V = k S L \]  \hspace{1cm} (7.1)

where $V$ is wear volume, $S$ is sliding distance and $L$ is the normal load.

From the data obtained through experimental investigation, another important secondary parameter termed as energy dissipated per unit sliding distance ($E/S$) can be derived. The value of $E$ can be calculated by multiplying the area under friction vs time graph ($f$ vs $t$) with the sliding speed $‘c’$. The formula is written in eq. 7.2 given by:

\[ E = \int_{t_1}^{t_2} (f \cdot dt) \cdot c \]  \hspace{1cm} (7.2)

[142]
$k$-value and $E/S$ in combination provide a good indication of the wear behaviour of laser generated steel specimen. These derived quantities are shown in Fig 7.9 as bar graphs:

![Specific wear rate and Energy dissipated for regimes 'C' to 'F'](image)

**Figure 7.9:** Specific wear rate ($k$-value) and energy dissipated per unit sliding distance ($E/S$) for 316L and H13 laser produced specimen for experimental regimes ‘C’ to ‘F’

The graphs for 316L and H13 illustrate a distinctly contrasting trend for sliding against the worn mild steel plate (regime ‘C’). For 316L pins ‘$k$’ is more with lesser value of $E/S$ while for H13 pins ‘$k$’ is halved with approximately 25% increment in $E/S$. It should be noted that $E$ is not the energy consumed in removing a given amount of volume from the specimen, instead it is the energy that manifests itself as friction and eventually dissipated into heat. Therefore, higher wear rate with lesser energy dissipation is an indication of softening of 316L pins and subsequently more wear at lower friction with increase of time and temperature at the interface. This trend is conspicuous for regime ‘C’ and continues for regimes ‘D’ and ‘E’, but almost diminishes for regime ‘F’. The main reason is that when DMD produced material is used as cladding then energy dissipation occurs from a much larger area than pin and hence softening effects are less pronounced.

The most important aspect to observe is the comparative $k$-values published for commercial carbon, stainless and tool steels so that wear characteristics of laser generated steels can be ascertained. The values observed are not far away from those reported for various commercial grades of steel thus illustrating the better quality of surface and bulk of laser generated specimen. H So [230] conducted numerous tests on various grades of steels with temperature variation and the results confirm well to those presented in this investigation. For instance he reported a wear rate of $7 \times 10^{-13}$ m$^3$/m at a contact pressure of 2 MPa for die steel pin (600 HV) rubbing against AISI 4340 disc at 4 m/s. This compares well to the wear rate of H13 laser cladded tool steel (~500 HV) tested under regime ‘C’ which is an extreme case in this investigation. The values reported in this research are $12.65 \times 10^{-13}$ m$^3$/m at a sliding speed of 1.57 m/s under a contact pressure of 2.5 MPa. Ramalho and Miranda [221] reported $k$-values to be $1.3 \times 10^{-13}$ m$^3$/N.m for mild steel and $3.99\times10^{-16}$ m$^3$/N.m for tungsten carbide when
tested for sliding wear against hardened high-speed-steel AISI M2. Lindroos et al [231] reported a worn volume of $5.5 \times 10^{-12} \text{m}^3/\text{N.m}$ for HV500 steel which has martensitic microstructure and is very similar in mechanical properties to commercial grade H13 tool steel. The tests were conducted as single and multiple scratch high stress two-body abrasion tests on a CETR UTM-2 tribometer under a normal contact load of 60 N.

7.4. Microscopic Analysis of worn surfaces
A closer look into the morphology of wear scars and the nature of wear debris provides a good insight into wear characteristics of the laser generated steels. Most important concern about the DMD generated 316L and H13 parts is the extent of surface damage and extent of material transfer between the materials in contact during severe sliding wear. This form of surface damage occurring between sliding solids and characterized by macroscopic roughening of the surfaces is termed as galling [232]. Although galling is a gradual process, but similar effects can be observed in the situations involving high contact loads at low sliding speeds between the materials having similar microstructure and mechanical properties. Karlsson et al [233] demonstrated the damaging effects related to galling for tool steels when tested at 50 and 500 N with sliding speeds as low as 2.5 mm/s under normal room environment. Fig 7.10 illustrates a number of scanning electron microscope (SEM) micrographs which are presented for DMD produced 316L and H13 specimen as well as the mild steel disc after wear.

Although a marked difference has been observed in the values of coefficients of friction among the several tested regimes, morphology of the wear surfaces reveals that the wear occurs in all cases by abrasion and accompanied by material transfer. In all cases noticeable scars, grooves, plastic deformation, flaking and ploughing features are visible on the worn surfaces. The worn surfaces are typical of the effects of sliding wear on carbon, stainless and tool steel surfaces [234, 235].

Out of all the experimental regimes examined for wear, most interesting case is that of composite pins (regime ‘E’). Micrographs from Figs 7.10 (a) to 7.10(d) represent the wearing effect on 316L and H13 composite pins and the mild steel disc under regime ‘E’. The mild steel core exhibits more wearing and deeper scars as compared to the 316L and H13 outer cladding shown in Figs 7.10(b) and 7.10(c) respectively. For H13 outer cladding we can observe more wear debris trapped in the shallow openings created in the surface due to wear which indicates a higher wear rate than 316L composite pins. Another important observation is that the wear marks and scars are not uniform, both on the pins and for the disc as shown in Fig 7.10(d).
Figure 7.10: SEM micrographs for worn DMD generated pins and mild steel disc (a) 316L composite pin - mild steel core (x 1.41k) (b) 316L composite - worn outside cladding (x1.78k) (c) H13 composite pin - worn outer cladding (d) Disc wearing against H13 composite pin (e) 316L Pin – regime ‘C’ galling and wear debris (x1.41k) (f) Disc after wear - regime ‘C’ (x656) (g) H13 Pin – regime ‘C’, galling and burnt debris (x2.35k) (h) 316L pin – flaking of surface layer (x1.07k) (i) Disc wearing for regime ‘D’ (x428)

The non-uniformity is more pronounced for H13 pins than 316L pins. This indicates an important fact that the difference of hardness between the core and outer cladded annulus results in non-uniform wear of the two regions and hence a non-uniform contact resulting in irregular pattern of wear scars.

Worst affected surfaces are observed for sliding severe wear on worn mild steel disc and for both 316L and H13 pins under regime C. Figs 7.10(e) to 7.10(g) show the wearing effects on the two types of laser cladded pins and the mild steel disc against which these slide. The existence of effects like whole layers being scratched away from the surface, deep ploughing into the surface and burnt fluffy debris collected into worn out pockets are clearly visible. It may be recalled that values of $\mu$ and $E/S$ were recorded as the highest for this experimental regime. Although this research does not take into account the temperature effects on wear in detail, but under regime ‘C’ involving severe wear on
surfaces with high roughness, temperatures were noted to be around 70 – 75°C within 30 seconds of finishing the experiment. High temperatures encourage wear due to adhesion and plastic deformation at the interface thus increasing the probability of creating deeply gouged surfaces with burnt particles broken away from the worn surfaces.

Figs 7.10(h) and 7.10(i) illustrate the conditions of 316L pin and the mild steel disc for experimental regime ‘D’. Since this series of experiments were carried out at high speeds (>1.5 m/s) therefore, the wearing marks are shallow scars and wear tracks. The worn surfaces are devoid of deep grooves and ploughing effects of one surface onto another and thus explain the recorded low values of $\mu$. For the laser cladded pin in Fig 7.10(h) the breaking up of surface and a resulting depression can be observed, which when observed in conjunction with Fig 7.10(g) highlights the somewhat inconsistent quality of laser cladded parts without heat treatment.

**7.5. Wear Debris**

The debris particles in a variety of shapes are formed as a result of sliding and rubbing action of alloy steel pins on the mild steel disc. Heilmann et al [236] mentioned that normally two types of debris are observed in sliding wear. In some cases the dark and very small particles are seen while in other cases, particularly for softer materials, the particles resemble long and thin flakes. Oxidized wear particles can significantly reduce the wear rates when they are compacted and form a protective oxide layer [225]. Rynio et al [237] described the metallic wear debris to be larger and darker particles while oxidized particles appear to be smaller and brighter. They also reported a decrease in wear due to presence of oxidized wear debris at the interface. But they also emphasised the importance of temperature in supporting this behaviour, because at ambient temperature, the metallic contact between pin and disc is never fully inhibited by oxides.

In this study Scanning Electron Microscopy (SEM) and Energy Dispersion Spectroscopy (EDS) analysis of wear debris accumulated after the high normal load experiments has been conducted and the results are presented in Figs 7.11 and 7.12. Fig 7.11(a) illustrates the removed and worn particles from laser cladded 316L pin on MS disc while Fig 7.11(b) shows the debris for H13 pin. In both the cases the debris includes fine dust like particles as well as long flakes. When ductile materials are sliding against each other then it is most likely to find long flakes within the debris generated due to ploughing action as can be observed from the highly magnified surface profiles shown in Fig 7.10. In addition to long flakes there are also fine powder like particles which can be seen to be attached to the flakes but will be easily detached if touched by a pin or needle.
As 316L is more ductile than H13, therefore, longer flakes can be observed in Fig 7.11(a) as compared to those in Fig 7.11(b). The results of EDS examination for both types of debris is presented in Fig 7.12 (a) to 7.12(d).

Fig 7.12 shows the results of EDS measurements for both H13 and 316L debris after sliding wear under high normal loads. It can be observed that there is intermixing of pin and disc particles like in Fig 7.12(a) where a higher percentage of carbon and oxygen is present, whereas in Fig 7.12(b) carbon...
and oxygen is absent. Similar behaviour can be observed in Figs 7.12 (c) & (d) for 316L debris. This implies that oxidation mainly occurs in mild steel particles and large number of un-oxidized alloy steel particles can be found interspersed within the mixture. It can also be safely concluded that majority of flakes found in the wear debris belong to MS disc while small and powder like particles wear out from 316L and H13 pins due to differences in strength and hardness of the materials in addition to ploughing action of pin into the disc surface.

7.6. Conclusions
The sliding wear tests under dry and severe conditions for 316L and H13 high strength alloy steels produced by laser assisted DMD process reveal important characteristics of these materials in terms of co-efficient of friction, specific wear rate and energy dissipated due to friction. The tests were conducted under several experimental configurations and regimes that take into account the variation of material, condition of surfaces and operating parameters like sliding speed, distance and normal contact load. The wide range of experimentation validates the wear quality of laser generated steels without heat treatment as comparable to the wrought and formed commercial grade steels. The co-efficient of friction is found to be on slightly higher side in comparison to commercial steel alloys and also as a significantly variable quantity. The high degree of variation may be due to severe testing conditions and may also relate to some inherent inconsistencies in the composition of DMD produced parts without post processing. Specific wear rates are also found to be in line with the values reported for commercial grade steel alloys which is a very encouraging sign and proves that the laser cladded structures are fully dense and dominantly homogenous. Composite pins offer a good option for material saving as it shows good wear behaviour with the quantity of cladded material being halved in comparison to completely cladded pins. Another important feature is less amount of flaking, ploughing and chipping from the surfaces of laser generated materials which in fact alleviates a serious concern regarding bulk properties of cladding and their bonding strength with the mild steel substrate. Therefore, it can be concluded that as-cladded specimen of 316L and H13 alloy steels exhibit encouraging sliding friction and wear behaviour which can be further improved by appropriate surface and heat treatment methods.
CHAPTER 8
FATIGUE BEHAVIOUR OF DMD CLADDED COMPOSITE STRUCTURES

8.1 Introduction and Research Objective
For additive manufacturing technologies in general and for DMD in particular, many results have been published regarding material and metallurgical characteristics augmented by investigations on the static mechanical properties of cladded parts, while the knowledge about their dynamic mechanical behaviour remains insufficiently low. Due to their peculiar method of manufacturing that covers an extremely vast scope due to very large number of possibilities owing to different material combinations, structural configurations and manipulation of process parameters; laser cladded structures need a thorough investigation with respect to dynamic behaviour in addition to metallurgical and pseudo-static properties. There is a significant gap in this area which this thesis has attempted to address and investigate. In the previous two chapters the investigated results pertaining to high strain rate compressive dynamic behaviour and sliding wear characteristics under un-lubricated conditions were presented in detail for 316L SS and H13 TS cladded structures. These are the two materials that have shown great workability under a laser processing system and very good bonding with mild and plain carbon steels. This chapter is related to the study of fatigue behaviour of DMD generated composite cylindrical specimen fabricated from 316L and H13 claddings upon mild steel core used as substrate in the high cycle fatigue range.

Study of fatigue behaviour is an extremely vast research area and even more typical for laser cladded structures due to greater probability of the presence of fatigue initiation defects and inclusions like micro-porosity, and non-homogeneity of microstructure due to unmelted and partially melted powder particles. Therefore, an observable fact which is well highlighted within the published research for fatigue properties of laser cladded steels is that this area has more questions than answers mainly because (i) the laser generated parts represent a set of “newly created” materials due to the manipulations that could be made to material properties through laser and machine parameters, and (ii) laser cladded parts have been scarcely employed as complete parts or machine components under heavy mechanical loads and stresses. That’s why this investigation, which is only a part of the whole research, is not particularly attempting to develop a comprehensive and elaborate SN-curves for DMD generated steel structures, but merely provides an insight into the fatigue performance of a specific laser generated structural configuration under different stress ratios and the manner in which the fatigue related fracture occurs.

This chapter deals with two main aspects regarding fatigue behaviour of DMD generated 316L and H13 composite specimen. The first section deals with investigating and comparing the fatigue
properties of these specimens at different stress ratios with non-zero mean stresses in tensile region. The tests were carried out according to the recommendations of ASTM E466, several versions of which relate to conducting axial force controlled fatigue constant amplitude tests to obtain the fatigue strength of metallic materials. The research also looks into the effects of frequency of cyclic loading upon the number of cycles to failure. The second important part of this chapter pertains to analysis of the fractured surfaces of laser cladded specimen through scanning electron microscopy (SEM) images in order to ascertain the locations of initiation of fracture and the manner in which fracture propagates once initiated. The observations illustrate some significant differences from the pattern observed after fatigue failure of rolled 316L rods.

When compared with the fatigue properties of rolled stainless steel bars, under similar testing conditions, the laser cladded specimen exhibited a fatigue life which is 25% lesser in terms of cycles to failure. As mentioned previously that this research investigates DMD generated high strength steel structures under as-cladded condition in order to ascertain the quality that can be obtained through a DMD process for near net shape parts. Therefore, this performance is not considered as discouraging, and obviously the fatigue performance and strength properties can be improved using a number of available heat treatment options, particularly those that help in reducing the residual stresses and making micro-structure more uniform and homogenous.

8.2 Sample Preparation and Experimental Methods

The composite cylindrical specimen fabricated by DMD cladding of 316L and H13 alloy steels upon mild steel bars were prepared for fatigue testing in the tensile range. This structural configuration is chosen because it can enable the use of DMD cladding to either repair machine parts and expensive equipment or produce independently functional parts with economic feasibility and optimization of mechanical properties. Mechanical components like gears, shafts, cams and tooling components like punches and dies, which can be manufactured as composite metallic parts, are subjected to intermittent and cyclic loading accompanied by vibrations during their operational life. These components can experience fatigue effects and subsequent failure particularly if the applied loading causes plastic straining during operational life cycle. Investigation into fatigue properties of composite steel structures will optimize the use of DMD setup in manufacturing such components with an inside tough core from low cost mild steel with deposited customized shapes from high strength alloy steels through cladding of alloy steels for superior performance.

Fig 8.1 (a) shows the condition of surface along gauge length during the process of cladding and final machined surface of DMD generated specimen obtained by laser cladding of two different high strength steel alloys i.e. 316L and H13 on the same substrate which is mild steel. The substrate material is a 4 mm rod of low carbon mild steel with a carbon content of 0.15 – 0.25 wt. % and other alloying
elements which include manganese (0.7 – 0.9 wt.%) and silicon (0.3 wt.%) and traces of sulphur (~0.005 wt.%). This steel offers a good combination of strength and toughness at low cost and above all exhibits excellent bonding with laser cladded alloy steels. The cladding was done as concentric layers from either 316L stainless steel or H13 tool steel powder material with mild steel rod held in a chuck and slowly rotating under the laser beam. Chemical composition of both the materials according to supplier’s catalogue has already been provided in chapter 4. The controlling DMD parameters for both types of cladded materials were 650 watts laser power, 40 mm/min as the laser scan speed with bar rotating at 2.5 rpm under the laser beam and a powder feed rate between 3 and 3.5 g/min. The laser cladding and scanning strategy with the final dimensions of composite cylindrical dog bone specimen is presented in CAD diagram as shown in Fig 8.1(b).

Figure 8.1: Cladding and scanning strategy for preparation of composite cylindrical dog bone samples on DMD machine using CO2 laser

As illustrated in Fig 8.1(b) that after completing one scan or depositing single annular track, the laser beam shifted rightwards with an overlap of 50%. The resulting track width was 1 mm with a track height or thickness in the range 0.6 to 0.9 mm. Therefore, in all, 3 tracks were needed to obtain
the layer height of 2 – 2.5 mm, which resulted in an overall diameter of composite cylindrical specimen to be approximately 8.5 mm. Later on the specimens were machined and ground to remove the top few rough layers and to reduce the possibility of any stress concentration on the surface while attaining a finished gauge diameter of 7 – 7.25 mm and gauge length as 45 mm.

The optical micrographs illustrating the microstructures of 316L composite and H13 composite specimens are shown in Fig 8.2. One of the most important thing to look at is the condition of heat affected zone (HAZ) which is developed in the substrate material mainly due to rapidly altering thermal cycles resulting from localized substrate heating by the laser beam, and formation of larger grains and distinct grain boundaries at the interface region between clad and substrate. As noticeable in both Figs 8.2(a) & (b), the HAZ area has a radius of about 200-300 µm and interfacial grains are neither lamellar nor dendritic but more or less equi-axed with size larger than the substrate grains and having well-defined grain boundaries. The outline of interfacial boundary between cladding and substrate is also clearly distinguishable. This observation illustrates a gradual transformation from mild steel to alloy steel in the radial direction across HAZ without any significant solid diffusion. Such a material structure can become vulnerable during fatigue loading particularly at the interface.

![Microstructure of laser cladded composite cylindrical specimen using an optical microscope (a) 316L specimen (x100) (b) H13 specimen (x200)](image)

Figure 8.2: Microstructure of laser cladded composite cylindrical specimen using an optical microscope (a) 316L specimen (x100) (b) H13 specimen (x200)

The microhardness values are in the same range as illustrated in Fig 5.7 in Chapter 5 for the DMD cladded 316L and H13 composite cylindrical specimen as calculated from substrate to cladding across the HAZ.

Both tensile and low-cycle fatigue tests were carried out on a commercial MTS closed-loop servo-hydraulic test machine at room temperature in ambient air. The machine has a maximum capacity of applying 100 kN force and generated cyclic loading in a sawtooth waveform. The samples were tension-tested and the results were compared to similar commercially available material as cold rolled
bars. The data recording system attached to the machine maintains the records of time elapsed, ongoing number of cycles, deformation in the specimen and load cycle steps. The frequency of all tests was maintained at 15 Hz except at $R=0.8$ when the test was conducted at 20 Hz to simulate the vibration scenario under high loads. The test was performed until failure which was considered as the instant when strain exceeded 0.1. The run-out condition was considered to occur if the number of cycles exceeds 1 million ($10^6$).

The fatigue testing regime was high cycle fatigue (HCF) because the intended use and application of laser cladded components or parts is similar to the practical applications in which components normally operate under the yield stress limit and thus testing under high cycle fatigue conditions is recommended. It should be noted that the criterion used to differentiate between high cycle and low cycle fatigue (LCF), as described by Ashby and Jones [130], mainly depends on the fact that whether the cyclically loaded specimen is subjected to plastic strain more than the elastic strain. For high strength metals like steel alloys, a generally accepted value of cycles to failure ($N_f$) is $10,000$ beyond which fatigue will enter into high cycle fatigue range.

After the failure of the fatigue test specimens, the fractured surfaces were analysed with scanning electron microscopy. The SEM images were obtained at low magnification ($<x500$) to illustrate the overall effect on the substrate and cladded side and particularly at the interface, and at higher magnification ($> x1 \text{k}$) to present the condition of fractured surface of the 316L and H13 cladding. On the substrate side crack propagation was found to be similar as in ductile fracture. But on the clad side, signs were evident for fracture by rupture and tearing away of the surface across the cladded layers by the tensile forces.

8.3 Results and Discussion

Research on DMD generated composite specimens of cylindrical configuration from high strength steel alloys has not been reported in literature for fatigue analysis either with or without pre testing notch or crack. Ganesh et al [238] have used 3.5 kW CO$_2$ laser with co-axial powder feeding nozzle, a system very similar to POM-DMD system which was used for our research, to produce a wedge shaped Inconel 625 cladding with a machined notch at the bottom of laser deposited region while using 304L SS as the substrate. Hutasoit et al [170] used 550 W Nd:YAG laser to generate 0.9 mm to 1.4 mm thick Deloro and Stellite-6 coatings with 12 mm length on a 4340 steel rod for rotating-bending fatigue test. Niederhauser and Karlsson [121] laser cladded Co-Cr alloy on low carbon steel plates for producing flat specimen with square cross-section for fatigue testing.

In this research, all the tests were conducted at room temperature (between 19 to 24 °C) with maximum stress in the range of 398–596 MPa and minimum stress within the values 0 – 398 MPa.
This range was selected keeping in view the static yield and ultimate tensile stresses for the three types of materials subjected to investigation. These values are given in section 8.3.1.

Since fatigue strength and the location of initiation of failure for a composite multi-material specimen is determined by the material of lower endurance limit, therefore, a higher probability for fatigue failure of the composite specimen exists in the mild steel substrate. But in laser cladded specimen, annular layers of cladded material can contain many defects and inclusions that can initiate fatigue failure. In addition, at substrate/clad interface and at HAZ, the fatigue characteristics of the material are unknown but are based on the development of microstructure in this region and with a combination of different phases, this may be the most vulnerable region for fatigue failure. Also it has been observed in section 5.6 during the discussion on the residual stress distribution in a composite cladded system that a transition takes place from compressive to tensile type at the substrate/clad interface. The existence of tensile stress in the HAZ region and first cladded track may also contribute to fatigue-damage initiation. Therefore, in order to determine which of these zones is more prone to fatigue failure, composite specimens were tested both below and above the endurance limit of mild steel. In addition to evaluating the vibration like effects at high stress ratio and high stress amplitudes, test was conducted at R=0.8 at 10 Hz for both 316L and H13 specimens.

8.3.1 Static Properties
For fatigue testing in the high cycle fatigue (HCF) range, it is necessary to determine the elastic limit and ultimate stress for the specimen. The tensile tests carried out by Soodi et al [239] on the same DMD setup as used for this research revealed that for pure or monolithic 316L and H13 specimen the ultimate tensile strength was recorded as 600 and 860 MPa respectively. From the same reference, yielding stress for the two alloy steels are found to be approximately 250 and 540 MPa respectively. Proximity to these values is also confirmed by the tensile tests conducted for composite cylindrical specimen produced for this investigation. Results reveal that the ultimate stress for 316L composite specimen was 623 MPa while for H13 composite bars the ultimate stress value was recorded as 805 MPa. Due to highly non-linear behaviour of high strength alloy steels in the elastic region there is no distinct value but according to the generally accepted criterion of 0.2% yield strength, values were 238.98 MPa for 316L specimen and 422.32 MPa for H13.

8.3.2 Fatigue Properties
The number of cycles to failure ($N_f$) for 316L and H13 composite cladded specimen as well as for 316L cold rolled bars were determined and are presented in Table 8.1 for comparison of fatigue properties. The value of stress ratio and frequency of cyclic loading was also varied to understand the effects of these two important parameters. Although the values of $N_f$ exhibit a large amount of scatter but the average values present some useful insight into the laser cladded steel specimen.
Table 8.1: Experimental Results for Fatigue testing at different stress ratio and amplitudes

<table>
<thead>
<tr>
<th>Experiment Designation</th>
<th>Stress (Max) ($\sigma_{max}$) MPa</th>
<th>Stress (Min.) ($\sigma_{min}$) MPa</th>
<th>Stress Amplitude ($\sigma_a$) MPa</th>
<th>Mean Stress ($\sigma_{av}$) MPa</th>
<th>Stress Ratio ($R$)</th>
<th>Frequency of Cyclic loading (Hz)</th>
<th>Number of Cycles to Failure ($N_f$)</th>
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<tr>
<td>316L Composite cylindrical DMD cladded specimen</td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td>L1</td>
<td>497.45</td>
<td>0.00</td>
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<td>248.73</td>
<td>0.0</td>
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<td>L2</td>
<td>596.94</td>
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<td>268.62</td>
<td>328.32</td>
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<tr>
<td>L3</td>
<td>397.96</td>
<td>79.59</td>
<td>159.18</td>
<td>238.78</td>
<td>0.2</td>
<td>15</td>
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</tr>
<tr>
<td>L4</td>
<td>517.35</td>
<td>258.68</td>
<td>129.34</td>
<td>388.01</td>
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<td>L5</td>
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<td>49.75</td>
<td>447.71</td>
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<td>H13 Composite cylindrical DMD cladded specimen</td>
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<tr>
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The values presented in the table are also plotted in Fig 8.3 as S-N plots showing variation in $N_f$ with stress amplitude ($\sigma_a$). Based on these observations, the following inferences can be obtained:

i) Since the microstructure of laser generated structures is similar to that of cast steels [240,241,242], which typically consists of relatively finer grain size and associated solidification texture [19], there is a high probability of defects like microscopic voids and inclusions within the cladded tracks, layers and inter-dendritic/grain boundaries. In addition, the rapid solidification associated with DMD cladding may result in significant residual stresses with variations in its type, orientation dependent strength properties and cracking in susceptible materials or brittle phase formation (e.g. martensite in ferritic steels) [238]. Presence of these defects in the laser generated parts may adversely affect their fatigue and fracture performance.
The performance of 316L composite cladded specimen is much inferior to the fatigue performance of cold rolled bar specimen obtained from the same material. With maximum stress approaching the ultimate tensile strength of the specimen i.e. 596.94 MPa at stress ratio 0.1, the specimen could last only up to 16386 cycles while its wrought counterpart sustained approximately quarter of a million cycles in the same loading range. On the other hand, H13 composite cladded specimen exhibited much better fatigue performance by surviving for slightly more than 125,000 cycles. Such behaviour corroborates the observation in Chapter 6 during high strain rate compressive dynamic testing of similar composite specimen, that the dynamic property for composite specimen is mostly governed by the high strength cladding provided the interface does not give in.

The maximum number of cycles to failure for both the specimen can be observed at R=0.2 under the experiment numbers L3 and H3, when the maximum stress is much below the ultimate strength of the cladded material. This is mainly due to the fact that the applied cyclic loading induced lesser amount of plastic strain particularly for H13 specimen. Therefore, in the regime within or near the elastic limit, longer fatigue life can be obtained for laser generated multi-material specimen without any strengthening and heat treatment. The main reason for this behaviour is that high cycle fatigue strength is dominated by plastic deformation of the material. As described by Ashby and Callister [121, 243 ], the initiation of the surface cracks governing the fatigue life is dominated by plastic deformation In the HCF regime, crack incubation and early stages of crack growth consume most of the total fatigue life in high strength steels.
(iv) For experiments H3 and H4, as shown in Table 8.1, in which the maximum stress remains under or slightly above the 0.2% yield strength of H13 composite specimen, a very definite high cycle fatigue behaviour can be observed. In HCF since the stress is below the general yield stress, therefore, most of the cycles are spent in initiating a crack. For commercially wrought or rolled steels, HCF can extend into giga-cycles regime [244] due to the reason that as a whole the specimen does not enter into the plastic range. But in case of specimen examined in this investigation, there is every likelihood of the presence of localized plastic zones that can act as a notch and where stress concentration can occur. An ensuing crack may initiate in such regions and propagate due to cyclic loading, initially at a slower rate over a long period of time and then suddenly at much faster rate, resulting in ultimate rupture apparently in an instant. This effect of highly rapid rate of propagation of crack can be observed in case of researched specimen, which showed increase in deformation from 1 to 5 mm within 3 loading cycles. This is a very important observation which focusses on reduction of abnormalities in the specimen’s structure and which can particularly benefit HCF performance by increasing the fatigue life many folds.

(v) The fatigue data not only depend on the stress ratio and amplitude but they also demonstrate substantial differences among the data recorded at different values of mean tensile stress. The data shown in Table 8.2 strongly strengthens the proposition that both the 316L and H13 composite cladded specimen exhibited higher tendency of fatigue failure when subjected to a large mean tensile stress in addition to having to cope with the repeated cycles of applied stress. The clear evidence can be obtained by comparing L3 and H3 experiments with L4 and H4 respectively. Similarly comparing L3 and H3 with L1 and H1 illustrates the strong influence of stress amplitude in reducing the fatigue life. It implies that fatigue life is affected by the combined effect of stress amplitude and mean tensile stress, and if the same fatigue life is needed at a lower mean stress and at higher average tensile stress, then the stress amplitude must be reduced to compensate for the higher mean stress.

(vi) A particular set of experiment designated as L5 and H5 pertains to fatigue behaviour at stress ratio, R =0.8 and mimics the exposure to low frequency vibrations under loadings greater than or comparable to yield strength of material. These experiments were carried out at very low stress amplitude (49.75 MPa) but very high mean stress (447.71 MPa). The effects are not so disastrous as observed in L2 but still both the 316L and H13 composite cladded specimen failed at substantially lower number of cycles (<10⁵). This failure is particularly significant for H13 specimen which has the ultimate tensile strength as almost double of the value of mean tensile stress.

(vii) An increase in frequency of the cyclic loading appears to have a favourable effect on the fatigue life but not as significant that may be comparable to the effects of stress ratio, stress amplitude and mean stress value. Although in these experiments, frequency was only varied from 10 to 20 Hz
but similar behaviour can be found in the S-N curves plotted for precipitation hardening and sintered steels by Sonsino [244] at frequencies of 30 Hz and 20 kHz.

(viii) The fatigue lives for the two types of tested DMD generated specimen does not show a distinct relationship with respect to change in stress ratio ‘R’. By carefully observing the point plots of Fig 8.3 it can be deduced that the fatigue life of both the 316L and H13 composite specimen shows an increase with the increase in stress ratio. But at the highest examined value of 0.8 it drops down to a value of about 25% to that at R=0.2. This observation is in contrast with what Huang et al [245] had described for HCF of welded 316L structures. But in current work, the high value of mean tensile stress is mainly influencing the occurrence of reduced fatigue life at R=0.8 thus supporting the notion that overall fatigue behaviour can only be completely explained through combination of R, $\sigma_a$ and $\sigma_{av}$.

(ix) The fatigue lifetime behaviour at higher mean tensile stress values for commercially available 316L bars was much superior as shown by the high value of $N_f$ under the experimental designation C2, which gets quite close to 1 million cycles. This value is approximately 4 times greater than the H13 composite cladded specimen as shown in H4. These differences amply highlight the need to put the laser cladded specimen under suitable mechanical and thermal treatment, particularly for improving their dynamic properties in the tensile region.

8.4 Fatigue Rupturing Behaviour using SEM Fractography

An adequate estimate of the causes and mechanisms of fatigue failure can never be complete without the analysis of fatigue fractured specimen under a microscope. In this section, the ruptured surfaces are analysed with the help of SEM microscopy, which is sometimes called as SEM fractography. The micrographs have been obtained for a number of cases including the 316L and H13 composite cladded specimen and the 316L solid cold rolled bar. The specimen that have been selected for analysis of fractured surfaces are the ones that have endured the maximum number of cycles for each material type i.e. L3, H3 and C2 according to the experimental designation of Table 8.1. The reason for this approach is to obtain a good picture of fatigue failure in the established HCF regime and the prime reason of failure after enduring more than $10^5$ cycles. It should be noted that in all the images shown in the following micrographs, the axis of applied tensile load is perpendicular outwards to the plane of image.

The influence of varying the stress ratio upon the nature of fractured surfaces for welded 316L specimen has been explained in detail by Huang et al [245]. A fatigue specimen under a lower applied maximum stress led to a larger fatigue damage region and smaller final rupture plastic region. The effect of stress ratio on the fracture features as described by them was that more negative stress
ratio should yield a larger fatigue damaged region. In the case of \( R = -1 \), the fatigue damaged region extended over the whole fracture surface. The specimen was fractured mainly by a fatigue rupture mode. The fatigue striations were the prevalent features.

Figure 8.4: SEM micrographs of the fractured surfaces after fatigue failure (a) 316L commercial rolled bar (x41) (b) 316L laser cladded composite specimen (x36)

In Fig 8.4 (a) for the commercially available wrought 316L bar, a typical ductile fracture can be observed, which is given by the name ‘cup-and-cone’ failure and mainly occurs in the following sequence: (a) Necking due to axial stretching, (b) Formation of microscopic voids, (c) Coalescence of
micro-voids to generate a finite sized crack, (d) Crack propagation by shear deformation along the maximum shear stress planes, and (e) Final fracture.

But a close examination of the fracture mechanism as illustrated by the micrograph of Fig 8.4 (b) presents a very different picture. The fracture mechanism shows three stages of fracture, which may be described as follows:

(i) Crack initiation in the substrate near the interface region in HAZ.
(ii) Opening up of cracks and voids along the circumference of interfacial layers
(iii) As the substrate gives in, the abrupt rupturing of the cladded portion occurs

Therefore, fractograph of Fig 8.4 (b) poses a very important argument which needs further investigation. That argument pertains to the question that whether the fatigue behaviour is totally governed by the weaker core of mild steel or the annular cladding which is analogous to cast stainless steel in tensile properties. This is very difficult to conclude from the limited experimental data but based on overall response of composite cladded specimen during compressive and tensile dynamic testing, the consideration of positive influence of cladding on improving the dynamic performance carries significant importance and must be thoroughly investigated.

Fig 8.5 (a) (b) & (c) examine the fractured surfaces of the cladded parts at high magnification and shows a comparison with commercially available specimen to reveal some important differences. In Fig 8.5(a) the grains are small and uniform in shape and size, while the fractured surfaces in Fig 8.5(b) shows a feathery network with long columnar grains, which commonly exist in the microstructure of laser cladded steels and has already been shown in optical micrographs in chapter 5. These columnar facets determine the macroscopic direction of crack propagation in the specimen [238]. In Fig 8.5(c) the fractured surface of H13 cladding shows fatigue striations and chunks of material removed from the surface. Another very important difference is that the fractured surfaces of laser cladded specimen show the effects of fatigue fracture along the depth as well as along the width and breadth. Or it can be said that the crack has penetrated through different layers of cladding thus giving the surface a very uneven texture with dark and bright regions. Another contrast in the nomenclature of two types of surfaces is that the columnar grains show different orientations across the cladded layers which indicate rapid rupturing and relative weakness of cladded specimen along an axis perpendicular to the cladding direction in contrast with the uniformly wrought grains along the loading axis in the cold rolled bar.
Figure 8.5: High magnification SEM images of the ruptured surfaces (a) 316L commercial rolled bar (x 1.39k) (b) 316L laser cladded composite (x 1.35k) (c) H13 cladding from composite cladded specimen (x 2.12k)
Fig 8.6 illustrates the condition of interface of the 316L and H13 composite cladded specimen after fatigue failure. As mentioned earlier that in these specimen, the performance of interfacial layer is crucial and during fatigue tests its bonding strength was severely tested by the cyclic tensile loading. As evident in Fig 8.6 (a) & (b) for both types of specimen, the interface has developed a crack, which is approximately 0.2 mm wide and it has been one of the major causes of fatigue failure.

Figure 8.6: SEM images illustrating the condition of post fracture interface (a) 316L composite cladded specimen (x341) (b) H13 composite cladded interface (x 326)
8.5 Summary

The study of dynamic fatigue behaviour under cyclic tensile loading with non-zero mean stress for laser cladded 316L and H13 specimen generated as composite structures with mild steel bar as substrate was presented along with the analysis of fractured surfaces under scanning electron microscope. This investigation was not an attempt to build S-N curves for laser cladded steel alloys like 316L and H13. The investigation also looked into the inter-relationship of stress ratio, stress amplitude and mean stress for determining the fatigue life. The emerging pattern demonstrates that the cladded specimen have good fatigue life when loaded within the yielding range, but their fatigue performance rapidly deteriorates when stretched into plastic range and particularly at higher values of mean tensile stress. In comparison to the commercial wrought stainless bar the fatigue life is very less under high mean stress values. When looked under SEM, two important aspects were revealed. First one shows the vulnerability of cladded layers against tensile loading by revealing the uneven fractured surface and columnar grains randomly oriented along the plane of view and into the depth. Second aspect relates to the weakness around the circular interfacial region with images showing 0.2 mm wide cracks developed as a result of fracture.
CHAPTER 9
CONCLUSIONS AND FUTURE WORK

9.1 General Remarks

As additive manufacturing technologies are advancing at a rapid rate in many aspects, therefore, any related research work needs to be well focussed and definitive to obtain some meaningful conclusions within the stipulated time. This research work has been conducted along two main streams of investigations as highlighted in the flowchart presented in Fig 1.1. The first major aspect of this research is dedicated to the characterization of Direct Metal Deposition (DMD) system whose features have the greatest potential, among all additive manufacturing technologies, to enable its induction as a valuable addition to the industrial manufacturing systems. During the course of research on DMD, the merits and demerits of the system have been thoroughly investigated in order to present a consolidated view regarding capabilities and constraints of DMD setup. Once the potential of DMD machine was fully realized, the second major line of investigation, which has its own offshoots, was pursued that was related to mechanical characterization of DMD generated or cladded parts. The prevalent approach in mechanical characterization, as evident from the cited published literature, has been to elucidate the material, metallurgical and static properties of the laser assisted metallic coatings or parts. Since a large portion of application of AM fabricated parts is directed towards biomedical scaffolds and implants, therefore, the just mentioned approach of mechanical characterization makes sense. But if the DMD generated parts have to become viable for construction, industrial and load bearing applications, then the most important aspect to investigate is the dynamic properties under compressive, tangential or wear and tensile loading. Hence for mechanical characterization, this research highlights the investigating stream pertaining to dynamic behaviour of DMD generated specimen with emphasis on high strain rate dynamic testing under compressive loads, investigation of wear behaviour and evaluation of co-efficient of friction under dry metal to metal contact and fatigue behaviour in high cycle range under cyclic tensile loading.

Before elaborating the research contributions and future research directions, couple of very important hallmarks of this research must be brought into consideration, so that the conclusions of this research may be viewed in the right perspective. These are:

A) From the outset, the investigation remained mainly experimental and focussed on DMD generated parts instead of laser generated coatings. Emphasis on experimental methods of investigation was due to the absence of well-established mechanical and material properties database for “laser generated materials”. So there is much need to build up the confidence,
through characterization, in practical utility of laser assisted DMD parts (not only coatings) for real life and industrial applications, which is a task that can only be accomplished through experimental investigation.

B) For the experimental investigation into dynamic properties of DMD generated parts, there may be two independent approaches. First is to perform investigation after applying surface finishing and heat treatment operations on the DMD generated parts, while second is to put the cladded parts to test just after removing the top rough layer without any polishing and thermal treatments. This research followed the latter course because one of the major claims of AM technologies is that these can accomplish the production of “near net shape” parts. So if the cladded parts could not prove their practical usefulness in as-cladded condition then there will be little encouragement in investing hugely for a complex setup like DMD machine.

9.2 Major Research Contributions
The major conclusions of this research pertain to thorough assessment of the potentiality and limitation of Direct Metal Deposition (DMD) process and the material and mechanical characterization of those parts to assess the possibility of producing fully functional parts. The prominent contributing aspects with respect to both streams of investigations are presented in the following discussion under separate headings for better clarity.

9.2.1 DMD Based Research Accomplishments
Although similar in operation to a LENS system, the DMD system, with its unique combination of large bed size, versatility in motion of laser beam and options for part manoeuvrability, generates an extremely suitable environment for coating and repair work. This gives DMD a distinct advantage in comparison to other AM systems for industrial applications. It is therefore, very important to approach the process from the correct perspective so that the offered advantages may be adequately tapped and explored. In chapter 4, a detailed account is given of the research that established the capabilities of DMD to process titanium alloy and steel alloy powders. DMD was found to be extremely suitable for processing of stainless and tool steel powders. For these materials, very good results were obtained not only from the aspect of cladding the profiles and building appreciable layer thickness but also from the aspect of achieving very good bonding strength when used with low cost mild steel as substrate.

Although all discussions on laser assisted processes have always taken into account the role of important process parameters, but this research has explored and optimized the controlling and controlled DMD parameters for processing different types of metallic powders and producing solid and porous parts. In the parametric study of DMD process, it has been concluded that the parameters associated with laser beam like beam power, S_focus and z-height play the most important role in
cladding. At the secondary level of importance, laser parameters combined with process parameters, which include powder feed rate, laser scan speed and inert gas flow, determine the quality of deposition, dimensional accuracy and bonding strength of the clad. It was also observed that increasing the laser power beyond a critical value does not improve the cladding process, a fact, which can greatly improve the cost and efficiency of laser cladding process. Conclusively, optimum cladding ranges were investigated and recommended for producing solid and porous parts.

While reviewing the quality of DMD processed parts, this research can be regarded as ending up with some positive and some negative results and outcomes. Negative results were obtained with respect to producing biomedical scaffold structures, which usually require pore sizes in the range of 0.1–0.5 mm, processing of titanium alloys due to very high affinity with oxygen and inconsistency of dimensions when part height was increased beyond 15 mm. With steel alloys, good quality porous structures were successfully developed but the pore size was limited to approximately 1 mm on the minimum side. Therefore, DMD process was ruled out as an unsuitable AM candidate for producing biomedical scaffolds and implants, especially in comparison to the results already obtained by selective laser sintering processes for polymers and metals.

This research has contributed significantly in bringing forth the potential of DMD setup in producing customized shapes with such dimensions that these claddings can be used as individual parts. The optimization of DMD parameters resulted in producing solid and porous parts, which exhibited material properties like hardness and strength to be comparable with commercially available cast and wrought alloys. The metallurgical investigations, as presented in chapter 5 revealed consistent microstructures and the cores largely free of microscopic pores and voids. The cladded surfaces were apparently very rough, but only after machining off 0.25 – 0.4 mm from the surface, clean and smooth surface free of cracks and non-homogeneities was obtained. The height of structures built on DMD may induce inconsistency in outside dimensions, but for solid parts an allowance of 0.5 to 0.7 mm can be easily maintained during cladding that can be later machined off to produce fully prismatic surfaces. Structures with heights up to 15 mm have been successfully produced in this manner.

9.2.2 Mechanical Characterization of DMD generated parts
This research concentrated on mechanically characterizing those DMD generated specimen which can be used as separate parts or components. In this respect, three different types of specimen were produced and tested, which were solid, porous and composite. The contribution of this research in ascertaining the dynamic characteristics of DMD generated steel alloy parts is significant and gradually proving its worth through publications in high impact factor journals. The three major lines of investigations pursued during the course of this research in the perspective of important achievements are presented as follows:
9.2.2.1 Dynamic Testing under Compressive load

One of the major features of this research is the detailed investigation of laser cladded 316L SS and H13 TS parts with solid, porous and composite configurations tested under high strain rate compressive loading. The equipment used was the Split Hopkinson’s pressure bar (SHPB) at strain rates greater than 2500 with all specimen in as-cladded condition without any post cladding heat treatment. The whole set of dynamic stress strain curves along with numerical modelling according to established stress strain relationships for alloy steels is presented in chapter 6 of this thesis. The important conclusions which are acknowledged as addition to the available research can be summarized in the following manner:

i. DMD generated specimen proved to be very efficient in absorbing the impact loads as every part out of 30 tested samples did not break or split apart or even developed macroscopic cracks. This observation was very encouraging as it exhibited a strong core and stable microstructure without any significant presence of defects or cleavage planes or inclusions that may initiate load induced fracture.

ii. When the compressive dynamic properties of laser generated parts were compared with the commercially available steel alloys, then the stress vs strain behaviour was found to be inferior to wrought and heat treated parts from the same materials but appreciably better than cast and hot worked specimen.

iii. A particularly important contribution is the strength shown by porous and composite parts against the striking force of very high momentum impact bar. These specimens showed good capability of absorbing impact energy through ductility and elastic+plastic deformation without rupturing. In composite specimen, a very important observation was the absence of separation or de-bonding at the interface layer between substrate and DMD cladding.

iv. Under compressive static loading, all types of DMD generated parts showed substantial work hardening and very large amount of compressive strains, with some samples being compressed to 70% of their original thickness. This behaviour coupled with the dynamic response rendered DMD produced steel alloy parts extremely suitable to be used in applications under compressive loading.
9.2.2.2 Abrasive Sliding Wear Testing

Another important segment of mechanical characterization investigation contributed in unfolding the behaviour of laser cladded stainless and tool steel parts when subjected to tangential forces at high speeds and normal loads. This brings forth the wear behaviour and friction related characteristics and the evaluated properties, which were duly presented and discussed in chapter 7. The database of generated information can usefully contribute, both quantitatively and qualitatively, in using DMD generated steel parts for developing or improving metallic machine components undergoing sliding motion. Few important conclusions obtained from this aspect of research can be presented as follows:

i. Under normal loads approaching 50 N and sliding distances nearly 5 km, no disintegration, deformation or distortion of the DMD cladded parts was observed. The SEM analysis of worn surfaces also did not reveal microscopic cracking although bruises and scars were present. This was a significant outcome for cladded parts without any post cladding operation for improving material quality.

ii. The acquired co-efficient of friction was apparently 7 to 10 % higher in comparison to the values mentioned in handbooks. But keeping in view the severity of conditions under which results were evaluated, this difference may be insignificant. Nevertheless, researchers have reported co-efficient of friction greater than 1 for most severe conditions while testing steel on steel counter faces.

iii. Wear rates were also slightly higher (<10%) than wrought steel counterparts, which is understandable for parts created by laser fusion of metallic powders. But the obtained values were highly encouraging for as-cladded parts and could be further improved by hardening of the surfaces.

9.2.2.3 Material and Metallurgical Investigation

Although there is abundance of research information available in relation to material and metallurgical characteristics of laser cladded metallic parts, but the scope of research is so enormous that some room for further enhancement always exists. Prime conclusions extracted from the results presented in chapter 5 of this thesis can contribute in throwing some fresh light as given below:

i. The difference in microstructure was observed at different laser powers used for cladding, which shows that microstructure of cladded steels evolve with the increment in laser power, not indefinitely but up to a certain level beyond which effects diminish. The metallurgical phases were not well developed at 500 watts which was the minimum beam power needed for producing solid and well bonded metallic parts. The microstructure
revealed grains, grain boundaries and phases much more clearly when laser power increased up to 1500 watts.

ii. Micro-hardness of 316L and H13 steels also exhibited slight increase with the power used during cladding. These values emphasize the importance of optimization of beam power and positioning with respect to substrate and powder stream for producing metallic specimen with superior material properties that can be comparable to wrought steels.

iii. Residual stress values obtained from neutron diffraction measurement for 10-12 mm thick composite slabs of DMD produced 316L and H13 specimens show important results that can be used in post cladding heat treatment and improving the fatigue behaviour of laser cladded parts. The concave curve was obtained within the substrate while a dominantly convex profile culminating into compressive stress at the top surface was exhibited within the cladding. This observed behaviour is very useful because the compressive residual stresses contribute in maintaining the integrity of cladding as noted under high velocity compressive loading impacts on SHPB.

iv. The maximum values of residual stresses do not exceed 250 MPa at any point in the substrate+clad system, which shows that the values are well below the yield stress of the material. Such a set of values clearly demonstrate the merits of DMD based laser cladding process in terms of localized heating, rapid solidification and absence of adverse thermal gradients.

9.2.2.4 High Cycle Fatigue under cyclic tensile loading

This line of investigation also revealed some important contributing pieces of information although the scope of research remained limited to composite DMD cladded specimen made from 316L and H13 claddings. The performance of laser generated composite specimen was not as impressive under cyclic tensile loading as was found under high strain rate and static compressive loading. Chapter 8 contains the quantitative data showing the S-N plots for two types of composite specimen at different stress ratios and variable mean tensile stress. The combined picture provides useful insight into the possible operating zones and estimated performance for DMD generated composite specimens. The important aspect, which is less highlighted in published research on fatigue and can contribute in defining fatigue behaviour is the significant contribution of mean tensile stress in combination with the stress ratio. For laser cladded composite parts, the research also exposed the vulnerability of bond between substrate and cladding which was not so inconsistent in composure under compressive loads.
9.3 Directions for Future Work

Since both the major streams that have been investigated during this research possess enormous scope, therefore, it is inevitable that at the conclusion of research many avenues remain unexplored and certain questions left unanswered. Although this research invested appreciable effort in ascertaining the capabilities and limitation of DMD process alongside experimentally assessing the dynamic mechanical behaviour of DMD generated parts, a good deal of work needs to be done in respect of attaining satisfactory performance for laser generated structures under tensile cyclic loading. In this research only composite structures were tested but there is a need for figuring out the optimized part shape and associated laser processing parameters. Another avenue open to research is the nature of heat treatment that can be applied on the laser cladded structures for improving the fatigue performance. It is beyond doubt that the applied heat treatment will improve the quality of material, but the combination and extent of these operations need to be carefully evaluated in order to keep the laser based manufacturing cost and time competitive.

On the basis of results obtained, it can be safely concluded that a very good option for further research is to produce lattice and porous structures on DMD that can be used in high impact energy absorption applications. The non-linear elasticity and higher work hardening rates in the plastic region, as shown in the plotted stress strain graphs make these high strength alloys in porous configurations ideally suited for absorbing energies of high velocity impacts. This is the avenue that can be safely pursued, since this research has amply demonstrated the capabilities of DMD cladded alloy steel parts under high strain rate compressive loads.

Another recommendation for further work is to produce profiled shapes like gears and cams with sizes up to 150 mm but thickness not more than 10 mm to avoid excessive machining owing to height related tapering effect. This option can be very handy for the robotics and mechatronic applications, due to very rapid conversion from CAD model to the manufactured part in a DMD setup. As it was clearly observed that roughness of DMD cladding does not penetrate deep into the bulk and strength of cladded parts is quite superior to castings, therefore, in moderately loaded applications such an effort may render functional machine components only after surface finishing operations.

A very important aspect of this research was to investigate the behaviour of composite parts beside solid and porous parts. The composite cylindrical parts with mild steel core and outside cladding may provide strong machine components that are economically produced and may be used as shafts and torque transmission components. Therefore, a useful set of investigations may be initiated that test the fatigue strength of these composite parts under cyclic tensile and torsional loads. The compressive strength has already been tested in this research which proves these parts to remain highly integrated.
(no separation at the interface layer) and strong with deformation values dictated by the cladding and not the core.

The final recommendation for furthering this research work is related to the investigation on residual stresses. It should be beneficial to probe into the residual stresses for composite cylindrical parts and the cladded parts that are separated from the substrate. Since the neutron diffraction and X-ray diffraction facilities are available in Australia, therefore, a thorough investigation, also taking into account the effects of stress relaxation as a result of heat treatment operations, may not be out of place. But this investigation may take appreciable time because both the DMD machine and diffractometer are one-off equipment and continuously remain in demand by the researchers.
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