Copper Based Bi-Metallic Tooling for High Pressure Die Casting Using Direct Metal Deposition

By

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In high pressure die casting (HPDC) cooling time greatly affects the total cycle time. As thermal conductivity is the main governing factor, a higher thermal conductive die material allows faster extraction of heat from the casting, thus resulting in shorter cycle time and higher productivity. This thesis presents a novel approach to replace a conventional steel die by a bi-metallic die made of copper alloy in which the cavity surface is coated with a protective layer of tool steel using laser based additive manufacturing technology, Direct Metal Deposition (DMD) for high pressure die casting application. Results obtained from the finite element heat transfer analysis showed that bi-metallic die offered superior thermal performance compared with monolithic steel die.

However, the main challenge lies in the deposition of tool steel on copper alloy due to the very different material properties of these two materials. Over the past decade, researchers have demonstrated interest in tribology and prototyping by the laser aided material deposition process. Laser aided DMD enables formation of a uniform clad by melting the powder onto a substrate material to form desired bi-metallic component. Thus, the principal objective of this research is to investigate the feasibility of deposition of a protective tool steel layer on copper alloy substrate material to form bi-metallic structure using DMD technology for HPDC tooling.

Heat transfer analysis, process and material characterization, tooling fabrication and industry evaluation have been investigated in this research to attain the stated aim. The following outcomes have been achieved from this thesis in order to develop an effective bi-metallic tooling for HPDC industries-

- Reduction of cycle time by employing bi-metallic die compared to that of tool steel for improvements to productivity of existing processes in HPDC.
• Establishment of feasibility of the application of DMD technique to deposit protective tool steel clad on completely different material i.e. copper alloy substrate.

• Investigation of microstructural characteristics of the DMD deposited tool steel clad on copper alloy substrate.

• Evaluation of mechanical properties namely, bond strength and fracture toughness of the bi-metallic structure.

• Analysis of thermal fatigue behaviour of the bi-metallic structure under conditions analogous to high pressure die casting environment.

• Fabrication and performance evaluation of bi-metallic core pin in a semi-industrial high pressure die casting machine.

The outcomes of this research offer innovative technologies based on leading edge research in laser based rapid tooling development for enhancing productivity and reducing lead time in die casting industry. This research has the potential to directly impact multi-million dollar die-casting industry through improved process efficiency, reduced cycle time and lower cost of component processing.
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DEDICATED

“To my father, Late Mohammad Abdul Quayum Mia, who was very passionate to see it, but, could not.”
DECLARATION

I hereby declare that I am the sole author of this dissertation and to the best of my knowledge, it contains no material that has been published by others previously except where references have been made. This is the true copy of the thesis that has no material accepted for the award of any other degree or diploma at any university.

I understand that my dissertation may be available to others electronically.

Mohammad Khalid Imran

April, 2012
LIST OF PUBLICATIONS

Refereed International Journals:


Peer Reviewed Conference Proceedings:


# TABLE OF CONTENTS

<table>
<thead>
<tr>
<th>Chapter</th>
<th>Title</th>
<th>Page</th>
</tr>
</thead>
<tbody>
<tr>
<td>1</td>
<td>INTRODUCTION</td>
<td>1</td>
</tr>
<tr>
<td>1.1</td>
<td>General</td>
<td>1</td>
</tr>
<tr>
<td>1.2</td>
<td>Cycle Time and Thermal Management in HPDC</td>
<td>2</td>
</tr>
<tr>
<td>1.3</td>
<td>Aims and Objectives of Research</td>
<td>3</td>
</tr>
<tr>
<td>1.4</td>
<td>Significance of Research and Major Contribution</td>
<td>4</td>
</tr>
<tr>
<td>1.5</td>
<td>Organization of Chapters</td>
<td>6</td>
</tr>
<tr>
<td>2</td>
<td>LITERATURE REVIEW</td>
<td>8</td>
</tr>
<tr>
<td>2.1</td>
<td>High Pressure Die Casting</td>
<td>8</td>
</tr>
<tr>
<td>2.2</td>
<td>Modes of Heat Transfer in HPDC</td>
<td>11</td>
</tr>
<tr>
<td>2.3</td>
<td>Factors Influencing HPDC Cycle Time</td>
<td>12</td>
</tr>
<tr>
<td>2.3.1</td>
<td>Effective Cooling System in HPDC</td>
<td>14</td>
</tr>
</tbody>
</table>

*ACKNOWLEDGEMENT* ........................................................................ iv  

*DEDICATED* ................................................................................ vi 

*DECLARATION* ............................................................................. vii 

*LIST OF PUBLICATIONS* ................................................................ viii 

*TABLE OF CONTENT* .................................................................... x  

*LIST OF ILLUSTRATIONS* ............................................................... xv 

*LIST OF TABLES* .......................................................................... xxiii 

*CHAPTER 1  INTRODUCTION* ............................................................... 1 

1.1. General .................................................................................. 1 

1.2. Cycle Time and Thermal Management in HPDC ......................... 2 

1.3. Aims and Objectives of Research ........................................... 3 

1.4. Significance of Research and Major Contribution .................. 4 

1.5. Organization of Chapters ...................................................... 6 

*CHAPTER 2  LITERATURE REVIEW* .................................................. 8 

2.1. High Pressure Die Casting ................................................... 8 

2.2. Modes of Heat Transfer in HPDC ........................................... 11 

2.3. Factors Influencing HPDC Cycle Time ................................... 12 

2.3.1. Effective Cooling System in HPDC ..................................... 14
<table>
<thead>
<tr>
<th>Section</th>
<th>Page</th>
</tr>
</thead>
<tbody>
<tr>
<td>2.3.2. Thermal Conductivity of Die Material in HPDC</td>
<td>16</td>
</tr>
<tr>
<td>2.4. Development of Die materials</td>
<td>17</td>
</tr>
<tr>
<td>2.4.1. Copper Alloys as Die Material</td>
<td>17</td>
</tr>
<tr>
<td>2.4.2. Functionally Graded Material</td>
<td>20</td>
</tr>
<tr>
<td>2.4.3. Bi-metallic Die Material</td>
<td>32</td>
</tr>
<tr>
<td>2.5. Various Techniques to Manufacture Bi-metallic Structure</td>
<td>33</td>
</tr>
<tr>
<td>2.5.1. Ion Implantation</td>
<td>35</td>
</tr>
<tr>
<td>2.5.2. CVD and PVD Process</td>
<td>35</td>
</tr>
<tr>
<td>2.5.3. Welding Process</td>
<td>38</td>
</tr>
<tr>
<td>2.5.4. Thermal Spraying Process</td>
<td>40</td>
</tr>
<tr>
<td>2.6. Solid Freeform Fabrication</td>
<td>49</td>
</tr>
<tr>
<td>2.6.1. Laser Aided Fused Metal Deposition</td>
<td>50</td>
</tr>
<tr>
<td>2.6.2. Direct Metal Deposition</td>
<td>54</td>
</tr>
<tr>
<td>2.7. Applications of Direct Metal Deposition</td>
<td>59</td>
</tr>
<tr>
<td>2.8. Summary</td>
<td>60</td>
</tr>
<tr>
<td><strong>CHAPTER 3  FINITE ELEMENT THERMAL ANALYSIS</strong></td>
<td>62</td>
</tr>
<tr>
<td>3.1. Introduction</td>
<td>62</td>
</tr>
<tr>
<td>3.2. Heat Flow and Governing Equations</td>
<td>62</td>
</tr>
<tr>
<td>3.3. Design of the Die</td>
<td>69</td>
</tr>
<tr>
<td>3.4. Meshing and Boundary Conditions</td>
<td>70</td>
</tr>
<tr>
<td>3.5. Results and Discussion</td>
<td>77</td>
</tr>
<tr>
<td>3.6. Summary</td>
<td>79</td>
</tr>
</tbody>
</table>
CHAPTER 6 EVALUATION OF MECHANICAL PROPERTIES

6.1. Introduction .................................................................122

6.2. Experimental Procedures ................................................123
  6.2.1. Materials and DMD Cladding ......................................123
  6.2.2. Metallographic Analysis .............................................124
  6.2.3. Mechanical Testing ....................................................124

6.3. Results ...........................................................................126

6.4. Discussion of Results .....................................................140

6.5. Summary .......................................................................143

CHAPTER 7 EVALUATION OF THERMAL FATIGUE

7.1. Introduction .....................................................................145

7.2. Experimental Details ..................................................146
  7.2.1. Preparation of Test Specimens ..................................146
  7.2.2. Thermal Fatigue Testing .........................................148
  7.2.3. Evaluation of Test Specimens .....................................150

7.3. Results ...........................................................................151

7.4. Discussion of Results .....................................................159

7.5. Summary .......................................................................166

CHAPTER 8 HIGH PRESSURE DIE CASTING TRIALS

8.1. Introduction .................................................................168

8.2. Experimental Details ..................................................168
  8.2.1. High Pressure Die Casting Machine ..............................168
8.2.2. Preparation of Core Pins ..................................................171
8.2.3. Evaluation of Core Pins ..................................................173

8.3. Results and Discussion ..........................................................173
  8.3.1. Analysis of Material Build-up ...........................................173
  8.3.2. Die Thermal Profile .......................................................179
  8.3.3. Evaluation of Tool Steel Clad ...........................................182

8.4. Summary ..............................................................................186

CHAPTER 9 CONCLUSIONS AND FUTURE RESEARCH ..........188

9.1. Overview ..............................................................................188

9.2. Major Conclusions ...............................................................188

9.3. Recommendation for Future Research ................................189

REFERENCES ............................................................................192
LIST OF ILLUSTRATIONS

Figure 2-1: Schematic presentation of various steps involved in a typical high pressure die casting cycle [11] .................................................................9

Figure 2-2: Various modes of heat transfer and the locations where it takes place in a high pressure die casting die (a) Die open for service and (b) Die closed after shot [18] ..........12

Figure 2-3: A one dimensional representation of heat flow resistance in a typical HPDC process [22] ......................................................................................14

Figure 2-4: Graphical representation of thermal resistance in HPDC process [35] .......16

Figure 2-5: Damages caused by washout and soldering on the copper pin used in HPDC [44] ...........................................................................................20

Figure 2-6: A basic sketch of direct metal laser remelting (DMLR) [56] ..................22

Figure 2-7: A basic sketch of selective laser fusion [46] ...........................................23

Figure 2-8: Images showing insufficiently molten and irregular shaped single layer coupons on stainless steel substrate produced with a laser power of 76.6W CW, 80 per cent beam overlap and laser scanning speeds varying from: (a) 10-50 mm/s in 10 mm/s increments; and (b) 25-125 mm/s in 20 mm/s increments [48] .........................24

Figure 2-9: Optical images showing porosity in various sections of 20 × 5mm builds; (a) longitudinal; (b) transverse sections produced with two scans per layer; (c) longitudinal; (d) transverse section of the build produced with three scans per layer; (e) longitudinal; and (f) transverse sections of the build produced with four scans per layer. The laser parameters for each were a laser scan speed of 100 mm/s, laser power of 76.6W CW and a beam overlap of 75 per cent [48] .........................................................24

Figure 2-10: Optical micrograph of cracking in 25% Cu, 75% H13 coupon processed by DMLR [49] .................................................................27

Figure 2-11: Optical images showing copper-rich solid phase at the austenite grain boundaries of cross-sectioned samples (laser scan speed 400mms⁻¹). Compositions: (a) 25% Cu, 75% H13; (b) 50% Cu, 50% H13 and (c) 75% Cu, 25% H13 [49] .................27

Figure 2-12: Optical images showing various defects in laser fused functionally graded H13/Cu materials (a) only polished, (b) and (c) etched [46] .........................29

Figure 2-13: Typical cross sections of specimens built with H13-Cu blends showing macro-porosity along the cross section of the specimens: (a) filling, (b) sequential, (c) alternated
and (d) refill strategy. Also the cross sections show an area reduction compared to what should be the height of the specimens [46]

Figure 2-14: Optical micrographs of FC-0208 cylindrical tubular samples sintered at 1260uC for 20 min soak time in reducing atmosphere. Sintered samples in unetched condition were employed for observing typical porosity distributions obtained (a) microwave sintering (b) conventional sintering [61]

Figure 2-15: Comparison of coating processes [73]

Figure 2-16: Principle of electric arc wire spray [73]

Figure 2-17: Casting simulation rig used to test thermal behavior of copper alloyed dies [121]

Figure 2-18: Pseudo die block used to test thermal behavior of copper alloyed dies [121]

Figure 2-19: Copper die block prior and after thermal spraying [121]

Figure 2-20: Optical image showing porosity in the sprayed coatings [121]

Figure 2-21: Principle of thermal spraying [73]

Figure 2-22: Schematic diagram of a thermal sprayed coating with various features present in the coating [73]

Figure 2-23: Cavity defects after casting trials [121]

Figure 2-24: Casting defects due to porosity in the thermal sprayed coating [121]

Figure 2-25: Steel runner inserted due to the damage in the sprayed coating [121]

Figure 2-26: Schematic representation of SFF based on fused metal deposition technique [147]

Figure 2-27: Schematic diagram of a closed loop DMD process

Figure 2-28: Schematic diagram of a DMD feedback system [185]

Figure 2-29: Multiple sensors layout in DMD feedback system (top view) [185]

Figure 2-30: Example of fabrication with height controller: (left) with height controller, (right) no height controller [174]

Figure 3-1: Three dimensional heat conduction through rectangular element [36]
Figure 3-2: Bi-metallic die with cooling channels showing copper alloy die block and tool steel clad (a) separately (b) assembled ................................................................. 70

Figure 3-3: Tool steel die with cooling channels ................................................................. 70

Figure 3-4: Meshing of (a) bi-metallic die (b) tool steel die ........................................ 72

Figure 3-5: Faces of the die through which heat transfer was considered negligible (a) bi-metallic die (b) tool steel die ................................................................. 74

Figure 3-6: Cavity surfaces of the die on which heat flux was applied (a) bi-metallic die (b) tool steel die ................................................................. 75

Figure 3-7: Cooling channel of the die on which convective heat transfer coefficient was applied (a) bi-metallic die (b) tool steel die ......................................................... 76

Figure 3-8: Temperature distributions on the die surfaces at the end of a thermal cycle (a) bi-metallic die (b) tool steel die ................................................................. 78

Figure 3-9: Transient temperature responses of the dies over entire solidification process ......................................................................................................................... 79

Figure 4-1: Beads and clads produced in random trials ................................................. 84

Figure 4-2: Beads and clads produced with low power laser ........................................ 86

Figure 4-3: Steel clads produced with various combinations of process parameters ...... 87

Figure 4-4: SEM image of the bead produced with laser power 5 kW, powder mass flow rate 3.15 gm/min, focus -15 and feed rate 150 mm/min ............................................ 89

Figure 4-5: SEM image of the bead produced with laser power 5 kW, powder mass flow rate 1.58 gm/min, focus -18 and feed rate 150 mm/min ............................................ 90

Figure 4-6: SEM image of the bead produced with laser power 5 kW, powder mass flow rate 0.8 g/min, focus -18 and feed rate 150 mm/min ............................................ 90

Figure 4-7: Quantitative analysis of porous area in DMD produced beads ................. 91

Figure 4-8: SEM image of the bead produced with laser power 4 kW, powder mass flow rate 0.8 g/min, focus -18 and feed rate 150 mm/min ............................................ 92

Figure 4-9: SEM image of the bead produced with laser power 3 kW, powder mass flow rate 0.8 g/min, focus -18 and feed rate 150 mm/min ............................................ 92

Figure 4-10: Burnt nozzle due to reflected heat at 3 kW laser power ......................... 93
Figure 4-11: Burnt during large area deposition with 3 kW laser power and tilted nozzle (a) back plate and (b) feedback tube .................................................................94

Figure 4-12: Vickers microhardness of clad deposited with 2 kW laser power ..........96

Figure 4-13: SEM image of the clad with 2 kW laser power and 2.3 gm/min powder mass flow rate showing incomplete melting .................................................................97

Figure 4-14: Vickers microhardness of clad deposited with 2.5 kW laser power ........97

Figure 4-15: Vickers microhardness of clad deposited with 2.5 kW laser power ........99

Figure 4-16: Vickers microhardness of clad deposited with 2.75 kW and 2.5 kW laser power ................................................................................................................................100

Figure 4-17: SEM image of the cladding with the combination of 2.5 and 2.75 kW laser power (a) fully melted layer without porous holes (b) interface between clad and substrate ................................................................................................................................101

Figure 5-1: Fe-Cu binary phase diagram showing miscibility gap [213] .................104

Figure 5-2: SEM image showing interface between H13 tool steel clad and substrate deposited using 2 kW laser power .................................................................108

Figure 5-3: SEM images of different interfaces between (a) H13 tool steel and substrate deposited using 2.5 kW laser power (b) 316 L stainless steel and substrate (c) H13 tool steel and 316 L stainless steel ................................................................................................................................109

Figure 5-4: SEM microstructure of H13 tool steel clad produced from 2 kW laser power (a) voids resulted from incomplete melting (b) dendritic microstructure ..................110

Figure 5-5: XRD spectrum of H13 tool steel clad deposited using 2 kW laser power ....111

Figure 5-6: SEM image showing microstructure of H13 tool steel clad deposited using 2.5 kW laser power .................................................................111

Figure 5-7: XRD spectrum of H13 tool steel clad deposited using 2.5 kW laser power ...112

Figure 5-8: SEM image showing microstructure of H13 tool steel clad deposited using 316L SS as buffer layer .................................................................114

Figure 5-9: XRD spectrum of H13 tool steel clad deposited using 316L SS as buffer layer ................................................................................................................................114

Figure 5-10: Vickers microhardness of clad deposited using 2 kW laser power ..........116

Figure 5-11: Vickers microhardness of clad deposited using 2.5 kW laser power ..........117
Figure 5-12: Vickers microhardness of clad deposited using 316 L stainless steel buffer layer

Figure 5-13: EDS line scan showing distribution of elements in the clad and substrate when H13 tool steel was directly deposited on copper alloy substrate

Figure 5-14: EDS line scan showing distribution of elements in the clad and substrate when H13 tool steel was deposited on copper alloy substrate with 316L SS buffer layer

Figure 6-1: Schematic drawings of the tensile test specimens prepared from (a) H13 tool steel directly cladded on substrate (b) H13 tool steel cladded with 316 L stainless steel as buffer layer (c) copper alloy substrate

Figure 6-2: Schematic drawings of the charpy impact energy test specimens prepared from (a) H13 tool steel directly cladded on substrate (b) H13 tool steel cladded with 316 L stainless steel as buffer layer

Figure 6-3: SEM images of different interfaces between (a) H13 tool steel and substrate (b) 316 L stainless steel and substrate (c) H13 tool steel and 316 L stainless steel

Figure 6-4: Tensile test performance of different clads with substrate material (half part of the specimens was clad and other half part was substrate)

Figure 6-5: Necking characteristic of different specimens after tensile test (a) 316 L buffered specimen seen from top (b) 316 L buffered specimen seen from side (c) directly cladded H13 specimen seen from top (d) directly cladded H13 specimen seen from side (e) copper alloy specimen seen from top (f) copper alloy specimen seen from side

Figure 6-6: Charpy impact energy of two different cladding specimens

Figure 6-7: Fracture morphology of the tensile test specimens (a) H13 tool steel directly cladded on substrate (b) Cu alloy (c) H13 tool steel cladded with 316 L stainless steel as buffer layer

Figure 6-8: EDAX analysis of fracture surfaces of tensile test specimens (a) H13 tool steel directly cladded on substrate (b) H13 tool steel cladded with 316 L stainless steel as buffer layer
Figure 6-9: Fracture surfaces at different regions of ‘directly cladded H13 tool steel’ charpy impact energy test specimen (a) top layer where crack initiated (b) 1st layer just above the interface (c) HAZ (d) substrate

Figure 6-10: Fracture surfaces at different regions of ‘H13 tool steel cladded with 316 L stainless steel as buffer layer’ charpy impact energy test specimen (a) top layer where crack initiated (b) 316 L stainless steel layer just above the interface (c) HAZ (d) substrate

Figure 7-1: Schematic drawing of the test specimen used in the experiment (a) H13 tool steel directly clad on copper alloy core material (b) H13 tool steel clad with 316 L SS as buffer layer on the copper alloy core material

Figure 7-2: Schematic diagram of the thermal fatigue test rig

Figure 7-3: Input energy from the induction coil applied in the thermal cycling

Figure 7-4: Corresponding surface temperatures of two different clads during thermal cycling

Figure 7-5: Optical micrographs showing the cross section of (a) H13 TS and (b) 316 L SS buffer specimen

Figure 7-6: Comparison of the crack density with number of thermal cycles on the surfaces of two different clads

Figure 7-7: Optical image showing large catastrophic cracks on the surfaces of H13 TS specimen (left side) and 316 SS buffer specimen (right side)

Figure 7-8: SEM images showing the propagation of fine and shallow thermal fatigue cracks in the clad of (a) H13 TS specimen and (b) 316 L SS buffer specimen

Figure 7-9: Optical image showing pull off feature on the surface of H13 TS specimen

Figure 7-10: SEM image showing (a) cracks in the 1st layer of H13 TS specimen and (b) crack free 1st layer of 316 L SS buffer specimen (cross sectional view)
Figure 7-11: SEM images showing (a) cracks in the 1st layer of H13 TS specimen and (b) crack free 1st layer of 316 L SS buffer specimen (longitudinal section) ..........................157

Figure 7-12: Propagation of the catastrophic crack into (a) H13 TS specimen (b) 316 L SS buffer specimen ........................................................................................................159

Figure 7-13: SEM image showing Carbon network in the cracks of 1st layer of H13 TS specimen (longitudinal section) .................................................................160

Figure 7-14: (a) SEM morphology and (b) distribution of Fe and Cu elements along the cross-section of 316 L SS specimen .................................................................161

Figure 7-15: (a) SEM morphology and (b) distribution of Fe and Cu elements along the cross-section of H13 TS specimen .................................................................162

Figure 7-16: XRD spectrum of the 1st layer of H13 TS specimen ..................................164

Figure 7-17: XRD spectrum of the 1st layer of 316 SS specimen ..................................164

Figure 8-1: High pressure die casting die used in this experiment at open position showing various features .................................................................169

Figure 8-2: Moving half of the die used in the experiment with core pin placed in the cavity .............................................................................................................170

Figure 8-3: A casting part produced in the HPDC trials, showing hole feature produced by the core pin ..........................................................................................171

Figure 8-4: Schematic drawing of core pins prepared for the HPDC trials (a) long bi-metallic (b) long tool steel (c) short bi-metallic and (d) short tool steel core pin .......173

Figure 8-5: Optical photograph of core pins after HPDC trials from left to right: bi-metallic short, bi-metallic long, tool steel long and tool steel short core pin .........................174

Figure 8-6: Optical photograph of casting parts showing the material build up trend on tool steel short core pin due to soldering .....................................................175
Figure 8-7: (a) Short tool steel core pin with a piece of casting part stuck after thirty five HPDC shots (b) casting part .............................................................176

Figure 8-8: Optical photograph of casting parts showing the material build up trend on bimetallic short core pin due to soldering .........................................................177

Figure 8-9: Optical photograph of casting parts showing the material build up trend on bimetallic long core pin due to soldering ..............................................................178

Figure 8-10: Optical photograph of casting parts showing the material build up trend on tool steel long core pin due to soldering ..............................................................178

Figure 8-11: Thermal profiles of core pins and fixed die during HPDC trials (a) tool steel short (b) bi-metallic short (c) bi-metallic long and (d) tool steel long core pin ..............180

Figure 8-12: Thermal profiles of core pins for a single cycle at equilibrium stage during HPDC trials ................................................................................................................181

Figure 8-13: SEM micrographs of the interfaces after 50 HPDC cycles (a), (b) long bi-metallic core pin and (c), (d) short bi-metallic core pin ......................................................184

Figure 8-14: SEM micrograph showing the cracks in the tool steel clad in long bi-metallic core pin .............................................................................................................185

Figure 8-15: SEM micrograph of the front face of bi-metallic core pin where copper alloy was exposed to aluminium casting material .....................................................186
LIST OF TABLES

Table 2-1: Properties of selected die materials [42] ..................................................18
Table 2-2: Principal coating processes and characteristics [73] .................................34
Table 3-1: Lists of tool steel properties [42] .................................................................76
Table 3-2: Density of Moldmax [42] ........................................................................76
Table 3-3: Specific heat of Moldmax [42] .................................................................77
Table 3-4: Thermal conductivity of Moldmax [42] ......................................................77
Table 4-1: Chemical composition of copper alloy (Moldmax) substrate .................83
Table 4-2: Chemical composition of H13 tool steel ..................................................83
Table 4-3: Summary of the random trials .................................................................84
Table 4-4: Summary of the trials to reduce reflection ..................................................85
Table 4-5: Summary of the parametric combinations to deposit blocks ..................87
Table 5-1: Chemical composition of 316L stainless steel .......................................106
Table 6-1: Bond strength of different clads with substrate .......................................129
Table 6-2: Fracture toughness of different clad and copper alloy substrate combinations .........................................................................................................................140
Table 7-1: Number of cracks (N), mean crack length (CL) and maximum crack length (CLmax) of the thermal fatigue test specimens .................................................153
Table 7-2: EDAX analysis on the black network confirming presence of C ..............162
Table 8-1: High pressure die casting parameters and material used in the trials ..........170
Table 8-2: Maximum and Minimum temperatures of the core pins ........................181
1.1. General

Metal can be shaped in many ways to fabricate a designed part. For instance, machining off unwanted material, joining two materials together, roll milling, squeezing heated or cold material to simple shape and pressing the plate material to a certain shape are some of the commonly used techniques in manufacturing industries. When people discovered that it was possible to melt metals, they have tried to form it into designed shapes by pouring the liquid metal into moulds whose shape it would retain during and after solidification. This process is known as casting and if permanent moulds are used, it is called die casting [1]. Die casting has now reached into a very sophisticated manufacturing process that plays an important role in providing products for today’s high standard of living. Today, die casting is a high volume manufacturing method widely used to produce components for automotive, industrial, transportation and household applications. In US, the annual market for die casting is about AUD 9 billion and in Australia about AUD 3 billion [2, 3].

Die casting by pressure injection first occurred in the middle of 19th century where manually operated machine was used for casting printing types [4, 5]. Mass production of many products began in the beginning of 20th century. The die casting technique has progressed from the original manual low-pressure injection method to today’s complicated high pressure die casting (HPDC), squeeze casting and even to semi-solid die casting techniques. Nevertheless, high pressure die casting is the most commonly used casting process and deals with a range of materials including both heavy and light metals. Particularly, aluminium and
magnesium are two widely used metals for fabricating parts with complex shapes using high pressure die casting.

1.2. Cycle Time and Thermal Management in HPDC

To produce a casting part, HPDC involves various processes such as injection of molten metal at high pressure and high velocity into a die cavity, holding the dies for solidification of molten metal, opening of die halves for ejection of the casting part and die cooling by lubricant spray. These processes are repeated in cyclic order and the time required to perform one cycle is known as cycle time in HPDC industries. The success of the HPDC industries depends largely on high volume production, lower production cost and enhanced mechanical properties of castings. Since all these factors are reliant on reduced cycle time, quick dissipation of heat from the casting is the most critical feature for the profitability of the process. Now that quick solidification of the casting is further dependant on design of cooling system and die material, these two factors require considerable attention.

Application of extensive cooling system has been useful in rapid heat extraction from HPDC dies [6]. As a result cooling time analysis in dies has received some attention in recent years [7, 8]. Accurate design and location of cooling channels provides effective cooling of the casting. In particular, design of conformal cooling channel has been very effective in the reduction of solidification time of casting part. Though cooling channels are used to remove heat from the dies, the governing factor that transfers heat from the casting to the cooling channels is the thermal conductivity of the die material. Therefore, thermal conductivity of the die material is the main factor in the solidification of the casting part. Use of a higher thermal conductive die material can significantly reduce the cycle time to increase the productivity. However, application of high thermal conductive die material in order to reduce cycle time in die casting cannot be found in the open literature.
Hence, the research area that needs significant attention for the quick solidification of casting parts is the selection of suitable die material.

The dies and inserts required to produce die castings are usually made of hot tool steels, such as H13. However, the low thermal conductivity of tool steel sacrifices the quick solidification of casting. Since a higher thermal conductivity die material allows faster extraction of heat from the casting, a high thermal conductive die material instead of tool steel would result in shorter cycle time and higher productivity. In addition, the faster cooling of the casting would be useful for improved mechanical properties resulting from the growth of fine microstructure due to rapid solidification.

Copper and its alloys are well known for their exceptional high thermal conductivity. Excellent productivity has been achieved with such copper alloy tooling in injection moulding of plastics. Among various important properties of copper alloys, the ones that make them to be chosen over other metals are good resistance to corrosion, excellent electrical and thermal conductivity, as well as strength, resistance to fatigue and ability to take a good finish [9]. Some copper alloys, particularly Moldmax are renowned for outstanding high strength property given by the age hardening. Moldmax HH for instance, can reach a tensile strength as high as 1280 MPa. However, copper moulds and inserts cannot be used in die casting of aluminium, because of the tendency of copper to dissolve in molten aluminium. The die casting industry is therefore continually striving to develop high thermally conductive die materials in order to increase productivity and profitability.

1.3. Aims and Objectives of Research

This research aims to develop a novel bi-metallic tooling for high pressure die casting of aluminium alloys which has superior performance compared with monolithic steel dies. The tooling involves a solid copper alloy substrate onto
which layers of H13 tool steel are deposited using the laser additive manufacturing process called Direct Metal Deposition (DMD) to protect the copper alloy surface from direct contact of molten aluminium alloys. This is a new approach to die design with a potential for significant improvement in heat extraction due to the use of copper alloy substrates while at the same time offering the protection and benefits of H13 tool steel when dealing with high pressure die casting of aluminium alloys. Unlike other layer by layer manufacturing approaches, in the proposed methodology, copper and steel are not pre-mixed rather steel in the form of powder is deposited on a solid copper alloy substrate using laser cladding technology to create a bi-metallic structure. A detailed investigation involving tooling fabrication, process characterisation, heat transfer analysis and industry evaluation has been followed to attain the stated aims.

The outcomes of this research offer innovative technologies based on leading edge research in laser based rapid tooling development for enhancing productivity and reducing lead time in die casting industry. This research has the potential to directly impact multi-million dollar die-casting industry through improved process efficiency, reduced cycle time and lower cost of component processing.

1.4. Significance of Research and Major Contribution

This research has developed a novel bi-metallic tooling for high pressure die casting of aluminium alloys which has superior heat transfer performance compared with traditional steel dies. The tooling, involving a solid copper alloy substrate onto which a layer of H13 tool steel is deposited in order to protect the copper alloy surface from molten aluminium is a new approach to die design. This approach has been made possible by the recent developments in the laser metal deposition technologies, in particular, the development of monitoring and control systems capable of adjusting in real time the deposition parameters in order to compensate for melt pool size variation due to substrate/structure’ thermal mass. The deployment of high end direct metal deposition system, computer aided
design technique and engineering simulation tools helps to develop an innovative engineering environment for faster delivery of such tooling.

The significance of the research lies in its potential to revolutionize the high pressure die casting industry both locally and internationally by discovering an innovative methodology for manufacturing more efficient cost effective tooling. It will lead to significant economic and environmental benefits as the cycle times are reduced. The innovation lies in the method of depositing a high melting point, partially soluble material, H13 tool steel powder, onto a significantly lower melting material, solid copper alloy with good metallurgical bonding and fine microstructure. It has been achieved through a detailed experimental program to generate the optimum deposition parameters such as clad thickness, microstructure, bond strength with substrate and thermal fatigue resistance in order to both withstand the harsh environment associated with high pressure die casting of aluminium alloys. Moreover, the investigation involves a comprehensive study of current trends in the area and the development of novel techniques and methodologies involving computer based and laser based state of the art technology of rapid fabrication.

The following outcomes have been achieved from this research in order to develop an effective bi-metallic tooling for HPDC industries-

- Demonstration of reduction of cycle time by employing bi-metallic die compared to that of tool steel for improvements to productivity of existing processes in HPDC.
- Establishment of feasibility of the application of DMD technique to deposit protective tool steel clad on completely different material i.e. copper alloy substrate.
- Investigation of microstructural characteristics of the DMD deposited tool steel clad on copper alloy substrate.
• Evaluation of mechanical properties namely, bond strength and fracture toughness of the bi-metallic structure.
• Analysis of thermal fatigue behaviour of the bi-metallic structure under condition that is analogous to high pressure die casting environment.
• Fabrication and performance evaluation of bi-metallic core pin in a semi-industrial high pressure die casting machine.

1.5. Organization of Chapters

Chapter 1 presents objectives, outcomes, significance and benefits of this research. It then outlines the structure of the thesis.

Chapter 2 presents an in depth literature review on high pressure die casting along with development trends of die materials for efficient heat transfer from the casting. Development of bi-metallic die and various techniques that could be employed to manufacture the bi-metallic structure has also been reviewed. Finally, direct metal deposition, a laser based additive manufacturing technique which has been used to fabricate the bi-metallic die has been reviewed thoroughly in this chapter.

Chapter 3 presents the finite element heat transfer analysis of both bi-metallic and conventional H13 tool steel die. This chapter provides the comparison of the heat transfer and cooling time of the bi-metallic tooling with the conventional steel tooling.

Chapter 4 presents the optimization of DMD process parameters to deposit tool steel clad on copper alloy substrate. The melting point of pure copper at 1083 °C is significantly lower than that of tool steel which melts over the temperature range of 1370-1460 °C. In addition, copper is highly reflective to the output wavelength of CO₂ laser used in this study. Selection of suitable process parameters was therefore a major challenge of this project to produce a metallurgically sound, crack and
pore free tool steel clad on copper alloy substrate and has been described in this chapter.

Chapter 5 presents the microstructural characterization of the tool steel clad when deposited on copper alloy substrate. Tool steel and copper are only partially soluble in the solid state and the concentration of copper within the steel clad should be minimized during the DMD processing. Therefore, a solution has also been introduced in this chapter to eliminate copper dilution in tool steel clad.

Chapter 6 presents evaluation of mechanical properties of the tool steel clad. The final characteristics and accuracy of a tooling fabricated through the DMD process are strongly dependent upon mechanical properties. Mechanical properties such as bond strength, impact energy and fracture toughness have been investigated in this chapter.

Chapter 7 presents the thermal fatigue behavior of the bi-metallic structure. Tool steel was coated on copper alloy substrate and was subjected to cyclic heating and cooling condition simulating high pressure die casting environment. This chapter describes the analysis of the cracks and evaluation of the integrity of tool steel clad.

Chapter 8 presents the industrial evaluation of bi-metallic core pins under die casting conditions. The casting trials were done in CSIRO’s accelerated high pressure die casting facility.

Conclusions along with recommendation of future work have been presented in Chapter 9.
CHAPTER 2
LITERATURE REVIEW

2.1. High Pressure Die Casting

High pressure die casting (HPDC) is a near net-shape manufacturing process employed to produce geometrically intricate components of light metal alloys [10]. HPDC is a process with high economic impact since it can produce castings at a high production rate with close dimensional tolerance, smooth surface finish and fine intricate details. It is a monolithic process in which molten light metal alloy is injected into water cooled metallic dies under high velocity and pressure. The whole process consists of the following steps-

1. Ladling or pouring the molten alloy from the furnace into the shot sleeve.
2. Injecting the melt into a die cavity at high speed (30-100 m/s) through the forward movement of the plunger. At this stage, molten metal flows through the runner to the die cavity where it solidifies to form specific shape.
3. Holding the die under high pressure (50-80 MPa) to allow the melt to solidify at a high cooling rate, 100-1000 °C/s, particularly to accommodate the formation of fine grained casting. Holding the die at high pressure reduces the porosity produced during filling and solidification and reduces the contact thermal conducting resistance between the die and the casting which increases the productivity and casting surface quality.
4. Ejection of the casting. At this step, various ejection pins push the casting parts to separate it from the die.
5. Die cooling and lubrication with water and lubricant mixture. Die spraying helps to cool the die, but more importantly spraying of water based
lubricant deposits a thin layer of lubrication material on the die surface which prevents soldering and makes the casting ejection easier.

Total time required to perform step 1 to 5 is designated as cycle time. The casting rate can be over a hundred parts per hour, with a cycle time of 30-60 seconds per part being very common. Figure 2-1 shows various steps involved in a typical high pressure die casting cycle.

Figure 2-1: Schematic presentation of various steps involved in a typical high pressure die casting cycle [11].
HPDC is a process of choice in many manufacturing industries- automotive, hardware, electrical and electronics, computers and many others since it produces aluminum, zinc and magnesium alloy components with satisfactory properties at competitive prices. The main advantages of HPDC are [12, 13] -

1. HPDC is able to provide complex shapes within close tolerances
2. High rates of production with little or no machining required
3. HPDC parts are durable, dimensionally stable and have a good appearance
4. HPDC process produces low scrap
5. HPDC is a monolithic process which can combine many functions in one complex shaped part.

The load conditions during each stage of the process make great demands on the quality of the die material and therefore the type of die material is critical to the quality of castings and continuous operation of the die [14, 15]. HPDC dies are usually manufactured from tool steel, hardened to increase longevity as they are required to withstand high mechanical and thermal loads associated with the process. Steel dies are fabricated predominantly by CNC machining. The steel of choice in Australia and the United States is H13, while in Europe H11 is more popular. These steels provide a good balance of toughness and strength required by the demanding conditions of high pressure die casting process.

However, the cost effectiveness of the process is dictated by the rate at which components can be produced and the working life of the die. A high production rate is always desirable since it has major impact on the profitability. Therefore, creating conditions for rapid heat extraction from the die surface, dissipation inside the material or rapid heat transfer to a ‘heat sink’ becomes attractive. Though, any material selected for HPDC application must withstand the harsh environment associated in the process without failing prematurely, high heat diffusivity die materials are also required to transfer heat rapidly from the casting.
This rapid heat transfer of the die material can contribute in the high production rate to increase profitability of the process.

2.2. Modes of Heat Transfer in HPDC

Heat transfer is that science which seeks to predict the energy transfer that may take place between material bodies as a result of a temperature difference. According to thermodynamics, this energy transfer is defined as heat [16]. As per basic rules of thermodynamics, heat can be transferred in three distinct modes namely-

1. Conduction heat transfer
2. Convection heat transfer and
3. Radiation heat transfer

In practice, heat is transferred in all three modes of heat transfer in a HPDC process. Figure 2-2 shows various modes of heat transfer and the locations where it takes place in a high pressure die casting die. Different heat transfer modes that seem to occur in the whole HPDC process randomly are: conduction within casting part and die material, convection between die material and coolant flowing through cooling channels and radiation throughout the process [17]. Conduction, the greatest factor, occurs when heat moves from a higher temperature to a lower temperature within casting and die component. In HPDC, heat is first transferred from the molten metal to the bulk die by conduction. In addition, the heat received from the molten metal is conducted to the coolant through the die material. Convection, another important mechanism for heat transfer, takes place when cold fluid passes through a hot die component. In HPDC the heat is carried away by the flowing coolant in cooling channels by convection heat transfer. Internal cooling within the die casting die occurs because the colder medium flowing through the cooling channels removes the heat conducted to the channel location by force. Since, the temperature difference is the driving force for heat transfer and in a
In a typical casting process, the temperature of the casting part is higher than its surroundings; the radiation heat energy is transferred from the casting to its surroundings. Radiation causes the die to lose some heat, but only from surfaces that are exposed to the ambient air. Usually, radiation is negligible compared to conduction and convection.

Figure 2-2: Various modes of heat transfer and the locations where it takes place in a high pressure die casting die (a) Die open for service and (b) Die closed after shot [18].

2.3. Factors Influencing HPDC Cycle Time

In order to increase productivity and profitability, the die casting industry is continually striving to improve the process. Cycle time is one of the areas that has received attention in recent years and technologies that can reduce it are of high priority for the die casting industry. Though determined by multiple variables, the cycle time is controlled to a large degree by the cooling or solidification time of the casting part. In general, the cooling or solidification time is depended upon the effective heat transfer from the casting and so is the profitability and cost.
effectiveness. HPDC dies are not only used to contain the melt to impose the final shape of the casting but also are required to extract heat effectively. Effective heat transfer requires removing heat quickly for faster solidification of the casting, thus making the production rate economically viable. Effectiveness is also determined by some other factors since removing heat too rapidly from casting may result in poor surface finish and other casting defects. In contrast, too slow extraction of heat may delay the solidification resulting in an unnecessarily long cycle time or bursting of the casting due to premature ejection. Therefore, heat transfer considerations are crucial in the design of the HPDC dies. Each component of the heat transfer resistance path should be considered appropriately. Research has shown that the transfer of heat from a solidifying casting is resisted by the casting itself, the thermal barrier at the casting-die interface and the die and the coolant boundary layer. Figure 2-3 depicts a one-dimensional representation of this resistance path.

Interfacial heat transfer resistance between the cavity surface and the casting part affects the solidification process. However, HPDC process is performed with numerous materials and under high casting pressure, as a result, the heat transfer resistance of the casting and the casting-die interface are relatively low. In addition, several studies have been carried out in order to investigate the interfacial heat transfer coefficient and its behaviors to minimize this resistance [19-21].
Figure 2-3: A one dimensional representation of heat flow resistance in a typical HPDC process [22].

From the magnitude of the different components of the total resistance for typical dimensions of a steel die (shown in Figure 2-3), it is evident that the main heat transfer resistance is from the coolant boundary layer and the steel die.

2.3.1 Effective Cooling System in HPDC

The cooling system of a HPDC is of great importance since it significantly affects both productivity and quality of the casting part. Considerable amount of the heat energy from the casting part is removed by the cooling channels. Therefore, design of the cooling channels is very influential and critical to the success of the HPDC process. Cooling channels must be strategically designed before the die materials are machined and hardened since it is not practical to change the cooling system after they are finished [23]. The heat transfer resistance of the coolant can be reduced effectively by careful shaping and positioning of the cooling channels. Matsumoto et al. [24] and Barone et al. [25] have shown how design sensitivity analysis can be applied to the thermal design of the moulds and dies. Park et al. [26] showed a method based on direct differentiation of the boundary integral equations to optimize the number, radius and position of cooling channels in an injection moulding die. Lin et al. [27] designed a cooling system for a casting die with free form surface and suggested optimum diameter and location of cooling channels for effective heat extraction. Boiling in cooling channels has recently been demonstrated as an effective rapid heat extraction mechanism for high pressure
die casting. Clark et al. [28] has shown that cooling rates can be significantly increased by radically redesigning the cooling channels so as to exploit surface boiling of the coolant. He showed that the cooling and solidification of a zinc or aluminium melt is often sufficient to cause not only sub cooled nucleate boiling but also transitional film boiling over portions of the cooling channel surface. The enhanced heat transfer due to sub cooled nucleate boiling of liquid coolant can greatly increase the efficiency of the cooling channels and thus reduce cycle times and correspondingly increase productivity. Davey et al. [29] showed that heat transfer can be increased by distributing the thin thermal boundary layer of almost quiescent water local to the surface of the cooling channels. Moreover, he has demonstrated that production rates can be significantly improved by optimizing the shape of the cooling channels in pressure die casting dies so that they fully exploit boiling heat transfer. Clark et al. [6] demonstrated that optimization of the cooling channel configuration can increase productivity of a die up to 60 percent whilst maintaining the surface quality of the castings. Hu et al. [30] has showed the effect of cooling water flow rates on local temperatures and heat transfer of casting dies in order to optimize the cooling water flow rate. Optimum cooling channel design through conformal cooling in both injection molding and die casting has also been subject to many researches [31-34]. The increased effect of conformal cooling helps to reduce cycle times, which in turn results in increased production rates.

Thus the information extracted from open literature clearly demonstrates that accurate design of shape and surface area, sufficient number and proper location of cooling channels and adequate flow of cooling water through the cooling channels allow greater heat dissipation from the tool to the water which in turn provides effective cooling performance. Therefore, if the cooling channels are well designed to reduce the heat transfer resistance of coolant boundary layer, the resistance due
to the die material becomes the most significant part of the heat transfer resistance as shown in Figure 2-4.

![Graphical representation of thermal resistance in HPDC process](image)

Figure 2-4: Graphical representation of thermal resistance in HPDC process [35]

### 2.3.2 Thermal Conductivity of Die Material in HPDC

The thermal conductivity is the quantity of heat transmitted, due to unit temperature gradient, in unit time under steady conditions in the direction of the temperature gradients. The thermal conductivity of a material is a measure of the ability of the material to conduct heat. A high value for thermal conductivity indicates that the material is a good heat conductor, and a low value indicates that the material is a poor heat conductor or insulator [36]. Though effective design of cooling channel improves the heat extraction rate, first, the heat needs to be transferred from the cavity surface to the cooling channels. Hence, thermal conductivity of the die material controls the heat transfer mechanism of HPDC process to a large degree. Nevertheless, it is of interest to note that almost all
HPDC processes involve placing the molten metal in a die of lower thermal conductivity. In addition to the thermal conductivity property, die materials for aluminum HPDC require being resistant to heat checking and to have good resistance to washout and soldering in a fast flow of molten aluminum. Therefore, in practice, H13 tool steel is the widely used material for aluminum HPDC dies. Thus, use of low thermal conductive H13 tool steel sacrifices effective heat transfer property which is one of the most influential factor in the production rate. In contrast, use of high thermal conductivity materials instead of tool steel would allow faster extraction of heat from the casting, with corresponding reduction in cycle time and increased productivity.

2.4. Development of Die materials

2.4.1. Copper Alloys as Die Material

Copper and nickel alloys and superalloys, titanium, molybdenum and tungsten and to some extent yttrium and niobium alloys, have all been considered as potential die materials for HPDC applications [37-40]. However, the use of these materials in HPDC applications is limited. The reasons for this are all too apparent as the materials of high thermal conductivity do not have the necessary mechanical and thermal properties to withstand the rigors of the HPDC process. Table 2-1 illustrates the critical properties including thermal conductivity of H13 tool steel together with selected copper alloys, suitable for die manufacturing applications [41]. Moldmax (HH, XL and LH) and PROtherm are copper-beryllium and copper-nickel tin alloys respectively. It is evident from Table 2-1 that the thermal conductivity of the copper alloys is much higher relative to steel.
Table 2-1: Properties of selected die materials [42]

<table>
<thead>
<tr>
<th>Material</th>
<th>Rockwell Hardness (HRC)</th>
<th>Thermal conductivity (W/m°C)</th>
<th>Thermal Expansion Coefficient ($10^{-6}$/°C)</th>
<th>Tensile strength (MPa)</th>
</tr>
</thead>
<tbody>
<tr>
<td>H13 Tool Steel</td>
<td>45-50</td>
<td>26</td>
<td>12.6</td>
<td>1720</td>
</tr>
<tr>
<td>Moldmax HH</td>
<td>40</td>
<td>104</td>
<td>17.21</td>
<td>1280</td>
</tr>
<tr>
<td>Moldmax XL</td>
<td>30</td>
<td>60.5-70</td>
<td>16.5</td>
<td>758</td>
</tr>
<tr>
<td>Moldmax LH</td>
<td>30</td>
<td>130</td>
<td>17.21</td>
<td>1170</td>
</tr>
<tr>
<td>PROtherm</td>
<td>20</td>
<td>250</td>
<td>17.64</td>
<td>793</td>
</tr>
</tbody>
</table>

Among various important properties of copper alloys, the ones that make them to be chosen over other metals are good resistance to corrosion, excellent electrical and thermal conductivity, as well as strength, resistance to fatigue and ability to take a good finish [9]. Particularly, copper-beryllium has superior high strength property given by the age hardening. Moldmax HH for instance, can reach a tensile strength as high as 1280 MPa. In Moldmax, spherical and uniformly dispersed beryllides (Cu,Co)Be are present in the matrix of equiaxed α-Cu as the hardening phase and tend to form bands of cellular precipitates at the grain boundaries at long aging times (over 8 h) or high aging temperatures (above 315°C) [43]. Copper alloys have made significant inroads in the manufacture of moulds for plastic injection molding. The shorter cycle times facilitated by copper mould allow increased production rates. A Brush-Wellman publication [30] highlights an injection moulding application at RT Technologies in Toronto, Canada. RT Technologies manufactures and sells 50,000 chair bases from Toronto plant on a weekly basis. By changing tool from steel to Moldmax, they decreased cycle time from 122 seconds to 98 seconds, which allowed them to increase annual production by 500,000 chair bases. They also reduced the number of presses from 20 to 16, providing enormous economic benefits. Beryllium-copper has also been considered as the material of choice for plunger tips in aluminum die casting [40].
Though, due to high thermal conductivity, copper alloys demonstrate excellent heat transfer property and thermal fatigue behavior, copper alloys cannot be used directly in HPDC of aluminum because of the strong chemical affinity between copper alloys and molten aluminum [44]. Soldering and washout are two prominent damages that restrict copper to be used as die material for aluminum HPDC.

The impinging jet of an incoming molten metal stream removes the die material and causes washout of the HPDC dies. In aluminum HPDC, washout is accompanied by the combination of corrosive and erosive wears. Corrosive wear is defined as the dissolution of die material in the molten aluminum. It takes place due to the solubility of various elements in liquid aluminum, oxidation of the die surface at high temperature and formation of intermetallic layers at die surface. On the other hand, erosive wear is defined as the gradual removal of material from the die with every liquid aluminum impingement as a result of the motion of aluminum melt. Liquid-impingement erosion to create pits on eroded surfaces, cavitation and solid erosion caused by the impact of solid particles during filling are the main erosion mechanisms [45]. Though the initial effects are not devastating, the continuous growth of washout results in the damage of the die surface and finally leads to the permanent failure of the die.

Soldering is defined as the adhesion of the cast metal to the die or core surface. During the filling and solidification stage of HPDC, chemical and mechanical reactions occur due to the affinity of the aluminum alloy atoms and die material atoms for each other. The chemical reactions often result in the formation of intermetallic layers at the die material and aluminum alloy interface. The consequent buildup of the aluminum alloy that occurs at the interface is called soldering. Sticking problems arise from soldering when the casting is ejected. In addition, it can give rise to adhesive wear when the casting is separated from the die [45]. Yulong et al. [44] reported that in their investigation on soldering,
washout, and thermal fatigue resistance of advanced metal materials for aluminum
die-casting dies, copper-based alloys got severe washout, soldering and were
broken due to the very high solubility of copper in molten aluminum. Therefore
they recommended not using copper alloys as die material. Figure 2-5 shows the
damages caused by washout and soldering on the pin that they used in their
experiment.

![Figure 2-5: Damages caused by washout and soldering on the copper pin used in HPDC](image)

2.4.2. Functionally Graded Material

Despite possessing excellent thermal conductivity, copper alloys exhibit many
draw backs to be used in HPDC application directly. However, efforts have been
made to develop new copper based die materials with high thermal conductivity
and resistant to soldering and washout. Studies carried out to date to develop
moulds with higher thermal conductivity copper alloys have concentrated on the
manufacture of functionally graded material (FGM) of H13/copper structures [46-
49]. FGM is the combination of two or more materials in the same component with
a variation of composition. This variation or gradient can be discrete or continuous
[50]. The application of FGM can improve the design of a part by local material
composition with variable properties, adding extra functionality to an integrated
part. FGMs have advantages in increasing the capability and the performance of a
part compared to other techniques. It can be achieved by combining distinctive materials to add new functions to the same component. In addition, even non-combining materials can be joined using this technique. However, the application of FGM to manufacture part is geometry limited because of the design complexity and implementation of fabrication technologies that can build FGM. Therefore, layered manufacturing technologies (LMT) such as, selective laser sintering, direct metal laser remelting, laser melting, selective laser fusion (SLF) etc. have been employed as an alternative for the fabrication of FGM components. These LMT approaches involve manufacturing 3D structures from a mixture of different powders. These methods are based on rapid prototyping where structures are built layer by layer from a CAD file using a laser beam to sinter or melt the mixture of the powders.

Selective laser sintering (SLS) is one of the leading commercial rapid prototyping (RP) processes. The process involves fabricating solid objects by selectively fusing powder of successive layers according to numerically defined cross-sectional geometry. Various thermoplastic, metal composite and ceramic composite powders can be used in SLS to produce prototype parts [51-54]. SLS is based on the laser sintering of fine powder layers spread over a platform. Parts are produced from a powder by selectively fusing powder particles layer by layer as opposed to traditional machining operations which involve material removal to define the final object geometry [55]. In the SLS method, a layer of powder is first deposited on a support platform and sintered through the energy delivered by a laser beam. Once sintering of the layer is finished, a new layer of powder is spread and sintered to add it with the previous layer. This process is repeated until the final product is built. Indirect SLS of metal parts uses a laser to create green parts from binder coated powder. A metallic powder coated with a thermoplastic binder is used to provide shape to the part through the melting of the binder. Binder
removal, pre-sintering, sintering etc are the subsequent steps associated with the process.

*Direct metal laser remelting (DMLR)* is a derivative of SLS. However, unlike SLS, DMLR uses a high power Nd:YAG laser to fully melt the metal powder bed. Nd:YAG lasers are advantageous because the shorter wavelength radiation (1.064mm), couples more efficiently with metals compared to CO2 laser radiation (10.64mm), reducing the power of the laser required to melt the metal powder [48]. A basic sketch of this manufacturing process is presented in Figure 2-6.

![Figure 2-6: A basic sketch of direct metal laser remelting (DMLR)](56)

*Laser melting (LM)* of metal powders is a LMT that enables freeform fabrication of complex three-dimensional metal parts utilizing numerical control to bond defined layer geometries in a powder bed by a moving laser beam [56]. The LM process is similar to the SLS process; only in the former the metal powder is completely melted by a continuous wave Nd:YAG laser to produce parts of near full density [57].

*Selective laser fusion (SLF)* is an example of LMT that works with fine powder layers spread over a platform and are then melted layer by layer by scanning the laser.
The powder particles are completely melted along with partial melting of the previous layer. This ensures better bonding between the layers. An advantage of this process is that virtually any kind of material can be processed and can produce high density parts [58]. A basic sketch of this manufacturing process is presented in Figure 2-7.

Figure 2-7: A basic sketch of selective laser fusion [46]

In order to achieve effective cooling with the combination of FGM and LMT techniques, research has been conducted on H13, WC and Cu powder laser fusion [59]. Pogson et al. [48] investigated the processing aspect of directly producing copper parts from powders using DMLR. In this experiment, concentration was directed specifically to the feasibility of the process for manufacturing copper parts, rather than focusing on the microstructural and materials issues raised, although these two were linked. They described the DMLR processing of copper powder to produce parts and compared the results to work on different materials and discussed how the laser parameters were optimized to produce parts with suitable mechanical properties. Numbers of coupons of Cu were produced on different substrates with various parametric combinations. Problems were however encountered in the samples since insufficient melting occurred in all
samples as shown in Figure 2-8 and Figure 2-9. These resulted in high level of porosity and low bonding of the powder particles. In addition, the substrate and its thermal properties had a significant effect on the process applicability, as, on copper substrate, the powder couldn’t be processed for any parameter set.

Figure 2-8: Images showing insufficiently molten and irregular shaped single layer coupons on stainless steel substrate produced with a laser power of 76.6W CW, 80 per cent beam overlap and laser scanning speeds varying from: (a) 10-50 mm/s in 10 mm/s increments; and (b) 25-125 mm/s in 20 mm/s increments [48]
Figure 2-9: Optical images showing porosity in various sections of 20 × 5mm builds; (a) longitudinal; (b) transverse sections produced with two scans per layer; (c) longitudinal; (d) transverse section of the build produced with three scans per layer; (e) longitudinal; and (f) transverse sections of the build produced with four scans per layer. The laser parameters for each were a laser scan speed of 100 mm/s, laser power of 76.6W CW and a beam overlap of 75 per cent [48]

With the ability of DMLR to produce highly complex parts having been developed, research has considered the processing of other materials and an extension of this is the ability to form components with functionally graded material properties or components produced from more than one material. The ability of DMLR to produce functionally graded parts influenced Pogson et al. [38] to investigate the
DMLR of mixed powders of copper and H13 tool steel to determine the problems associated with the production of tool steel components with either discrete copper regions or containing graded structures. The result, however, was not favorable to build molds from the DMLR processed mixture of tool steel and Cu powders since, the melting together of the powders lead to unfavorable microstructures that increased the risk of hot shortness and cracking. The detrimental effect of copper on the processing of H13 tool steel can be seen in Figure 2-10, where the component cracked on cooling, probably by hot tearing. During solidification of the laser scanned melt pool, the variant monotectic reaction between high-carbon iron-rich phase and low-carbon copper rich phase lead to the formation of austenite first (with a significant copper content), probably by the continued growth of the grains in the underlying layer rather than by the nucleation of new grains. The growth of austenite from underlying substrate, followed by the formation of copper-rich phase, showed the tendency to leave copper-rich region at the top of the track. When next layer was deposited, this copper-rich phase incorporated into the melt pool, to some extent, and again solidified at the top of the track. The accumulation of copper phase at the end of each track finally lead the top of the processed layer being copper-rich. Thus copper tended to sweep up the block in a similar way to zone refining to produce discrete regions of copper within the tool steel. Additionally, on further cooling, the amount of copper that was soluble in austenite reduced leading to its rejection and the formation of a copper-rich solid phase at the austenite grain boundaries (Figure 2-11). Thus, the incorporation of copper into the tool steel during processing formed a copper-rich region around the prior austenite grain boundaries, which most likely lead to hot tearing and cracking, as at higher temperatures they would be very weak compared to the surrounding H13 and significantly weakened the material. Therefore, a number of metallurgical challenges arose in the use of DMLR for fabrication of tool steel components with copper sections.
Figure 2-10: Optical micrograph of cracking in 25% Cu, 75% H13 coupon processed by DMLR [49]

Figure 2-11: Optical images showing copper-rich solid phase at the austenite grain boundaries of cross-sectioned samples (laser scan speed 400mms−1). Compositions: (a) 25% Cu, 75% H13; (b) 50% Cu, 50% H13 and (c) 75% Cu, 25% H13 [49]

Beal et al. [46] reported the use of FGM in which copper was dispersed to specific regions/volumes of a H13 tool steel mold to increase thermal efficiency. They used selective laser fusion technique to disperse different proportion of Cu powder in H13 tool steel and analyzed it. Using this technique, layers were added and fused
sequentially to form 3D objects with H13 tool steel and up to 50% Cu. However, cross-section of the graded regions between pure H13 tool steel and H13+50% Cu showed that cracks were visible in the microstructure of the materials. These cracks in the microstructure were one of the major defects that could affect the mechanical property. Figure 2-12 demonstrates various defects observed in their fabricated parts. High level of porosity was another defect that could deteriorate the mechanical property of the material. In addition, during the process, the contraction of material due to the solidification after localized laser fusion and the material lost due to spatter caused problems with dimensional accuracy (Figure 2-13). The amount of material lost due to the spatter effect was measured as between 17.6 and 19.8%. In order to produce an H13/Cu functional graded material part, it was essential to eliminate or reduce defects in the graded regions. Consequently, Beal et al. [60] focused on crack and porosity elimination using Response Surface Method (RSM) and factorial analysis tools. In their analysis they found that formation of cracks in the H13+Cu graded material occurred from solidification problem. The copper content increased thermal conductivity of the melt pool. The heat flowed through the parabolic melt pool solidifying finally at the centre of line of the bead and the restrained contraction produced cracks from the generated stress. Since, a FGM part is made with different materials and same laser processing parameters were not suitable to fuse it, they investigated the optimization of processing parameters using the RSM to eliminate crack formation. The RSM was proved to be useful for reducing the overall defects found in the processing of H13 and copper blends. The cracks however, couldn’t be eliminated completely and both transverse and longitudinal cracks were still present in the material. Moreover, the bonding between layers was poor and caused balling effects.
Figure 2-12: Optical images showing various defects in laser fused functionally graded H13/Cu materials (a) only polished, (b) and (c) etched [46]

Figure 2-13: Typical cross sections of specimens built with H13-Cu blends showing macro-porosity along the cross section of the specimens: (a) filling, (b) sequential, (c) alternated and (d) refill strategy. Also the cross sections show an area reduction compared to what should be the height of the specimens [46]

Anklekar et al. [61], reported the fabrication details of microwave sintered copper and nickel steel powder metallurgy (PM) and the in-house modified commercial microwave oven used for sintering. Microwaves can be defined as that part of the
electromagnetic radiation spectrum having a wavelength typically ranging from about 1 mm to 1 m in free space, and the frequency ranging from about 300 MHz to 300 GHz. However, only narrow frequency bands centred at around 915 MHz, 2.45 GHz, 28–30 GHz and 80–81 GHz are actually permitted for microwave research purposes. Microwave processing has gained a lot of significance in recent years for high temperature processing and synthesis of a diverse range of materials, mainly because of its intrinsic advantages such as rapid heating rates, shorter processing times, uniform temperatures with minimal thermal gradients, higher energy efficiency, improved mechanical properties, novel finer microstructures, reduced atmospheric interaction and lesser environmental hazards as compared with the conventional sintering process. Recent reviews on microwave processing by Clark et al. [62], Schiffman et al. [63], Katz et al. [64], Sutton [65, 66], and Agrawal et al. [67] described the potential uses of microwave technology for a wide range of diverse materials. Microwave heating of materials is fundamentally different from conventional heating, in that the heat is generated internally within the whole volume of the material, unlike from an external source in a conventional process, and subsequent heat transfer involving a thermal conductivity mechanism. Microwave heating is a very sensitive function of the material being processed and depends on a variety of factors such as size, geometry, mass and dielectric property of the sample. As a matter of fact, the sample itself becomes the source of heat during processing in a microwave field. In their investigation, Anklekar et al. [61] showed that microwave sintering resulted in higher sintered density and improved mechanical properties for both Cu and Ni steel PM parts as compared with that processed using conventional sintering under identical conditions. The improved mechanical properties were attributed primarily to more uniform distribution of the alloying elements, which resulted in greater material homogeneity at the nano- and micro-levels as revealed by the Cu and Fe X-ray maps. The microwave fabricated samples clearly showed development of novel sintered microstructures. The porosity however, couldn’t be
eliminated from the samples. The shape of the porosity became smoother and rounded pores with low stress concentration regions were produced against the sharp, triangular and wedge shaped pores with high stress concentration regions for conventional sintering (Figure 2-14). Therefore application of this process to fabricate graded material with high thermal conductive copper alloys for high temperature and pressure applications is constrained.

Figure 2-14: Optical micrographs of FC-0208 cylindrical tubular samples sintered at 1260°C for 20 min soak time in reducing atmosphere. Sintered samples in unetched condition were employed for observing typical porosity distributions obtained during (a) microwave sintering (b) conventional sintering [61]

Most of the layered manufacturing technologies described above use lasers to sinter the powder blend for producing parts. During melting of the powder, several complex issues arise from laser-material interaction, which involve various parameters and phenomena [58]. Also, in the case of laser melting, complex metallurgical interactions can occur if different powder types are used as the powders are fully melted and allowed to mix freely within the melt pool. Since, parts are produced additively in these layered manufacturing technologies, the energy provided by the laser beam must be sufficient to bond the powder particles. But laser processing is a short-period iteration process and the powder particles do not have enough time for the sintering mechanisms, which results in porous parts.
Post-processing in furnaces and infiltration with bronze/copper or even sealing the pores with epoxy based resin are some effective solutions to reduce the residual porosity [68]. But these post processing involves different consequent steps to make the whole process complex. Energy delivered by the laser can be increased to melt the particles completely and higher bonding can be achieved by the solidification of the melted pool. However, it will demand for complex scanning strategies. Even when the particles are completely melted, porosity due to trapped gas, poor laser penetration and the structure of fused beads will be present in the fabricated parts that will result in lower strength than a machined part to produce tooling for short-run production. Adjustments of the laser processing parameters can be useful to improve the densification. Nevertheless, porosity was not the only defect in the part and due to rapid solidification; cracks were also found in the final product. Moreover, the implementation of FGM in part fabrication is still geometry limited due to the complexity in design and only simple shapes can be fabricated. The layer by layer approach is time consuming and fabrication of large parts can take several hours. The high costs and long fabrication time are some of the other barriers that have limited a wider commercial use of these techniques.

2.4.3. Bi-metallic Die Material

Technological advances demand higher performances from materials. The limitations of functionally graded materials and their fabrication techniques influenced researchers to develop die materials without mixing of steel and copper powders. It has been demonstrated that high rates of heat extraction are possible with the use of copper-alloyed dies suitably protected with a tool steel layer. In this methodology copper and tool steel are not pre-mixed rather tool steel is coated on copper substrate to create a bimetallic structure. In the bimetallic structure, the purpose of copper is to transfer heat at faster rate and tool steel coating is used to provide necessary protection of the surface to withstand the rigors of the casting
process, in particular the abrasive action of the liquid melt. The basic idea is that the area which is subjected to aggressive die casting conditions can be protected by applying a protective H13 tool steel coating. Application of coating to provide necessary surface requirements of die casting die component is a common practice [69-72]. The range of surface requirements may vary considerably from sufficient protection against wear, corrosion resistance, thermal insulation, electrical insulation to even improved aesthetic appearance depending on its service environment. In practice, it is quite rare that die surfaces are only exposed to a single condition. Usually a combination of different conditions is present; for example, abrasive wear is combined with high thermal stress. Hence, wear and thermal fatigue resistance are the most frequent conditions the H13 tool steel coating must withstand. Since H13 tool steel coating will be subjected to several die casting environmental conditions, selection of coating process to apply best integrity coating is very important.

2.5. Various Techniques to Manufacture Bi-metallic Structure

There are quite a number of coating processes as well as a nearly unlimited number of coating materials. Nevertheless, the knowledge of specialists is usually required in order to select the correct combination for the respective application. Table 2-2 lists principal coating processes, the typical coating thicknesses attainable, common coating materials, and sample applications [73]. Some processes are not suitable for certain coating materials; also, the necessary coating thicknesses are not attainable with all methods. Beyond that, the equipment necessary for some processes can be quite complex and, therefore, costly. The use of cost analysis can determine whether a coating is a practical solution. Today’s regulations require that ecological criteria of the respective coating processes must also be examined, as not all methods are environmentally equal. Figure 2-15 demonstrates a comparative study on the applicability of various coating processes.
Table 2-2: Principal coating processes and characteristics [73]

<table>
<thead>
<tr>
<th>Coating Process</th>
<th>Typical Coating Thickness</th>
<th>Coating Material</th>
<th>Characteristics</th>
<th>Examples</th>
</tr>
</thead>
<tbody>
<tr>
<td>PVD</td>
<td>1-5 μm (40-200 μm)</td>
<td>Ti(C,N)</td>
<td>Wear resistance</td>
<td>Machine tools</td>
</tr>
<tr>
<td>CVD</td>
<td>1-50 μm (40-2000 μm)</td>
<td>SiC</td>
<td>Wear resistance</td>
<td>Fiber coatings</td>
</tr>
<tr>
<td>Baked Polymers</td>
<td>1-10 μm (40-400 μm)</td>
<td>Polymers</td>
<td>Corrosion resistance, aesthetics</td>
<td>Automobile</td>
</tr>
<tr>
<td>Thermal Spray</td>
<td>40-3000 μm (0.0015-0.12 in)</td>
<td>Ceramic and metallic alloys</td>
<td>Wear resistance, corrosion resistance</td>
<td>Bearings</td>
</tr>
<tr>
<td>Hard Chrome Plate</td>
<td>10-100 μm (40-4000 μm)</td>
<td>Chrome</td>
<td>Wear resistance</td>
<td>Rolls</td>
</tr>
<tr>
<td>Weld Overlay</td>
<td>0.5-5 mm (0.02-0.2 in)</td>
<td>Steel, Stellite</td>
<td>Wear resistance</td>
<td>Valves</td>
</tr>
<tr>
<td>Galvanize</td>
<td>1-5 μm (40-200 μm)</td>
<td>Zinc</td>
<td>Corrosion resistance</td>
<td>Steel sheet</td>
</tr>
<tr>
<td>Braze Overlay</td>
<td>10-100 μm (40-4000 μm)</td>
<td>Ni-Cr-B-Si alloys</td>
<td>Very hard, dense surface</td>
<td>Shafts</td>
</tr>
</tbody>
</table>

Figure 2-15: Comparison of coating processes [73]
2.5.1. Ion Implantation

Ion implantation involves the bombardment of a solid material with medium to high energy ionized atoms and offers the ability to alloy virtually any elemental species into the near surface region of any substrate [74]. This near surface alloying can be achieved without considering solubility and diffusivity. When these advantages are coupled with the additional possibility of low temperature processing, this technique can be explored into applications in which the limitations of dimensional changes and possible delamination of conventional coatings are a concern. The modified region is within the outermost micrometer of the substrate in almost all the cases. Applications of ion implantation in metals have expanded from the initial friction and wear resistance to areas such as corrosion, oxidation and fatigue [75, 76]. Polymers and ceramics have also been investigated with the principal aims of increasing the conductivity of polymers [77] and improving the fracture toughness and tribological properties of ceramics [78]. However, the limited thickness, typically less than a micrometer, is an indispensable limitation of the ion implantation process. It is a vacuum process that requires high background pressure. Hence, vacuum-compatible fixtures and thermal heat sinks are employed to ensure uniform dosage and adequate cooling to dissipate the imposed heat load on the component. Various limitations on the types of substrates arise from these requirements. Implanted atoms are always in an intimate mixture with atoms of the original substrate and form surface alloys. Ion implantation is a very costly process and requires intensive training compared to other surface treatment processes. In addition, limited commercial ion implantation facilities are available. As a result, application of ion implantation in coating high pressure die casting dies cannot be found in the open literature.

2.5.2. CVD and PVD Process

The Chemical Vapor Deposition (CVD) process can be defined as the deposition of a solid on a heated surface via chemical reaction from the vapor or gas phase [73].
The deposition species are atoms or molecules or a combination. Hence, CVD falls to the class of vapor-transport processes which are atomistic in nature. Numerous chemical reactions such as, thermal decomposition, reduction, hydrolysis, disproportionation, oxidation, carburization and nitridation are used in CVD. Thermodynamic, mass transport, kinetic considerations, chemistry of the reaction, processing parameters of temperature, pressure and chemical activity control a CVD process. CVD can be applied to deposit layers of nearly any metal, as well as nonmetallic elements [79]. Carbides, nitrides, oxides, intermetallics and many other compounds can also be deposited. This technology has become very important in coating of tools, bearings and other wear-, erosion-, and corrosion-resistant parts. CVD reactions require activation energy to proceed and depending on these energy sources, CVD can be of different types such as, thermal CVD, plasma CVD, laser CVD etc.

Like CVD, different Physical Vapor Deposition (PVD) techniques such as vacuum, evaporation, sputtering, ion implantation, ion-beam assist and arc are also vapor-transport process [80]. Nevertheless, unlike CVD processes, PVD processes do not rely on a chemical reaction in the gas phase to form the product that will be deposited and films can be deposited in a very pure form. For instance, vacuum deposition or vacuum evaporation is a physical vapor deposition process in which the atoms or molecules from a vaporization source reach the substrate without colliding with residual gas molecules [73]. Though CVD competes with PVD directly, recent trend shows significant merging of these two techniques since CVD now makes extensive use of plasma (a physical phenomenon), whereas PVD uses chemical environment (reactive evaporation and reactive sputtering). The distinction between these two basic processes becomes further blurred since both CVD and PVD are often processed in the same integrated equipment in a sequential fashion [81].
Both CVD and PVD have been extensively employed in modifying surfaces of the tools for various applications [82-91]. Since HPDC dies suffer from several surface failures, researchers have implemented these surface modification techniques to provide sufficient resistance to the various failure modes. Chellapilla et al. [71] reviewed a range of surface engineering treatments to improve the die surface with a view to establishing which of these were beneficial. They concluded that certain surface engineering treatments, e.g. physical vapor deposition (PVD), thermo-reactive diffusion (TRD), and chemical vapor deposition (CVD) had the potential to provide better surface quality; however, they did not offer any explanation as to why certain surface engineering treatments succeeded and others did not. Specifically, PVD surface coatings have been studied extensively in relation to evaluate the performance of these coatings on tools used for HPDC of aluminium alloys. Sundqvist et al. [72] used immersion test of core pins in molten aluminum to show that PVD coating of pins provides improved surface resistance in terms of reaction with molten aluminium. However, the immersion tests they used did not subject the coating to the thermal cycling and mechanical shearing that occurs during HPDC. Moreover, there are several difficulties associated with both CVD and PVD coating techniques. Using PVD, deposition of many alloys and compounds are not possible. High radiant heat loads are required during processing and the capital equipment costs are high relative to those of other deposition techniques. Vaporized materials are not used efficiently; hence, system loss is high. There are few processing parameters available to control the coating properties. The maximum achievable coating thickness is only 0.1 mm [73]. The chemical precursors in CVD are hazardous or toxic and many reactions either leave solid byproducts or generate solid byproducts with neutralizing solutions which are toxic and corrosive as well. In addition, a research group in the Ohio State University showed that single layer hard PVD and CVD coatings did not protect the die surface from cracking rather enhanced cracking [70]. Therefore,
PVD and CVD are not suitable for manufacturing bi-metallic die for HPDC application.

The same research group investigated a multi-layer duplex coating approach with a multi-layer and hard outer film applied by large area filtered cathodic arc deposition (LAFAD) technique that prevented reaction with the liquid melt, and altered the thermal fatigue behavior. In the LAFAD technique, arc sources can be used to extract highly energetic electrons and to ionize the gaseous plasma, such that a ‘plasma envelope’ that completely surrounds the part can be created in the coating chamber [92-95]. Very high ion currents can be obtained as compared to the other PVD techniques such as EBPVD and sputtering. The high degree of ionization of the gaseous plasma permits ion saturation levels suitable for ion nitriding. Moreover, when the substrate is strongly biased, significant ion implantation can be achieved. Application of LAFAD indicated that the multi-layer duplex coating system helped to reduce the density of the thermal cracks in the thermal cycling tests, though it could not be eliminated completely. Since the thermal cracks were not removed completely, application of LAFAD was not suitable to fabricate bi-metallic structure for HPDC.

2.5.3. Welding Process

The welding of various metals and alloys is a mature subject in metal joining [96, 97]. Many of the structural components in pressure vessels, transport vehicles, space crafts, earthmoving equipment etc. are made of welded joints [98]. Various mechanical performances of the welded joints have been investigated to analyze the feasibility of fabrication and construction of numerous structures [99-101]. Research has also been concentrated on analyzing the behavior of the welded joints under different loading conditions [102]. Although the emerging new technologies increasingly require dissimilar metals and alloys to be joined together, welding of dissimilar materials represents a major scientific and technical challenge. The reason is that, welding of dissimilar metals involves electromagnetic fields which
has the possibility of beam deflection [103-105] due to several electromagnetic fields and leads to missed joints in dissimilar couples like Ni-Cu. Nevertheless, welding and microstructural analysis of various laser welded similar and dissimilar metals and alloys have been reported in the open literature [106-108]. Since laser processing is free of electromagnetic fields, it is considered suitable for welding dissimilar couples. In particular, Phanikumar et al. [109] investigated the joining of Fe and Cu in solid form using CO2 laser. But the aim of the study was just to characterize the microstructure evolution of the weldments of a binary couple of pure metals and to delineate the unique features that needed to be understood. The application of the structure in the high pressure and temperature application was not evaluated. Fusion welding of those two classes of alloys yielded unsatisfactory joints. Though, unlike other surface coating techniques, welding provides metallurgical bonding between metals, this technique is not popular to provide surface protections. Kumar et al. [110] investigated the fatigue characteristics of gas tungsten arc welded bead-on-plate of commercially available aluminum and AA7020 grade aluminum alloys and found that due to bead-on-plate, the fatigue life reduced drastically in both the alloys. Analysis of the failure of weldments indicated that fatigue alone is accountable for most of the disruptive failures [111-115]. In general, vast fluctuation of fatigue strengths in welded joints of different metals is a result of variations in stress concentration due to designs of different welded joints and loading conditions. A welded joint is a stress raiser and different stress raisers in welded joints result in local stress concentration. Fracture is almost certain when these locally stressed joints are subjected to high pressure and temperature applications. Miki et al. [116] reported that even though the fatigue properties of the weld metal is good, problems may arise from an abrupt change in section caused by excess weld reinforcement, undercut, slag inclusion and lack of penetration and nearly 70% of the fatigue cracking occurs in the welded joints. Malarvizhi et al. [98] reported that the tensile and fatigue strength of as welded joints are much lower compared to base metal. The significantly large
gap between the strength of the base metal and welded joints forces the design
engineers to use thicker base metal plates which in turn increase the total weight of
the structure. In addition, welding is further restricted on the surfaces of complex
shape geometries. Hence, welding is not suitable to create bi-metallic tooling
structure.

2.5.4. Thermal Spraying Process

*Thermal spraying* is a group of processes in which finely divided metallic and
nonmetallic materials are deposited in a molten or semimolten state on a prepared
substrate [117, 118]. The heat source is a combustion flame, a plasma jet, or an arc
struck between two consumable wires. In this process, the substrate can be kept at
relatively low temperature by specific cooling devices. For instance, in the electric
arc wire spray, an arc is formed by contact of two oppositely charged metallic
wires, usually of the same composition. This leads to melting at the tip of the wire
material. Air atomizes the melted spray material and accelerates it onto the
substrate. The rate of spray is adjusted by appropriate regulation of the wire feed
as it is melted, so a constant arc can be maintained [73]. Figure 2-16 is a schematic
diagram showing the principle of electric arc wire spray. There are several
different processes used to apply a thermal sprayed coating. They are:

- Conventional flame spray
- Electric arc wire spray
- Plasma spray and
- High velocity oxy-fuel spray (HVOF)

A close observation of Table 2-2 and Figure 2-15 reveals that the coating process
having the greatest range of coating materials, coating thicknesses and possible
coating characteristics is thermal spray. Successful implementation of coating to
improve surface quality influenced researchers to investigate the feasibility of H13
tool steel and copper alloy bimetallic structure. A thermal spray process such as
thermal arc spray, followed by a hot isostatic pressing (HIP) [119] and heat treatment could be employed to coat tool steel on copper alloy substrate.

A group of researchers at the University of Manchester developed bimetallic dies for pressure die casting where they used arc spray process for coating tool steel on copper alloy substrate [35, 120-124]. Rasgado et al. [35, 122] developed a numerical model based on the boundary and finite element methods to investigate the thermal balance of the copper based bi-metallic die tool and the solidification pattern of the component. Their model was capable to predict both steady-state and transient temperatures in the alloy dies and casting which can be used to optimize the cycle time, while avoiding surface and structural casting defects. Davey et al. [125] developed boundary element thermal modelling for pressure die casting which was extended by Bounds et al. [125, 126] to incorporate the solidification of the casting using a FE based method. In their investigation, Rasgado et al. [35] extended the boundary and finite element models to incorporate the copper-steel layer in the thermal modelling of copper based bi-metallic dies. They considered the copper alloy block and the steel layer as two separate domains linked by suitable boundary conditions to model the copper-
steel bi-metallic structure. In practice, to reflect the physical situation, a high heat transfer coefficient is applied on the interfaces between the steel layer and the copper-alloy block. This model predicted that the average rate of energy entering the cavity surface from the casting was 1910 W [122]. Their theoretical model predicted that 80 cpm run was possible using the test die compared to 40 cpm for a conventional steel die. In the experimental validation of this model, it appeared that the surface temperature of the casting did not decay as rapidly as predicted after injection. They reported it as a higher thermal resistance at the copper/steel interface than used in the computer simulation. However, there was a good agreement between the predicted and measured temperature at the back of the tool steel coating. Moreover, the measured temperature in the copper die block, which was at 2.5 mm back from the casting surface, also showed good agreement with the predicted temperature. Though this agreement demonstrated the high heat extraction capability of copper based bi-metallic dies for rapid heat extraction and solidification of the casting part, the main challenge arose from the manufacturing of this tool.

Clark et al. [121] also investigated the feasibility of the bi-metallic die experimentally. Initially they tested a pseudo die block in a casting simulation rig that simulated pressure die casting process conditions. Figure 2-17 and Figure 2-18 show the test rig and die blocks used in their experiment. The test rig was capable of producing the heat transfer mechanisms involved in pressure die casting but avoiding the hazardous features of the actual process at the same time. The steel coated copper pseudo die block showed faster extraction of heat compared to steel pseudo die block in the test rig.
Figure 2-17: Casting simulation rig used to test thermal behavior of copper alloyed dies [121]
Figure 2-18: Pseudo die block used to test thermal behavior of copper alloyed dies [121]

However, significant trouble came up when a copper alloy test die was designed and manufactured comprising two die blocks and a stepped cavity which included core pin holes, a gate and a runner and was tested under real die casting conditions. In this experiment copper dies were sprayed with a layer of steel and tested in hot chamber die casting machine under industrial conditions. The steel coating was applied using thermal arc spray system. 95MXC was chosen as the spray coating material. Chromium copper C18200 and Ampcolloy 940 were used
as copper alloy block. To provide structural support and provision for ejection pins and cores, they incorporated steel inserts in the die. Presence of steel inserts resulted in surface cracks due to differential cooling between the two materials which required localized repair. After thermal spraying, differential cracks were observed in the coating as shown in Figure 2.19. Figure 2.20 shows the porous surface of the steel coating they used in this experiment. This porosity in the coating was generated from the thermal spray process and was detrimental for the coating strength. Figure 2.21 shows the principle of thermal spray process. In this process the coating material is melted in a heat source. This liquid or molten material is then propelled by process gases from a spray gun and sprayed onto a base material as splats, where it solidifies and forms a solid layer. This solidification leaves porous areas at the edges of the splats and results in high level of porosity in the coating. Figure 2.22 shows a schematic diagram of a thermal sprayed coating with various features present in the coating. This is why porous holes were detected in the tool steel sprayed coating on copper alloy block in the experiment carried out by Davey et al. [121]. The high level of porosity in the sprayed coating at the die-cavity surface resulted in poor surface finish on both the casting and die (Figure 2.23 and Figure 2.24). The bond strength between the coating and the copper die block was also very low. Among all the combinations tested, a chrome-copper die, sprayed using a single bonding agent of 10T produced the highest bond strength of 51 MPa. This low bond strength was not sufficient to withstand the rigors of the high pressure die casting conditions. Therefore cavity surface and runner gate were damaged severely. Particularly, significant damage of the runner gate required special inserts to be placed in the die in the runner region to withstand nozzle impact as shown in Figure 2.25. Application of the runner insert to ensure sufficient strength to withstand the forces made the die relatively more complex and expensive to manufacture. In addition, the bond strength of the spray coating was found inversely proportional to the thickness of the coating which limited the coating layer thickness to a
maximum 1 mm. Limitation in the deposited layer thickness was another prominent restriction to apply thermal sprayed coating on the die surfaces.

Figure 2-19: Copper die block prior and after thermal spraying [121]

Figure 2-20: Optical image showing porosity in the sprayed coatings [121]
Figure 2-21: Principle of thermal spraying [73]

Figure 2-22: Schematic diagram of a thermal sprayed coating with various features present in the coating [73]
Figure 2-23: Cavity defects after casting trials [121]

Figure 2-24: Casting defects due to porosity in the thermal sprayed coating [121]

Figure 2-25: Steel runner inserted due to the damage in the sprayed coating [121]
Though the bi-metallic die of Clark et al. [121] had severe limitations, the research demonstrated the feasibility of using copper as die material and left a strong impression for the future researchers. Their research demonstrated that higher rates of heat extraction is possible using copper based bi-metallic dies to shorten casting times. However, the defective wear resistant coating applied by thermal spray needed to be replaced by some other process to produce sound and solid tool steel layers for greater protection. In addition, application of steel base and columns to reinforce the low strength copper alloys indicated the need to employ high strength copper alloy base materials since use of these supportive base and columns made the die relatively more complex and expensive. Therefore, if highly protective, fully dense and metallurgically bonded tool steel layer can be applied on high strength copper alloy base material, it will be capable of withstanding the severity of high pressure die casting. Laser aided Solid Freeform Fabrication (SFF) can be an effective way to serve in this context since it is capable of producing fully dense, sound and metallurgically bonded metallic layers on various base metals.

2.6. Solid Freeform Fabrication

One of the most recognized problems in the die casting industries is the time taken to get technology and hence products to the market. The design and fabrication of moulds and dies of all types is the limiting time for many products in manufacturing. Large complicated dies may require weeks to months or even to almost a year before they are ready to manufacture product. Therefore, a process that can make a product directly from a CAD model with a desired macro and microstructure is demanded by most manufacturers. Repair procedures that do not change the material properties and do not result in reduced life for the product are also needed for the manufacturers. Hence, rapid realization of CAD models into 3D products has become a reality thanks to the development of solid freeform fabrication (SFF) or additive manufacturing techniques. These techniques can fabricate 3D items by progressively adding and consolidating controlled amounts
of feedstock materials at precise locations, layer by layer, without any supporting preform or mask. There are several SFF techniques that are capable of producing full strength and fully dense items made from structural engineering materials. Main commercially available SFF techniques that can make fully dense metal parts are: Selective Laser Sintering (SLS), Selective Laser Melting (SLM), Electron Beam Melting (EBM) [127, 128], Laser Engineered Lens Shape (LELS) and Direct Metal Deposition (DMD). However, laser-based Fused Metal Deposition takes the lead in the quality of the items produced, the range of materials that can be deposited and the number of process variants that have been developed [129].

2.6.1. Laser Aided Fused Metal Deposition

Various efforts of using pre-placed powder beds to develop graded materials have been discussed in earlier sections. Unlike laser consolidation of pre-placed powders, Weerasinghe et al. [130, 131] introduced another class of laser-based SFF techniques using fused metal deposition based on laser cladding. In this process, the feedstock powders are delivered to a laser beam generated melt pool on substrate material by a carrier gas instead of pre-placing the powder on a platform [132-146]. The molten material rapidly solidifies and forms a continuous track of dense material as the laser-material’s interaction zone moves along the part’s surface. A continuous layer of material is produced by the partial overlapping of individual tracks in a suitable pattern. When these layers are overlapped, fully dense 3D objects are generated. Since, the feedstock material fuses completely, metal powders directly transform into fully dense solid objects with metallurgically bonded tracks of material that require no subsequent infiltration step and exhibit mechanical properties equivalent to those of similar alloys in the wrought condition. A schematic of fused metal deposition based on laser cladding is depicted in Figure 2-26.
In recent years, several independent process development teams have explored the laser based fused metal deposition technique as [147]:

- Laser direct casting [148] developed at the University of Liverpool;
- Direct metal deposition (DMD) [135] developed at CLAIM, University of Michigan, and commercialized by Precision Optical Manufacturing (POM);
- Directed light fabrication (DLF) [136, 149] developed at Los Alamos National Laboratories;
- Laser forming or Lasform process [138] developed by the MTS Systems Corporation and commercialized until recently by AeroMet Corporation;
- Laser deposition of metals for shape deposition manufacturing [137] developed at Stanford University;
- Laser-engineered net shaping (LENS) [150] developed at Sandia National Laboratories and commercialized by Optomec Design Co.;
- Laser powder fusion [148] commercialized by Huffman Corporation (www.huffmancorp.com);
• Freeform laser consolidation [141] developed at the Integrated Manufacturing Technologies Institute of the National Research Council of Canada;

The research carried out in the last few years demonstrated the capabilities of laser based fused metal deposition, ranging from manufacturing of thin-walled structures to highly complex, bulk objects in a wide variety of metallic materials [135-138, 141, 151-159]. However, the density of the deposition paths and the thickness of each corresponding layer have to match the specific deposition capabilities of the system for dimensional accuracy and fully dense parts. To build a solid object, the corresponding CAD model is generally sliced into a series of layers followed by sequencing of deposition paths, describing the deposition scheme for each layer [160-164]. Nevertheless, a carefully programmed deposition scheme does not ensure that part build up will run smoothly nor assure the intended dimensional accuracy, since laser metal deposition process shows the tendency to deviate from prescribed deposition schemes in response to several non-linear effects, the most common of which are those resulting from the accumulation of heat during the deposition process.

For instance, Koch et al. [132] reported progressively deeper and wider melt pools and lower layer thicknesses due to the accumulation of heat during part build which forced subsequent production runs to be performed with progressive adjustment of the processing conditions along the part height so as to achieve a more consistent geometry. Wei et al. [165] and Hofmeister et al. [166] also observed similar effects of heat accumulation on the size of the melt pool during the deposition of thin wall structures during part build up. In addition, since conduction of heat to the cooler substrate is usually the dominant heat transfer mechanism by which heat is extracted from the melt pool, cooling of newly deposited material will depend strongly not only on the temperature of the underlying material but also on its geometry [166-168]. As the wall height
increases, the reduction in the thermal gradient between the melt pool and the substrate which is a consequence of the accumulation of heat during part build up and the increase of distance between the hot melt pool and the cool substrate causes an effective reduction of the heat extraction rate. Moreover, Vasinonta et al. [169] developed a 2D heat transfer model of the laser metal deposition (LMD) process for thin wall parts to investigate the influence of wall height, absorbed laser power, laser velocity and pre-heating (or heat accumulation) on the melt pool length. The outcome of this research was a process map that clearly revealed the effects of heat accumulation and wall height on the melt pool length. Thus, changes to the shape and size of the melt pool due to the changes in the heat extraction efficiency of the underlying material affect both the powder catchment efficiency and effective laser radiation absorption coefficient thereby, changing the overall material deposition rate and, most importantly, the extent of remelting of previous layers. Again, for thin walls, Gremaud et al. [139] noted that a progressive deviation of the actual build up from the pre-determined deposition scheme may significantly deviate the stand-off distance between the laser-powder deposition (LPD) head and the melt pool from its nominal value, which will alter the dimensions of the melt pool as predicted by Pinkerton et al. [170] using a 3D heat transfer model, and change the powder catchment efficiency as observed by Lin et al. [171]. Moreover, Gremaud et al. [139] reported that due to uncontrolled melt pool shape instabilities, which are sometimes periodic in nature, catastrophic breakdown of the deposition process can eventually take place.

To resolve these problems caused by the process non-linearities, automatic real-time adjustment of the processing parameter that can ensure a melt pool and a material deposition rate consistent with the pre-determined deposition scheme, is needed. Thus, this automatic real-time adjustment will ultimately maintain the stability of the process and dimensional accuracy of the parts. To date, several closed-loop feedback control systems have been successfully demonstrated [167,
These control systems are empowered by video or thermal imaging techniques, capable of monitoring the melt pool in real time. Krantz and Nasla [172] have presented a review of various independent R&D efforts on LMD closed-loop process control, including a discussion on the advantages and disadvantages of various sensing techniques capable of monitoring melt pool characteristics. Mazumder et al. [135, 174, 180, 181] reported an example of implemented closed-loop feedback control system in direct metal deposition (DMD) technique, which effectively expanded the SFF capabilities of a LMD system. Use of this closed loop feedback system in DMD differentiates it from other LMD processes.

2.6.2. Direct Metal Deposition

Originated at the University of Michigan and further developed and commercialized by Precision Optical Manufacturing (POM) Group Inc., Michigan, Direct Metal Deposition (DMD) is a proprietary Laser Aided Manufacturing (LAM) process. Solid freeform fabrication (SFF) of various metals is now feasible by using DMD and rapid prototyping process has now reached the stage of rapid manufacturing via DMD technique. Closed loop optical feedback system of DMD is capable to produce realistic components with very close dimensional accuracy that requires little or no machining depending on application requirements. It will provide substantial savings in post process machining cost for surface finish. Substantial cost reduction is possible and post process heat treatment is reduced since desired properties can be achieved through process control. Such capabilities of controlling macro and microstructure have created considerable interest in manufacturing industries. The initial work on direct metal deposition of Aluminium has been demonstrated and has shown properties equivalent to a wrought process [132]. AISI H13 tool steel trimming and injection molding dies with embedded copper chill blocks and conformal cooling channels have also been successfully manufactured using DMD [135]. Advances in DMD are continuing.
and offering the promise of manufacturing and design improvements that can make design and fabrication of metallic parts speedier and more cost efficient.

Figure 2-27: Schematic diagram of a closed loop DMD process

A schematic diagram of closed loop DMD is shown in Figure 2-27. In DMD technique, laser beam generates a melt pool on a substrate material while a second material in powder form from a feeder is injected into the melt pool with a concentric nozzle where laser beam and powder come through the same nozzle. The delivered powder melts and when solidified, forms metallurgical bond with the substrate material. Integration of lasers with multi-axis CNC machines, computer-aided design (CAD)/computer aided manufacturing (CAM), sensors and powder metal delivery through co-axial nozzles are the main innovations in DMD to fabricate components [132, 182-184]. CNC controls the movement of the laser beam and traces out the part layer by layer based on CAD geometry.
Computer controlled five-axis processing head allows deposition of complex geometries. Some of the important features of DMD are outlined below [185]-

- Patented closed loop optical feedback system that monitors and controls the melt pool height to obtain a near net shape part is used in DMD. It provides better control of microstructure and properties of the finished product and reduces post finishing operations significantly [132, 135, 186]. Figure 2-28 shows a schematic diagram of the DMD feedback system. The system uses high speed sensors with optics looking at the melt pool to collect the melt pool image. Figure 2-29 shows a layout of the multiple sensors used in DMD feedback system. This image is sent directly to a dedicated DMD controller which extracts the melt pool dimensional data from this signal. DMD controller software algorithm processes this dimensional data and creates a new control laser power within a very fast response time, which is sent to the laser generator. Thus, the deposition layer thickness is directly related to the melt pool dimension and the final product is near net shape. The usage of melt pool dimension allows distinct advantage of directly controlling part geometry compared to some other competitive processes, which use temperature measurement to control deposition. Moreover, feedback system controls heat input to the melt pool by controlling melt pool dimensions, which leads to minimal HAZ and better microstructure with superior properties. Figure 2-30 shows two cylindrical samples produced by DMD laser cladding with deposition path of overlapping circles. The sample at the right was produced without any height control and excess cladding build-up was detected due to the circles overlap. Since the laser cladding passed this point twice per layer, laying down the same amount of material during each pass, overall height at this location was approximately twice of the cylinder height. Overall height was also uneven in other areas besides the overlap. The sample on the left in Figure 2-30 was fabricated
using the height controller, which had the ability to sense that the cladding layer was building up higher in the overlapping region. Therefore it shut off the laser until it passed the area of excess build-up and produced near net shape part with close dimensional accuracy.

Figure 2-28: Schematic diagram of a DMD feedback system [185]

Figure 2-29: Multiple sensors layout in DMD feedback system (top view)[185]
The co-axial design of the nozzle to integrate laser, metal powder and shielding gas delivery provides full five-axis deposition capability compared to independent side powder delivery systems which can only deposit linearly. Shielding strategy provides delicate balance of powder delivery without causing excessive disturbance at the melt pool. In addition, it also supplies adequate pressure to drive away the ambient air which in turn prevents the deposit from oxidation. Instead of using expensive and time consuming inert gas chambers, DMD uses Argon as shielding gas to protect melt pool.

DMD is equipped with multiple powder feeders. Deposition of different materials simultaneously or consecutively at desired locations are possible through these feeders to create new alloys or functionally graded materials.

DMD-CAM is used as tool path software in DMD to translate CAD model into the nozzle motion for five-axis deposition. It is a comprehensive additive manufacturing software that integrates direct metal deposition technology database to allow process recipe as part of the software. It is capable of creating multilayer deposition paths in a single operation with contour, surface and volume deposition in 3D. Once a deposition path has been created, simulation and collision detection modules with ‘ready to use’
deposition path can be run to detect any possible collision of the processing head. It eventually enables the user to find alternate tool path strategy.

- Since DMD uses ‘moving optics’ with the part stationary, it can process large size heavy parts. Risk of powder accumulation in the rails and drives of a moving table is also eliminated by moving optics.
- DMD deposited materials are fully dense and precise process control yields a better microstructure in the product. Therefore, properties of DMD deposited materials have been found comparable or better than cast material properties [180]. A broad range of materials including tool steels, stainless steels, high speed steels, Ni alloys, Co alloys and Ti alloys can be processed using DMD.
- Various multifunctional and user friendly softwares are used in DMD to control deposition process. Recipe database system used in DMD stores all process parameters, which can be accessed automatically through machine commands.

2.7. Applications of Direct Metal Deposition

DMD is a value adding technology, which can be applied in a broad range of industrial applications covering automotive, aerospace, medical, injection molding, constructions and die casting tooling. Most of the other metal based processes result in sintered products due to oxides trapping and inability to bond totally. These need to try to fill the metal matrix and do not demonstrate strong properties provided by DMD. Therefore, it demands post processing for total bonding and no oxides or sintered properties. For instance, a weld type structure which is built up by repeated passes requires significant heat treatment to make the part useful and relieve stresses. DMD eliminates these problems as it creates a very small heat affected zone. DMD can be used on almost any surface and can provide required clad geometry through the given CAD model. Moreover, DMD provides wider range of deposition capabilities. Since DMD process is of additive nature and
creates less amount of waste compared to subtractive processes like machining, grinding etc., DMD has minimal impact on environment.

Forging/stamping industry, oil industry, construction/mining industry, automotive industries have used DMD to hard facing tools to provide improved resistance against wear [185]. Cermet material containing 60% WC in a Ni-alloy matrix has been coated on a down hole drilling tool for oil and gas industry [185]. It extends tool life which subsequently adds significant cost savings. Reduction in down time due to less frequent tool changing exerts a tremendous impact on the economy of these processes in addition to direct cost savings in terms of tool cost and inventory reduction.

When higher cost of some materials such as Ni, Co, Ti etc. prohibits their usage in many high temperature and wear/corrosion resistant applications, DMD offers excellent solution to this problem. With close loop feedback system and five-axis deposition capability, DMD has been employed to coat expensive metals onto low cost substrates of different geometries [185]. It enables a large savings on materials while still providing performance functionality of the produced part. In addition to coating parts, re-manufacturing of worn or damaged components are also possible using DMD.

2.8. Summary

From the above literature review, it emerges that very few researches have been carried out to fabricate bi-metallic dies. Moreover, the effort made by other researchers was not successful since the protective tool steel coating produced by using thermal spray process was not sufficiently strong due to high level of porosity. Also, the bonding between the coating and the substrate was not metallurgical and resulted in low bonding strength. Therefore, integration of design and manufacturing of closed-loop DMD has enormous potential to manufacture bi-metallic structure. Particularly, the ability of DMD in coating parts
directly from CAD model along with the ability of bonding two dissimilar materials with high precision directs it towards the fabrication of bi-metallic tooling for high pressure die casting application. However, to date, no efforts have been made to utilize DMD technique for the fabrication of high thermal conductive bi-metallic tooling for HPDC. Hence, this thesis has explored deposition capability of fully dense and metallurgically sound H13 tool steel clad on high strength and high thermal conductive copper alloy substrate using DMD to form bi-metallic structure for HPDC application. Since DMD progressively deposits relatively small amount of laser melted and resolidified H13 tool steel at the desired surfaces, it facilitates rapid cooling conditions for the clad. This rapid cooling of the tool steel clad can thereby favour the development of fine microstructure throughout the clad to convey excellent mechanical properties of the clad for application in harsh environment associated with HPDC. However, no detailed investigation on microstructure and mechanical properties of bi-metallic structure of DMD deposited tool steel on copper alloy substrate has been published in open literature. Therefore, a comprehensive analysis of microstructure and mechanical properties has also been described in this research. Finally, performance of DMD deposited tool steel clad on copper alloy core material has been evaluated in semi-industrial HPDC condition.
3.1. Introduction

In this chapter, a finite element analysis has been conducted that simulates the thermal behavior of bi-metallic dies in the high pressure die casting process. A comparison of transient thermal response of the bi-metallic die with an identical H13 tool steel die under identical die casting conditions has been investigated. Transient thermal response of the proposed bi-metallic die demonstrated faster extraction of heat compared with traditional tool steel die under high pressure die casting conditions.

3.2. Heat Flow and Governing Equations

It has long been recognized that the thermal state of the die reflects the die casting conditions. After injecting the molten metal into the die cavity, heat is extracted from the casting to the die and carried outside through various processes. The majority of the heat is transferred from the casting to the die block, then through the die block to the cooling channels and finally through the cooling medium to the outside. The foremost characterizing parameters for heat flow conditions in given die casting process are the thermal gradients on various surfaces, thermal conductivity and the flow of cooling medium of the die.

Once the heat is applied on die cavity surfaces, firstly it is transferred by conduction of die material due to the temperature gradients between hot cavity surface and cooler cooling channels. If we consider it as one dimensional heat flow, the heat flow equation becomes as follows according to Fourier’s law of heat conduction [36]:
\[
\dot{Q}_{\text{cond}} = -kA \frac{dT}{dx} \quad \text{................................................} (3.1)
\]

where,

\[\dot{Q}_{\text{cond}} = \text{Rate of heat conduction}\]

\[k = \text{Thermal conductivity of the material}\]

\[A = \text{Cross sectional area of the heat conduction}\]

\[dT = \text{Temperature difference between the surfaces}\]

\[dx = \text{Distance between the surfaces}\]

\[\frac{dT}{dx} = \text{Temperature gradients between the surfaces}\]

The constant of proportionality \(k\) is the thermal conductivity of the material, which is a measure of the ability of the material to conduct heat. The equation (3.1) indicates that the rate of heat conduction in a given direction is proportional to the temperature gradient in that direction. Heat is conducted in the direction of decreasing temperature, and the temperature gradient becomes negative when temperature decreases with increasing distance. The negative sign in equation (3.1) ensures that heat is transferred towards the low temperature surface. This is a one dimensional heat flow equation. The governing differential equation has been developed for a rectangular block considering multidimensional heat flow [36].
Let us consider a small rectangular element of length $\Delta x$, width $\Delta y$ and height $\Delta z$ as shown in Figure 3-1. If the density of the body is $\rho$ and the specific heat is $c$, the energy balance on this element during a small time interval $\Delta t$ can be expressed as

\[
\text{(Rate of heat conduction at } x, y \text{ and } z) - \text{(Rate of heat conduction at } x + \Delta x, y + \Delta y \text{ and } z + \Delta z) + \text{(Rate of heat generation inside the element)} \nonumber
\]

\[
= \text{(Rate of change of the energy content of the element) } \ldots \ldots \ldots \ldots \ldots \ldots \ldots (3.2)
\]

or

\[
\dot{Q}_x + \dot{Q}_y + \dot{Q}_z - \dot{Q}_{x+\Delta x} - \dot{Q}_{y+\Delta y} - \dot{Q}_{z+\Delta z} + \dot{E}_{\text{gen,element}} = \frac{\Delta E_{\text{element}}}{\Delta t} \ldots \ldots \ldots (3.3)
\]
Provided that the volume of the element is \( V_{element} = \Delta x \Delta y \Delta z \), the change in the energy content of the element and the rate of heat generation within the element become-

\[
\Delta E_{element} = E_{t+\Delta t} - E_t = mc(T_{t+\Delta t} - T_t) = \rho c \Delta x \Delta y \Delta z (T_{t+\Delta t} - T_t) \quad \ldots \quad (3.4)
\]

\[
\dot{E}_{gen,element} = \dot{e}_{gen} V_{element} = \dot{e}_{gen} \Delta x \Delta y \Delta z \quad \ldots \quad \ldots \quad (3.5)
\]

Substituting these values in Equation 3.2 we get,

\[
\dot{Q}_x + \dot{Q}_y + \dot{Q}_z - \dot{Q}_{x+\Delta x} - \dot{Q}_{y+\Delta y} - \dot{Q}_{z+\Delta z} + \dot{e}_{gen} \Delta x \Delta y \Delta z
\]

\[
= \rho c \Delta x \Delta y \Delta z \frac{(T_{t+\Delta t} - T_t)}{\Delta t} \quad \ldots \quad \ldots \quad (3.6)
\]

Dividing by \( \Delta x \Delta y \Delta z \), Equation 3.6 becomes-

\[
- \frac{1}{\Delta y \Delta z} \frac{\dot{Q}_{x+\Delta x} - \dot{Q}_x}{\Delta x} - \frac{1}{\Delta z \Delta x} \frac{\dot{Q}_{y+\Delta y} - \dot{Q}_y}{\Delta y} - \frac{1}{\Delta x \Delta y} \frac{\dot{Q}_{z+\Delta z} - \dot{Q}_z}{\Delta z} + \dot{e}_{gen}
\]

\[
= \rho c \frac{(T_{t+\Delta t} - T_t)}{\Delta t} \quad \ldots \quad \ldots \quad (3.7)
\]

Heat transfer areas of the element for heat conduction in the x, y and z directions are \( A_x = \Delta y \Delta z \), \( A_y = \Delta z \Delta x \), and \( A_z = \Delta x \Delta y \) respectively, and taking the limits as \( \Delta x, \Delta y, \Delta z \) and \( \Delta t \to 0 \) yields,

\[
\frac{\partial}{\partial x} \left( k \frac{\partial T}{\partial x} \right) + \frac{\partial}{\partial y} \left( k \frac{\partial T}{\partial y} \right) + \frac{\partial}{\partial z} \left( k \frac{\partial T}{\partial z} \right) + \dot{e}_{gen} = \rho c \frac{\partial T}{\partial t} \quad \ldots \quad \ldots \quad (3.8)
\]

Equation 3.8 is the general heat conduction equation in rectangular coordinates for multidirectional heat flow. If the thermal conductivity is constant, it becomes-
\[
\frac{\partial^2 T}{\partial x^2} + \frac{\partial^2 T}{\partial y^2} + \frac{\partial^2 T}{\partial z^2} + \frac{\dot{e}_{\text{gen}}}{k} = \frac{1}{\alpha} \frac{\partial T}{\partial t} \] (3.9)

Where, the property \( \alpha = k/\rho c\) is the thermal diffusivity of the material. This equation is called Fourier-Biot equation and reduces further to the following equations under specific conditions [36] -

1. Steady state-

\[
\frac{\partial^2 T}{\partial x^2} + \frac{\partial^2 T}{\partial y^2} + \frac{\partial^2 T}{\partial z^2} + \frac{\dot{e}_{\text{gen}}}{k} = 0 \] (3.10)

2. Steady state, no heat generation-

\[
\frac{\partial^2 T}{\partial x^2} + \frac{\partial^2 T}{\partial y^2} + \frac{\partial^2 T}{\partial z^2} = 0 \] (3.11)

3. Transient state, no heat generation-

\[
\frac{\partial^2 T}{\partial x^2} + \frac{\partial^2 T}{\partial y^2} + \frac{\partial^2 T}{\partial z^2} = \frac{1}{\alpha} \frac{\partial T}{\partial t} \] (3.12)

Coolant flowing through the cooling channels takes away the heat that reaches from the casting part through the conduction of die material. Since this heat transfer mechanism is governed by the flow of coolant, the transfer of heat is due to convection. Let us consider a fluid flowing through a tube. From Newton’s law of cooling, the rate of heat transfer to or from a fluid flowing in a tube can be expressed as [16]-

\[
\dot{Q} = h_c A_s \Delta T_{avg} = h_c A_s (T_e - T_i)_{avg} \quad (W) \] (3.13)
Where, \( \dot{Q} \) = rate of heat transfer to or from the flowing fluid

\( h_c \) = average convective heat transfer coefficient

\( A_s \) = heat transfer surface area which is equal to \( \pi DL \) for a circular pipe of length \( L \) and diameter \( D \)

\( T_e \) = average temperature of the fluid at exit

\( T_i \) = average temperature of the fluid at inlet

The convection heat transfer coefficient can be calculated based on Dittus-Boetler [187] correction equation for forced convective heat transfer by turbulent flow in a circular pipe. This equation is expressed as-

\[
h_c = 0.023 \frac{k}{D} Re^{0.8} Pr^{0.4} \ldots \ldots \ldots \ldots \ldots \ldots (3.14)\]

When, \( 2500 < Re < 125000 \) and \( 0.6 < Pr < 100 \)

The symbols in equation (3.14) denote-

\( h_c \) = convective heat transfer co-efficient

\( k \) = thermal conductivity of coolant (water)

\( D \) = hydraulic diameter of the cooling channel

\( Re \) = Reynolds Number

\( Pr \) = Prandtl Number

In Equation 3.14, Reynolds Number and Prandtl Number are two dimensionless numbers. Use of these dimensionless numbers is a common practice in convection studies to nondimensionalize the governing equations and to combine the variables in order to reduce the number of total variables.
Prandtl number is named after Ludwig Prandtl, who introduced the concept of boundary layer and can be defined as [36] -

\[ Pr = \frac{Molecular \ diffusivity \ of \ momentum}{Momentum \ diffusivity \ of \ heat} = \frac{\vartheta}{\alpha} = \frac{\mu c_p}{k} \quad \ldots \ldots \ldots \ldots \quad (3.15) \]

Where,
\[ \vartheta = \frac{\mu}{\rho} = \text{kinetic viscosity of the flowing fluid} \]
\[ \mu = \text{dynamic viscosity of the flowing fluid} \]
\[ \rho = \text{density of the flowing fluid} \]
\[ \alpha = \frac{k}{\rho c_p} = \text{thermal diffusivity of the flowing fluid} \]
\[ c_p = \text{specific heat of the flowing fluid} \]
\[ k = \text{thermal conductivity of the flowing fluid} \]

The flow regime depends mainly on the ratio of the inertia forces to viscous forces in the fluid and this ratio is called the Reynolds Number. Reynolds number is named after Osborn Reynolds. It identifies whether a flow is laminar or turbulent. For flow in a circular tube, the Reynolds number is defined as [36] -

\[ Re = \frac{Inertia \ forces}{Viscous \ forces} = \frac{V_{avg} D}{\vartheta} = \frac{\rho V_{avg} D}{\mu} \quad \ldots \ldots \ldots \ldots \quad (3.16) \]

Where,
\[ V_{avg} = \text{average flow velocity of the flowing fluid} \]
\[ \vartheta = \frac{\mu}{\rho} = \text{kinetic viscosity of the flowing fluid} \]
\[ \mu = \text{dynamic viscosity of the flowing fluid} \]
\[ \rho = \text{density of the flowing fluid} \]
\[ D = \text{hydraulic diameter of the cooling channel} \]
3.3. Design of the Die

In the thermal analysis, the die was made of Moldmax HH, which was a high strength copper alloy, capable of transferring heat at a higher rate than steel and also withstanding the harsh environment associated with HPDC. Figure 3-2 shows the CAD model of the bi-metallic die made of Moldmax copper alloy with the die cavity deposited with 2 mm thick H13 tool steel layer. In this figure, copper alloy bulk part and tool steel clad part has also been shown separately. Figure 3-3 shows the identical tool steel die used for reference and heat transfer performance comparison with bi-metallic die. The die was 200 mm in length, 100 mm in width and 40 mm in height and consisted of five water cooling channels each 10 mm diameters. These cooling channels were located 27 mm below the casting-die interface. Due to the symmetry of the die, only one half of the die was considered and it was designed to produce an aluminum casting plate of 130×30×5 mm dimension.
3.4. Meshing and Boundary Conditions

ANSYS simulation software has been applied to carry out finite element heat transfer analysis to study the transient thermal response of both conventional tool
steel die and bi-metallic die. This software is capable of simulating both the steady state and transient behavior when subjected to different heat loads. In this simulation, transient thermal analysis has been used as the temperature of the die changes with time for given heat loads. Figure 3-4 shows the meshing of the dies that was done for the model with quadrilateral elements (automatic mesh method). There were 39881 nodes and 23459 elements for steel die whereas the bi-metallic die was meshed with 44314 nodes and 25553 elements that were automatically created depending on the physical structure. Fine relevance centre and medium smoothing was also applied in the meshing. Welding joint was used as connecting mode between tool steel and copper block considering the welded nature of DMD cladding.
Boundary conditions were set to use the heat flux as heat input on the cavity surface only and convection film coefficient in the cooling channels that carried out heat from the die and the casting. All other surfaces were considered as insulated since they were not exposed to molten aluminium and heat flow through those surfaces were taken as zero (Figure 3-5). The initial temperature of the die was considered 100 °C. Figure 3-6 shows the cavity surface where heat flux was applied. Heat flux value was used from the work of Papai [188], who used an iterative procedure reported by Beck [189] that determines the best temperature fit by minimization of the least squares error between the measured temperatures and the thermal field as calculated via the solution of the one-dimensional planar Fourier heat conduction equation with unknown heat flux at the surface,

\[
\frac{\partial T}{\partial t} = \alpha \frac{\partial^2 T}{\partial x^2} \tag{3.17}
\]

where,

\( T \) = temperature at a given position and time
$t = \text{time}$

$x = \text{position relative to the die surface}$

$\alpha = \text{die thermal diffusivity} = \frac{k}{\rho c}$

where

$k = \text{die thermal conductivity}$

$\rho = \text{die density}$

$c = \text{die specific heat}$

The instantaneous heat flux, $q$ at the surface of the die (at $x = 0$) can be derived from equation (3.1) as given by,

$$q = -k \frac{\partial T}{\partial x} \quad \text{(3.18)}$$
Figure 3-5: Faces of the die through which heat transfer was considered negligible (a) bi-metallic die (b) tool steel die
Figure 3-6: Cavity surfaces of the die on which heat flux was applied (a) bi-metallic die (b) tool steel die

Figure 3-7 shows the cooling channel surfaces where convection heat transfer condition was applied. The convection film coefficient has been calculated based on Dittus-Boelter correction equation (equation-3.14) for forced convective heat transfer by turbulent flow in a circular pipe and was 9917 watt/m² °C.
Other properties that were used in the analysis were density, specific heat and thermal conductivity of the materials. Table 3-1 lists tool steel properties and these were constant throughout the analysis. In contrast, properties of Moldmax are temperature dependant and have been listed in Tables 3-2, 3-3 and 3-4 [42].

Table 3-1: Lists of tool steel properties

<table>
<thead>
<tr>
<th>Property</th>
<th>Value</th>
</tr>
</thead>
<tbody>
<tr>
<td>Density</td>
<td>7850 kg.m$^{-3}$</td>
</tr>
<tr>
<td>Specific heat</td>
<td>460 J.kg$^{-1}$.K$^{-1}$</td>
</tr>
<tr>
<td>Thermal conductivity</td>
<td>24.6 W.m$^{-1}$.K$^{-1}$</td>
</tr>
</tbody>
</table>

Table 3-2: Density of Moldmax

<table>
<thead>
<tr>
<th>Temperature (°C)</th>
<th>Density (kg.m$^{-3}$)</th>
</tr>
</thead>
<tbody>
<tr>
<td>20</td>
<td>8350</td>
</tr>
<tr>
<td>200</td>
<td>8275</td>
</tr>
<tr>
<td>300</td>
<td>8220</td>
</tr>
</tbody>
</table>
Table 3-3: Specific heat of Moldmax

<table>
<thead>
<tr>
<th>Temperature (°C)</th>
<th>Specific heat (J.kg⁻¹.K⁻¹)</th>
</tr>
</thead>
<tbody>
<tr>
<td>20</td>
<td>380</td>
</tr>
<tr>
<td>200</td>
<td>480</td>
</tr>
<tr>
<td>300</td>
<td>535</td>
</tr>
</tbody>
</table>

Table 3-4: Thermal conductivity of Moldmax

<table>
<thead>
<tr>
<th>Temperature (°C)</th>
<th>Thermal conductivity (W.m⁻¹.K⁻¹)</th>
</tr>
</thead>
<tbody>
<tr>
<td>20</td>
<td>110</td>
</tr>
<tr>
<td>200</td>
<td>145</td>
</tr>
<tr>
<td>300</td>
<td>155</td>
</tr>
</tbody>
</table>

3.5. Results and Discussion

Simulation result shows that for a given solidification time, bi-metallic die cools the casting part at a much higher rate than that of a traditional steel die. Norwood et al. [190] reported from their experimental work that the die surface temperature reduced to 150–200 °C after solidification of the casting part. Therefore, the final temperature of the die at the end of the cycle was considered around 175 °C in the analysis. Figure 3-8 shows the temperature distribution on the die surfaces after completion of a cycle. From the figure it is evident that die surface temperature of both dies reached close to 175 °C. Figure 3-9 depicts the overall transient thermal behavior of the steel die and bi-metallic die for the entire solidification process. For both dies the temperature rose to its maximum value immediately after molten metal injection and then decreased gradually to the ejection temperature. The steel die required 24.6 seconds to reduce the temperature to the ejection point where as the bimetallic die took only 7.8 seconds to reach the ejection temperature. It was thus confirmed that the solidification time of a casting part reduced to one third by
using a bi-metallic die in comparison to a conventional tool steel die. This is a significant reduction of the whole high pressure die casting cycle time.

Figure 3-8: Temperature distributions on the die surfaces at the end of a thermal cycle (a) bi-metallic die (b) too steel die
Figure 3-9: Calculated transient temperature responses of the dies over entire solidification process

3.6. Summary

Finite element analysis shows that high strength copper alloy coated with protective tool steel layers offers a potential alternative to replace tool steel as a die material in terms of thermal behavior since it can reduce solidification time to 1/3 of the time taken by tool steel die.
4.1. Introduction

Die coatings and surface treatments are used to improve the properties of the dies for many applications. Physical vapor deposition (PVD), chemical vapor deposition (CVD), thermo-reactive deposition and thermal spray process are some of the widely used techniques for application of coatings to molds that have shown promising results for increasing wear resistance, reducing die maintenance and machine down time [91, 191-195]. Depending on the applied process, these coatings often fail due to high level of porosity and poor bonding of the applied surface to the base material. In addition, these processes may entail a risk of distortion or embrittlement and produces toxic and corrosive gases. Laser aided direct metal deposition (DMD) can be used to enhance the property of the coating. DMD is a laser cladding process in which a particular part of the substrate can be coated with a material which has superior properties, producing a fusion bond between the two materials with minimal dilution of the clad by the substrate. Currently DMD is used to produce objects for tooling and mould production. The greatest potential of the DMD process is the capability of producing specialized and functional products such as bi-metallic structures through metallurgical bonding of the coating and substrate materials. Moreover, both similar and dissimilar materials can be coated using DMD, which holds major advantage over any other joining method for industrial applications.

In DMD process a laser beam is used to create a melt pool on the surface of a solid substrate into which a metallic powder is injected [135, 196, 197]. The laser melts the powder and fuses it on to the substrate creating a fully dense, metallurgically bonded bead. By overlapping the beads, usually by 50%, a continuous layer is produced. However, the properties of the clad do not entirely depend upon the
materials alone; laser processing parameters and chemical composition of the materials have significant impact on the properties of the clad. Particularly, when dissimilar material is clad on the substrate, the process mechanics involved in fusing two different materials at the same energy from a laser beam is very complex. Though DMD can be applied for coating dissimilar metals through metallurgical bonding, this surface treatment is limited for copper, due to the high reflectivity of copper to laser beam during laser cladding. Specifically, high reflective characteristic of copper to the infrared output wavelength of a CO$_2$ (10.6 µm) laser results in lower process efficiency and make the process more complex. Therefore, it is almost certain that production of H13 tool steel clad on copper alloy substrate is entirely parameter responsive. To succeed in producing quality bimetallic structure of steel on copper alloy substrate, DMD process parameters like laser power, focus position, scanning speed and powder mass flow rate need to be applied precisely. This can be done through investigation of the response of process parameters on clad. The potential benefit of investigating the process response is twofold. Firstly, understanding the build parameter effects on process response will aid development of further intelligent process control. Secondly, understanding the process response will allow users to select appropriate parameter values prior to processing in order to produce parts with desired physical properties such as hardness and porosity.

Several studies have been carried out in laser cladding of various materials which mainly focused on selecting appropriate materials to be processed, developing the process and general process parameter selection [198, 199]. Choi et al. [200] investigated the characteristics of steel deposited on steel substrate. Emamian et al. [201, 202] investigated the effects of laser parameters on the quality of an in-situ formed TiC–Fe based composite clad. Wang et al. [203] reported the effect of laser power and heat treatment process on microstructure and property of multi-pass Ni based alloy laser cladding coating. Wang et al. [204] investigated the effects of crucial process parameters, such as laser power, scanning speed and wire delivery
rate on the clad layer of direct laser deposition (DLD) using coaxial inside-beam wire feeding technique. Finite element analysis (FEA) has also been employed for metallurgical characterization and hardness measurements to quantify the effect of processing parameters on the variation of geometrical features (size and shape) of the clad, and the surface properties (hardness, resistance to heat, wear, and corrosion) [205]. Zhao et al. [206] investigated the influences of laser cladding parameters on nickel-based composite coatings on H13 steel surface. DMD processed tungsten and Inconel 718 on high strength copper substrate are studied as well [207]. Although these works provide knowledge of various aspect of laser cladding process, some important factors such as reflectivity of copper, influence of process parameters in manufacturing 3D structure of steel on copper substrate have not been well understood. Therefore, specific study of the scientific and technical aspects of Direct Metal Deposition of H13 tool steel on copper alloys is in great demand for fabrication of copper based bi-metallic dies for die casting and injection molding industries. These studies involve characterizing the influence of the various process parameters on the deposition process, determining difficulty associated with 3D manufacturing of bimetallic structure of these two materials using DMD and finally selection of a suitable set of process parameters that overcome the associated difficulty for a sound and functional structure.

This chapter closely investigates the influence of various process parameters such as laser power, focus position, scanning speed and powder mass flow rate on the deposition process of H13 tool steel on Copper alloy substrate. To find out the relationship between process parameters and cladding, various common characteristics in rapid prototyping, such as, distribution of porosity, microhardness and micro cracks in the layer have been investigated. Reflectivity, a common barrier associated with laser processing on copper with a CO₂ laser was also considered during the deposition process. In fact a set of process parameters has been determined to overcome high reflection and to produce good quality clad.
4.2. Experimental Work

H13 tool steel was deposited on a copper alloy substrate using POM DMD 505 laser cladding facility. A continuous wave CO2 laser with a maximum power of 5 kW was used to produce beads and clads. Steel powder and copper alloy plate were supplied by Robtec (Australia) Pty Ltd and Bohler Uddeholm (Australia) Pty Ltd respectively. The thickness of all the copper alloy plates was 10 mm. Chemical compositions of copper alloy plate and H13 tool steel powder are listed in Tables 4-1 and 4-2. The process responses are usually the physical properties such as distribution of porosity, strength and degradation of deposited clad. These properties are all influenced by process parameters. Whole experimental work was conducted in three steps through analysis of various process responses.

<table>
<thead>
<tr>
<th>Element</th>
<th>Co + Ni</th>
<th>Be</th>
<th>Cu</th>
</tr>
</thead>
<tbody>
<tr>
<td>Chemical composition</td>
<td>0.5%</td>
<td>1.9%</td>
<td>Balance</td>
</tr>
</tbody>
</table>

<table>
<thead>
<tr>
<th>Element</th>
<th>C</th>
<th>Mn</th>
<th>Si</th>
<th>V</th>
<th>Mo</th>
<th>Cr</th>
<th>Fe</th>
</tr>
</thead>
<tbody>
<tr>
<td>Chemical composition</td>
<td>0.35%</td>
<td>0.4%</td>
<td>1%</td>
<td>1%</td>
<td>1.5%</td>
<td>5%</td>
<td>Balance</td>
</tr>
</tbody>
</table>

4.2.1. Preliminary Laser Cladding Trials

Initially, for analysis purposes, bead, which is a single track produced in a single pass with one set of process parameters was constructed. With this research, the main intention was to select a set of parameters that are capable of sufficiently melting both materials to make a metallurgical bond. Therefore, in the early stage of the study, few trials were carried out by using a range of parameters. Table 4-3 shows summary of the combinations of parameters used in these trials. Figure 4-1 shows corresponding beads produced. After visual observation of the beads, one set of parameters with 3 kW laser power was selected to clad a 25×25 mm square
area (Figure 4-1, area 1). Laser beam overlap was set at 50% and oxidation was avoided by using Argon gas shielding. Area 2 in Figure 4-1 was deposited with the same process parameters but with tilted nozzle position. Reason for tilting nozzle has been explained in results and discussion section.

Table 4-3: Summary of the preliminary trials

<table>
<thead>
<tr>
<th>Trial order</th>
<th>Laser power (W)</th>
<th>Powder mass flow rate (g/min)</th>
<th>Focus position (mm)</th>
<th>Feed rate (mm/min)</th>
<th>Visual observation</th>
</tr>
</thead>
<tbody>
<tr>
<td>R1</td>
<td>1500</td>
<td>9.2</td>
<td>+20</td>
<td>500</td>
<td>No bonding</td>
</tr>
<tr>
<td>R2</td>
<td>2000</td>
<td>9.2</td>
<td>+20</td>
<td>500</td>
<td>No bonding</td>
</tr>
<tr>
<td>R3</td>
<td>2500</td>
<td>9.2</td>
<td>+20</td>
<td>500</td>
<td>No bonding</td>
</tr>
<tr>
<td>R4</td>
<td>4000</td>
<td>9.2</td>
<td>+20</td>
<td>250</td>
<td>Poor bonding</td>
</tr>
<tr>
<td>R5</td>
<td>5000</td>
<td>9.2</td>
<td>+20</td>
<td>250</td>
<td>Poor bonding</td>
</tr>
<tr>
<td>R6</td>
<td>5000</td>
<td>6.2</td>
<td>0</td>
<td>150</td>
<td>Poor bonding</td>
</tr>
<tr>
<td>R7</td>
<td>5000</td>
<td>3.15</td>
<td>-15</td>
<td>150</td>
<td>Poor bonding</td>
</tr>
<tr>
<td>R8</td>
<td>5000</td>
<td>1.58</td>
<td>-18</td>
<td>150</td>
<td>Better than previous</td>
</tr>
<tr>
<td>R9</td>
<td>5000</td>
<td>0.8</td>
<td>-18</td>
<td>150</td>
<td>Good bonding</td>
</tr>
<tr>
<td>R10</td>
<td>4000</td>
<td>0.8</td>
<td>-18</td>
<td>150</td>
<td>Good bonding</td>
</tr>
<tr>
<td>R11</td>
<td>3000</td>
<td>0.8</td>
<td>-18</td>
<td>150</td>
<td>Good bonding</td>
</tr>
</tbody>
</table>

Figure 4-1: Beads and clads produced in preliminary trials
4.2.2. Trials with Low Power Laser to Reduce Back Reflection

In these set of experiments, the possibility to construct tool steel bead on copper alloy substrate using low power laser that would create bead without high reflection of the laser beam was investigated. Initially the powder delivery nozzle was kept tilted to minimize back reflections from the samples. The feed rate was kept fixed at 150 mm/min, while powder mass flow rate and focus position were varied in the range of 1.58 to 4.6 g/min and (-) 18 to (+) 5 mm respectively. Two laser powers, 2 kW and 2.5 kW were used in these trials. Table 4-4 shows these parametric combinations and Figure 4-2 shows the produced beads with these parametric combinations (numbered 1 to 13). At the end of this step, to investigate the nature of reflection during large area deposition with these parametric combinations, three pads were deposited. Two blocks each 15×15 mm were deposited with 2 kW and 2.5 kW laser power (Figure 4-2, area 1 and 2), while in the next pad, the area was increased to 60×25 mm and was deposited with 2 kW laser power in single operation (Figure 4-2, area 3).

Table 4-4: Summary of the trials to reduce back reflection

<table>
<thead>
<tr>
<th>Trial order</th>
<th>Laser power (W)</th>
<th>Powder mass flow rate (g/min)</th>
<th>Focus position (mm)</th>
<th>Nozzle angle (˚)</th>
<th>Visual observation</th>
</tr>
</thead>
<tbody>
<tr>
<td>L1</td>
<td>2000</td>
<td>4.6</td>
<td>-18</td>
<td>20</td>
<td>Very poor bonding</td>
</tr>
<tr>
<td>L2</td>
<td>2000</td>
<td>3.15</td>
<td>-18</td>
<td>20</td>
<td>Very poor bonding</td>
</tr>
<tr>
<td>L3</td>
<td>2000</td>
<td>1.58</td>
<td>-18</td>
<td>20</td>
<td>Poor bonding</td>
</tr>
<tr>
<td>L4</td>
<td>2000</td>
<td>1.58</td>
<td>-15</td>
<td>20</td>
<td>Poor bonding</td>
</tr>
<tr>
<td>L5</td>
<td>2000</td>
<td>1.58</td>
<td>-13</td>
<td>20</td>
<td>Poor bonding</td>
</tr>
<tr>
<td>L6</td>
<td>2000</td>
<td>1.58</td>
<td>-10</td>
<td>20</td>
<td>Poor bonding</td>
</tr>
<tr>
<td>L7</td>
<td>2000</td>
<td>1.58</td>
<td>-5</td>
<td>20</td>
<td>Good bonding</td>
</tr>
<tr>
<td>L8</td>
<td>2000</td>
<td>1.58</td>
<td>+5</td>
<td>20</td>
<td>Good bonding</td>
</tr>
<tr>
<td>L9</td>
<td>2500</td>
<td>2.3</td>
<td>+5</td>
<td>20</td>
<td>Good bonding</td>
</tr>
<tr>
<td>L10</td>
<td>2000</td>
<td>2.3</td>
<td>+5</td>
<td>20</td>
<td>Good bonding</td>
</tr>
<tr>
<td>L11</td>
<td>2000</td>
<td>2.3</td>
<td>+5</td>
<td>10</td>
<td>No reflection</td>
</tr>
<tr>
<td>L12</td>
<td>2000</td>
<td>2.3</td>
<td>+5</td>
<td>5</td>
<td>No reflection</td>
</tr>
<tr>
<td>L13</td>
<td>2500</td>
<td>2.3</td>
<td>+5</td>
<td>0</td>
<td>No reflection</td>
</tr>
</tbody>
</table>
4.2.3. Preparing Clad Pads

For further analysis of clads, various pads of steel were deposited with different combinations of laser power (LP) and powder mass flow rate (PFR). The deposition area was 15×10 mm for all these deposited clads. Feed rate and laser focus position were kept constant at the values 150 mm/min and +5 respectively throughout the deposition as high values of these two parameters combined with other parametric combinations were found to produce insufficient heat energy to melt the powder. Since, reflection was a major obstacle in laser processing of the copper alloy and the 3 kW laser power resulted in tremendous reflected heat from the substrate, some clads were built by using different laser powers sequentially. In these clads, first few layers were deposited using 2.5 kW laser power followed by clads deposited using high laser power (3 kW) at the top progressively. Table 4-5 summarizes these parametric combinations and Figure 4-3 shows deposited layers.
Table 4-5: Summary of the parametric combinations to deposit clad pads

<table>
<thead>
<tr>
<th>Block No (in Figure 4-3)</th>
<th>Laser power (kW)</th>
<th>Powder feed (g/min)</th>
<th>Remarks</th>
</tr>
</thead>
<tbody>
<tr>
<td>1</td>
<td>2</td>
<td>1.58</td>
<td>6 layers all with same parameters</td>
</tr>
<tr>
<td>2</td>
<td>2.5</td>
<td>1.58</td>
<td>6 layers all with same parameters</td>
</tr>
<tr>
<td>3</td>
<td>2.75</td>
<td>2.3</td>
<td>6 layers all with same parameters</td>
</tr>
<tr>
<td>4</td>
<td>2.5</td>
<td>2.3</td>
<td>2 layers with 2.5 kW LP and 3 layers with 2.75 kW LP</td>
</tr>
<tr>
<td>5</td>
<td>2.5</td>
<td>2.3</td>
<td>2 layers with 2.3 gm/min PFR and 4 layers with 1.58 gm/min PFR</td>
</tr>
<tr>
<td>6</td>
<td>2.3</td>
<td>2.3</td>
<td>6 layers each 2 with 2.5, 2.75 and 3 kW LP respectively</td>
</tr>
</tbody>
</table>

Figure 4-3: Steel clads produced with various combinations of process parameters

4.2.4. Metallurgical Analysis

All samples were cut in the transverse direction and polished with fine diamond polishing machine up to 1 micrometer for porosity analysis and hardness measurement. Afterwards, the parts were rinsed with alcohol and dried with air in order to remove any contaminations. Finally the parts were etched using 2% nital to allow a precise microstructural view of the surfaces. SEM images were obtained using SUPRA 40 VP SEM and Philips XL30 FEG SEM. The microhardness
across the interface of the specimens was measured using CLARK Microhardness Tester (Model- CM 400AT) with 300 g load.

4.3. Results and Discussion

4.3.1. Preliminary Trials

The first experiments considered beads on plate surface to determine a set of deposition parameters that are capable of sufficiently melting both the tool steel powder and substrate material. Initial experiments were performed using a broad range of processing parameters. Figure 4-1 shows the quality of bead formation observed visually with various laser processing parameters. In the preliminary trials, copper alloy was thought to be melted with low energy due to its low melting temperature. Therefore these trials were explored with low laser power starting from 1.5 kW. But visual inspection of the beads revealed that, for each power setting no bead was produced up to 2.5 kW laser power with high values of powder mass flow rate (9.2 g/min), feed rate (500 mm/min) and focus position (+20). It meant that the energy produced with the combination of these parameters was not able to create a melt pool on copper substrate. The high thermal conductivity of copper that transferred heat from the surface to the other part of the substrate made the melting further difficult. In an effort to create a melt pool on substrate, the laser power was gradually increased to 4 kW and 5 kW while feed rate was decreased to 500 mm/min and 250 mm/min to increase intensity of the energy delivered.

The first bead was produced at 4 kW laser power and 250 mm/min feed rate though bonding of the materials was poor (bead R4 in Figure 4-1). The situation couldn’t be rectified even at maximum 5 kW power with all other parameters constant (bead R5 in Figure 4-1). These beads clearly demonstrated insufficient melting as lots of powder particles were present in the beads which were caused from excessive powder mass flow rate leading to insufficient melting of powder.
Therefore, powder mass flow rate was gradually decreased to 6.2, 3.15, 1.58 and 0.8 g/min along with focus position (0, -15 and -18) and feed rate (150 mm/min) to increase the energy intensity while laser power was fixed at 5 kW. These reductions vastly affected the deposition process since visual examination of these tracks revealed that in all cases a bead was produced and it was well bonded to the substrate. Bead R7 in Figure 4-1 produced with laser power 5 kW, powder mass flow rate 3.15 g/min, focus position -15 and feed rate 150 mm/min looked sound for further deposition. Nevertheless, SEM image shown in Figure 4-4 depicts the presence of numerous micro cracks in the bead. These micro cracks were well described by Simchi [208] who found that, at high laser energy input, delamination of sintered layers and formation of large cracks were feasible. The image also shows insufficient melting due to the large powder mass flow rate.

Figure 4-4: SEM image of the bead (surface view) produced with laser power 5 kW, powder mass flow rate 3.15 g/min, focus position -15 and feed rate 150 mm/min

The bead quality improved with the reduction of powder mass flow rate while all other parameters were fixed. Figure 4-5 illustrates that the powder was better
Figure 4-5: SEM image of the bead (surface view) produced with laser power 5 kW, powder mass flow rate 1.58 g/min, focus position -18 and feed rate 150 mm/min

Figure 4-6: SEM image of the bead (surface view) produced with laser power 5 kW, powder mass flow rate 0.8 g/min, focus position -18 and feed rate 150 mm/min
melted and bonded to substrate in the bead produced with 1.58 g/min powder mass flow rate. However, porous holes were present in the microstructure which reduced with the reduction of powder mass flow rate with all other parameters fixed (Figure 4-6). Figure 4-7 shows a comparison of porosity area fractions in the beads. Simchi [208] showed that at high laser intensity, delamination of laser processed layer caused thermal stresses, formation of gas pores during solidification, porosity formation due to material shrinkage and balling effects. Therefore, in the next trials laser power was reduced gradually keeping all other parameters fixed. Thus, the bead produced with 4 kW laser power showed reduced porous holes and beads produced with 3 kW laser power showed almost pore free microstructure as shown in Figure 4-8 and Figure 4-9 respectively. Though the microstructure showed low level of porosity, this set of parameters with 3 kW laser power was not suitable for large area deposition (Figure 4-1, area 1) as in this case, back reflection from the substrate was incident on the nozzle and burnt it (Figure 4-10).
Figure 4-8: SEM image of the bead (surface view) produced with laser power 4 kW, powder mass flow rate 0.8 g/min, focus position -18 and feed rate 150 mm/min

Figure 4-9: SEM image of the bead (surface view) produced with laser power 3 kW, powder mass flow rate 0.8 g/min, focus position -18 and feed rate 150 mm/min
Figure 4-10: Melted nozzle due to back reflected light when using 3 kW laser power

Making substrate surface rough, tilting nozzle head and reduction of laser power could be three effective ways to minimize the back reflection. In this experiment sand blasting at the copper alloy surface performed, but did not eliminate back reflection completely. The degree of roughness is restricted since highly rough surfaces cause potential defects at the interface of clad and substrate [180, 209]. Again, at the tilted position of the nozzle, reflection was high enough to melt nozzle back plate and feedback tube of the machine (Figure 4-11) during large area deposition (Figure 4-1, area 2). Tilting nozzle further could potentially damage some other part as tilting the nozzle was not actually reducing reflection rather directed reflected light to other machine parts instead of nozzle tip. Moreover, while reading the CAD model for 3D deposition of complex shaped products, DMD machine doesn’t operate from the tilting position since it creates own tool path from the vertical position of the nozzle relative to the substrate. As a result, the laser power was needed to be reduced further to avoid these consequences.
Figure 4-11: Melting during large area deposition with 3 kW laser power and tilted nozzle (a) back plate and (b) feedback tube
4.3.2. Low Power Laser Trials

The aim of reducing laser power was to minimize beam reflection while producing quality clad layers. Initially, laser power was reduced to 2 kW in these trials. The nozzle was kept tilted from the beginning to avoid any unexpected melting of nozzle tip from the reflected laser beam that might be produced. Keeping in mind that if more powder was used it could absorb most of the energy from the laser beam leaving less energy to be reflected, powder mass flow rate was also increased at the beginning of these trials. But, reduction of laser power in combination with increased powder mass flow rate resulted in insufficient melting since the bead produced with 2 kW laser power and 4.6 g/min powder flow rate showed very poor bonding (Figure 4-2, bead 1). It happened due to the fact that low energy of the laser beam couldn’t provide sufficient heat for the large amount of powder to be molten to create adequate bonding. Though bead quality improved gradually with reduction of powder mass flow rate, the bonding was not adequate even up to 1.58 g/min powder mass flow rate. Therefore, another process parameter i.e. focus position was considered to be altered to change the energy density. This change in the process parameter influenced the bead quality immediately since the bonding improved considerably with the decrease of focus position (Figure 4-2, beads 4-7). Bonding looked much better when focus position was changed from negative (-5) to positive (+5) with same set of process parameters (Figure 4-2, bead 8). Even increasing powder mass flow rate slightly to 2.3 g/min with this set of process parameters showed same quality bonding as the bead produced with 1.58 g/min (Figure 4-2, bead 10). In addition, one bead was successfully produced by increasing laser power to 2.5 kW. However, it was not considered for large area deposition since this power could be unfavorable in terms of reflection. As a result, 2 kW laser power was considered for further deposition and nozzle was shifted to vertical position gradually producing beads with good bonding and without reflection that could damage nozzle tip (Figure 4-2, beads 11-13). Successful deposition of two pads of 10×10 mm and 60×25 mm without any reflected beam
issues (Figure 4-2, area 1 and 3) showed further possibility of using this set of process parameters. However, Figure 4-12 shows that hardness was slightly above 200 HV through the entire clad layers which was too low to be applied in high pressure and temperature applications. The reason for low hardness values was the incomplete melting of the tool steel powder. Though the clad looked well bonded from outside, the microscopic observation revealed that there were lots of particles in the surface that didn’t melt sufficiently to create fully dense clad (Figure 4-13). In contrast, hardness was high in 10×10 mm pad (Figure 4-2, area 2) deposited using 2.5 kW laser power, 2.3 g/min powder mass flow rate and +5 focus position (Figure 4-14). The difference in hardness in clads with different laser power urged analysis of hardness of various pads prepared with various combination of laser power.

Figure 4-12: Vickers microhardness of clad deposited with 2 kW laser power
Figure 4-13: SEM image of the clad (surface view) with 2 kW laser power and 2.3 g/min powder mass flow rate showing incomplete melting

Figure 4-14: Vickers microhardness of clad deposited with 2.5 kW laser power
4.3.3. Analysis of Microhardness

Through microhardness analysis of the pads, the main intention was to achieve a set of process parameters that was capable of producing sufficiently strong clads without reflection. Therefore, microhardness of clads were correlated with laser power and powder mass flow rate. A 300 g load was used to measure the hardness of the specimens. Figure 4-12 showed that hardness increased slightly with decrease of powder mass flow rate though it was still not enough to be applied in high strength applications. Reduced powder mass flow rate facilitated more energy to be absorbed by fewer amounts of powder particles from molten pool of the substrate during cladding process and increased melting quality. In the same manner, increased laser power provided more energy in molten pool for powder to be melted completely which resulted in increased hardness of the clad. Figure 4-14 showed that microhardness was around 410 HV in the clad deposited with 2.5 kW laser power and 2.3 g/min powder mass flow rate. Though reduction of powder mass flow rate offered hardness close to 450 HV, this combination of parameters was not applicable as it was found reflecting heat to damage the nozzle tip of DMD machine. At high laser power like 2.5 kW with reduced powder mass flow rate (1.58 g/min), the heat energy in the molten pool couldn’t be absorbed fully by the fewer amount of powder particles and hence; remaining amount of energy was reflected from the substrate. The 2.75 kW laser power was reflecting heat even with 2.3 g/min powder mass flow rate due to the same reason. Figure 4-15 shows microhardness of clad deposited with 2.75 kW laser power. Though a small pad of 15×10 mm could be deposited and hardness of this clad was higher compared to other pads, 2.75 kW laser power was not suitable for large area deposition since it could potentially damage machine by reflected beam. The increase in hardness with increase in laser power however offered a method to combine different level of laser powers to clad different layers of steel on copper alloy substrate. This method could rectify process shortcoming associated with
reflection as well as provide sufficient strength of the clad. In this method, first two layers were deposited with 2.5 kW laser power without reflection and then on top of this layer further cladding was done. This way, the first layer produced with 2.5 kW may not provide sufficient strength and melting of the powder. However, it worked as substrate for the next layers which helped this layer to melt completely during deposition of next layers with higher laser power to offer increased hardness. Figure 4-16 demonstrates hardness of the clad found in this method. It showed that hardness was around 480 HV from the top to around 1.5 mm in the layer. The hardness then dropped below 450 HV in the region that was deposited with 2.5 kW laser power. SEM image in figure 4-17 (a) shows complete melting of powder particles. Unlike layers produced with low laser power, proper melting of the powder particles in the layers with high power laser left no porous holes in the clad that could affect strength of the clad. The hardness reduced largely in heat affected zone (HAZ), just below the interface up to around 1 mm in the substrate plates in all specimens. Heat affected zone (HAZ) is a common feature in laser cladding which changes the grain size of the substrate material to change material properties. Applied laser caused grain coarsening of copper alloy substrate plate to reduce hardness. The fully melted interface of the cladding shown in Figure 4-17 (b) proved the further applicability of this method.

Figure 4-15: Vickers microhardness of clad deposited with 2.5 kW laser power
Figure 4-16: Vickers microhardness of clad deposited with 2.75 kW and 2.5 kW laser power
One of the major challenges of the research was to deposit tool steel layers on copper alloy substrate due to very different physical properties of these two materials. Nevertheless, through an extensive experimental investigation this objective was attained and following conclusions were drawn from the DMD parametric investigation to deposit H13 tool steel on copper alloy substrate:

- Though high intensity of energy is required for DMD deposition of steel on copper substrate due to the higher thermal conductivity of copper alloys, laser power more than 2.5 kW is not suitable as it reflects tremendous heat to cause damage to the machine. In contrast, laser power less than 2.5 kW produces clads with low hardness. Therefore, 2.5 kW laser power is recommended for this particular application.
• Laser scanning speed should be applied within 150 mm/min since scanning speed more than 150 mm/min cannot produce sufficient intensity of heat to create a complete melt pool on the substrate.

• Powder mass flow rate should be within the range of 1.58 - 2.3 gm/min for any combination of the parameters.

• Increase in laser power increases the hardness. Nevertheless, more than 2.5 kW laser power can’t be used directly on copper alloy substrate due to reflection of laser beam. Therefore, to achieve high strength property of the clad, first layer should be deposited with 2.5 kW laser power and on the top of first layer further depositions can be made with high power laser.

• For all clads on copper alloy, hardness decreases significantly in heat affected zone (HAZ).
CHAPTER 5

MICROSTRUCTURAL CHARACTERIZATION

5.1. Introduction

Microstructural homogeneity is a pre-requisite for the superior mechanical properties of the laser deposited clads. Conventional processing of tool steels, such as casting and hot working gives rise to coarse-grain microstructure and carbide segregation at grain boundaries due to inherently high alloy contents [210]. This coarse microstructure and carbide segregation reduce toughness which consequently increases risk of premature failure through chipping or breakage of tools during highly stressed service conditions. Due to rapid solidification (RS), DMD is able to overcome the metallurgical challenges associated in conventional processing methods and can produce highly refined and homogeneous microstructure. Depending on cooling rate the matrix microstructure of a rapidly solidified tool steel varies from ferrite to martensite. Therefore, high hardness combined with good toughness is achievable through subsequent control of the cooling to precipitate fine carbides [211].

However, the phase diagram of Fe-Cu reveals a large solidification temperature range coupled with low solid solubility, both of which promote solidification cracking (Figure 5-1) [212]. Noecker et al. [213] experimentally correlated solidification cracking of steel-copper deposits as a function of copper concentration in the clad. He showed that when copper concentration ranges from 7.5 wt% to 55 wt%, copper wets the interdendritic region, which decreases the ability to withstand solidification stresses, thereby forming cracks. Hence, to avoid solidification cracking in steel-copper clad, the nominal composition of copper in the first deposited layer must be at least 55 wt% which is practically difficult to control in laser metal deposition. Moreover, to provide more copper in melt pool,
substrate needs to be melted more and too much melting of the substrate will increase the dilution of copper in the steel clad which subsequently degrades clad quality to withstand high temperature and pressure application. Since, nickel has excellent solubility in copper and very good solubility in iron, placing a high nickel stainless steel (316L SS) buffer layer between the copper substrate and steel deposits can be a potential solution to the solidification cracks induced from miscibility gap of Fe-Cu.

Since microstructural homogeneity is an essential element to determine the quality of the product, many investigations have been carried out to characterize tool steel [214-219]. Most of the earlier studies focused on fundamental issues about process optimization of various processes for defect prevention and microstructure optimization. However, from a thorough literature survey, it appears that a deep knowledge on the microstructural aspect of H13 tool steel when deposited on copper alloy substrate is not fully achieved. Moreover, use of intermediate 316L stainless steel layer needs to be evaluated in terms of microstructure as it might

![Fe-Cu binary phase diagram showing miscibility gap](212)

Figure 5-1: Fe-Cu binary phase diagram showing miscibility gap [212]
have effect on the clad characteristics. Hence, a contribution to fill this lack of literature data has been presented in this investigation. In the previous chapter, among various DMD process parameters, laser power was found to affect the hardness of the tool steel clad. Therefore, an approach has also been taken to correlate laser power, microstructure and microhardness of the clad.

This chapter concerns direct metal deposition of H13 tool steel powder onto high strength copper alloy substrate. It presents detailed microstructural study of H13 tool steel clads when deposited both directly as well as using high nickel 316L stainless steel as buffer layer onto copper alloy substrate. In addition, effect of laser power in the microstructure and microhardness of the tool steel clad has also been investigated.

5.2. Experimental Details

5.2.1. Materials and DMD Cladding

POM DMD 505 machine equipped with a CO2 laser capable of producing maximum 5 kW power was used for closed loop experimental laser metal deposition. Commercially available tool steel powder of grade H13 supplied by Robtec Australasia Pty. Ltd. was used as deposition material due to its excellent mechanical behaviour and extensive use in HPDC. A 10 mm thick plate of high strength copper alloy (Moldmax) was used as the substrate material. Moldmax plate was supplied by Bohler Uddeholm Australia Pty. Ltd. The 316 L stainless steel powder (trade name Metco 41C stainless steel) was used as buffer layer between substrate and tool steel deposits. The 316 L stainless steel powder was supplied by Sulzer Metco Australia Pty. Ltd. Table 5-1 shows the chemical compositions of 316 L stainless steel (316L SS) powder. Chemical compositions of copper alloy substrate and H13 tool steel powder have been listed in the previous chapter. During the experiment, two 60×40×6 mm pads were deposited on copper alloy substrate. First pad was of H13 tool steel directly deposited on substrate and
this specimen is designated as *H13 TS* specimen throughout the text. Other block was deposited with 1 mm thick 316L SS buffer layer deposited immediately on substrate, followed by 5 mm thick H13 tool steel layer deposited on 316 L SS and this specimen is designated as *316 L SS buffer* specimen throughout the text. A 2.5 kW laser power with 1 mm laser beam diameter, 150 mm/min laser scanning speed and 50% overlapping were used for these two depositions. In addition, one 15×10×3 mm pad of *H13 TS* specimen was prepared by depositing H13 tool steel directly on copper alloy substrate using the same set of process parameters but reduced laser power of 2 kW. This pad was used to compare the characteristics with the pad prepared using 2.5 kW laser power in order to correlate laser power, microstructure and microhardness of tool steel clad. These settings were selected based on the previous work shown in previous chapter on process parameters optimization to manufacture bimetallic dies by DMD.

<table>
<thead>
<tr>
<th>Element</th>
<th>Mn</th>
<th>Si</th>
<th>Mo</th>
<th>Ni</th>
<th>Cr</th>
<th>Fe</th>
</tr>
</thead>
<tbody>
<tr>
<td>Chemical composition</td>
<td>1%</td>
<td>2%</td>
<td>2%</td>
<td>12%</td>
<td>18%</td>
<td>Balance</td>
</tr>
</tbody>
</table>

### 5.2.2. Metallographic and Microhardness Analysis

All specimens were cut transversely and were ground and polished to 1 µm finish. A solution of FeCl₃ (20 gm), HCl (50 ml) and H₂O (100 ml) was used for etching the polished clads. Scanning electron microscope (SEM, Philips XL30FEG) equipped with energy dispersive X-ray spectroscopy (EDAX®) technique were used for all metallographic examinations of specimens. X-ray diffraction (XRD) was performed using Rigaku rotating anode X-ray diffractometer with Cu Kα radiation operating at 40 kV and 100 mA. Specimens were scanned in the standard θ-2θ range of 30-120° and data were collected at every 0.01° interval. The microhardness across the interface of the specimens was measured using CLARK Microhardness Tester (Model- CM 400AT) with 300 gm load and dwell time of 15 seconds.
5.3. Results and discussion

5.3.1. Microstructure

Clad/substrate bonding at the interface is a critical requirement for the applicability of both clad and substrate material. Figures 5-2 and 5-3 show the interface microstructures of all clads prepared for this investigation. Figure 5-2 shows defective clad/substrate interface produced from 2 kW laser power, which resulted in incomplete melting of the materials. The clad/substrate interface quality improved in the clad produced using 2.5 kW laser power (Figure 5-3a). This showed that 2.5 kW laser power was capable of melting the materials completely. Figure 5-3 (b) and (c) showed that the interfaces between substrate/316L SS and 316L SS/H13 tool steel were also free from any physical discontinuities. SEM microstructure revealed numerous voids in H13 tool steel clad produced with 2 kW laser power (Figure 5-4a). These voids particularly occurred along boundaries of two adjacent tracks, and were irregularly shaped. Kobryn et al. [220] described it as lack of fusion porosity, which is caused by insufficient/incomplete melting between adjacent tracks. A certain portion of this clad showed dendritic microstructure with the axis of the dendrites approximately parallel to the build direction of the deposit (Figure 5-4b). The growth morphology of the dendritic microstructures can be determined by their solidification conditions. The substrate acts as a heat sink in laser deposition process. Therefore, during the rapid melting and solidification in the laser deposition process, mainstream of the heat is always transferred from melted materials to bulk substrate. This directional solidification forms columnar dendritic microstructure parallel to the heat transfer direction. XRD spectrum revealed some intermetallic particles such as Chromium Iron Carbide (Cr$_2$Fe$_{14}$C), and solid solution matrix phases such as Molybdenum Iron (Mo$_{0.03}$Fe$_{0.97}$) and Chromium Iron (Cr$_{0.5}$Fe$_{0.5}$) in the clad (Figure 5-5).
Figure 5-2: SEM image showing interface between H13 tool steel clad and substrate deposited using 2 kW laser power
Figure 5-3: SEM images of different interfaces between (a) H13 tool steel and substrate deposited using 2.5 kW laser power (b) 316 L stainless steel and substrate (c) H13 tool steel and 316 L stainless steel
Figure 5-4: SEM microstructure of H13 tool steel clad produced using 2 kW laser power (a) voids resulted from incomplete melting (b) dendritic microstructure
Figure 5-5: XRD spectrum of H13 tool steel clad deposited using 2 kW laser power

Figure 5-6: SEM image showing microstructure of H13 tool steel clad deposited using 2.5 kW laser power
Figure 5-7: XRD spectrum of H13 tool steel clad deposited using 2.5 kW laser power

Unlike the metastable intermetallics formed in the clad layers produced using 2 kW laser power, clad formed using 2.5 kW laser power consists of stable δ ferrite (bcc) phase along with chromium carbide (Cr₇C₃) in the microstructure. Figure 5-6 shows the SEM microstructure of H13 tool steel clad deposited using 2.5 kW laser power. A lath martensitic microstructure could be seen in the clad [219]. The XRD spectrum of this clad is shown in Figure 5-7 which revealed stable δ ferrite (bcc) phase along with chromium carbide (Cr₇C₃) in the microstructure. Differences of energy intensity in the molten pools formed by two different laser powers created two different microstructures. Probably due to lower laser power in case of 2kW samples, the H13 tool steel powder was partially melted during delivery into molten metal pool. As a result of low energy intensity of the beam and high cooling or solidification rate of the melt pool, higher melting metallic constituents such as Cr, Mo etc. were trapped in iron to form different intermetallics. In contrast, 2.5 kW laser beam provided sufficiently high intensity of energy to the molten pool and also completely melted the H13 powder. The high intensity of energy of the molten pool provided sufficient time to form δ ferrite (bcc) through
primary crystallization during solidification. Primary crystallization was followed by a peritectic reaction between carbon-rich liquid and δ ferrite to form austenite. This reaction transformed some of the original δ ferrite into austenite leaving behind δ ferrite at the core of the dendrites. Subsequently, through the eutectoid reaction carbide was formed as $\text{M}_x\text{C}$ (where, $\text{M}$ is denoted as metallic constituent). Finally austenite transformed into martensite during further cooling [221, 222].

Figure 5-8 shows the SEM microstructure of tool steel clad where 316 L stainless steel was used as a buffer layer. The SEM micrograph revealed presence of fine and disoriented dendritic structure. The dendritic growth direction was not exactly aligned with the heat flow direction since dendritic growth was significantly controlled by a growth-crystallography consideration [106]. The SEM morphology of this clad did not resemble any of the usual microstructures attributed to martensite. It is thus believed that the long films present in the microstructure are austenite while fine dendrites are ferrite lathes. Figure 5-9 shows the XRD spectrum of this clad. XRD spectrum also revealed presence of austenite (fcc) phase along with ferrite (bcc) phase in the microstructure. Formation of austenite is due to underlying high nickel stainless steel layer. During laser cladding, nickel from the buffer layer dissolved in the tool steel clad to some extent and promoted the stability of austenite through locally retaining austenite. Thus the dilution of high nickel stainless steel proved to be very useful by stabilizing the formed austenite phase which is stronger and more stable at higher temperatures then ferrite.
Figure 5-8: SEM image showing microstructure of H13 tool steel clad deposited using 316L SS as buffer layer

Figure 5-9: XRD spectrum of H13 tool steel clad deposited using 316L SS as buffer layer
Although H13 tool steel is one of the most difficult alloys for deposition because of residual stress accumulation due to martensitic phase formation during cooling, no martensite was observed in the XRD spectrum of any of the clad in this investigation. Niu et al. [221] reported that tetragonality of martensite was difficult to be detected from the XRD pattern due to broadening of the diffracting peaks caused by a fine grain size and lattice strain. They however assumed the bcc phase in the rapidly solidified high speed steel to be either δ ferrite or martensite or a combination of both, depending on cooling rate. This may be the cause why martensites were not detected in this investigation. Due to rapid solidification during laser deposition, probably the grain size and lattice strain became very fine and hence martensite phase could not be detected by XRD. Moreover ferrite and martensite have very close lattice parameters which also make it difficult to resolve by XRD. This may be the reason why martensite phase was not detected in the XRD investigation. Therefore, the bcc phase in the tool steel clad is assumed to be either ferrite or martensite or a combination of both.

5.3.2. Microhardness

Figure 5-10 shows microhardness distribution on the cross-section from the H13 tool steel layer through the copper alloy substrate of the H13 TS specimen deposited using 2 kW laser power and revealed different steps corresponding to clad layer, heat-affected zone and substrate. The average hardness was between 220 and 250 HV in the tool steel clad, which reduced to around 100 Hv in the heat affected zone (HAZ) just below the interface between the clad and the substrate. Reduction of hardness in HAZ was expected since it is a common feature in laser cladding. Particularly Moldmax is susceptible to laser power and hardness drop. Hardness was also lower than expected in the tool steel clad. It was probably due to the partial melting of the deposited powder. The hardness was found to be in between 440 and 460 Hv in the tool steel clad deposited using 2.5 kW laser power (Figure 5-11). This increase in hardness can be attributed to the complete melting of
the deposited powder and formation of ferrite-martensite microstructure in the clad. The heat affected zone depth was wider in this clad which can be attributed to more power applied on the substrate since HAZ size is a function of the amount of heat generated at the clad/substrate interface, which is further dependent on the processing parameters [219]. The microhardness of tool steel clad when deposited using 316L SS was similar to the microhardness of the directly deposited tool steel clad (Figure 5-12). The microhardness of 316L SS specimen was found to be 250 HV. Nickel content in stainless steel is the cause of increased ductility while sacrificing hardness. High nickel content in 316L SS specimen is thus believed to be the cause of hardness reduction and to increase the ductility of the buffer layer.

Figure 5-10: Vickers microhardness of clad deposited using 2 kW laser power
Dilution of elements into the clad from the underlying substrate is a common characteristic of laser metal deposition. Dilution takes place since the laser beam melts both substrate and clad material together and allows mixing to various
extents. In this investigation, energy dispersive spectroscopy (EDS) was performed to quantify the composition of different elements in the clad. Line scans from the substrate into the clad showed the variation in composition of various elements and their composition along the line. Figure 5-13 shows the EDS line spectra of H13 tool steel clad when deposited directly on copper alloy substrate. From the spectra it is evident that copper diffused in the tool steel clad up to certain depth. Considering copper dilution area in the spectra, it is believed that the dilution of the previous layer into the next layer was minimal. Dilution of copper in tool steel can be explained by the work of Pogson et al. [49] in which they showed incorporation of copper into the tool steel during laser processing leading to the formation of a copper-rich region around the prior austenite grain boundaries. During DMD, steel powder is injected into the molten copper pool from which the partially melted powder takes latent heat and melts down. The variant monotectic reaction \((L_1 \rightarrow L_2 + \gamma)\), between tool steel and copper forms copper rich low carbon liquid \((L_2)\) and austenite \((\gamma)\) from iron rich high carbon liquid \((L_1)\) during solidification [223]. On further cooling, \(L_2\) becomes more rich in copper together with the formation of austenite \((\gamma)\) also with significant copper content. Due to the immiscibility of copper with iron, on further cooling, copper is rejected from austenite \((\gamma)\) and leads the formation of copper rich region around austenite grain boundaries. Finally, the liquid phase \(L_2\) solidifies as copper dominant phase along with considerable iron component [224]. Thus, the first phase to freeze out is iron before solidification of copper rich phase and allows diffusion of copper into the tool steel layer.
Figure 5-13: EDS line scan showing distribution of elements in the clad and substrate when H13 tool steel was directly deposited on copper alloy substrate.

In contrast, when 316L SS was used as a buffer layer the dilution of copper into the clad could not be traced (Figure 5-14). Nickel is a well known austenite stabilizer and can retain austenite by reducing the solidification temperature of austenite in stainless steel. During solidification of 316L SS layer and copper molten pool, high nickel content in the stainless steel retained austenite and allowed the copper phase to be solidified first. Thus, the last phase solidified in the layer was austenite and left copper phase at the bottom of the layer to minimize dilution of copper in the clad. Figure 5-14 also revealed that nickel was diluted in H13 tool steel matrix to some extent. This nickel content is believed to be the cause of formation of austenite phase in the H13 tool steel clad.
Microstructure plays an important role in the properties of materials. Since the behaviour of the tool steel clad on copper alloy substrate also depends largely on the texture and microstructure formation, the microstructure of tool steel clad was characterized. From the investigation, it was found that delicate control of the laser power was critical in obtaining desired material properties since hardness of deposited H13 tool steel varied considerably with the increase of the laser power from 2 kW to 2.5 kW. The microstructure of directly deposited H13 tool steel on copper alloy substrate was also found to be slightly different from that of H13 tool steel clad deposited using a 316L stainless steel buffer layer. The microhardness of H13 tool steel clad was identical in both deposits. Though the parts exhibited full interlayer bonding with virtually no porosity, copper was found to be diluted in tool steel when deposited directly on substrate. In contrast, no trace of copper dilution was detected in tool steel clad when deposited using a 316L stainless steel buffer layer. Therefore, use of a high nickel stainless steel buffer layer was
recommended in between tool steel clad and copper alloy substrate for better performance of the clad in high pressure and temperature applications.
CHAPTER 6

EVALUATION OF MECHANICAL PROPERTIES

6.1. Introduction:

Unlike porous microstructure, poor adhesion and low fatigue resistance of other coatings, DMD layers are metallurgically bonded to the substrate material with minimal dilution. DMD coatings also cause minimal degradation of the mechanical properties and reduced thermal-stress induced distortion, cracking and delamination of the processed materials compared to other welding-based methods [147]. Both similar and dissimilar materials can be deposited using DMD which holds major advantage over any other joining method for industrial applications. However in case of deposition of tool steel on copper alloy substrate, iron and copper are immiscible in liquid state and form several intermetalics when solidified, while nickel and copper form complete solid solutions at all compositions. Therefore, a buffer layer of high nickel steel such as, 316 L stainless steel (SS) can be used as a bonding agent between copper and tool steel.

The closed loop DMD is capable of combining optimal powder and energy through feedback system to produce effective cladding thus it is more useful for bonding between dissimilar substrate and cladding materials. The wear properties and characteristics of Co-Ni duplex laser cladding on copper have been reported [225, 226]. Bond strength of a laser-clad iron-base alloy coating on Al-Si alloy substrate and its fracture behavior and the abrasive wear behavior of laser-clad tool steel coatings has also been reported [227, 228]. But the metallurgical characteristics and mechanical properties of the DMD deposited tool steel on copper alloy substrate are not found in the open literature.

This chapter presents a detail investigation of the mechanical behaviour of the DMD deposited tool steel on copper alloy substrate material. The primary focus
was on interface behaviour as interfaces are critical in cladding of two dissimilar materials. Thus H13 tool steel was deposited both directly and with 316 L SS as buffer layer on the copper alloy substrate. Fractures and pull offs are two common characteristics of the HPDC tools [229]. Therefore bond strength and toughness of the bimetallic structure were determined and characteristics of fracture surfaces were analysed.

6.2. Experimental Procedures

6.2.1. Materials and DMD Cladding

POM DMD 505 with 5 kW CO₂ laser was used for closed loop experimental cladding. Due to excellent mechanical behaviour and extensive use in HPDC, H13 tool steel powder was used as the cladding material on copper alloy substrate. Chemical compositions of copper alloy substrate, 316 L SS and H13 tool steel powder have been listed in previous chapters. Two sets of test sample pads were prepared for both tensile and Charpy impact energy test, to compare the characteristics of directly deposited tool steel and steel deposited with 316 L SS buffer layer, were performed. For the tensile test, two pads of 20×20×25 mm were deposited on copper alloy substrate; one pad was of H13 tool steel directly deposited on substrate whereas another one was with 1 mm 316 L SS buffer layer deposited immediately on substrate. Moreover, to compare the strength of the substrate material, one set of tensile test specimens was prepared with the copper alloy material. For the Charpy impact energy test pads, similar deposition paths as tensile test blocks were followed, except the pad dimensions were 60×40×6 mm. A 2.5 kW laser power with 1 mm laser beam diameter, 250 mm/min laser scanning speed and 50% overlapping were used for all depositions. These settings were selected on the basis of previous work on parametric investigation of DMD parameters to manufacture bimetallic dies.
6.2.2. Metallographic analysis

All specimens were cut transversely and were ground and polished to 1 µm finish. A solution of FeCl₃ (20 gm), HCl (50 ml) and H₂O (100 ml) was used for etching the polished deposits. Scanning electron microscope (SEM, Philips XL30FEG) with energy dispersive X-ray spectroscopy (EDAX®) technique were used for all of the metallographic examinations of specimens.

6.2.3. Mechanical Testing

In order to determine the bond strength for both cases, five tensile test specimens from each 20×20×25 mm DMD pad were prepared following ASTM E8/E8M standard. As it was a comparative study and for manufacturing advantage, sub size specimens (dimensions proportionately reduced) were prepared for tensile tests. The drawings of the tensile test specimens are given in Figure 6-1. Two sets of three Charpy impact energy test specimens were prepared from remaining two 60x40x6 mm pads following ASTM standard E-23. In the HPDC dies the cracks initiate from the cavity surface and propagate inside the dies. Considering the nature of initiation and propagation of cracks in the HPDC tools, V-notch was cut on the H13 tool steel side of the specimen. The schematic drawing of the Charpy impact energy test specimens is shown in Figure 6-2. Tensile tests were carried out using 10kN universal static testing system, type Z010 manufactured by Zwick/Roell. The Charpy impact energy test was conducted using Sonntag Universal Impact Machine (Model SI-1) with a maximum capacity 325.44 J (240 ft-lb). All tests were conducted at room temperature.
Figure 6-1: Schematic drawings of the tensile test specimens prepared from (a) H13 tool steel directly cladded on substrate (b) H13 tool steel cladded with 316 L stainless steel as buffer layer (c) copper alloy substrate
6.3. Results

SEM micrographs (Figure 6-3) of the specimens showed that both H13 tool steel and 316 L stainless steel were successfully deposited on copper alloy substrate. Sound deposits with crack and pore free interfaces were observed in both the deposits. Both the interfaces show the characteristic of sharp transition (distinct layer boundaries) between cladding and substrate materials.

Tensile test observations showed that fractures occurred in the mixed layer (immediate layer on top of the substrate), between substrate and clad, which indicates that the tensile testing was effectively used to measure the bond strength between the clad and the substrate [230]. The tensile test behaviour of the specimens is shown in Figure 6-4. These results represent the behaviour of the specimens produced according to the drawings provided in Figure 6-1. The ultimate tensile strength of the copper alloy substrate was measured as 1170 MPa.
The average bond strength of the interface between H13 tool steel and copper alloy substrate was measured as 673 MPa whereas the bond strength reduced to 579 MPa when 316 L stainless steel was used as buffer layer. Table 6-1 shows the bond strength of different clads with copper alloy substrate. The characteristic of necking was observed in both tensile test specimens as shown in Figure 6-5. The necking behaviour was prominent in the substrate part. However, no necking was observed in copper alloy substrate specimen.
Figure 6-3: SEM images of different interfaces between (a) H13 tool steel and substrate (b) 316 L stainless steel and substrate (c) H13 tool steel and 316 L stainless steel

Figure 6-4: Tensile test performance of different clads with substrate material (half part of the specimens was clad and the other half was substrate)
Table 6-1: Bond strength of different clads with substrate

<table>
<thead>
<tr>
<th>Type of specimen</th>
<th>Bond strength (MPa)</th>
</tr>
</thead>
<tbody>
<tr>
<td>H13 tool steel directly clad on copper alloy substrate</td>
<td>673</td>
</tr>
<tr>
<td>H13 tool steel clad with 316L SS buffer layer on copper alloy substrate</td>
<td>579</td>
</tr>
</tbody>
</table>

(a) Clad and Substrate

(b) Clad and Substrate
Figure 6-5: Necking characteristic of different specimens after tensile test (a) 316 L buffered specimen seen from top (b) 316 L buffered specimen seen from side (c) directly cladded H13 specimen seen from top (d) directly cladded H13 specimen seen from side (e) copper alloy specimen seen from top (f) copper alloy specimen seen from side
Figure 6-6 shows the mean Charpy impact energy of two different clad materials. There was not much variation in the Charpy impact energy between two different types of specimens tested. In the test specimen where H13 tool steel was directly deposited on the copper alloy substrate, the Charpy impact energy was measured as 67.12 J. When H13 tool steel was deposited using buffer layer of 316 L stainless steel on the substrate, the Charpy impact energy was measured as 67.8 J. In all the test specimens the cracks initiated in the V-notch made in the clad and propagated through the substrate. As none of the cracks propagated through the interface between the clad and the substrate, it could be concluded that the clad-substrate has good bond strength.

![Graph showing Charpy impact energy of two different cladding specimens](image)

**Figure 6-6: Charpy impact energy of two different cladding specimens**

After both the tensile and Charpy impact energy tests were performed, fracture surfaces were characterized using SEM, for morphology and compositional variations across the interfaces. Figure 6-7 shows the fracture surfaces of the tensile test specimens, which appear to be predominantly ductile. EDAX® analysis in Figure 6-8 shows the composition of copper, iron, chromium, nickel in directly deposited H13 tool steel and H13 tool steel deposited with 316 L SS specimen
respectively. More detailed discussion of all the results of this section has been given in the next section (Section 6.4).
Figure 6-7: Fracture morphology of the tensile test specimens (a) H13 tool steel directly cladded on substrate (b) Cu alloy (c) H13 tool steel cladded with 316 L stainless steel as buffer layer
Figure 6-8: EDAX analysis of fracture surfaces of tensile test specimens (a) H13 tool steel directly cladded on substrate (b) H13 tool steel cladded with 316 L stainless steel as buffer layer

Figure 6-9 shows the morphology of the fracture surfaces of directly deposited H13 tool steel Charpy impact energy specimens. Intergranular fracture by quasi-cleavage along with dimple fractures is observed at the top layer where the crack initiated (Figure 6-9a) [231]. The fracture appears to be absolutely ductile near the interface (Figure 6-9b). Just below the interface, the fracture surface shows elongated grains in the heat affected zone (HAZ) of the substrate (Figure 6-9c). The fracture surface of the substrate also shows dimple morphology or ductile fracture. Figure 6-10 shows the SEM micrographs of the Charpy impact energy test specimen with 316 L SS as buffer layer. The H13 tool steel clad region shows cleavage fracture along with ductile fractures (Figure 6-10a) and the 316 L SS region of the deposit shows absolutely ductile fracture (Figure 6-10b). However the fracture surfaces of the substrate including HAZs of these specimens show no difference between the substrate and HAZ of the specimens where H13 tool steel was directly deposited on substrate.
Figure 6-9: Fracture surfaces at different regions of ‘directly cladded H13 tool steel’ charpy impact energy test specimen (a) top layer where crack initiated (b) 1st layer just above the interface (c) HAZ (d) substrate
Figure 6-10: Fracture surfaces at different regions of ‘H13 tool steel cladded with 316 L stainless steel as buffer layer’ charpy impact energy test specimen (a) top layer where crack initiated (b) 316 L stainless steel layer just above the interface (c) HAZ (d) substrate
The fracture toughness of these two types of specimens were measured using the following equation- [232]

\[
K_{IC} = 0.804 \times \sigma_{ys} \left[ \frac{CVN}{\sigma_{ys}} - 0.0098 \right]^{0.5} \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots \ldots (6.1)
\]

Where,

- \( K_{IC} = \) Fracture toughness (\( MPa.m^{1/2} \)) at room temperature
- \( \sigma_{ys} = \) Tensile yield strength (\( MPa \))
- \( CVN = \) Charpy V Notch Impact Energy (\( Joules \))

The fracture toughness value was calculated as 162.3 \( MPa.m^{1/2} \) for the directly deposited H13 tool steel specimens and 152.5 \( MPa.m^{1/2} \) for the specimens of H13 tool steel deposited with 316 L SS as buffer layer. These values have been summarised in the Table 6-2.

### Table 6-2: Fracture toughness of different clad and copper alloy substrate combinations

<table>
<thead>
<tr>
<th>Type of specimen</th>
<th>H13 tool steel directly clad on copper alloy substrate</th>
<th>H13 tool steel clad with 316 L SS buffer layer on copper alloy substrate</th>
</tr>
</thead>
<tbody>
<tr>
<td>Fracture toughness ( (MPa.m^{1/2}) )</td>
<td>162.3</td>
<td>152.5</td>
</tr>
</tbody>
</table>

### 6.4. Discussion of Results

The first experiment considered examination of the bond strength between the clad and substrate materials. The achieved bond strength is much lower compared to the tensile strength of copper alloy substrate material. The EDAX® analysis of the fracture surfaces shows mostly copper and some iron (see Figure 6-8). Since both iron and copper are present in the fracture surface, the EDAX® observation confirms that the failure of the specimen took place in the mixed layer just above the transition region between the clad and the substrate. This can be explained by
the work of Pogson et al. [49] in which they showed incorporation of copper into the tool steel during laser processing leading to the formation of a copper-rich region around the prior austenite grain boundaries that significantly weaken the material. During DMD steel powder was injected into the molten copper pool from which the powder takes up the latent heat of melting. During solidification, the variant monotectic reaction (L₁→L₂ + γ) between these two liquids forms copper rich low carbon liquid (L₂) and austenite (γ) from iron rich high carbon liquid (L₁) [223]. On further cooling L₂ becomes more rich in copper together with the formation of austenite (γ) also with significant copper content. Due to the immiscibility of copper with iron, on further cooling copper is rejected from austenite (γ). Therefore formation of copper rich region around austenite grain boundaries takes place which deteriorates the strength of the clad. However the liquid phase L₂ finally solidifies as copper dominant phase along with considerable iron component [224]. Thus formation of austenite (γ) before solidification of copper rich phase allows formation of copper rich region at the top of the clad layer. When subsequent layers are deposited, copper gets incorporated in these succeeding layers from the preceding copper rich layers. Though the copper content decreases gradually in the succeeding layers, presence of copper was significant up to certain height and has been shown in previous chapter. These infiltrated copper behaves as weak regions when the deposits are subjected to mechanical stresses. As a result, in the tensile test the failure took place at the copper rich regions. Due to the high solubility with copper, both nickel and nickel rich irons have been widely used in the laser processing of copper and steel [225, 233, 234]. But interestingly in this experiment, use of high nickel stainless steel as buffer layer reduced the bond strength in the tensile test compared to directly deposited H13 tool steel specimen as shown in Figure 6-4 (the fracture surfaces of the tensile specimens). Nevertheless, the results of bond strength obtained for steel/copper have confirmed that the proposed bimetallic die has sufficiently
higher bond strength compared to other coating process such as thermal spray [121].

The fine distribution of voids or dimples in Figure 6-7 depicts that the tensile test specimens failed by a fibrous mode of fracture though the size of the dimples varied considerably among the specimens. The fracture surface topography of copper alloy specimen is fine dimples, which indicate the ductile failure behaviour. The void or dimple mechanism is also observed in other two types of specimens but the dimple size is coarser in the specimen where H13 tool steel was directly clad on copper alloy compared to specimen that was clad with 316 L SS as buffer layer. Presence of necking in the specimens also depicts the ductile behaviour of the tensile specimens (see Figure 6-5). However in both directly clad H13 and H13 clad using buffer layer of 316 L SS, necking deformation was prominent in the substrate part of the specimens compared to the clad part. The deformation was greater than that of the specimen prepared from the substrate material only. This behaviour can be attributed to the heat affected zone (HAZ) of the substrate material. During DMD process copper alloy substrate was heated below the first layer (mixed layer) which significantly softened the substrate material.

In this study, the Charpy impact energy was found to be 67.12 J in the directly clad H13 tool steel specimen that was comparable to the 316 L SS buffered specimen. Both cleavage and dimple ruptures were observed in the fracture surfaces at the top layer where cracks initiated (Figure 6-9a and 6-10a). The surfaces turned ductile in nature showing only dimple ruptures in the clad near the interface as shown in Figure 6-9b and 6-10b. Tear ridge around the dimples was found in the 316 L stainless steel buffered layer (Figure 6-10b). This can be attributed to the high Nickel content which retained austenite locally and increased ductility [14]. In HAZ, tearing topography was observed with elongated grains. During DMD process significant softening in the HAZ occurred that may cause increase of the ductility. As a result, when the HAZ material was subjected to tensile stress during
Charpy test, the grains elongated and appeared as torn rather than fractured (Figure 6-9c and 6-10c). The copper alloy substrate showed ductile dimple fractures.

In this research, another intention was to evaluate the strength of the interface between clad and substrate in the V-notch Charpy impact energy test condition. In the impact condition, crack initiated at the V-notch and propagated through the weak zone. In this test no crack propagated along the interface which shows good bond strength at the interface.

The fracture toughness value was greater for the directly clad H13 tool steel specimens compared to the specimens of H13 tool steel clad with 316 L SS as buffer layer. Though Ni works as an austenite stabilizer in the liquid pool and austenite maintains ductility of the clad, in this research, fracture toughness of the specimens was found to reduce when 316 L stainless steel was used as a buffer layer. Clearly it is due to the low strength in the tensile test which suggests that 316 L SS when clad on copper alloy substrate reduces the strength of the bonding.

6.5. Summary

The interface between copper alloy and tool steel clad was critical considering the high pressure and temperature nature of HPDC process. Therefore, nature of bonding between these materials was evaluated based on various mechanical properties and following conclusions were illustrated-

- The interfaces were uniform and showed distinct transition between substrate and deposits.
- The bond strength of the deposits and substrate were lower compared to the ultimate tensile strength of the substrate material. The bond strength reduced further when 316 L SS was used as a buffer layer between H13 tool steel and copper alloy substrate.
- The tensile fracture surfaces showed dimple ductile fracture morphology.
• The impact toughness was found slightly higher when 316 L SS was used as buffer layer compared to directly deposited H13 tool steel on copper alloy substrate.

• Both brittle and ductile fracture morphology was present in the upper cladding region, but in the lower region near the interface only dimple ductile fracture was observed.

• Though using 316 L SS as buffer layer reduced the fracture toughness compared to directly clad H13 tool steel, it was observed in the ductile region for both the cases. The reduction in the fracture toughness with 316 L SS buffer layer was mainly due to the lowering of bond strength observed in the tensile test.
CHAPTER 7

EVALUATION OF THERMAL FATIGUE

7.1. Introduction

Although DMD deposited H13 tool steel on copper alloy substrate can be applied to high pressure die cast tooling application, the scientific and technical aspects, when applied in the high temperature and pressure environment, have not been well studied and understood. Researchers have shown that in HPDC, the primary in-service tool failure modes are attributed to the damage mechanisms such as (1) heat-checking or thermal fatigue (2) gross cracking (3) corrosion and soldering and (4) erosion due to melt flow [82, 83, 235]. Heat checking or thermal fatigue is one of the most critical failure mechanisms in the HPDC application. The rapid alteration of surface temperature induces high stresses on the surface, which cause plastic strain thus resulting in network of fine thermal fatigue cracks. Usually thermal fatigue behavior of the HPDC tooling depends mainly on the thermal expansion coefficient and high temperature strength of the die materials. Due to very different thermal expansion coefficient of copper alloy \((17.2 \times 10^{-6} \, \text{C}^{-1})\) and H13 tool steel \((12.6 \times 10^{-6} \, \text{C}^{-1})\), heat checking or thermal fatigue behavior is very critical for bi-metallic tooling particularly in the high temperature application. Moreover iron and copper are immiscible in the liquid state and form several intermetallics when solidified, which can further degrade the interface quality. On other hand, nickel and copper form complete solid solutions at all compositions. Therefore a buffer layer of high nickel steel such as; 316 L stainless steel (SS) can be used as a bond layer between copper and tool steel.

Although the thermal fatigue behaviour of a copper based bi-metallic tooling demands much more attention, no information could be found in the open literature on this particular area. Therefore the main objective of this chapter is to
investigate the thermal fatigue behaviour using an induction heating thermal fatigue test rig and correlate the microstructure of the tool steel clad with fatigue behavior. The crack initiation and propagation behavior has been discussed and a comparative study of the thermal fatigue characteristics between H13 tool steel clad directly deposited on copper alloy and H13 tool steel clad with a 316 L SS buffer layer has also been investigated in terms of microstructure and stress-strain analysis. This investigation provides relative insights into the thermal fatigue behaviour of a copper based bi-metallic tooling to facilitate and improve its use in the HPDC application. It also provides necessary information on whether a buffer layer of 316 L SS is needed during DMD of H13 tool steel on a copper alloy substrate for improved thermal cycling performance.

7.2. Experimental Details

7.2.1. Preparation of Test Specimens

Figure 7-1 shows the schematic drawings of 70 mm long and 35 mm diameter test specimens used in the thermal fatigue experiment. One 10 mm diameter and 25 mm deep axial hole was drilled in each specimen to connect with the pneumatic actuator of the thermal fatigue test rig. Initially, two cylindrical cores each 31 mm diameter and 70 mm long were prepared from copper alloy. Two different types of coatings were applied on these core surfaces using POM DMD 505 with 5 kW CO\textsubscript{2} laser. In the first specimen 2.2 mm thick H13 tool steel coating was deposited directly on the copper alloy cylindrical core and this specimen is designated as H13 TS specimen throughout the following text. On the other core, 0.6 mm 316 stainless steel buffer layer was deposited on the core surface, followed by 1.6 mm H13 thick tool steel layer and this specimen is designated as 316 L SS buffer specimen throughout the following text. The chemical compositions of the powders and metals used in the experiment are given in chapters 4 & 5. After DMD both specimens were machined in order to remove the surface roughness resulting from the DMD process and the final machined coating thickness was 2 mm.
Figure 7-1: Schematic drawing of the test specimen used in the experiment (a) H13 tool steel directly clad on copper alloy core material (b) H13 tool steel clad with 316 L SS as buffer layer on the copper alloy core material
7.2.2. Thermal Fatigue Testing

A fully automated and innovative thermal fatigue test rig was employed in the experiment that enabled cyclic HF induction heating and cooling of the thermal fatigue specimens according to a pre-programmed heating and cooling conditions simulating the conditions of an aluminum HPDC cycle in production [229]. Figure 7-2 shows the schematic of the thermal fatigue test apparatus. The test rig consisted of a 30 kW induction furnace and programmable logic controller (PLC) which was capable of applying an identical net energy input to all specimens during each heating cycle. The specimens were heated for 13 seconds and immersed into the
cooling bath for 10 seconds to produce high temperature gradient between outer and inner surface. The thermal cycles were repeated until 5000 number of cycles was reached. This number of cycles was equivalent to several thousand of industrial HPDC cycle that can produce significant fatigue cracks in dies. Unlike other thermal fatigue test rigs described in the literature, which apply energy to raise the surface temperature to a certain value regardless of the thermal conductivity of the testing material [236-238], this test rig used an identical computer controlled energy in every cycle. For producing cyclic heating and cooling effect, a pneumatic actuator was used to place the specimens alternately into the induction coil for heating and immersing into lubricant bath for cooling. A computer data acquisition system through a thermocouple recorded the energy input and surface temperature in the heating and cooling cycles throughout the experiment. The input energy applied and corresponding surface temperature developed in a thermal cycle are given in Figure 7-3 and 7-4 respectively.

Figure 7-3: Input energy from the induction coil applied in the thermal cycling
7.2.3. Evaluation of Test Specimens

Performance of the thermal fatigue specimens, the crack initiation and propagation for each coating was characterised using optical microscopy (Nikon™) and scanning electron microscopy (SEM) equipped with EDAX® (Philips™, XL30 FEG scanning electron microscope). X-ray diffraction (XRD) was performed using Rigaku rotating anode X-ray diffractometer with Cu Kα radiation operating at 40 kV and 100 mA. Specimens were scanned in the standard θ-2θ range of 30-120° and data were collected at every 0.01° interval. Image analysis was done using ImageJ® software. In a specific interval, the surface was polished and observed in the optical microscope to investigate the crack density, mean and maximum crack length. The H13 TS specimen was removed from the thermal fatigue test after 3000 cycles due to severe surface degradation caused from numerous long cracks that initiated and propagated on the entire circumference where as the 316 L SS buffer specimen survived 5000 cycles. Each of the specimens was sectioned and polished at the end of the experiment to inspect the crack depth and the condition of the interface between core and coatings. The optical micrographs of the cross sectional
surfaces are shown in Figure 7-5. The first layer of both specimens was sectioned longitudinally for precision microstructural analysis.

![Figure 7-5: Optical micrographs showing the cross section of (a) H13 TS and (b) 316 L SS buffer specimen](image)

**7.3. Results**

Crack number, mean crack length, maximum crack length and crack density (percentage of cracks for a given area) measurements were recorded every 1000 thermal fatigue cycles from the image analysis results. During induction heating, heat was concentrated mainly in the middle part of the specimen. So, only the cracks in the middle of the specimens were taken into account. Figure 7-6 shows the crack density versus thermal fatigue cycle of both specimens. In the thermal fatigue testing with subsequent heating and cooling, both longitudinal and
circumferential cracks initiated. As the number of cycles increased, both types of cracks propagated and got connected with each other to develop numerous continuous fine crack networks known as thermal fatigue cracks. In the H13 TS specimen thermal fatigue cracks were observed within 1000 cycles though the crack density was only around 2%. The crack density increased to approximately 4.5% at 3000 thermal fatigue cycles when the specimen was finally removed from the test rig. On the other hand, thermal fatigue cracks were only detected at 2000 cycles in the 316 L SS buffer specimen. Since the samples were checked only visually during thermal fatigue cycling but thoroughly under microscope every 1000 cycles, the cracks might have initiated few hundred cycles earlier. The crack density was approximately 1.3% which was much lower compared to the crack density of the H13 TS specimen. After 2000 thermal fatigue cycles, crack density increased rapidly and was approximately 10% after 5000 cycles.

Figure 7-6: Comparison of the crack density with number of thermal cycles on the surfaces of two different clads

The crack evaluations from optical micrographs revealed large longitudinal cracks on the surface of both test specimens (as shown in Figure 7-7). These cracks severely deteriorated the clad surfaces and have been designated as catastrophic
cracks. Table 7-1 summarises a comparative observation of the number of catastrophic cracks (N), mean crack length (CL) and the maximum crack length (CL\textsubscript{max}) detected in each 1000 cycles during the thermal fatigue cycling of both specimens. The catastrophic cracks were dominant in the H13 TS specimen since a number of prominent cracks appeared on the surface within 1000 cycles. These cracks only propagated along the longitudinal direction on the surface as the number of cycles increased. Within 3000 cycles the longest crack propagated all through the specimen and the specimen was removed from further thermal fatigue cycling. These catastrophic cracks not only propagated rapidly but also increased in number with the increased number of thermal fatigue cycles. In contrast, the 316 L SS buffer specimen provided better resistance to the catastrophic cracks since only one catastrophic crack was detected after 2000 cycles. The length of the crack was also smaller than the largest crack of the H13 TS specimen. Though the number of cracks did not increase, this crack preferentially propagated along the longitudinal direction with subsequent thermal cycling. However the propagation rate was slower compared to the propagation rate of cracks in the H13 TS specimen as the crack did not propagate all through the specimen even after finishing 5000 cycles.

Table 7-1: Number of cracks (N), mean crack length (CL) and maximum crack length (CL\textsubscript{max}) of the thermal fatigue test specimens

<table>
<thead>
<tr>
<th>Number of cycles</th>
<th>Type of specimen</th>
<th>N</th>
<th>CL (mm)</th>
<th>CL\textsubscript{max} (mm)</th>
</tr>
</thead>
<tbody>
<tr>
<td>1000</td>
<td>H13 TS</td>
<td>6</td>
<td>20</td>
<td>30</td>
</tr>
<tr>
<td></td>
<td>316 SS</td>
<td>-</td>
<td>-</td>
<td>-</td>
</tr>
<tr>
<td>2000</td>
<td>H13 TS</td>
<td>9</td>
<td>29</td>
<td>40</td>
</tr>
<tr>
<td></td>
<td>316 SS</td>
<td>1</td>
<td>18</td>
<td>18</td>
</tr>
<tr>
<td>3000</td>
<td>H13 TS</td>
<td>10</td>
<td>31</td>
<td>61</td>
</tr>
<tr>
<td></td>
<td>316 SS</td>
<td>1</td>
<td>23</td>
<td>23</td>
</tr>
<tr>
<td>4000</td>
<td>H13 TS</td>
<td>-</td>
<td>-</td>
<td>-</td>
</tr>
<tr>
<td></td>
<td>316 SS</td>
<td>1</td>
<td>31</td>
<td>31</td>
</tr>
<tr>
<td>5000</td>
<td>H13 TS</td>
<td>-</td>
<td>-</td>
<td>-</td>
</tr>
<tr>
<td></td>
<td>316 SS</td>
<td>1</td>
<td>42</td>
<td>42</td>
</tr>
</tbody>
</table>
Figure 7-7: Optical image showing large catastrophic cracks on the surfaces of H13 TS specimen (left side) and 316 SS buffer specimen (right side)
Figure 7-8: SEM images showing the propagation of fine and shallow thermal fatigue cracks in the clad of (a) H13 TS specimen and (b) 316 L SS buffer specimen

Figure 7-9: Optical image showing pull off feature on the surface of H13 TS specimen

After finishing 3000 and 5000 thermal fatigue cycles respectively for H13 TS and 316 L SS buffer specimen, the cross-section of both specimens was examined. The cross-sections showed the propagation depth of different cracks in the clad layers. Shallow and fine thermal fatigue cracks were detected in the clad of both
specimens as shown in Figure 7-8. The SEM micrographs revealed that most of the thermal fatigue cracks of the H13 TS specimen appeared perpendicular to the surface (Figure 7-8 a). Few cracks changed propagation direction from perpendicular to horizontal inside the clad to be connected with adjacent perpendicular crack and could be the cause for the pull off that was observed in the H13 TS specimen (Figure 7-9). In the 316 L SS buffer specimen, all the thermal fatigue cracks propagated in the direction perpendicular to the surface (Figure 7-8 b). The maximum crack depth was approximately 50 µm after 5000 cycles in the 316 L SS buffer specimen, which was much smaller than 100 µm deep cracks detected in the H13 TS specimen after 3000 cycles. This suggested that the propagation rate was slower in 316 L SS buffer specimen.

Figure 7-10: SEM image showing (a) cracks in the 1st layer of H13 TS specimen and (b) crack free 1st layer of 316 L SS buffer specimen (cross sectional view)
Figure 7-11: SEM images showing (a) cracks in the 1st layer of H13 TS specimen and (b) crack free 1st layer of 316 L SS buffer specimen (longitudinal section)

The first layer of the H13 TS specimen experienced severe crack generation within 3000 thermal fatigue cycles, while no crack was detected in the 316 L SS buffer
layer even after 5000 thermal fatigue cycles (Figure 7-10). Most of these cracks followed perpendicular propagation direction with few exceptions that followed horizontal propagation direction. The longitudinal section of the 1st layer also detected networks of cracks in the H13 TS specimen, while no crack was observed in the 316 L SS buffer layer (shown in Figure 7-11). The propagation direction of the catastrophic cracks was also perpendicular to the surface in both specimens (Figure 7-12). However the propagation depths of the cracks were different in these specimens. The crack propagated more in the 316 L SS buffer specimen compared to the H13 TS specimen. Once the crack initiated, it propagated inside the clad with increasing cycle number. Since the H13 TS specimen was removed after 3000 thermal fatigue cycles, crack propagation could not be monitored further and compared to that in the 316 L SS buffer specimen. None of the specimens showed crack propagation along the interface between the clad and the core.
7.4. Discussion of Results

During cyclic heating and cooling, the temperature of both the core and coating materials varies periodically. In the heating phase, the surface temperature of the coating becomes higher than that of the inner core and vice versa when the specimen is cooled in the cooling bath. This temperature gradient between the outer surface and the inner core develops thermal stresses in the specimen. When the specimen outer surface temperature is higher than the core temperature, the surface expansion is constrained by the cooler core and undergoes a compressive stress. On the other hand, during cooling, the temperature of the outer surface
drops rapidly and the contraction of the cooler outer surface is constrained by the inner core generating a tensile stress on the outer surface. In general, if any of these stresses exceeds the material’s yield stress, plastic deformation occurs in the material. The basic assumption is that the crack generates when the specimen can no longer absorb the plastic deformation due to the exhausted material ductility during the cyclic compressive and tensile stress condition. In order to gain new knowledge about the mechanisms that causes the thermal fatigue cracking, calculation of thermally induced stress and strain has recently been performed [229, 236-239]. Based on the thermal cycling induced stress analysis, all of these investigations showed that the material underwent a compressive plastic strain and developed a tensile stress in the stabilized thermal cycles and caused crack initiation. The thermal fatigue cracks observed on the surfaces of the specimens in this investigation are the result of these thermally induced cyclic stress and strain.

Figure 7-13: SEM image showing Carbon network in the cracks of 1st layer of H13 TS specimen (longitudinal section)
Figure 7-14: (a) SEM morphology and (b) distribution of Fe and Cu elements along the cross-section of 316 L SS specimen
Figure 7-15: (a) SEM morphology and (b) distribution of Fe and Cu elements along the cross-section of H13 TS specimen

Table 7-2: EDAX analysis on the black network confirming presence of C

<table>
<thead>
<tr>
<th>Elements</th>
<th>C</th>
<th>Fe</th>
<th>Cl</th>
<th>Cu</th>
<th>Si</th>
</tr>
</thead>
<tbody>
<tr>
<td>Wt%</td>
<td>83.65</td>
<td>9.76</td>
<td>3.2</td>
<td>2.56</td>
<td>0.62</td>
</tr>
</tbody>
</table>
Another objective of this investigation was to investigate the behavior of the interface between the clad and the core material under cyclic heating and cooling condition. Though interface remained unchanged in both specimens, the first layer of the H13 TS specimen was damaged badly showing numerous cracks just above the interface. The reason for these cracks is the formation of a carbon network in the first layer of H13 TS clad (Figure 7-13) during direct metal deposition on the copper alloy core. The EDAX analysis on this network is summarized in Table 7-2. Another reason for the cracks in the first layer of H13 TS specimen is the likely higher stress developed due to the dilution of substrate material in the clad. Figure 7-14 and Figure 7-15 show the morphology and dilution of copper content in 316 L SS and H13 TS specimens respectively. A comparison of these two figures showed that the dilution of copper content in the 316 L SS specimen was much less compared to the H13 TS specimen. During laser metal deposition, copper alloy substrate material is diluted in the tool steel clad. Pogson et al. [49] reported that this dilution of copper in the tool steel increases the risk of hot tearing and significantly weaken the material. Zeng et al. [234] showed that addition of nickel in the melt pool increases the solubility of copper by locally retaining the austenite phase. Thus nickel content in the 316 L SS prevented the formation of copper-rich region around the austenite grain boundaries and provided sufficient strength at the elevated temperature. The XRD analysis has shown presence of martensitic phase in the 1st layer of H13 TS specimen (Figure 7-16). Since the operating temperature of the specimen was much below the austenitizing temperature, entrapped carbon in the martensitic phase and the residual stress in the structure could not be released. The brittle behavior introduced by the martensitic phase along with the stress generated from thermal cycling instrumented the fracture in the 1st layer of the H13 TS specimen. The face centered cubic (FCC) crystal structure austenite phase was dominant in the XRD spectrum of the 316 L SS buffer layer (Figure 7-17). The Ni content in the 316 L SS buffer layer helped to stabilize the FCC structure and increased its toughness to absorb thermal cycling induced
stress. Formation of Cr$_7$C$_3$ carbides due to the high Cr content was also believed to be another reason for the absence of cracks in the 316 SS buffer layer since it was less voluminous compared to other types of carbides at elevated temperature and thus reduced the risk of interdendritic cracks [240].

Figure 7-16: XRD spectrum of the 1$^{st}$ layer of H13 TS specimen

Figure 7-17: XRD spectrum of the 1$^{st}$ layer of 316 SS specimen
Temperature gradient that causes cracking highly depends on the thermo-physical properties of the materials. Materials with high heat capacity and heat conductivity transfer heat quickly and, therefore, there is not much increase in temperature gradient [241]. Inspite of favorable thermo-physical properties of the copper alloy core material, the catastrophic cracks in both specimens propagated through the clad into the core, instead of propagating along the interface. The reduction in ductility of the core material by plastic deformation caused by the dislocation movement was the reason that these catastrophic cracks did not propagate along the interface. Minotto et al. [242] reported that in ductile materials, favorably oriented crystals and planes undergo plastic flow along with additional dislocations developed from thermal cycling deformation and move through crystals under stresses to pile-up at grain boundaries and other dislocations. Thus a complex dislocation matrix is formed, which strain-hardens the material by opposing further dislocations movement to reduce the ductility or ability to absorb further deformation of the material and causes cracks.

The catastrophic cracks observed on the surface were the consequence of the stress–strain–temperature loadings coupled with the difference in thermo-physical properties between the clad and core material. The difference in the thermal expansion coefficients caused the thermal expansion mismatch between the core and the coating. Due to the higher thermal expansion coefficient, the copper alloy core material experienced higher expansion compared to the H13 tool steel coating material. This expansion mismatch induced additional tensile stress on the surface and developed plastic strains. Due to the decrease of material strength at higher operating temperature, the plastic strain appeared at tight surface regions of the clad and accelerated surface crack generation [22]. The longitudinal propagation direction of the catastrophic cracks in both specimens indicated that the stress was induced in axial direction. This axial stress was another reason why the catastrophic fractures followed propagation in the direction perpendicular to the
surface and grew through the core instead of growing along the interface. Though the stress thus developed should be equal in both the H13 TS and 316 L SS buffer specimens, the H13 TS specimen suffered from severe damages compared to the other specimen. In the previous chapters, it was shown that copper particles got incorporated in the DMD deposited H13 tool steel clad on copper alloy substrate up to certain thickness of the layer [24]. The incorporated copper could deteriorate clad properties at elevated temperature. The dilution of copper in the H13 TS specimen was thus believed to be instrumental in the catastrophic crack generation. Moreover the crack generated in the 1st layer of the H13 TS specimen made the upper layers vulnerable and accelerated the generation and propagation of the catastrophic cracks.

7.5. Summary

Considering the cyclic heating and cooling nature of HPDC, the thermal fatigue behaviour of bi-metallic structure was evaluated using an induction heating thermal fatigue test rig and the microstructure of the tool steel clad was correlated with fatigue behaviour. The results of this investigation revealed significant difference in the thermal fatigue resistance of directly clad H13 tool steel on copper alloy core material and H13 tool steel clad with 316 L SS buffer layer on copper alloy core material. The 316 L SS buffer layer made a favorable impact on the thermal fatigue resistance showing superior performance compared to the directly clad H13 tool steel specimen. The surfaces of both clads suffered from two types of cracks (1) networks of fine and shallow cracks and (2) large catastrophic cracks. Though the resistance to the networks of fine and shallow cracks was almost similar to both clads, the damage due to the large catastrophic cracks was much more prominent in the directly clad H13 tool steel specimen. Moreover the 1st layer of the directly clad H13 tool steel showed numerous cracks that made the clad vulnerable to the thermal cyclic environment. Different types of cracks were detected and use of 316 L SS buffer layer improved the thermal fatigue resistance
of the clad. Thus the knowledge of this investigation can be used for further development of highly conductive bi-metallic tooling for application in high pressure die casting.
CHAPTER 8
HIGH PRESSURE DIE CASTING TRIALS

8.1. Introduction

From the thermal fatigue test described in the previous chapter it was evident that tool steel clad on copper alloy substrate provided a level of strength against cyclic heating and cooling condition. It proved resistance to the mismatch of co-efficient of thermal expansion between copper alloy substrate and tool steel clad and promoted the bi-metallic structure as a promising candidate for HPDC. However, induction heating type thermal fatigue test did not subject the clads to the severe conditions that exist during HPDC, such as high velocity cavity filling, high pressure, solidification, casting ejection, die cooling, and lubrication. Therefore, it is necessary to evaluate DMD clads under conditions, which more clearly resemble those occurring in HPDC trials.

In this chapter accelerated die casting trials were performed by using a specially designed die fabricated from H13 tool steel (TS) that had the provision to investigate core pins under same conditions [243, 244]. The objective of this trial was to evaluate the performance of bi-metallic core pins in the specially designed die under semi-industrial HPDC conditions. Conventional P20 tool steel core pin was also used to compare its performance with that of bimetallic core pin under accelerated HPDC trials. These comparisons provided better understanding of the behavior of bi-metallic core pins in HPDC of aluminium alloys.

8.2. Experimental Details

8.2.1. High Pressure Die Casting Machine

In the experiment, a 250 tons Toshiba cold chamber high pressure die casting machine at Commonwealth Scientific and Industrial Research Organization
(CSIRO) was used. An accelerated high pressure die casting condition was applied during the casting trials that included extreme operating conditions of injection speed (50-55 m/s), holding pressure (70-75 MPa) and die holding time (30s). The conditions are listed in Table 8-1. Aluminum alloy CA 313 was used as casting material and the melt temperature was maintained at 680°C in the crucible. The cycle time was 60 seconds, in which die holding time was 30 seconds and the remaining time was for molten metal injection, casting ejection and coolant spray. Both casting ejection and coolant spray was performed manually on die surfaces and core pins after die opening in every cycle. Water cooling below 25 mm of the cavity surface was supplied only in the fixed half of the die. In first three shots, injection speed was slow and was increased to high speed in next two shots but without intensification. Finally from sixth shot and onward the castings were performed with high speed and intensification. Figure 8-1 shows a high pressure die casting die used in this experiment at open position with various features. This die had two halves namely fixed and moving half and the core pin was placed in moving half.

Figure 8-1: High pressure die casting die used in this experiment at open position showing various features
Table 8-1: High pressure die casting parameters and material used in the trials

<table>
<thead>
<tr>
<th></th>
<th>Aluminium alloy CA 313</th>
</tr>
</thead>
<tbody>
<tr>
<td>Casting material</td>
<td></td>
</tr>
<tr>
<td>Injection speed</td>
<td>50 - 55 m/s</td>
</tr>
<tr>
<td>Holding pressure</td>
<td>70 - 75 MPa</td>
</tr>
<tr>
<td>Die holding time (except TS long core pin)</td>
<td>30 s</td>
</tr>
<tr>
<td>Avg. cycle time (except TS long core pin)</td>
<td>62 s</td>
</tr>
<tr>
<td>Die holding time (TS long core pin)</td>
<td>60 s</td>
</tr>
<tr>
<td>Avg. total cycle time (TS long core pin)</td>
<td>100 s</td>
</tr>
</tbody>
</table>

Figure 8-2 shows moving half of the die that was used to perform the die casting trials in this investigation [11, 245, 246]. This die was specially designed to produce 115×50×50 mm casting part. This part has a hole feature at the centre produced from core pin (Figure 8-3). The high thickness of the casting brick generated accelerated casting environment by providing higher heat intensity to the core pins compared to the heat intensity provided by thin wall casting parts in industrial casting shots. Another important feature of this die was that when the core pin was placed facing the injection gate, it produced immense impingement on the core pin surface, which was used to evaluate the integrity of the coating.
8.2.2. Preparation of Core Pins

Two sets of core pins, in which the basic difference was in the length, were prepared for HPDC trials. In each set, one core pin was made from tool steel and another one was made from copper alloy. Tool steel core pins were used in the trials to allow reference to current industrial practice and to compare its performance with bi-metallic core pins. Each of the copper alloy core pins was coated with protective tool steel clads by direct metal deposition. The front faces of copper alloy core pins were not coated in order to examine how copper alloy performs in the molten aluminium without protective layer. A 316L stainless steel (SS) buffer layer (0.5 mm thick) was also deposited in between copper alloy substrate and tool steel clad. Figure 8-4 shows the schematic drawing of core pins prepared for the HPDC trials. Short core pins were prepared to avoid the direct impingement of the injection of molten aluminium. The long core pins were specially designed so that it could be positioned in front of gate entry in the die cavity. Positioning the core pins in front of the die cavity gate provided the performance evaluation under direct liquid aluminium impingement condition.
Provisions were made to insert a 1 mm diameter K-type thermocouple down the centre axis of the core pin up to 5 mm from the front face. These thermocouples were placed to record the temperature profile for each core pins during each HPDC cycle.
8.2.3. Evaluation of Core Pins

All core pins and casting parts were visually examined in every 10 HPDC cycles. In addition, optical photographs of the surfaces of core pins and casting parts were collected in order to investigate any failure of the tool steel clad and to evaluate soldering development. The bi-metallic core pins were cut and cross-sections were polished with fine diamond polishing machine up to 1 μm finishing to observe the integrity of the interfaces between clad and substrate material after finishing designated numbers of HPDC cycles. Polished surfaces were etched using a solution of FeCl₃ (20 g), HCl (50 ml) and H₂O (100 ml). Metallographic examinations of the cross-section of core pins were done using a SUPRA 40 VP scanning electron microscope (SEM).

8.3. Results and Discussion

8.3.1. Analysis of Material Build-up

Gulizia [244] has reported that fifty HPDC cycles of the die used in this HPDC machine was equivalent to several thousand cycles in an industrial HPDC machine. Therefore, each of the four core pins was designed for fifty HPDC shots. Figure 8-5 shows the core pins after finishing HPDC trials. During trials, the short
tool steel core pin, however, survived thirty five HPDC shots. Both visual observation and optical photograph of the casting hole that generated from the core pin revealed that the build-up of aluminium initially started just after 10 shots. This build-up was observed at the side of the core pin and eventually increased rapidly in next 20 shots. Figure 8-6 shows the material build-up trend on the short tool steel core pin surface. The material build-up at the surface was so rapid that at thirty fifth shot aluminium casting part stuck on the core pin surface and a piece of casting peeled off from the part and remained stuck with the core pin after forced ejection (Figure 8-7). This catastrophic failure particularly substantiated that the die holding time was not long enough for the casting part around the core pin to be solidified completely.

Figure 8-5: Optical photograph of core pins after HPDC trials from left to right: bi-metallic short, bi-metallic long, tool steel long and tool steel short core pin
Figure 8-6: Optical photograph of casting parts showing the material build up trend on tool steel short core pin due to soldering.
In contrast, short bimetallic core pin survived designed number of HPDC shots without any catastrophic failure. Though it did not experience serious failure as in the case of short tool steel core pin, it appeared to have been coated with silvery aluminium layer. Figure 8-8 shows the material build-up trend of the short bimetallic core pin. It was evident that unlike short tool steel core pin, soldering or material build-up initially started within five HPDC shots at the front face of the short bimetallic core pin where copper was exposed to the molten aluminium. There was no sign of soldering on the side of the core pin before thirty shots. The material build-up however began on the side of the core pin after thirty HPDC shots and became thick gradually. It was also apparent from Figure 8-8 that front face of this core pin suffered more severely from soldering and exaggerated the material build-up at the side of the core pin.
Figure 8-8: Optical photograph of casting parts showing the material build up trend on bimetallic short core pin due to soldering

Long bimetallic core pin also survived designed number of HPDC cycles without experiencing any catastrophic failure. Figure 8-9 shows optical photograph of the material build-up trend of bimetallic long core pin. Similar to bi-metallic short core pin, material build-up initially occurred at the front face within 10 shots. In addition, the direct impingement of molten aluminium caused significant soldering on the area tangent to the injection gate within 20 HPDC shots. Conversely, long tool steel specimen endured only 11 HPDC shots before the casting stuck on it due to severe soldering. Figure 8-10 shows optical photograph of the material build-up trend of tool steel long core pin. It experienced soldering so harshly within 12 shots that the holding time was required to be increased by 30 seconds for proper ejection of the casting part. The extended holding time allowed sufficient solidification of the casting part around the core pin so as not to stack with it (Figure 8-10 14th shot). But it lengthened the die holding time by 100% which in turn increased the cycle time to large extent. The inevitability of increase in die holding time for tool steel core pin particularly confirmed that bi-metallic core pin was able to transfer heat at a faster rate compared to tool steel core pin. The quick extraction of heat from the casting part around the core pin surface
provided rapid solidification and unlike tool steel core pin, the part could not get stuck with the bi-metallic core pin.

Figure 8-9: Optical photograph of casting parts showing the material build up trend on bimetallic long core pin due to soldering

Figure 8-10: Optical photograph of casting parts showing the material build up trend on tool steel long core pin due to soldering
8.3.2. *Die Thermal Profile*

Figure 8-11 shows the thermal profiles of both core pin and bulk die (fixed half) obtained during HPDC shots. The temperature of core pins increased sharply during injection of the molten metal. There was a gradual decrease of temperature in all core pins after lubricant spray. Initially, maximum temperatures of the core pins were slightly lower, which increased exponentially with the increase of number of HPDC shots. It happened since at the beginning of the shots the die was cold and as the time passed the minimum temperature increased gradually, which lead to the reduction of heat transfer of the core pins. It was however evident from these graphs that all core pins reached thermal equilibrium within a fifty cycle HPDC period.

![Graph (a)](image1.png)

![Graph (b)](image2.png)
Figure 8-11: Thermal profiles of core pins and fixed die during HPDC trials (a) tool steel short (b) bi-metallic short (c) bi-metallic long and (d) tool steel long core pin.

Figure 8-12 shows one typical temperature profile for each core pin recorded at equilibrium condition. Table 8-2 lists the maximum and minimum temperatures of the temperature profiles shown in Figure 8-12. The maximum temperature of the short bi-metallic specimen was slightly lower than the short tool steel specimen. The temperature of the long bi-metallic core pin was higher compared to short one. During HPDC, outer part of the casting transmitted heat quickly to the bulk die leaving large intensity of heat at the inner core. Therefore, the core pin that was inserted more into the centre of casting part was more exposed to the high intensity of heat and eventually showed higher temperature. Since long tool steel
specimen ceased to operate before it reached thermal equilibrium condition and the die holding time was increased by thirty seconds which allowed it to cool down for extended period of time, it was not possible to draw a comparison of temperature profile with other specimens.

Table 8-2: Maximum and Minimum temperatures of the core pins

<table>
<thead>
<tr>
<th>Specimen</th>
<th>Maximum Temperature (°C)</th>
<th>Minimum Temperature (°C)</th>
</tr>
</thead>
<tbody>
<tr>
<td>Tool steel (short)</td>
<td>554.92</td>
<td>401.72</td>
</tr>
<tr>
<td>Bi-metallic (short)</td>
<td>518.37</td>
<td>409.81</td>
</tr>
<tr>
<td>Tool steel (long)</td>
<td>549.49</td>
<td>340.60</td>
</tr>
<tr>
<td>Bi-metallic (long)</td>
<td>550.37</td>
<td>380.01</td>
</tr>
</tbody>
</table>

Figure 8-12: Thermal profiles of core pins for a single cycle at equilibrium stage during HPDC trials

In general, the maximum temperatures in bi-metallic core pins were anticipated to be much lower compared to that of the tool steel core pins since copper was expected to transmit heat quickly. However, the difference in the maximum temperature between these two groups of specimens was very small. The main reason was that, the size of the core pin was too small compared to the bulk amount of heat in the casting and resulted insufficient heat conduction through the
core pin. Moreover, the amount of heat it transferred from the casting could not be transmitted further to any other auxiliary medium since there was no cooling medium that could carry out the heat conducted by copper alloy. As a result, though bi-metallic core pin transferred heat quickly from the surface to some extent, this heat concentrated in the inner part and the hot inner part eventually restricted further transfer of heat from the surface. Thus the bi-metallic core pins showed lower temperature difference with tool steel core pins.

8.3.3. Evaluation of Tool Steel Clad

Since interfaces were most susceptible to cracking due to different material properties of two different materials, after finishing 50 HPDC cycles, the cross-section of both short and long bi-metallic core pins were examined to evaluate the integrity of copper alloy-316 L SS and 316 L SS-H13 TS interfaces. Figure 8-13 shows the SEM micrographs of the interfaces for the four core pins after designated numbers of HPDC cycles. From the micrographs, it was evident that there was no sign of failure in the interfaces and this proved the applicability of the bi-metallic structure under HPDC environments. Though there was no catastrophic failure in the interfaces, few cracks were observed in the tool steel clad in few areas. Figure 8-14 presents the cracks that were visible in the tool steel clad in bi-metallic long core pin. These cracks were believed to be the consequence of insufficient clad thickness since, due to the geometry limitation of the core pin, more than 2 mm thick clad could not be applied.
(a) H13 tool steel clad

316L SS buffer layer

(b) 316L SS buffer layer

Copper alloy substrate
Figure 8-13: SEM micrographs of the interfaces after 50 HPDC cycles (a), (b) long bi-metallic core pin and (c), (d) short bi-metallic core pin
Figure 8-14: SEM micrograph showing the cracks in the tool steel clad in long bi-metallic core pin

In addition, the HPDC trials also evaluated the factor of erosion of copper alloy due to soldering reaction with aluminium alloy during HPDC. Figure 8-15 shows SEM micrograph of the front face of bi-metallic core pin, where copper alloy was exposed to aluminium casting material. Within 50 HPDC cycles, front face was eroded for 0.35 mm and the erosion particularly occurred in the copper alloy core material. This phenomenon of copper erosion can be explained by the work of Zhu et al. [44], where they showed copper washout in molten aluminium. In this experiment, since copper was exposed to molten aluminium at the front face of the core pin, it washed out severely due to the very high solubility of copper in molten aluminium. This erosion of copper substantiates that copper alone cannot be applied as HPDC die material.
8.4. Summary

Induction heating type thermal fatigue test described in the previous chapter did not subject the tool steel clads to the severe conditions that exist during HPDC. Therefore, performance of bi-metallic core pins were evaluated in semi-industrial environment that included all the features present in HPDC industries. This investigation revealed that though each of the four core pins was designed for fifty HPDC shots, tool steel core pins could not survive intended number of HPDC shots during trials and resulted in catastrophic failure due to severe soldering. This catastrophic failure particularly substantiated the view that the die holding time was not long enough for the casting part around the core pin to be solidified completely. Unlike tool steel, bi-metallic core pins facilitated quick solidification of the casting material around the pin that in turn reduced the rate of soldering and survived the designed number of HPDC shots without any severe failure. It was evident from the thermal profiles that all core pins reached thermal equilibrium within the designated HPDC period. The maximum temperature of short bi-
metallic specimen was slightly lower than short tool steel specimen. Since long tool steel specimen ceased to operate before it reached thermal equilibrium condition, it was not possible to draw a comparison of temperature profile with other specimens. There was no sign of failure in the interfaces between copper alloy and 316 L SS layer and 316 L SS and H13 TS. The intact integrity of the clad interfaces certainly proved the applicability of the bi-metallic structure under HPDC environments.
CHAPTER 9
CONCLUSIONS AND FUTURE RESEARCH

9.1. Overview

The objective of this research was to develop a bi-metallic tooling of copper alloy coated with protective tool steel layers for high pressure die casting application. Various investigations were conducted to explore the fabrication feasibility of such tooling using direct metal deposition technique. Also, finite element analysis using ANSYS simulation software was employed in order to examine the quick heat transfer capability of such tool. This chapter presents the conclusions reached after conducting all these experiments and some recommendations are provided for further research in the area of developing bi-metallic tooling in HPDC industry.

9.2. Major Conclusions

This thesis has found following major conclusions after performing investigation on bi-metallic tooling to reduce cycle time for high pressure die casting application-

- Finite element analysis revealed that copper based bi-metallic die reduced the casting solidification time to 1/3 of the time taken by tool steel die. Thus, high strength copper alloy coated with protective tool steel layers is a potential alternative of tool steel as a die material and offers significant reduction in the total cycle time of a casting part in the HPDC industries.

- One of the major challenges of the research was to deposit tool steel layers on copper alloy substrate due to very different physical properties of these two materials. However, the extensive parametric investigation demonstrated that by combining various suitable DMD process parameters, metallurgically sound tool steel clad can be deposited on copper alloy substrate material. Parametric
investigation substantiated that high laser power was required to melt the copper alloy substrate. Also, the hardness of deposited H13 tool steel increased considerably with the increase of the laser power up to certain limit. In contrast, high laser power was found disadvantageous in this instance due to the high reflectivity of the copper alloy substrate. Therefore, delicate control of the laser power was required to deposit tool steel on copper alloy substrate.

- Microstructural characterization revealed that delicate control of the laser power was critical in obtaining desired material properties since the microstructure of deposited H13 tool steel varied considerably with the variation of the laser power. When tool steel was deposited immediately on copper alloy substrate, copper was found to be diffused in the tool steel clad. Use of 316 L SS as a buffer layer in between clad and substrate materials became very effective to eliminate copper diffusion in tool steel clad. The microstructure of directly deposited H13 tool steel on copper alloy substrate was also found to be slightly different from that of H13 tool steel clad deposited using a 316 L SS buffer layer.

- Investigation on mechanical properties demonstrated that the bonding between tool steel clad and copper alloy substrate was much higher compared to other commercially available techniques to join these two materials. Considering the application environment, the die materials required for HPDC should be ductile. This particular requirement was met through the fracture toughness analysis since the bi-metallic structure was within the ductile region. Investigation on fracture surfaces also showed ductile fractures in majority of the area in the bi-metallic structure.

- Although the bond strength between tool steel clad and copper alloy substrate reduced when 316 L SS was used as a buffer layer, the thermal fatigue test showed that use of buffer layer was advantageous considering the application environment of bi-metallic structure in HPDC. In the cyclic heating and cooling condition, the diffused copper in tool steel clad worked as weak region and
initiated cracks in the thermal fatigue specimen. In contrast, since 316 L SS buffer layer eliminated this copper diffusion in the clad, specimen with 316 L SS buffer layer showed better resistance to such cracking in thermal cycling.

- Semi-industrial HPDC trials revealed that bi-metallic core pins facilitated quick solidification of the casting material around the core pin compared to that of tool steel and reduced the rate of soldering. The interfaces between copper alloy and 316 L SS layer and 316 L SS and H13 TS layers, which were most susceptible to fail in the HPDC conditions also survived without any damage. Furthermore, the intact integrity of bi-metallic core pins without any failure in the tool steel clad substantiated that bi-metallic dies are applicable in HPDC environment.

9.3. Recommendation for Future Research

This thesis has provided promising results on viability and manufacturing capability of bi-metallic tooling for high pressure die casting applications. Though the outcomes of these investigations show numerous potential, some aspects of this research need to be explored further. These future works are described briefly in the following sections.

During both thermal fatigue testing and industrial trial, the tested specimens showed some cracks in the tool steel clad. These cracks need to be eliminated completely since it will have detrimental effect on the clad integrity if run for long time in the industrial environment. One effective way that can eliminate these cracks can be to increase tool steel layer thickness which will certainly provide more strength to resist crack generation due to cyclic thermal loading. Eventually, increase in layer thickness will have some adverse effect on HPDC cycle time. Therefore, FEA can be employed to investigate the effect of layer thickness on thermo-mechanical properties and optimum thickness can be obtained.

Due to the small size compared to a bulk size die, bi-metallic core pins could not prove effective heat transfer compared to tool steel core pins. Therefore, a bulk bi-
metallic die with proper design of cooling channels needs to be investigated under industrial conditions to demonstrate sufficient reduction in cycle time. In addition, various cooling channel design can be investigated in bi-metallic dies in order to get more effective heat transfer.

The actual dies have sharp edges and tight corners on the cavity surface that will challenge DMD cladding. Therefore manufacturing viability of these complex features on cavity surface using DMD should be investigated further.

It is expected that the properties of casting parts produced using bi-metallic part will be changed due to rapid solidification of the molten metal. In general, the casting part will have fine microstructure due to quick heat transfer through high thermal conductive die material, which, in turn, will provide better physical properties. However, the properties of the cast parts made of bi-metallic tooling should be analyzed and correlated with bi-metallic die capabilities.

Bi-metallic core pins provided resistance to soldering to some extent compared to tool steel core pins. However, the soldering could not be eliminated completely. Therefore, PVD coating which is widely used to hinder soldering on tool steel dies can be explored in order to eradicate soldering on bi-metallic tooling.

Finally, cold spray technology can be explored to deposit tool steel on copper alloy substrate and its capability can be compared with DMD.
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