MACROSTRUCTURE AND MECHANICAL BEHAVIOR OF INCONEL 718 SUPERALLOY FABRICATED BY A BLOWN POWDER LASER DEPOSITION ADDITIVE MANUFACTURING PROCESS

by

CHRIS HILL

A THESIS

Submitted in partial fulfillment of the requirements for the degree of Master of Science in Engineering in The Department of Mechanical and Aerospace Engineering to The School of Graduate Studies of The University of Alabama in Huntsville

HUNTSVILLE, ALABAMA

2018
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Timothy C. Hill  3/2/18
(Timothy C. Hill) (Date)
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We, the undersigned members of the Graduate Faculty of The University of Alabama in Huntsville, certify that we have advised and/or supervised the candidate on the work described in this thesis. We further certify that we have reviewed the thesis manuscript and approve it in partial fulfillment of the requirements for the degree of Master of Science in Engineering with an option in Mechanical Engineering.

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ABSTRACT
The School of Graduate Studies
The University of Alabama in Huntsville

Degree: Master of Science in Engineering  College/Dept. Engineering/Mechanical Aerospace.

Name of Candidate: Timothy C. Hill

Title: Macrostructure and Mechanical Behavior of Inconel 718 Superalloy Fabricated by a Blown Powder Laser Deposition Additive Manufacturing Process.

Additive manufacturing (AM) has become a process of great interest to the aerospace world for its potential of fabricating near net shape parts with unique complex geometries. As certain components are being proven to be able to be manufactured with suitable material properties on smaller scale systems, engineers are looking to scale up the size of manufactured components for larger systems. While powder bed AM is limited to the size of components that can be built in the box, other methods can be used to print larger components. This study looks at blown powder AM which is scalable to larger structures.

This thesis evaluates the material property data and characterization of an Inconel 718 superalloy fabricated using a blown powder laser deposition (BPLD) AM process. A detailed metallographic evaluation was made of the resulting macrostructure. Tension and fatigue tests were used to determine the mechanical properties of the material.
Abstract

Additive manufacturing (AM) has become a process of great interest to the aerospace world for its potential of fabricating near net shape parts with unique complex geometries. As certain components are being proven to be able to be manufactured with suitable material properties on smaller scale systems, engineers are looking to scale up the size of manufactured components for larger systems. While powder bed AM is limited to the size of components that can be built in the box, other methods can be used to print larger components. This study looks at blown powder AM which is scalable to larger structures.

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CHAPTER I

Introduction

Additive manufacturing (AM) has been gaining a lot of attention from multiple industries around the world. A quick search of the word on a search engine returns hits numbering from the tens of thousands to millions. AM technology is being used to print components for use in applications ranging from the home do-it-yourselfers to aerospace industries. This technology can be applied to various materials such as plastics, metals, and ceramics. AM technology is even being used to print human organs [1].

One significant area of modern industry were AM is being used and researched is in the aerospace field. Various aerospace companies including: General Electric (GE), Boeing, Lockheed Martin, and the National Aeronautics and Space Administration (NASA), are currently using AM technology [1]. The NASA-Marshall Space Flight Center (MSFC) has recently demonstrated the capabilities of the technology by printing and testing various liquid rocket engine (LRE) components including: a turbo pump, an injector, an igniter, and a regeneratively cooled chamber [2-3].
1.1 Background and Motivation

Material selection is vital part of engineering design and fabrication. Without the proper material for the application, even the best designs can fail to meet performance expectations. Some of the most advanced engineering concepts have relied upon materials that remain their strength in extreme environments. One example is the fabrication of the RS-25 engine, or the Space Shuttle Main Engine, which used Inconel alloy 718 for structural integrity of the main combustion chamber (MCC) [4]. Since Inconel alloy 718 retains its high strength at elevated temperatures, in addition to resisting corrosion, it is used in a wide variety of applications ranging from rocket engine components and jet turbines to nuclear reactor components.

The technologies used to process materials are an important element in manufacturing. Utilizing advanced manufacturing technologies is critical to the ability of the NASA to produce launch vehicles for space exploration [5]. As new technologies become available, their ability to reduce the cost of fabrication continues to be explored in an effort to make space exploration more affordable. When considering the fabrication of LREs, many components are designed with current manufacturing technologies in mind such as conventional subtractive machining. AM is changing the way designers and engineers think about component manufacturing by enabling new designs with geometries and features that were traditionally very expensive or near impossible to manufacture. AM also allows printing monolithic components that would otherwise be traditionally manufactured as multiple pieces requiring assembly or welding. One example of this is a shrouded rocket turbomachinery impeller. Figure 1 shows an unshrouded (a) versus a

2
shrouded (b) impeller where it can be seen that machining a monolithic shrouded impeller would be very difficult due to the shroud covering up the complex geometry of the fluid flow channels. This type of LRE pump component can be designed to operate at speeds as high as 90,000 RPM and pump cryogenic propellants [6]. One traditional method of manufacturing this type of impeller is to subtractively machine the impeller, and then braze on a shroud [7]. With AM, engineers now have the option to deposit the impeller as a single monolithic piece.

![Unshrouded (a) vs Shrouded Impeller (b) [8]](image)

**Figure 1: Unshrouded (a) vs Shrouded Impeller (b) [8]**

Another example of how AM is reducing time of manufacturing with more monolithic deposited components is the NASA Additive Manufacturing Demonstrator Engine (AMDE) project. This program is designed to gather data on the ability of AM to reduce the time and cost of rocket engine development [9]. Figure 2 shows the engine being developed for this project and reduction in the part count achieved by using AM as compared to traditional subtractive manufacturing techniques.
The individual components on this demonstrator engine are small enough to be built in a powder bed. Currently the largest production powder bed AM machine gives a build box of approximately 800 x 400 x 500 mm³ (31.5 x 15.7 x 19.7 in³) [10].

Figure 2: Liquid Rocket Engine (LRE) being developed under the AMDE program showing reduced part count achieved using AM vs conventional subtractive manufacturing [9]

Figure 3 illustrates the range in size of LRE and their components. While the powder bed process has shown great potential for reducing the part count for complex assemblies, it is currently limited in size for these larger scale LRE components.
An alternative to size limited powder bed techniques is the use of direct metal deposition (DMD) processing. DMD techniques are not confined to a box but utilize either robotic or computer numerical control (CNC) platforms to directly deposit the metal. RPM Innovations, Inc. out of Rapid City, South Dakota, USA has a DMD machine with a 1.5m x 1.5m x 2.1m (5ft x 5ft x 7ft) work envelope. Another company, DM3D out of Auburn Hills, Michigan USA, has a robotic arm DMD system that offers a 3.2m x 3.7m x 360 degrees (10.5ft x 12.1ft x 360 degrees) work envelope.

The objective of this study is to characterize the mechanical properties of Inconel 718 specimens deposited using a DMD process. The specimens deposited for this study were performed on a DMG MORI Lasertec 4300 prototype machine at their Advanced Solutions Department in Hoffman Estates, Illinois, USA. After deposition, these
specimens were heat treated (HT) and then machined for testing. Representative samples were also prepared for metallurgical evaluation including grain size and void content.

1.2 Overview of Additive Manufacturing

The creation of tools have been a vital part of humanity’s survival and quality of life from as far back as 2.6 million years ago during the Paleolithic age and is still just as vital present day [11]. With the invention of machines and more advanced tools to aid in manufacturing, processes can be categorized into three main categories: forming, machining, and casting. On March 11th, 1986 a patent was published that added a fourth category to aid in manufacturing. An Inventor by the name of Charles W. Hull filed a patent with the United States Patent office entitled, “Apparatus for production of three-dimensional (3D) objects by stereolithography” [12]. It was this invention that led to the development of AM in plastics, followed later by metals. Initially the AM process utilized photopolymerization by which an ultraviolet (UV) laser is focused into a vat of photopolymer resin to cure the resin. The laser follows the path generated by slicing a 3-dimensional (3D) computer aided design (CAD) model. As each cross-sectional slice of the 3D component is cured, a new layer of resin is added and the process repeats. It is this repetitive layer wise curing that leads to the creation of a 3D part.

AM was expanded to include metals in the mid 1997 when inventors Wilhelm Meiners, Konrad Wissenbach, and Andres Gasser filled a patent with the United States Patent Office entitled, “Selective laser sintering at melting temperature” [13]. Their invention, illustrated
in Figure 4, uses a laser to sinter the metal powder in a vat, or “bed”, in locations defined by slicing a 3D model. As each layer is sintered, a new layer of powder is deposited and the process repeats to build a part layer by layer. The process has since been adapted by many companies under names including selective laser sintering (SLS) and selective laser melting (SLM).

![Figure 4: SLM illustration from Meiners, Wissenbach and Gaser patent [13]](image)

With the developments in metal AM in the 1990’s, Jyoti Mazumder and Justin Koch, took an interest in creating another means of rapid prototyping via laser cladding [14]. Laser cladding, as defined by the patent owner, adds a metal coating to a metallic substrate [15]. This process, devised by Mazumder and Koch, blows metal powder into the focal point of a laser where it is melted. This new process came to be known as “Directed Metal Deposition” (DMD). This type of AM process is more commonly called blown powder direct laser deposition (BPDL), directed energy deposition (DED), or blown powder laser
deposition (BPLD). In 2000, Mazumder and Koch had their patent published and their system is shown in Figure 5.

Figure 5: DMD Image from Mazumder and Koch patent [16]

1.3 Research Goals and Approach

This thesis presents the material macrostructure and property characterization of the superalloy Inconel 718 produced using a blown powder laser deposition AM process. The effect of various deposition environments and heat treat conditions were also evaluated.
With the growing interest in the utilization of AM to manufacture components, the relationship between the macrostructure and mechanical properties must be understood to guide safe engineering design practices. The material characterization and properties undertaken in this study is representative of, but not limited to, the information needed to support design and modeling efforts by industry to better understand the resulting material properties of AM processes.
CHAPTER II

Literature Review

2.1 Inconel 718 Overview

Inconel alloy 718 is a precipitation strengthened austenitic nickel (Ni) based super alloy. It is primarily used in applications where retention of strength is needed at high temperatures or in corrosive environments, or a combination of the two. Table 1 summarizes the elemental composition of Inconel 718 and Table 2 summarizes the phases present and their chemical formula.

Table 1: The Chemistry of Inconel 718 (%) [17]

<table>
<thead>
<tr>
<th>Element</th>
<th>Composition</th>
</tr>
</thead>
<tbody>
<tr>
<td>Nickel (Ni)</td>
<td>50.00 – 55.00</td>
</tr>
<tr>
<td>Chromium (Cr)</td>
<td>17.00 – 21.00</td>
</tr>
<tr>
<td>Iron (Fe)</td>
<td>Balance</td>
</tr>
<tr>
<td>Niobium (Nb)</td>
<td>4.75 – 5.50</td>
</tr>
<tr>
<td>Molybdenum (Mo)</td>
<td>2.80 – 3.30</td>
</tr>
<tr>
<td>Titanium (Ti)</td>
<td>0.65 – 1.15</td>
</tr>
<tr>
<td>Aluminum (Al)</td>
<td>0.20 – 0.80</td>
</tr>
<tr>
<td>Cobalt (Co)</td>
<td>1.00 Max</td>
</tr>
<tr>
<td>Carbon (C)</td>
<td>0.08 Max</td>
</tr>
<tr>
<td>Manganese (Mn)</td>
<td>0.35 Max</td>
</tr>
<tr>
<td>Silicon (Si)</td>
<td>0.35 Max</td>
</tr>
<tr>
<td>Phosphorus (P)</td>
<td>0.015 Max</td>
</tr>
<tr>
<td>Sulfur (S)</td>
<td>0.015 Max</td>
</tr>
<tr>
<td>Boron (B)</td>
<td>0.006 Max</td>
</tr>
<tr>
<td>Copper (Cu)</td>
<td>0.30 Max</td>
</tr>
</tbody>
</table>
Table 2: Summary of Inconel alloy 718 phases

<table>
<thead>
<tr>
<th>Phase Names</th>
<th>Structure</th>
<th>Composition</th>
<th>Primary Function</th>
</tr>
</thead>
<tbody>
<tr>
<td>Gamma ((\gamma))</td>
<td>FCC</td>
<td>See Table 1.</td>
<td>Matrix of the alloy</td>
</tr>
<tr>
<td>Gamma Prime ((\gamma'))</td>
<td>FCC</td>
<td>(\text{Ni}_3\text{(Al,Ti)})</td>
<td>Strengthening</td>
</tr>
<tr>
<td>Gamma Double Prime ((\gamma''))</td>
<td>BCT</td>
<td>(\text{Ni}_3\text{Nb})</td>
<td>Main Strengthening</td>
</tr>
<tr>
<td>Delta ((\delta))</td>
<td>Orthorhombic</td>
<td>(\text{Ni}_3\text{Nb})</td>
<td>Amount limited to avoid Detrimental effect on Mech. Properties</td>
</tr>
<tr>
<td>Laves</td>
<td>HCP</td>
<td>((\text{Ni,Fe,Cr})_2\text{(Nb,Mo,Ti)})</td>
<td>Detrimental to Mech. Properties</td>
</tr>
</tbody>
</table>

The material consists of an austenitic Ni matrix, or \(\gamma\) phase, comprised of 52.5wt% Ni-19.0 wt% Cr-3.0 wt% Mo-18.0 wt% Fe in solid solution. Cr is present to provide oxidation resistance by forming a passive layer of chromium oxide \((\text{Cr}_2\text{O}_3)\) [18]. Various phases are present in Inconel 718 which form either during solidification or during subsequent aging heat treatments.

The high strength of alloy 718 is obtained during heat treatments to precipitate the strengthening phases. Nb is the main strengthening alloying element which precipitates to form two phases, gamma prime \((\gamma')\) and gamma double prime \((\gamma'')\). Gamma double prime is a body centered tetragonal (BCT) precipitate phase with composition \(\text{Ni}_3\text{Nb}\) [19]. This precipitation strengthening is due to the coherency strains induced by the precipitate \(\gamma''\) forming in the nickel \(\gamma\) matrix. The \(\gamma''\) precipitate creates a strain on the \(\gamma\) matrix causing the lattice to become distorted. This distortion inhibits dislocation movements, thus
increasing the strength of the material [20]. Figure 6 shows the \( \gamma'' \) precipitates (white) in the \( \gamma \) matrix (black).

![Figure 6: Micrograph of \( \gamma'' \) precipitates (white) in \( \gamma \) matrix (black) [21]](image)

Identification of the primary strengthening phase has been questioned in the literature and evolved over the years of research [20 - 22]. A parallel was found between the identification of the main strengthening phase and the advancements in testing equipment and procedures. Literature in the mid 1960’s and earlier state that the primary strengthening phase was the \( \gamma' \) \( \text{Ni}_3(\text{Al,Ti}) \) phase. Eiselstein analyzed the residue left from electrolytic dissolving of a specimen using X-ray diffraction which revealed a face centered cubic (FCC) crystalline structure. The residue was then analyzed chemically and the age hardening precipitate was identified as \( \gamma' \) [21].
Later work suggests that the $\gamma''$, a BCT structure, is the primary strengthening mechanism and can be detected via x-ray diffraction and transmission electron microscopy (TEM) dark field analysis [20] [22]. The advancement in testing equipment and technologies appear to advance the understanding of the material. It is also worth noting that the $\gamma'$ phase has been found to strengthen the material as well but at a lesser amount due to the smaller volume fraction of 4% in comparison to the 15% of $\gamma''$ [20].

Along with strengthening phases, there are two phases that can exist in alloy 718 than can be detrimental to its mechanical properties. These phases are called delta ($\delta$) and laves. Delta phase is an orthorhombic structure with Ni$_3$Nb composition and its microstructure is shown in Figure 7 where in (A) the delta phase is present only at the grain boundaries and (B) where the delta phase is present everywhere in the structure (inter/transgranular). Although the carbides and the $\delta$ phase deplete the matrix of the Nb needed for the strengthening precipitates, they are noted to pin grain boundaries to retain a fine grain size during high temperature heat treatments [23]. Studies have shown that $\delta$ phase can have a detrimental effect on the yield strength with a 10% reduction in comparison to standard alloy 718, although some retention of the $\delta$ phase is also noted to improve stress-rupture properties [24].
The major detrimental phase on the mechanical properties of the alloy is the Laves phase. Laves is a brittle intermetallic hexagonally closed packed structure with \((\text{Ni,Fe,Cr})_2(\text{Nb,Mo,Ti})\) composition. The mechanical properties of alloy 718 can be reduced by Laves phase in several ways. The most dominant being the fracture of the phase due to its inherent brittleness. This phase also depletes the matrix of Nb needed to precipitate the alloy’s strengthening phases. Laves is usually a result of segregation occurring during solidification, although it can also form in solid state [25]. Figure 8 shows scanning electron microscopy (SEM) images microstructure of Laves phase forming at the core of the dendrites.

Since Inconel 718 is precipitation strengthened, it is referred to as a heat treatable alloy. There are several standard heat treatment for this particular alloy. Table 3 lists several of the heat treatment standards for various applications in which a customer can
directly purchase the material in a variety of forms. Note that all these forms consist of Inconel 718 in the wrought form in which it has been mechanically worked prior to heat treatment.

![Figure 8: Laves Microstructure [33]](image)

### Table 3: Heat Treatment Specifications for Inconel alloy 718 [17]

<table>
<thead>
<tr>
<th>Standard</th>
<th>Product</th>
<th>Annealing temperature range</th>
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<tbody>
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<td>AMS 5596</td>
<td>Sheet, Strips, and Plates</td>
<td>927-1010°C Anneal and Age</td>
</tr>
<tr>
<td>AMS 5662 &amp; 5663</td>
<td>Bars, Forgings, and Rings</td>
<td></td>
</tr>
<tr>
<td>AMS 5589</td>
<td>Seamless Tubing</td>
<td></td>
</tr>
<tr>
<td>AMS 5664</td>
<td>Bars, Forgings, and Rings</td>
<td>1038-1066°C Anneal and Age</td>
</tr>
<tr>
<td>AMS 5597</td>
<td>Sheet, Strips, and Plates</td>
<td></td>
</tr>
<tr>
<td>AMS 5590</td>
<td>Seamless Tubing</td>
<td></td>
</tr>
<tr>
<td>NACE MR 0175</td>
<td>Oil Field Applications</td>
<td></td>
</tr>
</tbody>
</table>
Heat treatments are critical for alloy 718 to reach its full strength potential. Figure 9 shows a time-temperature-transformation (TTT) curve of Inconel 718. Both $\gamma'$ and $\gamma''$ are metastable, transforming into $\delta$ at temperatures in excess of 650°C (1200°F) which limits its high temperature usage [26]. Laves phase has been reported to form at temperatures in excess of 980°C (1800°F) [25].

Figure 9: TTT diagram of Inconel 718 [26]

2.2 Additive Manufacturing

AM is defined by the American Society for Testing and Materials (ASTM) standard F2792 as, “A process of joining materials to make objects from 3D model data, usually layer upon layer, as opposed to subtractive manufacturing methodologies.”
Directed energy deposition is a blown powder type AM process in which a stream of flowing powder is focused into the focal point of a laser and melted onto a substrate. Figure 10 illustrates this process and its components. The powder is delivered from a hopper by an inert carrier gas and into a nozzle. The nozzle acts as the delivery vessel in which the powder is focused to a set point above the substrate material. A conical style nozzle is illustrated in this scenario. The laser beam is focused to the same point as the powder in order to melt the powder in order to start the deposition process.

Figure 10: DMD Process and components
CHAPTER III

Experimental Procedure

3.1 Additive Manufacturing Technique and Parameters

The samples were deposited with a DMG Mori Lastertec 4300 3D prototype machine at their solutions development center in Hoffman Estates, Illinois. The Lasertec 4300 prototype machine is a hybrid AM machine and is shown in Figure 11.

Figure 11: DMG Mori Seki Lasertec 4300 3D prototype machine at their Hoffman Estates Location
The term hybrid means that the machine has the capability to perform traditional subtractive machining and AM all in one machine. The Lasertec prototype is a turn mill CNC platform with a blown powder deposition head added. The machine was equipped with a 2000W Laserline diode laser system, a custom in house manufactured powder nozzle, and an Oerlikon Metco dual powder hopper feeder.

Rectangular prism shaped Inconel 718 samples, as shown in Figure 12 were deposited onto 1045 steel substrates. Table 4 shows the parameters used to perform the deposition of these samples. A total of 28 samples were deposited for this study using Micro-Melt® 718 powder with average size of 100 μm (-140+325M). These powders were produced by argon gas atomization.

<table>
<thead>
<tr>
<th>Deposition Parameters</th>
<th></th>
<th></th>
</tr>
</thead>
<tbody>
<tr>
<td>Laser Power</td>
<td>1550 W</td>
<td></td>
</tr>
<tr>
<td>Laser spot size</td>
<td>4 mm</td>
<td></td>
</tr>
<tr>
<td>Powder Feed Rate</td>
<td>30 g/min</td>
<td></td>
</tr>
<tr>
<td>Layer Height</td>
<td>0.85 mm</td>
<td></td>
</tr>
</tbody>
</table>

Table 4: Deposition Parameters

The test matrix is shown in Table 5 which lists the samples, their heat treatment conditions, and tests conducted. When depositing these samples, some were deposited with the argon shielding and others without to investigate the effect on the material properties. The post processing of these samples included stress relief, hot isostatic pressing (HIP), simulated HIP process, followed by solution and aging heat treatments.
Table 5: Test matrix for the Inconel 718 specimens from DMG

<table>
<thead>
<tr>
<th>Process</th>
<th>Number of specimens</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td>Wo argon shield</td>
</tr>
<tr>
<td>Stress relief</td>
<td>15</td>
</tr>
<tr>
<td>HIP</td>
<td>HIP</td>
</tr>
<tr>
<td>wo HIP</td>
<td>wo HIP</td>
</tr>
<tr>
<td>Lot designation</td>
<td>Lot 2</td>
</tr>
<tr>
<td>Solution treat/age</td>
<td>8</td>
</tr>
<tr>
<td>Tensile specimens</td>
<td>4</td>
</tr>
<tr>
<td>Fatigue specimens</td>
<td>4</td>
</tr>
</tbody>
</table>

The simulated HIP processed samples are labeled as without HIP and were exposed to the same time and temperature profile as the HIP specimens, but without pressure. This was to keep the temperature history of each sample constant to allow the results to reflect the effect of the HIP process. A small sample was sectioned off of the one with argon and one without argon specimen in an as-built state before they were sent off for post processing.
This was done to enable the comparison of the HIP and HT specimens to as-built conditions so that any effects the shielding gas may have on the material can be analyzed during void analysis and metallography.

### 3.2 Heat Treatments

The specimens were machined into cylinders prior to heat treatment in accordance with the American Society for Metals (ASM) standard 5664. Although not in ASM 5664, additional steps of SR and HIP were added to the heat treatment. In processing of AM components, a stress relief cycle is usually implemented to reduce residual stresses that could distort the part. A HIP process is used to reduce voids inherent in the process. As noted in Table 5, not all of the samples went through the HIP process prior to the HT. The ones that did not undergo HIP used a simulated HIP process using the same temperature and time profile, but without the pressure. Table 6 lists the HIP and HT parameters used on the specimens. Note where the environment could not be controlled, the specimens were wrapped in stainless steel foil to minimize oxidation. As the cylinders were processed, specimens were retained after each step of the HTs to allow metallographic characterization to document macrostructural changes.

<table>
<thead>
<tr>
<th>Stress Relief</th>
<th>1066°C ± 14°C for 90 minutes -5 ± 15 minutes, inert atmosphere</th>
</tr>
</thead>
<tbody>
<tr>
<td>HIP</td>
<td>1121°C to 1163°C for 4 hours ± 1 hour, 102 ± 2 MPa, cool in inert atmosphere to below 427°C</td>
</tr>
<tr>
<td>Solution Treat</td>
<td>1066°C ± 14°C for 1 hour in inert atmosphere, cool at a rate of air cooling or faster</td>
</tr>
<tr>
<td>Aging</td>
<td>760°C ± 8°C for 10 hours ± 0.5 hour, furnace cool to 649°C ± 8°C and hold until total aging time is 20 hours, furnace cool</td>
</tr>
</tbody>
</table>

Table 6: HIP and HT parameters used on the specimens
3.3 Metallographic Preparation

All the specimens were sectioned for metallography using a Buehler Abrasimet 250 saw and Allied blades. The sections were mounted in Allied black phenolic mounting powder (#135-10007) using a Simplimet 1000 automatic mounting press machine with parameters shown in Table 7.

<table>
<thead>
<tr>
<th>Specimen Mounting Parameters</th>
<th></th>
</tr>
</thead>
<tbody>
<tr>
<td>Heat Time</td>
<td>1 min</td>
</tr>
<tr>
<td>Cool Time</td>
<td>5 min</td>
</tr>
<tr>
<td>Pressure</td>
<td>4100 psi</td>
</tr>
<tr>
<td>Temperature</td>
<td>120°F to 360°F in 20°F Increments</td>
</tr>
</tbody>
</table>

The samples were cut in two orientations to show the transverse (1-6) and longitudinal (7-12) orientations of the build as illustrated in Figure 13. The mounted specimens were ground and polished using a Struers Tegramin-20 automatic grind and polisher. Silicon Carbide (SiC) papers and Alumina (Al₂O₃) powders were used per the parameters listed in Table 8. Images were taken of the as-polished specimens for an analysis of the void content.
After imaging for the void analysis, the specimens were etched using waterless Kalling’s to reveal the macrostructure. Prior to etching, each sample was freshly polished with the same final polishing parameters from before to remove the passivation layer. After polishing, the specimens were immediately rinsed with water and then isopropyl alcohol followed by a quick drying with heated air before going into the etchant. The process time of cleaning off the polishing compound after the fresh polish and getting the sample into the etchant took less than ten seconds. A submerge and swab technique was used to etch the samples. The swabbing was done with a cotton swab. Each sample was etched for a total of three minutes with swabbing every 30 seconds.
Table 8: Grinding and polishing parameters used

<table>
<thead>
<tr>
<th>Microscopy Grind and Polish Parameters</th>
<th>Grit</th>
<th>Paper/Pad</th>
<th>Force [N]</th>
<th>Time [min]</th>
<th>Head &amp; Table Speed [RPM]</th>
<th>Table Rotation Direction</th>
<th>Water Flow</th>
</tr>
</thead>
<tbody>
<tr>
<td>Initial Grind</td>
<td>240</td>
<td>SiC</td>
<td>20</td>
<td>3</td>
<td>150/300</td>
<td>Counter Rotating</td>
<td>Heaviest laminar stream</td>
</tr>
<tr>
<td>2nd Grind</td>
<td>600</td>
<td>SiC</td>
<td>20</td>
<td>3</td>
<td>150/150</td>
<td>Co-Rotating</td>
<td>Smallest steady laminar stream</td>
</tr>
<tr>
<td>3rd Grind</td>
<td>800</td>
<td>SiC</td>
<td>20</td>
<td>3</td>
<td>150/150</td>
<td>Co-Rotating</td>
<td>Smallest steady laminar stream</td>
</tr>
<tr>
<td>4th Grind*</td>
<td>1200</td>
<td>SiC</td>
<td>20</td>
<td>3</td>
<td>150/150</td>
<td>Co-Rotating</td>
<td>Smallest steady laminar stream</td>
</tr>
<tr>
<td>1st Polish</td>
<td>1 μm</td>
<td>Final A pad with Al₂O₃</td>
<td>20</td>
<td>6</td>
<td>150/150</td>
<td>Co-Rotating</td>
<td>~ 1 drip/sec</td>
</tr>
<tr>
<td>Final Polish</td>
<td>0.5 μm</td>
<td>Final A pad with Al₂O₃</td>
<td>20</td>
<td>6</td>
<td>150/150</td>
<td>Co-Rotating</td>
<td>~ 1 drip/sec</td>
</tr>
</tbody>
</table>

* Preformed this grind twice, changing the pad each time.

3.4 Void analysis

An evaluation was done to determine if the microscope magnification selected would bias the void analysis due to their resolution of certain size voids and field of view effects on void count. Images were taken of the un-etched samples. Images of an as-built and a HIP processed specimen were compared to investigate the effects on void density frequencies taken over a range of magnifications. Six images of each transverse and longitudinal orientation at the same relative locations of both samples, as seen in Figure 13, were taken at 5x, 10x, and 20x magnification using a Zeiss Vert A1 light microscope.
The images were analyzed using Image J, an image processing software, to identify the void count and size of each void detected from 4 pixels in size up to infinity. The void size frequencies were normalized and each magnification compared to one another as shown in Figures 14. The 10x magnification was chosen after reviewing the results based on range of sizes and count of voids that were detected.

![Normalized Data Comparison](image)

*Figure 14: Normalized void size frequencies*

Each of the ten samples listed in Table 9 had both transverse and longitudinal orientations imaged in relatively the same location. The locations on the macros that the images were taken are shown in Figure 13.
Table 9: List of specimen types analyzed for void analysis

<table>
<thead>
<tr>
<th>Void Analysis Specimens</th>
<th></th>
</tr>
</thead>
<tbody>
<tr>
<td>As built with Argon Shield</td>
<td></td>
</tr>
<tr>
<td>As Built without Argon Shield</td>
<td></td>
</tr>
<tr>
<td>With HIP with Argon Shield</td>
<td></td>
</tr>
<tr>
<td>With HIP without Argon Shield</td>
<td></td>
</tr>
<tr>
<td>With HIP with Argon Shield with Heat treat</td>
<td></td>
</tr>
<tr>
<td>With HIP without Argon Shield with Heat treat</td>
<td></td>
</tr>
<tr>
<td>Without HIP with Argon Shield</td>
<td></td>
</tr>
<tr>
<td>Without HIP without Argon Shield</td>
<td></td>
</tr>
<tr>
<td>Without HIP with Argon Shield with Heat treat</td>
<td></td>
</tr>
<tr>
<td>Without HIP without Argon Shield with Heat treat</td>
<td></td>
</tr>
</tbody>
</table>

The macro layout shown in Figure 13 from section 3.4 show locations 1 through 6 are the transverse orientation and 7 through 12 are the longitudinal orientation. Image J software was utilized to aid in the count and sizing of the voids in the specimens.

3.5 Macrostructure Characterization

After completing the void analysis, the specimens were etched as discussed in section 3.4. Images of the specimens were taken using a Zeiss Vert A1 light microscope. Representative macrostructures from those listed in Table 9 were imaged in the longitudinal and transverse orientations as shown in Figure 13. Images and findings are discussed later in the results section 4.2.
3.6 Tensile Testing

Samples were machined at NASA MSFC into round sub-sized specimens in compliance with ASTM E8 as shown in Figure 15.

![Figure 15: Round sub-sized tensile specimen mechanical drawing](image)

There were four kinds of samples tested as summarized in Table 4. Half the samples used argon purge during the build. Of these, half of each subsequently went through a HIP process. All 28 samples were SR and subsequently HT per ASM 5664.

The tensile axis of the specimens was aligned with the build plane (X-Y). All tensile tests were conducted on an Instron 5985 electro-mechanical machine. Stress measurements were based on loads obtained from the machines 250 kN load cell and specimen dimension measurements. All tests were run in displacement control at a constant crosshead velocity of 0.00196 mm/min (0.050 in/min) in accordance with ASTM standard E8.
3.7 High Cycle Fatigue Testing

Out of the 28 total samples that were deposited for this study, 15 were machined at NASA Marshall Space Flight Center in compliance with ASTM E466 to be high cycle fatigue (HCF) tested as summarized in Table 4. The specimen geometry is shown in Figure 16.

The samples were designated to “Lots” to designate whether they had an argon purge during the build and whether they went through a HIP cycle. All 28 samples were SR and subsequently HT per ASM 5664.

The HCF testing was performed on a MTS Landmark servo-hydraulic machine equipped with a 100kN load cell. The HCF tests were run in axial force control at 10 Hz with $R = 0.1$ in a room temperature environment. The stress levels tested were 689.5 MPa, 551.6 MPa, and 275.8 MPa. These test values were chosen based from the tensile results.
to not exceed the yield strength of the material and to also attempt a life time endurance measurement.

Fracture surfaces of all HCF specimens were imaged using a Keyence VHX-6000 system. This optical microscope has a large depth-of-field to obtain 3D images in addition to the height measurements.
CHAPTER IV

Results

4.1 Void Analysis

The void count or frequency, and sizes of the specimens were analyzed for each post process, HIP and simulated HIP, as well as in the as-built condition. Statistical analysis was performed to compare the void size ranges to each type of post process the specimens encountered in order to gain insight into the impact the post processes have on void size and frequency. Figure 17 shows the transverse orientation of the specimen and provides the normalized relative frequency of voids in range bins from zero $\mu m^2$ to greater than $1000 \mu m^2$. This graph shows that the majority of the void sizes are in the five to nine $\mu m^2$ range. This trend can be seen for the longitudinal orientations as well as shown in Figure 18 indicating spherical voids. Figure 19 shows the two orientations on the same plot to see how the void size differ in orientation.
Figure 17: Transverse orientation of void area relative frequencies

Figure 18: Longitudinal orientation of void area relative frequencies
As shown, the void sizes do not seem to be affected by orientation as each of the orientations, when comparing like samples, only differ within 3% of each other except in the 0-4μm² range where there is a 5% difference between the non-HIP longitudinal specimens and the longitudinal specimens that underwent HIP. This graph also shows that the majority of the voids in the samples analyzed are in the five to nine micrometer range. Further statistical analysis was performed in order to gain better insight into the effects of the post processes on the void sizes. Figure 20 shows a box and whisker distribution of the void’s areas in each orientation.
It can be seen that the samples that have gone through the HIP process show a smaller mean area and overall smaller size distribution in comparison to as-built and the simulated HIP specimens. It is interesting to note that the as-built mean size and HIP mean void sizes are very similar.

As summarized in Table 5, some of the specimens analyzed were deposited with Argon shielding gas and some without. The void size and relative frequency of the shielded vs. non-shielded specimens were statistically analyzed to see if there were any differences in the voids that may be attributed by the shielding gas. As shown in Figure 21, the void sizes and frequencies appear to be within 5% of each other in every bin.
4.2 Metallography

Metallography analysis of the as-built macrostructure show that the macrostructure has a large portion of dendritic formations in it. These dendritic formations are irregular throughout each position of both orientations and can be seen highlighted within the yellow circle of Figure 22. This dendritic macrostructure is seen in both orientations of the as-built samples. Figure 23 shows the as-built sample in the longitudinal orientation.
Figure 22: As-built no shielding gas macrostructure in transverse orientation at position 3 showing dendritic morphology

Figure 23: As-built no shielding gas macrostructure in longitudinal orientation at position 11 showing dendritic morphology
When comparing the two previous images which were deposited without argon shielding gas to the macrostructure of a sample that was deposited with the shielding gas, the macrostructure appears different. Figure 24 and Figure 25 show images of the macrostructure from the same position as Figure 22 and Figure 23 but on samples deposited with argon shielding gas.

![Figure 24: As-built with shielding gas macrostructure in transverse orientation at position 3](image)

Comparing the images, it can be seen that there is a significant difference in the macrostructure especially in the longitudinal orientation. There appears to be a reduction in the dendritic morphology in the macrostructure when the samples were deposited with the shielding gas.
Further reduction of this dendritic morphology can be seen in the specimens that underwent the HT process. Figure 26 shows a transverse orientation of a specimen that only underwent the HT process and was deposited with shielding gas. The HT process shows to have refined the material’s grain structure to a point where the grains are more pronounced in comparison to the as-built macrostructure. Figure 27 shows the longitudinal orientation of a non-HIP with shielding gas and HT sample.
Figure 26: No HIP with shielding gas with HT transverse orientation at position 3

Figure 27: No HIP with shielding gas with HT longitudinal orientation position 10
The images would suggest that grain refinement has taken place as a result of the HT process in the specimens. When comparing the non-HIP samples that underwent HT and were deposited with shielding gas to those that underwent the same post processing but were deposited without shielding gas, negligible differences were observed with the macrostructure. Figure 28 shows a non-HIP with HT and deposited without shielding gas.

![Image](image.png)

**Figure 28**: No HIP no shielding gas with HT transverse orientation position 3

Other than the cracks that are evident in the image, the appearance of the macrostructure is comparable to that of the ones that were deposited with shielding gas. Some samples were processed with HIP only. A representative image of the macrostructure of those samples can be seen in Figure 29.
Without the HT, it can be seen that the grain structure is not as defined in the transverse orientation as those in HT samples. Although, in the longitudinal orientation, there is some grain refinement. Figure 30 shows the grain structure that formed as a result of the HIP process.
The dendritic structure as seen prominent in the as-built macrostructure has been reduced after the HIP and HT processes.

The specimens that went through the HIP and heat treatment show the same reduction of the dendritic morphology as seen in the HIP only specimens. Figure 31 shows disperse circular structures scattered throughout the image. The longitudinal orientation shows more pronounced grain structure and less of the circular structures observed in Figure 32.
Figure 31: With HIP with shielding gas with HT transverse orientation position 1

Figure 32: With HIP with shielding gas with HT longitudinal orientation position 7
4.3 Tensile Testing

Samples with different post and build processes were tensile tested to evaluate their mechanical properties. The Ultimate Tensile Strength (UTS), yield strength (YS) and elongation results are shown in Figures 33, 34, and 35 respectively. Each plot compares the different processed specimens against the reported wrought properties [17].

![Ultimate Tensile Strength](image)

*Figure 33: Ultimate tensile strength results compared to wrought properties*

The results show that the HIP process combined with argon shielding gas are close to near net wrought strength properties. The samples without the HIP process and that were deposited without shielding gas produced the lowest UTS after testing.

The yield strength of the material after testing shows again that the samples that were deposited with the shielding gas and underwent HIP as part of the HT process have a greater yield strength then the other samples, comparable to wrought properties [17].
The elongation results of the material are shown in Figure 35. The samples that did not go through the HIP post process have a reduction of about 6% in elongation in comparison to the samples that did receive the HIP post process. The use of Argon purge during the build did not appear to affect the elongation. The samples that received the HIP post process show elongation results in line with wrought properties from the literature.
Table 10: Summary of tensile test results

<table>
<thead>
<tr>
<th>Blown Powder</th>
<th>UTS (MPa)</th>
<th>Elongation (%)</th>
<th>Yield (MPa)</th>
</tr>
</thead>
<tbody>
<tr>
<td>With HIP</td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td>No Argon</td>
<td>1169.7 ± 13.6</td>
<td>12.3 ± 0.26</td>
<td>952.3 ± 24.9</td>
</tr>
<tr>
<td>Argon</td>
<td>1264.0 ± 24.9</td>
<td>13.0 ± 1.87</td>
<td>1092.1 ± 33.1</td>
</tr>
<tr>
<td>No HIP</td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td>No Argon</td>
<td>1139.7 ± 44.4</td>
<td>6.8 ± 3.21</td>
<td>1003.4 ± 34.3</td>
</tr>
<tr>
<td>Argon</td>
<td>1185.7 ± 29.8</td>
<td>5.9 ± 0.78</td>
<td>1043.2 ± 34.8</td>
</tr>
<tr>
<td>SLM with HIP</td>
<td>1395.7 ± 4.2</td>
<td>23.6 ± 0.36</td>
<td>1110.9 ± 7.4</td>
</tr>
<tr>
<td>Wrought</td>
<td>1275.5</td>
<td>29</td>
<td>1034.2</td>
</tr>
</tbody>
</table>

4.4 High Cycle Fatigue Testing

HCF testing was conducted on the four lots of samples previously described in Table 5. A stress vs number of cycles (SN) curve was produced from the results of the HCF testing as show in Figure 36. As shown in the SN curve, the samples that received the HIP process are trending to higher cycle counts before failure or the intentional cut off of
10^7 cycles. It can also be seen that there are more specimens tested at 690 MPa than the 550 MPa or 275 MPa mark.

**Figure 36**: SN curve generated from the HCF results

**Table 11**: Summary of HCF results

<table>
<thead>
<tr>
<th>Shielding Gas</th>
<th>Stress Level at R = 0.1</th>
</tr>
</thead>
<tbody>
<tr>
<td>Lot 2 With HIP None</td>
<td>1.5 x 10^5 - 4.0 x 10^5</td>
</tr>
<tr>
<td>Lot 1 Argon</td>
<td>3.4 x 10^4 - 1.0 x 10^5</td>
</tr>
<tr>
<td>Lot 4 No HIP None</td>
<td>6.0 x 10^4 - 1.2 x 10^5</td>
</tr>
<tr>
<td>Lot 3 Argon</td>
<td>1.9 x 10^4</td>
</tr>
<tr>
<td>SLM with HIP</td>
<td>2.4 x 10^5 - 1.0 x 10^7*</td>
</tr>
</tbody>
</table>

* Intentionally terminated
The decision was made to use the samples that were to be tested at the 415 MPa mark and test them at 690 MPa due to the Lot 1 sample tested failing at around where the non-HIP samples did. With the limited number of samples to test it was important to gain more confidence in the HIP process specimen’s trends after seeing the other stress levels trends.

When looking at the Lot 1 HIP samples and comparing the 690 MPa stress level to the 550 MPa level, the two fracture surfaces seem very similar in appearance. Figure 37 and 38 show the fracture surfaces of specimens built with argon shielding and underwent the HIP process that were cycled at 690 MPa and 550 MPa, respectively.

Figure 37: Lot 1 690 MPa HCF fracture surface
The small black dots scattered on the surfaces of Figures 37 and 38 are voids in the material from the deposition process. The surfaces have a jagged topology indicative of a moderately ductile failure. Figure 39 shows an orthographic view of the fracture surface of a specimen that was deposited with argon shielding gas, underwent the HIP process, and tested at the 690 MPa stress state.
Figure 40 shows the fracture surface of the Lot 3 690 MPa specimen of which also had argon shielding gas during the build, but was not subjected to the HIP process. There is a clear line in the fracture surface in Figure 40. This surface had a very different topology as shown more clearly in an orthographic view of the surfaces topology in Figure 41. The 690 MPa Lot 3 fracture surface shown in Figure 40 was the only specimen displaying this type of fracture surface. The other specimens that failed had a fracture surface very similar to that Figure 39.
Figure 40: Lot 3 690 MPa HCF fracture surface

Figure 41: Orthographic surface plot of Lot 3 690 MPa fracture surface
CHAPTER V

Discussion

5.1 Void Analysis

The void analysis of all the specimens analyzed show that roughly 43% of the void sizes are in the range of 5 to 9µm². The analysis also shows that there is a reduction in void size when the specimens go through a HIP post process. Figure 42 coincides with what the statistical data shows that there is a reduction of void size with the specimens that underwent HIP when comparing to the as-build specimens.

Figure 42: As-built transverse orientation position 3 (a) vs HIP transverse orientation position 3 (b)
While there is a reduction in void size for the specimens that underwent the HIP post process, there is a slight enlargement to the void sizes in the specimens that only underwent the HT post-process. It can be seen in Figure 20 that the non-HIP specimens, which went through the HT only, show a 10% increase in their median void sizes. It is speculated that the increase in the void size for these specimens is related to the expansion of trapped gas inside the voids. This speculation was drawn from investigating the void analysis images. It can be seen in Figure 43 that some of the voids have what appears to be white spots in the bottom center. This is the reflection of light from the microscope and means that the void has a relatively smooth interior. Voids with smooth interior surfaces typically result from trapped gas. These reflections have been reported to correlate with the presence of trapped gas [27].

*Figure 43: No-HIP with shielding gas no HT transverse orientation position 1 showing light reflection indicating trapped gas*
To explore the sources of trapped gas in the specimens, the void sizes and relative frequencies of the samples that were deposited with and without argon shielding gas were analyzed. Figure 21 shows about a 5% difference between the void sizes in the shielded vs. un-shielded specimens, thus it doesn’t appear that gas shielding is related to the voids observed. It has been reported in the literature that argon gas can become trapped in the powder stock itself due to the atomization process used to manufacture the powders [28-29]. The size of the void in Figure 43 was measured to determine if the powder used could have contained trapped gas as a result of the gas atomization manufacturing process of the powder. The void size in Figure 43 was 26µm in diameter. Thus it is feasible that this void could result be from trapped gas in the starting powder feedstock. As shown in Figure 44, SEM images of mounted and polished Inconel 718 feedstock from the literature show interior voids roughly 10µm in diameter which are believed to contain argon. Figure 44 (a) shows the results of the author’s investigations and that voids were already in the powder feedstock. Figure 44 (b) also show that the spherical voids in the feedstock powders can be different sizes where in some cases 38.6 vol.-% of the powder particle is hollow [30]. The powder size used for the deposition of the specimens in this study ranged from 46 to 106µm in diameter. With this larger powder size distribution (PSD) relative to ones used in SLM (15 to 56 µm), there is greater potential for the feedstock to contain voids due to the increased surface area [31]. Knowing that any gas trapped inside a void would expand during HT, this gas expansion is believed to be related to the increase in size after the HT process without HIP. Thus it is believed that the void from Figure 43 is most likely due to trapped gas inherent in the argon gas atomized feedstock used. Voids in the specimens that had undergone HIP most likely were decreased in size and then grew during
the subsequent HT. This raises a concern that the standard practice to use HIP to reduce void size will be countered by expansion during subsequent heat treatments. Thus it is important to understand the effect of the void size distribution on the resulting mechanical properties as voids can occur in AM fabrication.

![SEM image of Inconel 718 feedstock powder mounted in resin and polished (cross-section)](image)

**Figure 44: SEM image of Inconel 718 feedstock powder mounted in resin and polished (cross-section)** [30]

Additionally, if there is some level of trapped gas in the specimens, this could lead to increased internal stresses due to thermal expansion during the HT. Thus both the larger void size in the HT only specimens in addition for the potential for increased internal stresses could both be responsible for the decreased elongation property.
5.2 Macrostructure

The as-built macrostructure of the specimens show a predominant dendritic morphology in both orientations of the build. This dendritic morphology is reduced in the specimens that underwent the HIP and other HT processes, even though the processing time and temperatures were held constant. It is also interesting to note that the specimens deposited with argon shielding gas also had reduced dendrite morphology. This could be due to the quenching effect of the argon which would result in faster solidification. Following the HT processes, evidence of the grain structure becomes more pronounced.

5.3 Mechanical Properties

The tensile testing results show that the effects of not using argon shielding gas can have adverse effects on the material properties of the specimens. The specimens that were deposited with shielding gas and received the HIP post process have improved material properties when comparing across the specimens to the wrought published properties. When analyzing the elongation however, there was a ~5 to 6% drop in elongation of the non-HIP specimens in comparison to the ones that received the HIP post process. It is speculated that this drop in elongation is due to the increased void size of the specimens who only went through the HT post process. Void analysis shows that the void size increased in the specimens who only went through the HT post process. This increase in void size could be the cause of the drop in elongation due to its effects of reducing the effective total area inside the material. The results of the tensile testing also show that the
specimens that were deposited with the argon shielding gas and when through the HIP and HT post process have strength, yield, and elongation properties that are on par with those of published wrought material.

It has been shown in the literature that the particle size distribution (PSD) may play a factor in the mechanical properties of components processed with a BPLD system. The material’s mechanical properties can vary with PSD in SLM processed components and results from experiments show that a narrower PSD shows significant improvements to both mechanical properties and surface finish [31-32]. However, it should be noted that the powder size used for manufacturing the material specimens for this study were larger than that typically used in SLM. So, the presented comparison between the mechanical properties of the samples manufactured by BPLD and SLM does not consider the effect of PSD on mechanical properties.

The results of the HCF test show a trend that specimens who have received the HIP post process have longer cyclic lives than those which did not receive the HIP post process.
CHAPTER VI

Conclusions

The following conclusions regarding the material characterization of Inconel alloy 718 can be drawn from the results and discussion chapters 4 and 5 respectfully prior to these conclusions.

1. Void analysis shows that the HIP post process causes a 4% reduction in void size, while the specimens who only underwent HT showed an increase of 5% in void size when compared to as-built specimens.

2. The as-built macrostructure of the material has predominant dendritic morphology in both the transverse and longitudinal orientations. These dendritic structures are reduced and the material’s grains start to become defined after the HT processes.

3. The ultimate strength, yield strength, and elongation of the specimens that had undergone HIP as part of the HT process, have material properties that are on par with wrought properties published.
4. Use of argon shielding in the fabrication of the specimens also had higher UTS and YS, but not much change in the elongation.

5. Elongation of specimens that did not undergo HIP were roughly 6% lower than the specimens that did undergo HIP and wrought properties.

6. Some of the voids in the specimens were observed to have smooth interior surfaces as shown by the reflection of light from the microscope during imaging. Smooth interiors usually result from trapped gas which could result from the gas shielding or initial feedstock. Review of the void size and relative frequencies of the argon shielded vs. non-shielded specimens did not show a correlation. However, since voids have been reported in the literature for larger argon atomized powder feedstock, this is believed to have persisted during AM processing and is related to the voids observed.

7. It is speculated that the drop in the elongation property of the HT only specimens correlates with the larger void size. In addition, expansion of trapped gases within these voids could also increase the internal stresses. Statistical data shows the mean average void size of the HT only specimens is 10% larger than that of the HIP specimens and 7% larger than that of the as-built specimens.
8. The HCF data starts to show a trend that suggests that the specimens that underwent HIP have a higher cycle count before failure in comparison to the specimens that did not receive the HIP post process.

9. The blown powder metal laser deposition or DMD AM process is a very feasible process for the manufacturing of larger components in comparison to SLM build box sizes. The review of the material properties in this thesis that came off a prototype machine with early development build parameters shows great potential for improvement and coupled the process’s larger build box, DMD is a great tool for the AM toolbox.
CHAPTER VII

Future Work

Although using light microscopy is generally a good first step into being able to characterize a material, it is limited in the depth of resolution needed to make definitive conclusions on what is being seen. More in depth microscopy such as with the use of a transmission electron microscope (TEM) and x-ray diffraction would be the next steps to take in order to better understand and characterize the material and definitively record the phases present with this process. Also using techniques such as Electron Backscatter Diffraction (EBSD) could be used with in situ heating to study the transformations in the material such as, recrystallization and precipitate formation. The knowledge gained from this study could aid in the understanding of the microstructural evolution of the material and by knowing how the material behaves at certain temperatures, the data could be used to aid in improved deposition parameters.

With only a limited number of samples in which to fatigue test, more specimens to HCF test would be great in order to draw more definitive conclusions about the trends of the material’s behavior when comparing the post processes. Early on with the HCF testing it was seen that some of the HIP specimens failed with cycle times that were fewer than those that did not receive the HIP post process at higher stress states. This finding caused
the samples that were slated for 413.7 MPa testing to be used at 689.5 MPa in order to gain more data points at this higher stress state in the attempt to gain more insight into why some HIP specimens failed before the non-HIP specimens. With groups such as the turbomachinery designers as NASA Marshall Space Flight Center developing turbo pumps that exceed 90,000 rpm and are made with AM processes, more HCF data could potentially give those designers better design limitations in order to refine their designs. More in depth fracture surface analysis could also aid in determining any effects that void sizes had on the material during fatigue testing and potentially aid in better deposition parameters.

Powder characterization such as particle morphology, PSD, density, and flowability would be a great follow up study to this thesis. BPLD typically uses larger powder sizes than that of SLM and seeing as the PSD has effects on the mechanical properties, density, and surface roughness of SLM processed components, experiments looking into the effects of PSD on BPLD would lead to knowing whether or not BPLD is affected by the varying sizes and if so, how.

Finally, with the roughly 6% reduction in elongation between the HIP and the HT only specimens during tensile testing, an investigation into why the reduction took place would be beneficial. After reviewing images that showed there was gas trapped in some of the voids and researching its effects on mechanical properties as well as how other AM process suffered due to varying PSDs, determining if one or both had effects on the material would be beneficial to future studies.
References


