CHARACTERIZATION OF PROCESS INDUCED DEFECTS IN LASER POWDER BED FUSION PROCESSED ALSi10MG ALLOY

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CHARACTERIZATION OF PROCESS INDUCED DEFECTS IN LASER POWDER BED FUSION PROCESSED ALSI10MG ALLOY

by
Edward J. Stugelmayer

A thesis submitted in partial fulfillment of the requirements for the degree of

Masters of Science in Metallurgical and Materials Engineering

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Abstract

Additive manufacturing using laser powder bed fusion (AM-LPBF) methods have recently experienced rapid growth and development, having the potential to replace manufacturing by plastic deformation, precision machining, or casting. AM offers advantages such as the freedom to design highly complex geometries, time and cost savings through material usage efficiency and shortened production cycles, and the potential for improved mechanical properties. Process induced defects, however, result in degradation and scattering of mechanical properties and hinder the widespread adoption of AM-LPBF in industry. This investigation focuses on the effects of varying energy density and build orientation on the evolution of process induced defects within an AlSi10Mg alloy produced using AM-LPBF. The area percentages and morphologies of the porosity is then related to the tensile and fatigue properties exhibited by the AlSi10Mg specimens. The area percentage of porosity within specimens was found to increase as the energy density of the build increased. The morphology of pores at higher energy densities also became more spherical in shape, suggesting excessive energy density caused the entrapment of gas bubbles within the melt. Lower energy densities resulted in a lower percentage of porosity although the general morphology of these pores appeared jagged or faceted in shape; likely resulting from lack-of-fusion defects during the build. Tensile properties including ultimate tensile strength (UTS) were found to increase as energy densities and resulting porosity decreased. Fatigue properties, however, were highest in specimens with the highest percentages of porosity despite having lower tensile strengths. This may derive from the spherical morphology of entrapped gas porosity within these specimens producing less of a stress concentration when compared to faceted lack-of-fusion porosity.

Keywords: Additive Manufacturing, AlSi10Mg, process parameters, porosity characterization, mechanical properties, fatigue
Dedication

I’d like to thank my friends and family for supporting me throughout my education. I would also like to thank the TRiO offices at Montana Tech. If not for the TRiO Upward Bound program, continuation of my post-secondary education would not have likely been possible.
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My graduate committee consisted of: K.V. Sudhakar, Bruce Madigan, Ronda Coguill, and Brahmananda Pramanik. Bryce Abstetar, Penn Rawn, Luke Suttey, and Steven Keckler were my fellow graduate researchers. Ronda Coguill and Taylor Winsor conducted mechanical testing.
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1. **Introduction**

Conventional manufacturing methods for metal materials can produce a wide variety of components for numerous applications. However, several limitations are inherent to traditional manufacturing methods which can constrain process efficiencies and design freedoms. Recently developed techniques known as additive manufacturing (AM) have the potential to replace manufacturing by plastic deformation, precision machining, or casting [1]. AM is a method where material is “added” or built up until the component takes on the desired shape; whereas subtractive manufacturing is the more conventional method involving removal of material from some initial mass to reach a desired shape.

Additive manufacturing methods have recently experienced rapid growth and development, especially utilizing polymeric materials in processes known as “3D-printing”. In 3D-printing, polymeric material is heated until it becomes fluid and deposited as a layer with some predetermined shape. The process is repeated layer-by-layer until a desired 3D structure is formed. Utilizing metal materials in additive manufacturing has always presented a challenge, however, due to metals typically requiring much higher temperatures to either be deformed or melted [2]. The materials of construction for the building process must be able to handle the high temperatures achieved when melting metals. The deposition mechanism must also maintain the accuracy and resolution required by the operator.

One method of additively manufacturing metal materials is known as laser powder bed fusion (LPBF). In the LPBF process, a thin layer of metal powder is spread across a level bed where it is selectively melted using a laser light source. The laser scans the bed until a specified 2D shape is solidified within the powder bed. Similar to conventional 3D-printing, this process is repeated layer-by-layer until the agglomerated layers represent the final 3D shape. This process
is analogous to processes known as selective laser melting (SLM), direct metal laser sintering (DMLS), and electron beam melting (EBM) [3].

AM of metal materials offers a multitude of advantages to designers and manufacturers that make it an attractive alternative to conventional manufacturing techniques. One of the biggest advantages arising from AM-LPBF is the ability to design components with highly complex geometries previously impossible to produce using conventional methods [1,4-20]. Geometries such as hollow structures, internal lattices, integral cooling channels, and those generated through topographical optimization can all be achieved [8,15,16]. These near net shape designs also allow for time and cost savings as product development cycles are shortened and components can be manufactured without part specific tooling [2,7,11,18,21]. Another advantageous aspect of AM-LPBF is the efficient use of materials arising from minimized raw material requirements and scrap production [2,4,6,11]. Unused metal powders within the build chamber can even be recycled if the proper precautions are taken to remove agglomerates and prevent deterioration of the powder’s properties [2].

One exciting advantage that AM-LPBF could facilitate is improved mechanical properties compared to conventionally manufactured wrought alloys. These properties derive from the novel, highly-refined microstructure resulting from exceedingly high temperature gradients (G) and solidification rates (R) reached using laser melting [1,2,4,8,9,17,18,22]. Microstructural features including grain refinement, extended solid solubility, chemical homogeneity, metastable phase formation, and phase segregation reduction can all be observed within components produced using AM-LPBF [2,8,14].

However, several disadvantages exist with AM of metals materials that inhibit its widespread application in producing engineering components. Components produced in LPBF
processes typically have poor surface quality when compared to other production methods; often exhibiting a surface with attached powder particles lacking complete fusion and a ridged surface arising from subsequent scan tracks [3,15,17,19,21-24]. Dimensional accuracy can also be a disadvantage if the component experiences distortion caused by excessive temperature gradients, large thermal expansions, and the accumulation of internal stresses [2,12,15,17,25,26].

Possibly the biggest disadvantage to AM of metal materials is the propensity to generate process induced defects; where internal defects dictate the resulting physiochemical properties of a metal material [2-4,6,7,9-14,18-22,25-28]. The term “defect” is used in this publication synonymously with the term “discontinuity” to describe internal features found within the specimens. Deleterious features such as entrapped gas porosity, lack-of-fusion porosity or bonding defects, secondary phase particles, or oxide inclusions can all be found in AM-LPBF specimens [21,25]. These defects often result in degraded or unpredictable physiochemical properties which present one of the largest challenges preventing widespread implementation of AM of metal materials [2,20,23-25,29].

Process induced porosities are some of the most prevalent defects found in metal components produced by AM-LPBF. The resulting mechanical properties of AM specimens are sensitive to the degree, shape, and location of pores; as they reduce the load bearing area, act as stress concentrations, and crack initiation sites [19,21]. Porosity is particularly significant to the fatigue performance of AM components as fatigue is highly sensitive to the presence of defects [3,4,7,19,21,25]. Numerous investigators report porosity as well as surface roughness to be the most critical factors controlling the fatigue performances in AM-LPBF specimens [4,19,21,25]. The term “porosity” is used throughout this publication to describe internal hollow spaces within the specimens, where lack of fusion porosity is synonymous with the term “void”.
In this study, metallographic image analysis was employed in order to characterize the evolution of porosity in AlSi10Mg alloy specimens produced using AM-LPBF. The effects of varying build parameters such as energy density, orientation, and location on the total evolution and morphology of pores within AlSi10Mg specimens was of primary interest. The overall degree of porosity was determined as area percentages to approximate the overall densification achieved during the build. The overall morphology of the porosity found with AM-LPBF specimens was also characterized; which can give insight to melting modes, porosity generation mechanisms, and previous processing regimes. The characteristics of porosity found in the AlSi10Mg specimens constructed using varying build parameters is then related to both the tensile and fatigue performances exhibited by these materials.
2. Background

Some of the first techniques for AM, such as stereolithography, became commercially available in the late 1980’s and were used for producing models and prototypes [26]. One of the early methods for AM of metal materials was known as selective laser sintering (SLS), where powder densification occurred via laser heating of powders followed by neck formation at the contact points [21,26]. Early attempts of SLS were not very successful due to high melt viscosities, resulting in capillary instability or balling caused by limited liquid formation. Laser melting was subsequently developed by the demand to produce fully dense parts comparable to bulk materials while minimizing post processing requirements [26]. Evolving from SLS, processes such as SLM or LPBF became possible due to improvements in powder qualities, laser technology, and machine systems. These advancements and the continual development of AM of metal materials have allowed a shift from rapid prototyping to rapid manufacturing [7,21,26].

In AM-LPBF, a thin layer of powder in uniformly spread over a substrate where it is selectively fused using a laser energy source. The laser scans over a 2D contour, corresponding to a 2D slice from a CAD design, to melt and fuse the desired layer. The build plate can then be lowered approximately one layer thickness, a new layer of powder is distributed, and the process repeats until the desired 3D components have been produced [7].
Complex interactions and coupling mechanisms between the laser, powder bed, and atmosphere must be understood to maximize the potential for AM of metal materials. Some of the physical phenomena taking place during LPBF processes include [27]:

- heat transfer
- chemical reactions
- phase transformations
- absorption and scattering of laser radiation
- evaporation and emission of material
- fluid flow driven by surface tension gradients (Marangoni effect)

In addition to the numerous phenomena occurring, AM-LPBF employs a wide variety of process variables that can be adjusted in attempt to optimize the build. The influencing effects must be quantified to interpret process errors and part failures [24].

2.1. AlSi10Mg Alloy

Aluminum alloys are the second most utilized metal in the world behind steel, being used extensively in the automotive and aerospace industries [1,4,21]. These alloys are desirable due to properties such as: excellent strength to weight ratios, corrosion resistance, thermal and electrical conductivities, formability, and aesthetics [1,4,21,25]. The AlSi10Mg alloy is similar to an A360 Al casting alloy (9.5%Si-0.5%Mg-<1.5%Fe); exhibiting high fluidity, low shrinkage, and good weldability. These alloys are considered as hypoeutectic alloys, with silicon contents below the Al-Si eutectic point at 12% Si [25]. The AlSi10Mg alloy also contains magnesium, which can form Mg2Si precipitates and potentially improve ductility and strengthen the matrix. These precipitates are generally not seen in AM-LPBF specimens due to rapid solidification rates, but can be formed via a T6 heat treatment [5]. The near eutectic composition of these alloys imparts
low solidification temperature ranges, high fluidity, and therefore good weldability. These characteristics make this alloy system quite suitable to processing via AM-LPBF.

2.2. Powder Material

Powders produced for AM-LPBF significantly differ from those typically produced for use in powder metallurgy (PM). AM-LPBF typically employ finer particles (<50μm) than those used in PM (>150μm), as well as employing a narrower size distributions [30]. AM powders are nearly spherical in shape with relatively smooth surfaces, providing improved flowability and bed packing densities [2,8,21,27,30,31]. Differences in bed packing densities can affect the transmission of energy into the surrounding areas and the resulting melt pool characteristics [21,31].

Metal powders are also highly susceptible to contamination by moisture, organics, absorbed gasses, or oxide and nitride surface films; arising from exceedingly high surface areas per unit volume [21]. Surface contaminants are often complex, exhibiting sesquioxides, carbonated hydroxides, and moisture [4]. Moisture on the powder surfaces results in poor flowability which can inhibit uniform layer deposition [2,4,6]. The powder surface chemistry has also been suggested to be a significant factor in reducing porosity within build specimens. One mechanism involves the melt becoming supersaturated with hydrogen gas from the powder surface contaminants, which then solidifies as entrapped gas pores [4,6]. Another porosity formation mechanism is through the introduction of oxides into the melt, which are thought to prohibit wetting and initiate capillary instability or balling [2,4]. These issues are typically mitigated by employing a powder drying regime as well as processing under an inert atmosphere [2,4,6,21].
2.3. Energy Density and Melting

There has been a great effort in past years to reduce the porosity within specimens produced via AM-LPBF; mainly focusing on some of the most influential processing parameters such as: laser power, hatch spacing, scanning speed, layer thickness, and scanning strategy [4,27]. Similar to fusion welding, these parameters are used to calculate the energy input into the system in the form of a global energy density (GED), seen below in Equation 1 and Equation 2

\[
GED_V = \frac{P}{vht} \\
GED_A = \frac{P}{vh}
\]

where GED\(_V\) is the global energy density by volume using laser power (P), scan speed (v), hatch spacing (h), and layer thickness (t); and GED\(_A\) is the global energy density by area which omits the layer thickness [1,9,10,13,18,21,26]. This provides an initial means of quantifying the building process which can then be used in process optimization. However, these parameters can exhibit interactions with each other as well as process parameters not included in the energy density calculation such as process atmosphere and pressure, scanning strategy, substrate preheating, or laser beam size [2,4,6,10,13,21,26,27].

The laser melting process can be broken into two steps, initial surface melting followed by heat conduction to the powder particles below the liquid. The thermal conduction competes with the penetration of the liquid metal, as liquid penetration is driven by capillary forces from the top of the powder layer [2]. Numerous researchers have reported an increased densification of AM components resulting from an increase in energy density; and the presence of lack-of-fusion porosity at insufficient GED’s [1,4,7,9,21]. Aluminum alloys also require high laser powers for melting due to losses from high reflectivity and thermal conductivity [27]. However, excessive energy densities can result in selective vaporization of alloying elements. Selective vaporization coupled with scan speeds can generate chemical gradients that may
change the sign of the surface tension coefficient of the melt (Marangoni flow) and the resulting mode of convection [6]. Marangoni flow is “the initiation of thermocapillary forces for fluid flow as a consequence of the temperature gradient in the melt pool, giving rise to differential surface tension between the edge and center of the melt pool” [21]. During Marangoni flow, the temperature dependent surface gradient will drive melt flow from the hotter center to the cooler rear and sides of the melt pool [28].

The presence of selective vaporization typically signifies the transition from conduction mode melting, where melt pools are relatively hemispherical, to keyhole mode melting which exhibits deeper conical melt pools. This transition from conduction mode to keyhole mode melting is governed by an energy balance between absorptivity and conductivity [27]. In keyhole mode melting, the selective vaporization of alloying elements generates recoil vapor pressures which add extra forces to the surface of the melt. These additional forces result in the formation of a depression below the laser [28]. The depression formed by the recoil vapor pressures can then form a vapor cavity that enhances laser absorption, allowing the beam to “drill” into the bed and achieve far deeper penetration than achievable with conduction mode melting. The collapse of the vapor cavity can then result in porosity entrapment in the tail of the melt pool [32].

The Marangoni and recoil forces present at excessive energy densities are considered as the main driving forces for melt flow instabilities like balling or spattering [28]. Generally, bonding defects or lack-of-fusion porosity is observed at lower energy densities while entrapped gas porosity is exhibited with excessive energy densities [7,26].

2.4. Processing Atmosphere

A controlled process atmosphere is typically used to shield the build substrate from reactive gasses and deleterious reaction products in order to prevent undesirable reactions such
as oxidation or nitration of powder surfaces [21,33]. This may be advantageous in some situations such as processing aluminum alloys under a nitrogen atmosphere to form aluminum nitrides [21]. The process atmosphere can change during the build process, however, arising from the vaporization of metals or gas impurities evolving from the melted powders [26,33,34]. If the gas flow rate is insufficient to remove by-products in the laser path, welding plume by-products including metal vapors and a plasma plume can have a significant effect on laser beam attenuation and scattering. These by-products can absorb incident laser energy, resulting in lack-of-fusion and balling on the surface of the build [33].

Evaporation of metal can also cause entrainment of material into a smaller volume [34,35]. This entrainment is induced by the extensive evaporation occurring within the laser spot and the pressure drop inside the associated vapor jet resulting from the Bernoulli effect [35]. This vapor driven entrainment has the capability to physically transport powder particles near the laser spot. The vapor flow induces inward ambient gas flow that entrains particles within the melt and results in what is known as denudation [35]. However, powder particles can also be transported outside the track in multiple directions by this gas flow. These powder denudation effects can lead to void defects and layer non-uniformity as denudation zones leave subsequent scan tracks relatively powderless [35]. Insufficient inert gas pressures typically result in only slight remelting, where outward gas expansion causes the ejection of particles and hinders stable track formation. Excessively high inert gas pressures, however, results in inward particle motion arising from the metal vapor jet and Bernoulli-effect driven gas flow which leads to denudation and a lack of powder material in subsequent melt tracks [34,35].
2.5. Scanning Strategy and Residual Stress

AM-LPBF inherently generates large thermal gradients which lead to non-uniform thermal expansions/contractions within the heat-affected-zone (HAZ); resulting in significant residual stresses within the final component [15,21,36-39]. Uneven thermal gradients and residual stresses can also lead to part distortion, susceptibility to crack formation, and the generation of porosity [15,21]. Varying scanning strategies and the local heat distribution can be employed to tailor processing times, microstructures, residual stresses, porosity, and ultimately the final physiochemical properties exhibited by a material [15,20,21,36].

A wide variety of scanning strategies have been employed during the development of AM-LPBF processing, with chessboard and stripe raster scanning being among the most prevalently used strategies [24]. Numerous investigators have observed the maximum residual stress occurring parallel to the scanning vector arising from higher temperature gradients parallel to the scan direction [15,37]. Minimizing the scan lengths for all orientations has been suggested when designing scan strategies in order to mitigate these residual stresses [15,21,37]. Additional contour scanning can also be employed in fatigue critical locations of a component in order to amend defects and minimize porosity. The secondary scanning and remelting of material can approach full densification within a component but will significantly increase production times [7,10,19,21,29,24].

These residual stresses can ultimately result in distortion of the built components and often require the incorporation of support structures or anchoring to prevent distortion and total build failure [15,39,40]. Support structures are metallurgically fused to the component and must be strong enough to resist deformation while also facilitating easy removal for dimensional tolerances [38,39]. The addition of support structures can be costly since they require additional
material and time to produce. Their removal can also prove to be difficult and time consuming, especially when producing complex geometries where structures may be inaccessible [38,39].

Techniques such as build chamber preheating, incorporating eutectic alloy systems, and reduced densification have all been suggested to minimize the requirement for support structures during AM-LPBF [15,39,40]. Preheating the build chamber can reduce the build-up of residual stresses by facilitating diffusional relaxation throughout the component [38]. Utilizing eutectic alloy mixtures as the feedstock material has also been suggested; where preheating temperatures are below the melting point of the feedstock powders but sufficient to prevent rapid solidification and resulting residual stresses within the eutectic melt [15,39]. Another suggested technique involves producing support structures with reduced densification, similar to SLS, which will provide enough strength to prevent distortion while enabling easy removal via sandblasting or similar methods [39].

2.6. Microstructure of AM-LPBF Components

Similar to fusion welding of metals, the size and morphology characteristics of a microstructure obtained using AM-LPBF is governed by the thermal gradients (G) and solidification rates (R) found in the melt pool [5,21]. The morphology of the resulting microstructure is determined the ratio of G/R; where the morphology will change from equiaxed dendritic, to columnar dendritic, to cellular dendritic, to planar grain structures as the G/R ratio increases, respectively [5,21]. The size or fineness of the resulting microstructure is governed by the cooling rate (GxR), where high cooling rates promote undercooling and finer grain sizes [5,21]. The microstructures of components produced using AM-LPBF are characterized by very fine grain structures deriving from high cooling rates (10³-10⁸ K/s) [21,26] and rapid solidification of the melt pool [2,9,18,22,27,29,31]. The rapid solidification observed in
AM-LPBF also inhibits diffusion, allowing the homogenous distribution of alloying elements as well as supersaturation of these elements within the host matrix [22,27].

Solidification and growth in AM-LPBF is also quite similar to that observed in fusion welding; where epitaxial growth at the fusion boundary has minimal free energy requirements and is thermodynamically favorable [21]. Epitaxial growth at the fusion boundary is initiated by arranging atoms in the liquid phase onto the existing crystalline substrate, extending the base metal while maintaining the pre-existing crystallographic orientation [21]. However, grain growth in fusion welding is most favorable along the maximum thermal gradient and therefore any existing crystallographic orientations preferentially aligned with this gradient will dominate [1,21]. Maximum temperatures during laser melting in AM-LPBF are achieved at the center of the top of the melt pool, resulting in competitive growth toward the centerline of the melt pool following initial epitaxial growth [21]. These competitive mechanisms result in a fine cellular-dendritic structure that becomes coarser towards the melt pool boundaries [5,7,21,22]. Heating from subsequent scan tracks will also increase diffusivity and promote coarsening in the melt pool boundaries, forming heat affected zones (HAZ’s) [21].

Microstructures observed in AlSi10Mg alloys produced via AM-LPBF exhibit very fine cellular grains consisting of α-aluminum with a continuous, fibrous network of silicon forming within the cellular matrix [5,7,22,27]. This intercellular network of silicon is interrupted in the HAZ as idiomorphic silicon crystals form due to increased diffusivity upon reheating [5,22]. Numerous other mechanisms can take place during the solidification and growth process, including: dendrite fragmentation, grain detachment, and heterogenous and/or surface nucleation [2].
2.7. Process Induced Defects

One of the biggest challenges facing AM-LPBF is the generation of process induced defects that hinder mechanical properties from meeting engineering quality standards [2,20,23-25,28,29]. Defects can include [21,25,27]:

- porosity
- oxide inclusions
- balling
- loss of alloying elements
- cracking

Process induced porosity is of very high concern as it is the most common defect observed in final components produced via AM-LPBF; significantly affecting the resulting mechanical properties, particularly fatigue performances [3,21,25]. Several porosity formation theories have been suggested, such as [4,28]:

- unmelted powders or satellite introduction
- balling effects or capillary instability
- keyhole formation
- gas entrapment via turbulent fluid flows or nucleation upon reheating

The evolution of porosity requires its own characterization in order to fully understand the effects porosity has on an AM components physiochemical properties. Archimedes method is a very simple method of non-destructive characterization that measures the density of an AM-LPBF component, where relative densification is indicative of the total porosity percentages [20]. However, this method does not provide information regarding the morphology of pore structures and cannot assess individual defects. Similar to Archimedes method, gas pycnometry utilizes gas displacement to measure a components density but is more expensive and experiences similar disadvantages [20].
Microscopic analysis can be a very valuable tool in characterizing process induced porosities. Despite being a destructive method and relatively time consuming, microscopic image analysis can better characterize area percentages, sizes, and shape distributions of porosity within the interior of a component [6,20]. One of the most powerful porosity characterization methods involves x-ray micro-computed tomology (μ-CT), where a set of 2D x-ray projections are acquired by rotating the specimen and used to generate a 3D model of its interior [19,20]. This non-destructive method provides analysis of the size, shape, volume, and distribution of porosity throughout the entire specimen; but is hindered by time and cost demands and having resolution limits inherent to construction of the 3D model [19,20].
3. Research Objectives

Specific research objectives in this investigation of AlSi10Mg alloy components produced using AM-LPBF include:

- Characterize internal defects using Image Analysis software
  - Determine effects of varying process parameters on the generation of process induced defects
    - Effects of build plate location on porosity generation
    - Effects of varying specimen orientation on porosity generation
    - Effects of varying energy densities on porosity generation
  - Determine effects of total porosity and morphology on mechanical properties
    - Effects of varying porosity on tensile strengths
    - Effects of varying porosity on fatigue strengths
4. Experimental Methods

4.1. Materials

Specimens for this investigation were constructed from gas atomized AlSi10Mg alloy powder having a size distribution from 15.5-50.6μm with an average composition reported in Table 1. Specimens were built using an EOS M290; which utilizes a Yb-fiber laser, a 67° scanning regime, and a 9.85 x 9.85 x 12.8-inch build chamber. Specimens were constructed with orientations of 0°, 30°, 45°, 60°, and 90°; where 0° is parallel to the build plate (horizontal) while 90° is perpendicular to the build plate (vertical). An example of the specimen orientations as they appear on the build plate is shown in Figure 1. Specimens with varying orientation were all replicated utilizing differing build parameters and energy densities, reported in Table 2. These energy densities were calculated using Equations 1 and 2, mentioned previously. Specimens were precision machined to final testing dimensions after removal from the build plate.

<table>
<thead>
<tr>
<th>Table I. Average Composition of AlSi10Mg Feedstock Powder</th>
</tr>
</thead>
<tbody>
<tr>
<td>Element</td>
</tr>
<tr>
<td>Wt.%</td>
</tr>
<tr>
<td>Al</td>
</tr>
</tbody>
</table>

<table>
<thead>
<tr>
<th>Table II. Build Parameters and Energy Densities</th>
</tr>
</thead>
<tbody>
<tr>
<td>Build ID</td>
</tr>
<tr>
<td>Stripes Power (W)</td>
</tr>
<tr>
<td>Stripes Speed (mm/s)</td>
</tr>
<tr>
<td>Stripe Hatch Spacing (mm)</td>
</tr>
<tr>
<td>Layer Thickness (mm)</td>
</tr>
<tr>
<td>Global Energy Density (J/mm³)</td>
</tr>
<tr>
<td>Global Energy Density (J/mm²)</td>
</tr>
</tbody>
</table>
Figure 1. As-received build plate showing AlSi10Mg specimens at varying orientations

4.2. Tensile Testing

Tensile testing is an invaluable method for characterizing mechanical properties such as elastic modulus, yield strength, ultimate tensile strength, or total elongation. Ultimate tensile strengths are of particular interest in this investigation to observe the effects of varying processing parameters on the evolution of process induced porosity and resulting mechanical properties. Tensile testing was performed using an MTS Landmark Servohydraulic Universal
Test Frame in accordance to ASTM standard E8/E8M [40]. During testing, computer software recorded stress and strain values at a frequency of 50 Hz. Specimens had a gage diameter of 8.89 mm and a gage length of 44.45 mm. All tensile tests were performed until final fracture [41].

4.3. Fatigue Testing

Fatigue is a form of failure that occurs in structures subjected to dynamic and fluctuating stresses, where it is possible for failure to occur at stresses considerably lower than the tensile or yield strength for a static load [44]. Fatigue failure is attributed to approximately 90% of all metallic failures and is characterized by three steps: crack initiation at near-surface stress concentrations, crack propagation and advancement with each stress cycle, and final failure or fast fracture [44]. Rotating beam fatigue testing was incorporated in this investigation, utilizing a reversed stress cycle with a stress ratio, R = -1. Fatigue tests were performed until final fracture or until the test reached the fatigue limit of ten million (10^7) cycles, classifying as high cycle fatigue [41].

4.4. Porosity Characterization

Before performing micrography and image analysis, portions of the tension and fatigue specimens for each build and orientation were sectioned along both the x-y and z planes; where the x-y plane is transverse to the long-axis of the specimen while the z plane is considered to be longitudinal to the specimen. These cross-sections were hot-mounted using a phenolic thermoset and polished using 120, 320, 480, 600 grit abrasive paper, successively. The specimens were then fine polished using 5 μm and 1 μm alumina slurry and final polished using 0.25 μm diamond paste.

Optical microscopy was performed using a Leica DM750P optical microscope. Five micrographs with a magnification of 100X were taken at various locations of each specimen’s
surface in order to provide a representative area and number of defects for image analyses. Once a micrograph was captured, and identical micrograph was captured with no scale bar and an applied color filter in order to aid identification of differing phases (pores/bulk) during image analysis. An example of this process is depicted in Figure 2. Image analysis was performed using the Leica Application Suite V4.8 software, allowing for the quantification of a wide variety of features and properties such as:

- total area percentages
- equivalent circular diameters
- individual locations and anisotropy
- degrees of roundness or compactness
- statistical data for each of the desired properties.

Figure 2. Example of micrograph manipulation performed during image analysis
Total area percentages of porosity were determined which can be used as a measure of densification and process optimization. The individual and overall compactness of porosity was incorporated in quantifying the morphology of porosity found within AM-LPBF specimens. Compactness provides a good measure of the circularity or sphericity of internal defects, giving insight into possible melting modes and processing histories. Compactness is calculated using the equation:

\[
C = \frac{(4\pi A)}{P_r^2}
\]

where \(A\) is the area of the feature and \(P_r\) is the perimeter. The compactness of a perfect circle will equal one and approach zero as a feature deviates from circularity [43]. Figure 3 gives an example of the compactness of two different pores found within the AlSi10Mg specimens, one being relatively circular while the other is irregular and faceted.

Figure 3. Compactness values for pores with differing morphologies
Results

4.5. Location on Build Plate

The porosities of specimens was tracked with respect to the specimen’s location on the build plate to observe for any variations that may arise from the F-theta lens and Galvano-mirror scanning system utilized by the EOS M290 machine. Figure 4 reports the total area percentage of porosity found within ALB1 specimens compared to the specimen’s radial distance from the center of the build plate, where the F-theta lens is centered.

Figure 4. Total area percent porosity in specimens at various build plate locations
4.6. Specimen Orientation

The porosity of specimens was also tracked with respect to the specimen’s orientation on the build plate. Figures 5, 6, and 7 depict the total area percentage of porosity for the ALB1, ALB2P, and ALB2M builds with respect to specimen orientation during the building process. Although no strong correlation appears to exist, 90° specimens do have slightly higher porosities than 0° specimens. The 90° specimens from the ALB2M build were saved for future investigations.

![Porosity vs. Orientation for ALB1 Build](image)

Figure 5. Total area percent porosity in ALB1 (GEDa = 1.50 J/mm2) specimens built with differing orientations
Figure 6. Total area percent porosity in ALB2P (GEDa = 1.36 J/mm²) fatigue specimens built with differing orientations.

Figure 7. Total area percent porosity in ALB2M (GEDa = 1.11 J/mm²) fatigue specimens built with differing orientations.
4.7. Energy Density

The ALB1, ALB2P, and ALB2M builds were all created using varying energy densities, achieved by varying laser power, scanning speeds, and hatch spacings; as reported in Table II. Representative micrographs of ALB1, ALB2P, and ALB2M specimens are depicted in Figures 8, 9, and 10; respectively. The total area percentages of porosity within specimens built using varying energy densities is depicted in Figure 11. The morphology of the porosity was also characterized using the property of compactness with respect to varying energy densities between the builds. Average compactness values were reported since the normal distribution curves of pore compactness did not show any significant skew; with these distributions referenced in Appendix A. Figure 12 represents the relationship of the porosity morphology with differing energy density. Figure 11 shows that the total area percentage of porosity increases with increasing energy density. However, the general morphology of pores will become more circular/spherical as the energy density for building is increased; seen in Figure 12.
Figure 8. Representative micrograph of ALB1 specimen (GEDa = 1.50 J/mm²) exhibiting relatively high porosity and high degree of circular/spherical pores
Figure 9. Representative micrograph of ALB2P specimen (GEDa = 1.36 J/mm²) exhibiting lower porosity and a lesser degree of circular/spherical pores.
Figure 10. Representative micrograph of ALB2M specimen (GEDa = 1.11 J/mm2) exhibiting the lowest porosity and the prevalence of irregular, faceted pores
Figure 11. Average total area percent porosity in specimens created using varying energy densities.

Figure 12. Compactness (circularity/sphericity) of porosity found in specimens built using varying energy densities.
4.8. Mechanical Properties

4.8.1. Tensile Properties

Ultimate tensile strength (UTS) was used as a measure of tensile properties for comparison with the degree and morphology of porosity within AlSi10Mg alloy specimens built via AM-LPBF. The average UTS for each set of build parameters is reported in Figure 13.

![Average UTS of Each Parameter Set](image)

**Figure 13.** Average UTS of specimens from each parameter set
Figures 14, 15, and 16 display the UTS for specimens of varying orientations which were built using the ALB1, ALB2P, and ALB2M parameter sets; respectively. Within the ALB1 build with the highest degree of porosity, the 0° specimen exhibited the minimum UTS of 349 MPa while the 90° specimen had the maximum UTS of 378 MPa. The ALB2P build showed slightly higher UTS values with 45° having the minimum UTS of 358 MPa while the 90° again has the maximum of 391 MPa. The ALB2M build contained the smallest area percentage of porosity and also exhibits the highest UTS values, where the 45° specimen had the minimum UTS of 389 MPa while the 30° showed the maximum of 399 MPa.

Figure 14. Average UTS of ALB1 (GEDa = 1.50 J/mm²) specimens having varying orientations
Figure 15. Average UTS of ALB2P (GEDa = 1.36 J/mm²) specimens having varying orientations

Figure 16. Average UTS of ALB2M (GEDa = 1.11 J/mm²) specimens having varying orientations
4.8.2. Fatigue Properties

S-N curves were generated to report the fatigue behavior of AlSi10Mg alloy specimens built using AM-LPBF. Having only three to four specimens available for the generation of the S-N curves exacerbates scatter in the fatigue data and limits this investigation as a preliminary study of the fatigue performances. S-N curves of specimens built using the ALB1 parameter set with varying orientations are displayed in Figure 17. The estimated fatigue strengths attained in ALB1 specimens ranged from approximately 103-126 MPa. Figures 18 and 19 report the preliminary S-N curves for the ALB2P and ALB2M parameter sets, respectively. Although incomplete, it can be seen in Figures 18 and 19 that specimens are readily failing below 100 MPa; markedly lower than specimens produced in the ALB1 build. Fatigue specimens built with the ALB2M parameter set exhibit very short fatigue lives, even when compared to the ALB2P specimens, with stresses not exceeding 100 MPa.

![S-N curve for Aluminum Fatigue Test Specimens](image)

Figure 17. Preliminary S-N curves for ALB1 (GEDa = 1.50 J/mm²) specimens with varying orientations, having an average UTS of 349MPa
Figure 18. Preliminary S-N curves for ALB2P (GEDa = 1.36 J/mm²) specimens with varying orientations, having an average UTS of 377MPa

Figure 19. Preliminary S-N curves for ALB2M (GEDa = 1.11 J/mm²) specimens with varying orientations, having an average UTS of 387MPa
5. Discussion

5.1. Effect of Specimen Location on Porosity

Seen in Figure 4, the radial distance from the center of the build plate did not show a significant correlation with the evolution of porosity within AlSi10Mg alloys produced via AM-LPBF. However, the specimens analyzed only fell within a certain region (3.5-5.5 in.) while specimens built directly underneath or perpendicular to the laser beam may exhibit differences as the angle for laser reflection is minimized. An investigation could be designed that would produce an array of components, possibly simple cubes, across the build plate that would allow for in-depth analysis of properties such as: densification, microstructure, hardness, or microhardness with respect to the radial distance from the center of the build plate. Effects from varying distances to the base-plate center may be hindered by the non-uniform surface of the powder bed; where spherical powder particles will always exhibit a surface that is normal to the laser beam and act in mitigating any absorptivity and reflectivity differences exhibited at base plate boundaries.

5.2. Effect of Specimen Orientation on Porosity

Although porosity levels between specimens with differing orientations were quite similar, specimens built at 0° orientations exhibited slightly less porosity compared to specimens built at 90° in Figures 4 and 5. Mower and Long reported comparable results; stating that although pores are developed in both orientations they appeared more numerous in specimens grown with a vertical orientation [3]. Olakanmi et al. also reviewed the prevalence of irregular shaped pores developing along layer boundaries found in the x-y plane of a specimen [21]. Similar results were also reported by Tang and Pistorius, stating the tendency for lack of fusion porosity to be aligned perpendicular to the building direction (x-y plane) [25]. One mechanism
for the propensity for lateral pore formation is suggested by Khairallah et al., stating that partially melted powders within the denudation zones will adhere to the melt track while introducing voids already present between particles [28]. Numerous pore formation mechanisms have been suggested, however, that may contribute to elevated porosity in vertically built specimens; such as: balling effects, entrapped gasses, keyhole formation, and flow patterns exhibited by the melt pool [4,21,28]. These mechanisms could very well take place as the ALB1 build had the highest degrees of porosity, suggesting that excessive energy density was used and possibly facilitated keyhole melting and entrapped gasses via extended solubilities.

5.3. Effect of Energy Density on Porosity

From Figures 8-11, the total area percentage of porosity increases as the applied energy density increases. This is contrary to numerous investigators who have reported increased densification as applied energy density is increased [1,7,9,18,21,25]. However, multiple researchers have also reported that excessive energy densities can overheat the melt which will promote balling, entrapment of supersaturated gasses, spattering, and excessive thermal stresses [7,12,21,26-28]. This implies that excessive energy densities are being incorporated in this investigation and porosity may still be reduced by approaching the transition from conduction to keyhole mode melting [28]. By observing Figures 8-10 and Figure 12 it can be seen that the morphology of pores within the components becomes more circular/spherical as the applied energy density increases. The change in morphology of the pores found between differing parameter sets suggests that varying process parameters and energy densities will promote differing porosity formation mechanisms.

Specimens built with the ALB2M parameter set exhibited the lowest total porosity; while the pore morphology exhibits lack of fusion porosity with numerous irregular, faceted pores.
Lack of fusion porosity can derive from several mechanisms, including: unmelted particles or satellites, balling, shrinkage porosity, and oxide inclusions [4,21,26-28]. Unmelted particles can adhere to the melt track from the denudation zones and introduce pre-existing voids found between particles [28,35]. Balling can induce lack of fusion porosity by inhibiting energy conduction to the substrate, drastically reducing the remelted depth of the scan track and prohibiting interlayer bonding [27]. Oxide inclusions act to inhibit wetting and consolidation of liquid metal which can result in lack of fusion [10,21,25,26].

Previous fractographic investigation of the AlSi10Mg specimens used in this study were performed by Abstetar [41] using a scanning electron microscope (SEM) and energy dispersive spectroscopy (EDAX). Analysis revealed the presence of unmelted powder and/or satellites within the structure, but elemental analysis did not suggest significant oxide inclusions within the defects [41]. This suggests that lack of fusion porosity found in this study likely formed from previous void spaces between particles and satellites. Localized instances of balling likely occur but are not as significant; as increases in energy density resulted in increased circularity/sphericity of pores whereas higher energy densities should aggravate balling and eventually prevent fusion between layers.

AlSi10Mg specimens produced using the ALB1 parameter set displayed the highest total porosity while the pore morphology was predominantly circular/spherical, suggesting the prevalence of entrapped gas porosity. Entrapped gas porosity can be generated by mechanisms including [2,4,6,21,28]:

- Turbulent fluid flow
- Collapse of scan tracks
- Entrapped gas within powder particles
- Rejection of supersaturated gasses within melt pools and HAZ’s
Upon cooling after keyhole melting, the recoil pressure maintaining the melt pool depression is overcome by an increase in surface tension as temperatures decrease [28]. The melt pool will then rapidly reverse directions with the chance to trap gas bubbles towards the bottom of the melt. Gas bubbles can also become trapped within vortices formed in the tail of the melt pool [28]. Another mechanism of gas porosity is by the nucleation and growth of hydrogen pores as the local solubility limit is reached within liquid metal or HAZ’s [2,6].

Siddique et al. reported similar results to this investigation using an AlSi12 alloy; where lower energy densities resulted in irregular shaped bonding defects while increased energy densities exhibit only circular/spherical pores [7]. This trend was also reported by Khairallah et al. when analyzing the physics of complex melt flows and defect formation [28].

5.4. Effects of Porosity on Tensile Properties

The resulting porosity in the AlSi10Mg specimens decreased as the applied energy density decreased, seen in Figure 11. Figure 13 displays the resulting increase in average UTS with increasing densification from reduced energy densities. During process optimization, it is widely reported that tensile properties such as ultimate tensile strengths and yield strengths will increase as the densification of components increases [1,7,8,10,13,26]. Figures 14-16 depict the average UTS values for specimens with differing orientations produced using the ALB1, ALB2P, and ALB2M respectively.

Mechanical properties can be greatly sensitive to pore shape and placement; where tensile strengths are typically dependent on the fractional density (ie. area percent porosity) achieved within the final component [21]. This accounts for the increase in UTS values as densification increased, despite the irregular pores acting more severely as stress concentrations. During tension testing, plastic strain increased to a point where further plastic deformation becomes
localized at an area with a smaller cross section or is less work hardened [44]. The cross sectional area at this location will become locally reduced and serve as either an internal or surface notch, promoting crack initiation within the region [44]. The extent of spherical pores in specimens built with higher energy densities significantly decreased the cross sectional loading area of samples and also provided many more locations susceptible to localized plastic deformation and crack initiation.

5.5. Effects of Porosity on Fatigue Properties

Despite advances in AM-LPBF of metal materials allowing relative densities >99.5%, remnant porosity proves to be detrimental to fatigue strengths and scatter [19]. These small levels of remaining porosity continue to impede further development and application of AM alloy systems [4]. Porosity and surface roughness have been attributed as having the most deleterious effects on the fatigue properties of metals produced via AM-LPBF [3,21,25]. Specimens produced for this study undergo precision machining after removal from the build plate which should minimize surface roughness effects and accentuate the effects of internal porosity.

AlSi10Mg specimens produced using the ALB1 parameter set exhibited the best fatigue strengths ranging from ~103-126 MPa; despite having the highest total percent porosities of the varying builds. Fatigue strengths for both the ALB2P and ALB2M appear in Figures 18 and 19 as though they will be markedly lower than those obtained by the ALB1 specimens. Lower fatigue strengths are likely since the specimens regularly undergo failure below 100 MPa before reaching the fatigue limit of ten million cycles. Specimens produced via the ALB2M build appear to experience failure before those produced via the ALB2P parameter set; possibly arising from porosity within ALB2M specimens having more of an irregular pore morphology.
For many common cyclical loading situations such as rotating bending, stress increases from the beam’s neutral axis to the maximum stress located at the components surface [42]. This results in the majority of crack initiation occurring near a components surface. These characteristics make fatigue inherently sensitive to surface conditions and the presence of stress concentrations near the surface [42]. The severity of the stress concentration depends on the sharpness or radius of curvature of a discontinuity; where the sharper the discontinuity, or the smaller the radius of curvature, the more severe the stress concentration [42].

This provides a viable explanation for the fatigue behavior exhibited by the AlSi10Mg specimens utilized in this investigation. Although the total percentage of porosity decreases with decreasing energy densities, fatigue behavior is highly sensitive to the morphology of porosity generated during the building process. Specimens produced using the ALB1 parameter set generally exhibit circular/spherical pore morphologies, which have large radii of curvature when compared to pores exhibited by the ALB2P and ALB2M built specimens. Porosity acts as increasingly severe stress concentrations as the morphology deviates from circular/spherical to irregular and faceted with small radii of curvature. Akita et al. reported similar findings when investigating fatigue behavior of type 630 stainless steels produced via AM-LPBF, where specimens exhibited much shorter crack initiation lives and lower crack initiation resistance [11].
6. Conclusions

- A specimen’s location and radial distance from the center of the build plate did not have any significant effects on the evolution of porosity within AlSi10Mg components produced via AM-LPBF
  - Only a small range of distances and specimens were observed, effects may not be seen in this investigation
- Porosity levels between specimens with differing orientations were quite similar, specimens built at 0° orientations exhibited slightly less porosity compared to specimens built at 90°
  - May arise from generation of porosity between successive build layers as constructing the 90° specimens requires the most layers
- The total area percentage of porosity increases as the applied energy density increases
  - Implies that excessive energy densities are being incorporated in this investigation and porosity may still be reduced
- The morphology of pores within the components becomes more circular/spherical as the applied energy density increases
  - The change in morphology of the pores found between differing parameter sets suggests that varying process parameters and energy densities will promote differing porosity formation mechanisms
• Increase in average UTS with decreasing energy density and improved densification
  o Tensile strengths are typically dependent on the fractional density (ie. area percent porosity) achieved within the final component
• Specimens produced using the ALB1 parameter set exhibited the best fatigue strengths ranging from ~103-126 MPa; despite having the highest total percent porosities of the varying builds
• Fatigue strengths for both the ALB2P and ALB2M appear as though they will be markedly lower than those obtained by the ALB1 specimens
  o Fatigue is inherently sensitive to surface conditions and the presence of stress concentrations near the surface
  o Porosity acts as increasingly severe stress concentrations as the morphology deviates from circular/spherical to irregular and faceted
7. Recommendations

- An investigation could be designed that would produce an array of components, possibly simple cubes, to observe effects with respect to the radial distance from the center of the build plate.
- Integration of current build parameters into an experimental test matrix, such as Taguchi-based experimental design, in order to provide a detailed understanding of the effects of varying process parameters and any interaction between variables.
- Additional contour scanning of specimens outer surface to reduce porosity in fatigue critical zones.
- Continued production of fatigue specimens for statistically viable S/N study in order to reduce scatter and more accurately approximate fatigue strengths of AM-LPBF components.
- Implementation of a powder feedstock drying regime to minimize the introduction of surface contaminants.
- Collaboration with the Army Research Laboratories (ARL) to perform μ-CT of specimens produced using differing build parameter sets in order to analyze the size, shape, volume, and distribution of porosity throughout the entire specimen.
References


http://dx.doi.org/10.1016/j.ijfatigue.2016.06.002


http://dx.doi.org/10.1016/j.addma.2016.05.007


Bibliography


Appendix A

Figure 20. Normal distribution of compactness for pores observed in ALB1 specimens

Figure 21. Normal distribution of compactness for pores observed in ALB2P specimens
Figure 22. Normal distribution of compactness for pores observed in ALB2M specimens
SIGNATURE PAGE

This is to certify that the thesis prepared by Edward J. Stugelmayer entitled “Characterization of Process Induced Defects in Laser Powder Bed Fusion Processed AlSi10Mg Alloy” has been examined and approved for acceptance by the Department of Metallurgical and Materials Engineering, Montana Tech of The University of Montana, on this 18th day of April, 2018.

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