Research on metallurgical bonding of selective laser melted AlSi10Mg alloy

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Keywords: selective laser melting, AlSi10Mg alloy, metallurgical bonding, porous microstructure, mechanical property

Abstract
The densification behavior, surface morphology and attendant microstructural characteristics of the selective laser melting (SLM) processed AlSi10Mg alloy affected by the processing parameters were systematically investigated. Increasing the laser scanning speed or hatch spacing will deteriorate the metallurgical bonding between melt pools, resulting in the increase of irregular pores. Scanning speed and hatch spacing affect the liquid metallurgical bonding of melt pools in different ways. By manipulating scanning speed, the shape of the melt pool changes, resulting in different extents of metallurgical bonding. Whereas hatch spacing influences the resultant metallurgical bonding by controlling the overlapping rate between neighboring scan tracks simply. The formations of the hierarchical microstructures which discriminated by the Si phase are elucidated. Coarse zones are formed by the instantaneous existence of extremely high ratio of thermal gradient (G) and solidification rate (R) at the melt pool boundary, where solidification microstructure grows planar. Fine zones are formed by columnar-dendritic growth of microstructure. During the solidification process, the contraction forces that generated by the trapped gas in the pores and gravity are applied to the liquid around irregular pores and, forms the porous microstructures different from that in dense areas eventually. The tensile tests reveal that the tensile properties of SLM-processed samples are significantly affected by the formation of porosity.

1. Introduction

Freedom of design and flexibility of production are very important to modern industry, including aerospace, biomedical and automotive industries [1, 2]. Many industries also need to reduce production time and cost, and are committed to the research of production systems that can produce complex products. Selective laser melting (SLM), as one of the manufacturing technologies of powder bed fusion (PBF), has attracted extensive attention due to its ability to perform near net shape forming on customized and complicated metal parts [3, 4]. At present, a series of metals have been applied to SLM research and production, such as Ti-based alloy [5–7], stainless steel [8, 9], nickel-based alloy [10, 11], Al-based alloy [12–14], etc. The as-built parts can’t be used directly as end-use parts without any surface post-processing yet due to its high surface roughness. As processing parameter is one of the influential factors on the characteristics of a as-built sample, appropriate combinations of the processing parameters are required to achieve the qualified samples. Up to now, there are lots of researches have been done in this area already [9, 15, 16]. However, most researchers focus on the optimization of SLM process parameters and basic principles of metals, and lack of elaboration on the metallurgical bonding mechanisms in the forming process.

SLM forming equipment uses a fiber laser device with high power density. The diameter of laser spot is generally less than 0.1 mm, melting metal powder at a very fast speed, and heat is rapidly conducted along the substrate, causing an extremely high cooling speed of $10^5 \sim 10^7$ K s⁻¹ [4]. These characteristics in SLM lead to the following physical phenomena: absorption and scattering of laser radiation, heat transfer under high-
temperature gradient, phase transformation, evaporation of metal, fluid flow within the melt pool caused by surface-tension gradient, and chemical reactions [17]. Because of this, that makes the metallurgical bonding mechanism of SLM different from the traditional forming process and more complex, thus making the evolution mechanisms of density, surface morphology and microstructure of fabricated parts differently. Since the metal products are built by line-by-line and layer-by-layer laser scanning during SLM, the quality of as-built samples are significantly determined by the metallurgical bonding of scan tracks. Inaccurate metallurgical bonding of molten metals in SLM process will cause pores, balling and other defects in products, seriously affecting density and surface quality of parts and reducing mechanical properties. Moreover, the microstructure evolution during solidification plays a critical role in determining the mechanical properties of SLM-processed (SLMed) parts. Due to ultra-rapid cooling rate, the microstructure of SLMed materials exhibited significant differences with as-cast ones [18].

Hypoeutectic AlSi10Mg alloy has become one of the most widely and maturely used aluminum alloys for SLM process due to its excellent welding performance, good melt fluidity and mechanical properties [19]. However, strong affinity with oxygen at high temperatures, low melting point, high thermal conductivity and laser reflectivity of aluminum alloys make the metallurgical bonding mechanism more complicated, and created more difficulties in successfully fabricating AlSi10Mg parts. Therefore, the study of AlSi10Mg alloy is even more important.

In this paper, AlSi10Mg alloy was utilized to produce samples for research. The metallurgical bonding mechanisms in the forming process were explored by observing the relative densities and surface morphologies of as-built samples formed by SLM under different process parameters. Microstructural characteristics of SLMed AlSi10Mg samples were disclosed. Tensile tests and microhardness tests were carried out to reveal the effects of metallurgical bonding on mechanical properties of the SLMed samples.

2. Materials and methods

2.1. Material

The AlSi10Mg powder used in this study is commercial, and it was purchased from PMG3D Technologies (Shanghai) Co. Ltd. The chemical composition of AlSi10Mg powder is shown in Table 1. The powders are consisted of spherical particles and attached small satellite particles, with a diameter of about 15 – 53 µm, as shown in figures 1(a) and (b). Some powders were encapsulated by epoxy and polished to reveal their internal characteristics, as shown in figures 1(c) and (d), pores were observed in partially powders under a microscope.

| Table 1. Chemical composition of AlSi10Mg (Wt%). |
|-------|-------|-------|-------|-------|-------|-------|-------|-------|-------|-------|
| Si    | Mg    | Fe    | Cu    | Mn    | Ni    | Zn    | Pb    | Sn    | Ti    | O     | Al    |
| 9.6   | 0.3   | 0.16  | 0.004 | 0.002 | 0.006 | 0.005 | 0.006 | 0.012 | 0.0208 Bal |

2.2. SLM process

Cubic bulk samples with a dimension of $10 \times 10 \times 10$ mm$^3$ were fabricated a YLM -120 powder bed machine equipped with a 200 W single-mode fiber laser. In order to prevent oxidation, the process was carried out under an argon atmosphere with an oxygen level of 100 ppm. The diameter of the focused laser spot size was about 70 µm. As shown in figure 2(a), an island laser-scan strategy with 67° rotation and 1 mm shift between the adjacent layers was employed to balance the residual stresses, the island size was chosen to be $5 \times 5$ mm$^2$. The platform temperature was preheated and maintained at 150 °C. After selectively melting the islands, contour scans were adopted around the perimeter of the layer to improve the surface finish. As shown in figure 2(b), two different cross-sections were methodically considered in the observation due to the anisotropy of the building process. One was parallel to the building direction (Side section) and the other one was perpendicular to the building direction (Top section). Figure 2(c) shows the as-built cubic samples. Processing parameters used in this study was shown in table 2.

2.3. Sample characterization test

The relative densities of samples were measured by Archimedes method and the results of each processing condition were indicated as the arithmetic means of three different measurements. All samples were polished and etched by Keller’s reagent (95 ml H$_2$O, 2.5 ml HNO$_3$, 1.5 ml HCl, and 1.0 ml HF) for 10 ~ 15s before microstructure observation. The microstructure was observed using an optical microscopy (OM, CLOOS ROMAT 350, Germany) and a scanning electron microscopy (SEM, S-3400N, Japan) that equipped with energy dispersive spectroscopy (EDS). The SEM was also used to observe the top surface morphologies of the as-built
Figure 1. Powder characteristics. (a), (b) SEM image of AlSi10Mg powders; (c), (d) Cross section of polished AlSi10Mg powders under OM.

Figure 2. Method of sample building. (a) Schematic illustration of the island scan strategy; (b) Building direction and view cross-sections of samples fabricated by SLM; (c) As-built samples.
samples. The tensile tests of the as-built AlSi10Mg samples were carried out using the testing machine (AG-25TA, Japan) at a strain rate of \( 1 \times 10^{-4} \text{s}^{-1} \) and room temperature. 64 × 14 × 3 mm\(^3\) rectangular samples were built horizontally (perpendicular to building direction) by each scanning speed during the SLM progress. Then the as built rectangular samples were mechanically cut from the platform and manufactured by line cutting. Before mechanical properties testing, all of the test specimens were polished with a 2000# sand paper to remove the surface layer. The final tensile specimen is illustrated in figure 3. Microhardness were tested by HXD-1000TMC/LCD (Germany) tester at a load of 200 g for a holding time of 15 s.

### 3. Results and discussion

#### 3.1. Relative density

Figure 4 shows the effect of scanning speed (\(v\)) on the relative density (R) of SLMed AlSi10Mg samples. With increasing the scanning speed, the internal porosity of the sample increases gradually and the decrease in relative density happens. When the scanning speed is 950 mm s\(^{-1}\), the relative density reaches the highest value of 99.1%. At a higher scanning speed (\(v = 1400 \text{ mm s}^{-1}\)), the densification level of the SLMed sample decreases significantly to 93.3%, and large irregular pores are observed on the side section. The decrease in densification level is caused by poor metallurgical bonding among adjacent melt pool tracks and layers due to the increase in scanning speed [20]. The increase in scanning speed leads to the increase of capillary instability of the melt pool tracks, and results in poor metallurgical bonding among adjacent melt tracks and layers.

Hatch spacing (\(h\)) affects the relative density of SLMed AlSi10Mg samples by controlling the overlap rate between adjacent melt pool tracks. As we can see from figure 5, it is obvious that the relative density of the as-built samples decreased with the increase of hatch spacing. When the hatch spacing increases from 70 um to 130 um, the relative density decreases from 99% to 92.8%. This is because that the increase in hatch spacing increases the gap size between the melt pool tracks, and the molten liquid cannot completely fill the gap, thus leaving some residual irregular pores.

#### Table 2. Processing parameters used in this study.

| Parameters                  | Value   |
|-----------------------------|---------|
| Laser power (W)             | 190     |
| Scanning speed (mm s\(^{-1}\)) | 950, 1100, 1250, 1400 |
| Hatch spacing (um)          | 70, 90, 110, 130 |
| Layer thickness (um)        | 30      |
| Laser spot size (um)        | 70      |

Figure 3. Size of as-built tensile sample.
As illustrated in figures 4 and 5, the deterioration of relative density of SLMed AlSi10Mg sample is mainly caused by three different types of pore defects, namely gas pore, keyhole pore and irregular pore. The size of gas pore is very small and round in shape. Gas pore is mainly caused by the inert Ar gas and the residual gas in original powders (figure 1 (d)), which have no time to escape from the molten liquid during solidification progress [21]. Keyhole pore is also round in shape, with a size larger than the gas pore but generally less than 100
um. The evaporation of metal elements in the melt pool forms a steam cavity, which is extremely unstable and easy to collapse, resulting in the formation of the keyhole pore [22]. Keyhole pore is more likely to generate when laser scanning speed is low. Irregular pore is generally large in size, and its direct cause of formation can be regarded as poor metallurgical bonding of melt pool tracks. However, the formation of poor metallurgical bonding in liquid contains different indirect factors. The employment of improper process parameters is one of the considerable factors. As shown in figure 4, with increase of the scanning speed, a gradual change of pore type from keyhole pore to irregular pore happens. It is because that the amount of molten liquid is inadequate and the cooling speed is fast when the scanning speed is too high, resulting in insufficient filling of the gap between melt pool tracks, thereby forming irregular pores. In addition, as shown in figure 5, the increase of hatch spacing leads to the decrease of overlap ratio between melt pool tracks, causes poor metallurgical bonding. Therefore, most of the pores formed in figure 5 are irregular pores.

In order to explore whether there are other factors leading to the poor metallurgical bonding during SLM progress, EDX chemical element analysis was carried out around the irregular pores, and the results are shown in figure 6. A large amount of oxygen are detected on the wall of irregular pore, with a content of 7.18 Wt%. However, the oxygen content on the matrix are in a rational content. This indicates that although the oxygen content in the forming chamber was controlled below 100ppm, the residual oxygen could still react with Al element to generate oxide and lead to the formation of irregular pore.

Oxidation usually occurs at the edge and surface of the melt pool, forming an oxide film, causing abnormal Marangoni convection, thus hindering the diffusion of molten liquid and affecting metallurgical bonding [19, 23]. Within individual melt pools, Marangoni convection is caused by the surface tension gradient, i.e. liquid always flows from a region with low surface tension to a region with high surface tension [24]. As shown in figure 7(a), in an ideal non-oxidation condition, the Gaussian laser heat source results in the highest temperature (lower surface tension) in the center of the melt pool and lowest temperature (higher surface tension) in the edge of the melt pool. Therefore, liquid will flow from the center to the edge of the melt pool. Under this ideal condition, liquids are easy to spread and have a good wettability, then promoting the combination of the melt pool tracks and improving the metallurgical bonding. In reality, however, an ideal Marangoni flow situation seldom exist. As oxygen is a surface active element, oxidation can dramatically reduce the surface tension of melt pool edge. The direction of the surface tension gradient is changed and results in a reverted flow of the liquid in
the edge (lower surface tension) toward the center of melt pool (higher surface tension), as shown in figure 7(b), liquid contractions are promoted by oxidation, causing the lack of overlap of neighboring laser tracks.

3.2. Surface morphology

Figure 8 shows the evolution of surface morphology of as-built samples at different scanning speeds. Balling and spatter particles are observed on the surface of the samples, and the balling phenomenon becomes more and more serious as the scanning speed increases.

Figure 9 shows the evolution of surface morphology of as-built samples at different hatch spacing. It can be seen that different from the scanning speed, the change of hatch spacing has not gave rise to a significant change on the surface morphology of as-built samples, i.e. no obvious increase in balling phenomenon and spatter particles.

The reason why the change of scanning speed will have a greater impact on the surface morphology of the as-built samples, but the hatch spacing is not, is that the two different parameters have different mechanisms on affecting liquid metallurgical bonding. In SLM process, laser energy density (LED, J mm\(^{-3}\)) is commonly used to represent the comprehensive effect of several different process parameters on laser heat input, as shown in equation (1) [25]:

\[
\text{LED} = \frac{p}{vhd}
\]

where: \(p\) is the laser power (W); \(v\) is the laser scanning speed (mm s\(^{-1}\)); \(h\) is the hatch spacing (mm); \(d\) is the layer thickness (mm). High-energy laser heat source will melt metal powders and matrix into a semi-cylindrical melt pool, and form a melt pool track with the movement of the laser. As shown in equation (1), the input of LED will be affected by adjusting different process parameters.

Figure 10 illustrates the difference between scanning speed and hatch spacing on the mechanism of metallurgical bonding. As shown in figure 10(a). When the scanning speed is low, the LED input is high and the melt pool has a high depth-to-width ratio in shape. Melt pools in this shape can produce sufficient liquids which have good wettability and can be effectively spread. Therefore, smooth and continuous melt pool tracks are formed and metallurgical bonding between melt pool tracks are favorable [17, 26]. When the scanning speed increases, the LED input decreases and the melt pool that formed under low LED possess a low depth-to-width ratio in shape. Insufficient liquid melting limits the contact area between the melt pool and the matrix, resulting in unfavorable flow and wetting, and the melt pool shrinks radially into spheres under the action of surface tension [27]. Besides, in the axial direction, Rayleigh-Plateau instability breaks the melt pool track into

![Figure 7](image-url)
discontinuous droplets, forming discontinuous melt pool tracks [17]. All of these make the metallurgical bonding between melt pool tracks extremely poor, resulting in corresponding balling phenomenon. To sum up, different scanning speeds will form different melt pool shapes, which will give rise to successive effects on liquid wettability and metallurgical bonding. However, when hatch spacing is changed, the shape of melt pool will not
be changed and metallurgical bonding is affected by the distance between adjacent melt pool tracks, as shown in figure 10(b).

3.3. Microstructure

Figure 11 shows the SEM images of microstructures viewed along the Side section of SLMed AlSi10Mg samples. Figures 11(b) and (c) are magnified diagrams of the melt pool boundary and center in figure 11(a) that indicated by red rectangles, respectively. An alternate structure of fine zone, coarse zone and heat affected zone (HAZ) can be distinguished by the transition from very fine cellular-dendritic to a coarser dendritic structure going from the centre to the boundary of the melt pool. Figure 11(c) reveals the presence of a very fine cellular-dendritic microstructure of the as built material: a primary $\alpha$-Al matrix phase decorated with eutectic Si. It is clearly visible in figure 11(b) that the fineness of the dendritic is not constant, the size of eutectic Si in coarse zone is coarser than that in fine zone. The Si particles in the HAZs are substantially refined and dispersed nearly homogeneously in the $\alpha$-Al matrix. The EDS data of $\alpha$-Al (point 1) and Eutectic Si (point 2) phase obtained from the areas arrowed in figure 11(c), suggesting that oversaturated Si element was present in Al matrix due to the limited diffusion time.

The ratio $G/R$ determines the mode of solidification, where $G$ and $R$ are the thermal gradient and solidification rate, respectively. With increasing $G/R$ values, The solidification microstructures can be equiaxed dendritic, columnar-dendritic, cellular, or planar [28, 29]. At the instant of the formation of a high-temperature melt pool, the edge of the melt pool contacts with the solid matrix which possesses a lower temperature, leads to a great difference in temperature between the melt pool boundary and the matrix. The value of $G$ reaches the maximum and $R$ is the minimum. The ratio $G/R$ is extremely high and results in planar growth of the solidification microstructure that existed in a very small region, corresponding with the coarse zones. However, planar growth will not last long. As the heat dissipation proceeds, the solid matrix around the melt pool will be heated, and a gradual change of $G$ and $R$ appears. Therefore, the extremely high $G/R$ at the boundary just stayed in an instant. With the transition from the boundary to the center of the melt pool, $G$ decreases and $R$ increases gradually. The $G/R$ ratio changes continuously cross the melt pool, gives rise to the change of solidification

**Figure 10.** (a) Laser scanning speed affects metallurgical bonding by changing the melt pool shape; (b) Hatch spacing affects metallurgical bonding by controlling the overlapping rate of the adjacent scan tracks.
mode from planar to columnar-dendritic, corresponding with the fine zones which account for the majority of the whole melt pool. The HAZ belongs to the solid matrix of the former layer, caused by the heat treatment from the laser heat source. Thermals are accumulated near the melt pool boundary, resulting in increase of the diffusion rate of Si, and precipitation from supersaturated α-Al solid solution.

The existence of two different kinds of microstructures are revealed in figure 12. In dense regions of the observed sample, the microstructure shows unity, and its morphology will be consistent with that observed in figure 11. However, different from the dense regions, porous microstructures are observed above some of the irregular pores, as shown in figures 12(a) and (b). A more comprehensive observation found that not all the irregular pores have porous microstructure above them, as shown in figures 12(c) and (d). Figure 13 illustrates a further look of the porous microstructure under SEM. The porous microstructure distributes in a network and shows a certain directionality. As it gradually moves away from the edge, the network becomes coarser and disappears at the parting of dense zone.

From figure 12, we can find that irregular pores which have porous microstructures above them are different in shape from the ones which have no porous microstructures. This is attributed to whether the solidification feeding behavior was generated by the molten liquid above the irregular pore. The formation mechanism of porous microstructure is shown in figure 14. A certain amount of gas is captured in irregular pores formed by poor metallurgical bonding during solidification. In the initial stage of irregular pore formation, trapped gas are heated by the high-temperature liquid and expand, resulting in a larger initial size of irregular ellipsoid pore with a ellipsoid shape. As the solidification process progresses, the temperature around the irregular pores gradually decreases, and the gas in the pores also begins to contract. Therefore, under the combined function of gas contraction force and gravity, feeding behavior are generated by unsolidified liquid above the initial ellipsoid pore. When the feeding is completed, the shape of the finally observed irregular pore will be formed. The part of liquid used for feeding is stretched by, which leads to the formation of porous microstructure.

3.4.Mechanical properties

Figure 15(a) shows the influence of scanning speed on the tensile strength and ductility of SLMed AlSi10Mg samples at h = 70 um. At the scanning speed of 950 mm s^{-1}, the maximum tensile strength and tensile strain are 386 MPa and 5.6%, respectively. The tensile strength decreases gradually from 386 MPa to 330 MPa as the scanning speed increased from 1100 mm s^{-1} to 1400 mm s^{-1}, the tensile strain decreased to 4% at the scanning speed of 1400 mm s^{-1}. This is mainly attributed to the formation of irregular pores caused by the poor metallurgical bonding and unmelted particles at high scanning speed, which decreases the tensile properties of SLMed samples. The tensile stress-strain curves for samples built at different scanning speeds are shown in figure 15(b).
Figure 12. OM images of microstructure of AlSi10Mg sample formed by SLM (Side section).

Figure 13. Porous microstructure of solids around the keyhole pore under SEM.

Figure 14. Schematic diagram of the formation mechanism of porous microstructure.
Figure 16 shows the effect of scanning speed and hatch spacing on microhardness of SLMed AlSi10Mg samples. Microhardness tests were carried out on both of the Top sections and Side sections, and the results of each sample were indicated as the arithmetic means of 30 measuring points. Measuring points were avoided to fall around pores. It is revealed that the microhardness values fluctuate slightly when changing scanning speed or hatch spacing, indicating that microhardness is rarely affected by the quality of metallurgical bonding. In addition, the microhardness results also indicate that the SLMed AlSi10Mg samples show anisotropy in their mechanical properties, which the hardness values of the Top section is always higher than that of the Side section. This discrepancy could be attributed to the different grain morphologies in different cross-sections. In the Side section, the microstructure is mainly composed of elongated grains which growth towards the center of the melt pool. While in the Top section, only equiaxed grains can be seen [30].

4. Conclusions

In this work, the AlSi10Mg samples were fabricated by SLM using various processing parameters including scanning speed and hatch spacing. By analyzing the evolution of density and surface morphology of as-built samples under different process parameters, it is found that scanning speed and hatch spacing have different mechanisms on affecting liquid metallurgical bonding. When increasing the scanning speed, the size of the melt pool decreases, accompany with the decrease of the amount of liquid, leading to a poor metallurgical bonding. Hatch spacing influences the resultant metallurgical bonding by controlling the overlapping rate between the neighboring scan tracks simply. Poor metallurgical bonding leads to the formation of irregular pores. Besides, oxidation show great influence on the form of irregular pores. The oxide film formed at the melt pool boundary will change the Marangoni convection direction of the liquid in the melt pool and hinder the diffusion of the liquid, thus affecting the metallurgical bonding of the melt pools.
Three different zones are differentiated by the morphology of Si phase across the melt pool of SLMed AlSi10Mg samples. Coarse zones are formed by the instantaneous existence of extremely high G/R at the boundary of the melt pool, where the solidification microstructure grows planar. In addition, porous microstructures are observed above some of the irregular pores. The formation of porous microstructures can be ascribed to liquid feeding behavior which caused by the combined effect of gas contraction force and gravity. The formation of porosity has a significant effect on the tensile properties, but has little effect on the microhardness of the SLMed AlSi10Mg samples.

## Acknowledgments

This work was financially supported by National Natural Science Foundation of China under Grant No. 51805313 and Industry-University-Research Collaboration Project between Shanghai University of Engineering Science and PMG 3D Technologies (Shanghai) Co., Ltd

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