The influence of divergent laser beams on the laser powder bed fusion of a high reflectivity aluminium alloy

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Abstract

The laser powder bed fusion (LPBF) of aluminium alloys is associated with numerous challenges when compared to other commonly used alloys (e.g., steels and titanium alloys) due to their higher reflectivity and thermal conductivity. This leads to a higher defect density in the final parts, commonly related to melt pool instabilities in the transition and keyhole melting modes.

In this work, a laser beam defocusing strategy using a divergent beam is proposed to achieve a stable conduction mode microstructure in AlSi10Mg, a eutectic Al composition that is most studied in LPBF. The effects of conduction, transition, and keyhole melting modes on the final part are studied in detail using processing diagrams, metallography, and X-ray computed tomography. The conduction mode LPBF of AlSi10Mg leads to parts with densities of over 99.98%, with close to no porous defects in the subsurface regions, which are known to directly affect the fatigue life of the final parts. The threshold between conduction and transition mode melt pools is also observed to be at a melt pool aspect ratio (ratio of melt pool depth to width) of about 0.4, which differs from the conventionally assumed 0.5. Additionally, a significant difference is observed in the standard deviation of the melt pool depths for the transition and keyhole mode melt pools, when compared to the conduction mode melt pools. This points to the large differences in laser absorptivity between melting modes, that are exaggerated by the onset of vaporization expected for transition and keyhole mode melting during LPBF of high reflectivity aluminium alloys.
Keywords:
Additive manufacturing, Laser powder bed fusion, Selective laser melting, Aluminium alloy, Defects, AlSi10Mg

1. Introduction

The potential of powder bed fusion (PBF) technologies to produce high-quality complex geometries and the ability to pursue assembly consolidation have been shown to increase technology adoption in the aviation [1]–[7] and automotive industries [8]–[12] in particular. Laser powder bed fusion (LPBF) is a PBF technology with a high industrial uptake, owing to the minimum feature size, resolution, and surface finish possible with the reduced beam spot sizes commonly used in LPBF, when compared to the electron beam powder bed fusion (EB-PBF) technology [13]. The use of lower achievable beam spot sizes, combined with the advances in design for additive manufacturing (DfAM) competencies, help the industry in realizing the true potential of LPBF for light weighting and design optimization of critical components.

With respect to materials, aluminium alloys are commonly used in aerospace and automotive applications which require a combination of high strength performance and low weight. AlSi10Mg has been adopted widely in the LPBF community for these applications [14]–[16]. The lower beam spot sizes used in LPBF, however, lead to the lowering of the vaporization threshold of aluminium [17], which can be a disadvantage due to the ease of porosity formation in the keyhole mode melting of aluminium alloys [18]. It has hence been previously assumed in literature that keyhole mode melting is the dominant melting mechanism in LPBF of aluminium alloys such as AlSi10Mg, while working with a beam spot radius of 10 μm [19]. Similarly, for another experimentally intensive effort on processing AlSi10Mg with a beam spot radius of 35 μm, keyhole mode porosity defects were observed in three-dimensional coupons at laser power settings ranging from 88 to 390 W, and scan speeds ranging from 250 mm/s to 2500 mm/s [20]. Based on the literature, there are challenges in identifying stable process parameter windows for obtaining defect-free components for AlSi10Mg due to the rapid onset of transition/keyhole melting mode driven by the interaction of a highly focused energy source
acting on a high reflectivity and high thermal conductivity material system [21]–[23]. The significant differences in absorptivity characteristics for aluminium alloys in the conduction and transition melting modes when compared to titanium, ferrous, and nickel alloys [24] also contributes towards the difficulties in obtaining defect-free parts as the rapidly increasing vapour cavity in the melt pool would make it difficult for a pore to escape during solidification. One path towards exploring such defect-free process windows is by exploring the effects of beam defocusing on achieving stable conduction-mode process windows [25], [26], as proposed in the present work.

The motivation behind exploring defect-free stable conduction mode melting in AlSi10Mg is reinforced by the body of literature studying the effect of conduction, transition, and keyhole modes on microstructures, porosity and resulting mechanical properties of multiple other alloys [26]–[31]. Qi et al. [26] observe a lower crack density in the keyhole melting mode for Al7050 but a higher and more uniform nano-hardness across the melt pool in conduction melting mode. Higher vaporization of Zn and Mg was observed in the keyhole melting mode for Al7050. Aggarwal et al. [27] observe a higher hardness, higher elongation, and finer cellular grains in stable keyhole mode coupons of 316L stainless steel, when compared to conduction mode coupons. Yang et al. [28] report a wider processing window for Ti-6Al-4V in the conduction melting mode and similar tensile properties for conduction and keyhole melting modes, but report a higher elongation for keyhole melting mode coupons. Using micro-scale simulations, Wang and Zou [29] report a more uniform thermal distribution during multi-track LPBF in the conduction mode when compared to keyhole melting mode, leading to a more uniform microstructure in Ti-6Al-4V. Patel et al. [30] reported reduced effects of adhered partially fused powder particles leading to low side-skin surface roughness for LPBF of Ti-6Al-4V using keyhole melting mode parameters when compared to conduction melting mode parameters. For a top-hat shaped laser beam profile, Tenbrock et al. [31] reported >99.95% density components in both the conduction and keyhole melting modes for 316L stainless steel with a gradual transition between the melting modes. A comparably uniform thermal distribution during multi-track printing leading to a uniform microstructure is hence a common characteristic observed in the stable conduction melting mode; with the potential for a finer microstructure,
improved side-skin surface finish, and improved tensile properties for the stable keyhole melting mode.

In this work, for the first time, it is shown that a beam defocusing strategy can help in achieving stable conduction mode melting in LPBF of AlSi10Mg, particularly for LPBF systems with lower beam spot sizes. This approach helps ease the process parameter optimization efforts for material systems with high reflectivity and high thermal conductivity, especially for the goals of reducing porosity. Deploying a process parameter strategy resulting in conduction mode LPBF of AlSi10Mg leads to a drastic reduction in pore defects, specifically in the subsurface regions, which are known to be the most important defects affecting fatigue life, as porous defects are the most likely site for crack initiation [32]–[36]; an example of a subsurface pore (diameter of ≈0.2 mm) leading to the fatigue crack in AlSi10Mg is shown by Plessis et al. [37]. The results from this present work demonstrate that stable conduction mode melting can be achieved in high reflectivity and high thermal conductivity alloys, with a drastic reduction in core and subsurface pore defects.

2. Materials and Methods

AlSi10Mg cubes of side-length 10 mm (for microstructure evaluation) and cylinders of diameter 5 mm and height 9 mm were printed on the reduced build volume (RBV) of a modulated LPBF system (AM 400, Renishaw, UK). Six sets of processing parameters, shown in Table 1, were investigated in this work, which were planned based on melting mode prediction from previous work [23]. For the AM 400 system, the beam spot radius at the focal point is given by \( r_0 = 35 \) μm, and the wavelength of the laser beam used is \( \lambda = 1070 \) nm. For samples A & B, the laser beam was defocused to a distance, \( z = 4.2 \) mm, above the build plate, to obtain a divergent beam. This leads to a beam spot radius \( (r) \) of 54 μm, obtained from the equation for a Gaussian distribution of a laser beam [38]:

\[
r = r_0 \left[ 1 + \left( \frac{z\lambda}{\pi r_0^2} \right)^2 \right]^{1/2}
\]  

(1)
The processing parameters listed in Table 1 were selected such that sample A and sample B would be expected to lie in the conduction melting mode from previous work on studying melt pools [23], while samples C, D, E, and F would be expected to lie in the transition melting mode. The powders used were plasma atomized AlSi10Mg with a size distribution of 15-63 µm and a \(D_{50}\) of 28 µm, which was used to determine the powder layer thickness value of 30 µm, as the steady state powder layer thickness obtained is effectively larger than the amount by which the build plate drops during each layer [39]. The hatching distance was kept constant at 100 µm for all the samples. The scan order was set such that the hatch volume (core) was scanned first, followed by the border; the border scans having the same set of processing parameters as the core. Border scans are commonly used in LPBF to improve the dimensional accuracy and surface roughness of LPBF coupons [40]; surface topography optimization was beyond the scope of this present work. The meander scan strategy was used with a 67° rotation between each layer, to reduce residual stresses, anisotropy, surface roughness and promote lower rates of defect propagation by the virtue of the scan vector direction not repeating until 180 layers [41]–[43].

Table 1. Processing parameters for LPBF part production.

| Sample Code | Power (W) | Point Distance (µm) | Exposure Time (µs) | Beam Spot Radius (µm) | Expected Melting Mode |
|-------------|-----------|---------------------|--------------------|-----------------------|-----------------------|
| A           | 300       | 55                  | 60                 | 54                    | Conduction            |
| B           | 350       | 55                  | 80                 | 54                    | Conduction            |
| C           | 150       | 55                  | 60                 | 35                    | Transition            |
| D           | 180       | 55                  | 70                 | 35                    | Transition            |
| E           | 200       | 55                  | 90                 | 35                    | Transition            |
| F           | 240       | 55                  | 100                | 35                    | Transition            |

The AlSi10Mg cylinders were analyzed for porosity characteristics by a 3D X-ray computed tomography (XCT) scanner (ZEISS Xradia 520 Versa) using a 6 µm voxel size. To visualize the defect distribution within each sample, the CT scanned files were analyzed using an image processing software (Dragonfly 3.0, Object Research Systems Inc., Montreal, QC). The AlSi10Mg cubes were sectioned, polished, and etched with diluted phosphoric acid (9 g phosphoric acid
and 100 ml H₂O) for studying their microstructure. Micrographs were taken at various locations including the top edge and core of the cubes. Five different melt pools were analyzed for each cube to obtain the melt pool depth and half width measurements as shown in Figure 1.

![Micrograph](image)

*Figure 1. Measurement of the melt pool depth and half-width for one of the melt pools from sample E.*

### 3. Results and Discussion

In this work, laser beam defocusing was studied primarily with the intention of achieving stable conduction mode LPBF process parameters for AlSi10Mg. Based on literature discussed in section 3.1, it is observed that during the LPBF of AlSi10Mg, conduction mode melting is not reported for beam spot radiiuses below 50 µm. Since the Renishaw AM 400 has a beam spot radius of 35 µm at the focal plane, a transition or keyhole melting mode is hence expected, unless the beam is defocused. The defocusing of the beam was kept to a position above the build plate to create a divergent beam (to an effective beam spot radius of 54 µm) at the laser-material interaction plane, instead of convergent beams obtained by defocusing to positions below the build plate.
3.1. Defect space outcomes across melting modes based on micrographs

![Microstructure Diagram](image)

Figure 2. Typical microstructure of AlSi10Mg obtained with the focused beam (left, sample E) and with the defocused beam (right, sample B). The focused laser beam microstructure consists of numerous defects related to excessive vaporization in the transition/keyhole modes, while a few hydrogen solubility related defects are observed by using a diverging defocused beam.

The effect of the divergent beam defocusing strategy is illustrated in Figure 2, wherein numerous keyhole type defects are observed for the samples built with the focused beam with a beam spot radius of 35 µm, whereas a conduction mode type microstructure with a few tiny defects are observed for the samples built with the divergent beam, attributed to hydrogen induced defects in aluminium alloys as reported by Weingarten et al. [44].

The defocusing strategy is particularly effective for systems with lower beam spot radiuses (10-35 µm) [19], [20], wherein melt pool aspect ratios (melt pool depth/width) of greater than 0.5 have been reported for all ranges of powers of 100 – 400 W for AlSi10Mg, which is considered to be the primary process variable driving the threshold between conduction and transition/keyhole melting modes. For AlSi10Mg melt pool datasets reported with systems such as EOS M290 with a higher beam spot radius of 50 µm at the focal point, most melt pools had an aspect ratio lesser than 0.5, except for power settings above 275 W [45], [46]. Divergent beams help in reducing keyhole pores by reducing the effective beam power density as the
melt pool formation progresses and have been used successfully in the laser welding of aluminum alloys [47] and LPBF of 316L stainless steel [25]. The cause of this could be associated to a deviance of the beam profile from a Gaussian distribution to resemble more closely a top-hat distribution during the divergence of the beam as shown by Nie et al. [48]. Assuming this deviance of the beam profile, the threshold power required for surface vaporization (commonly assumed to be the threshold between the conduction and transition modes) would be higher for divergent beams when compared to focused beams, when the other variables are kept constant. This assumption is based on the temperature prediction models proposed by Graf et al. [49] for predicting the threshold of surface vaporization for Gaussian and top-hat beam profiles for materials with high thermal conductivity and low surface tension such as aluminium and copper alloys. Graf et al. [49] show that in a Gaussian beam, when all other variables are held constant, lesser power is needed for initiating surface vaporization due to the higher peak intensity in Gaussian beam profiles, when compared to the top-hat distribution. To better understand the effect of the melt pool morphologies on porosity, measurements of the melt depth, widths, and the melt pool aspect ratios from the six samples are given in Table 2.

| Sample Code | Depth | Width | Melt Pool Aspect Ratio | Inferred Absorptivity |
|-------------|-------|-------|-----------------------|----------------------|
| A           | 47.04 | 157.56| 0.3                   | 0.20                 |
| B           | 53.25 | 165.78| 0.32                  | 0.17                 |
| C           | 56.62 | 143.01| 0.4                   | 0.59                 |
| D           | 53.63 | 136.05| 0.39                  | 0.43                 |
| E           | 80.44 | 187.49| 0.43                  | 0.51                 |
| F           | 81.67 | 170.00| 0.48                  | 0.41                 |
As observed in Table 2, samples A and B have a melt pool aspect ratios of ≈0.3 whereas samples C, D, E, and F have melt pool aspect ratios of ≈0.4 and higher. By these melt pool aspect ratios, based on existing literature, it would mean that all the six sample types would be expected to lie in the conduction melting mode; however, the optical micrographs of the six samples along the build direction (Z-axis) shows otherwise. As shown in Figure 3, the optical micrographs of samples E and F, in particular, reveal numerous rounded defects representative of keyhole instabilities implying that these sets of processing parameters likely lie in the transition/keyhole melting mode. Hence, a melt pool aspect ratio of 0.4 might be likely more representative of the threshold between the conduction and transition/keyhole melting modes for AlSi10Mg. Samples C and D seem to have a similar porosity level to samples A and B in the microstructural images shown in Figure 3, but there are qualitative differences between the melt pool morphologies. The qualitative differences between the melt pool morphologies of samples C and D with respect to samples A and B are apparent by virtue of more variability in the melt pool layer-by-layer organization. The quantitative differences between the melt pool morphologies of samples A and B when compared to the rest are better represented in Table 2 by the lower standard deviations of their melt pool depths when compared to samples C, D,
and E. Additionally, samples C and D seem to have melt pool depths comparable to samples A and B even when the laser power settings used for them were close to half, which would imply much lower heat inputs. This is due to the onset of vaporization in these samples due to the focused beam, as predicted in the normalized processing diagram in Figure 4. $E^*$ and $v^*$ in Figure 4 are given by equations (2) and (3) respectively.

$$E^* = \frac{AP_{\text{eff}}}{2l_4\lambda(T_m - T_0)}$$  \hspace{1cm} (2)$$

$$v^* = \frac{v_f}{\alpha}$$  \hspace{1cm} (3)$$

In equation (2), $E^*$ is the dimensionless heat input, $A$ is laser absorptivity, $P_{\text{eff}}$ is the effective laser power [W], $l$ is the powder layer thickness [m], $\lambda$ is the thermal conductivity [W/(m.K)], $T_m$ is the melting temperature [K], and $T_0$ is the initial (or powder bed) temperature [K], taken as 293 K. In equation (3), $v^*$ is the dimensionless beam velocity, $v$ is the laser beam velocity [m/s], $r_b$ is the beam spot radius used [m], and $\alpha$ is the thermal diffusivity [m$^2$/s]. The material properties used for equation (2) and (3) are taken at the solidus temperature from [50] and are

![Normalized Processing Diagram for Six Sample Types](image)
given in Table 3. For a modulated LPBF system used in the present study, the effective laser power, $P_{\text{eff}}$, and the effective beam velocity, $v$, are given by equations (4) and (5) respectively.

$$P_{\text{eff}} = \frac{p \cdot t_e}{t_e + t_d}$$

$$v = \frac{p_d}{t_e + t_d}$$

Table 3. Thermo-physical properties of AlSi10Mg taken at the solidus temperature.

| Properties                | Material (AlSi10Mg) |
|---------------------------|---------------------|
| Density [kg/m$^3$]        | 2670                |
| Thermal Conductivity [W/(m.K)] | 113            |
| Specific heat [J/(kg.K)]  | 565.29              |
| Solidus temperature [K]   | 831                 |
| Liquidus temperature [K]  | 867                 |
| Vaporization temperature [K] | 2740         |

In equation (5), $t_e$ is the time when the laser is acting on the material (exposure time) [s] and $t_d$ is the time when the laser is turned off and is repositioning to the next exposure point (drill delay time) [s], taken as 10 µs for the Renishaw AM 400 system. In equation (5), $p_d$ is the distance between two consecutive laser exposure points (point distance) [m]. The temperature prediction model used for predicting the surface vaporization (conduction mode) threshold and the transition mode threshold in Figure 4 has been derived in previous work [23]. The absorptivity values used for equation (2) and given by the inferred absorptivity columns in Table 2 were obtained inversely by comparing the predicted melt pool depths with experimental measurements. The standard deviation bars for $E^*$ are based on the standard deviation for the absorptivities driven by the variation in melt pool depths for a given sample.

The temperature prediction model has some limitations such as assumptions of a 2D heat source, temperature independent material properties, oversimplification of powder layer thickness effects, and ignorance of heat loss by refraction in the vapour plume [23], which
would contribute to the uncertainty margins in the identified surface vaporization threshold. Additionally, the latent heat of fusion, thermo-capillary phenomena (Marangoni effect) and varying laser power absorptivity due to the its angle of incidence (Brewster effect) are not incorporated into this modelling approach which could add to uncertainties [51], [52]. The use of standard deviation bars for $E^*$ as inferred inversely via the melt pool datasets are also a reflection of some of the limitations in experimentally validating the precise location of each experimental point in the process map.

The onset of surface vaporization brings about a more pronounced change in laser absorption and thereby in melt pool behaviour for high reflectivity materials such as aluminium alloys. This is because the onset of surface vaporization adds to additional absorptivity (A) of the laser beam in the material that is equal to $1 - R^N$, where is R is the reflectivity a material, and N is the number of reflections occurring in the vaporized cavity of the melt pool [49]. Materials such as aluminium alloys with higher reflectivity values compared to titanium, ferrous, and nickel alloys would thereby be expected to have differences in melt pool behaviour (melt pool dynamics and thereby solidified melt pool geometry) after the onset of surface of vaporization is crossed. This points towards the differences in absorptivity values that were obtained for samples A to F from the effective absorptivity obtained through the depths reported in Table 2, and the temperature prediction model proposed in [23]. The absorptivities values obtained for samples A and B are between 0.15 and 0.2, whereas samples C, D, E, and F have absorptivity values ranging from 0.4 to 0.7, which corresponds to the conduction and transition mode absorptivities for aluminium, based on in situ measurements of laser absorptivity during LPBF [24], [53].

In terms of porous defects, samples A and B have a small population of defects as shown in Figure 3, with the smallest defects being rounded and commonly attributed to hydrogen-induced defects observed in conduction LPBF of AlSi10Mg; this is also observed by Weingarten et al. [44]. The presence of moisture on the powder surfaces, is one of the main causes attributed to the reduction of hydrogen solubility in aluminium alloys during the resolidification of liquid aluminium [54]. For samples E and F, additional larger defects were observed as seen
in Figure 3, particularly at the bottom of melt pool, close to the melt pool boundaries. The major source of these defects is expected due to the excessive vaporization of metal expected in transition and keyhole mode melt pools [23]. In conduction mode LPBF of aluminium, where vaporization is not expected, the measured absorptivity values for LPBF were ≈0.15 for a beam spot diameter of 60 ± 5 μm [24], [53]. However, the high reflectivity in such materials would be expected to aid the overall absorptance significantly once vaporization initiates due to increased number of reflections of the laser beam inside the vaporized region, as observed for aluminium discs in transition mode [24]. High-speed and high-resolution X-ray imaging of two aluminium alloys (AlSi10Mg and Al6061) during LPBF has shown that fluctuations in their vaporized areas of melt pools lead to instabilities and thereby to the formation of porous defects, even with a shallow depth of the vaporization regions in transition/keyhole mode melting due to an increased number of laser beam reflections in the melt pool [55]. A few of the excessive vaporization-related defects are also observed in sample D, as pointed by the red arrows in Figure 3.

3.2. Defect space outcomes across melting modes based on XCT

![Figure 5. A three-dimensional visualization of the porous defect (above 4 voxels) space along the build direction (Z) from the XCT data of the six sample types, along with the density values obtained based on the XCT data.](image)

To further understand the effects of transition and conduction mode on defect formation in LPBF of AlSi10Mg, a visualization of the three-dimensional porous defect space (obtained by
XCT) for all six samples are shown in Figure 5. Segmented defects with sizes below 5 interconnected voxels (voxel edge dimension is 6 μm) have been truncated out from the defect visualization and defect aspect ratio assessments since it is not possible to accurately separate features below this size due to instrument noise. The defect aspect ratio parameter is the ratio between the minimum and the maximum Feret diameter, where the minimum Feret diameter is the shortest length of a given feature, while the maximum Feret diameter is the longest span of a given feature, as described in [56], [57]. Defects with aspect ratios above 0.7 were considered as rounded defects in Figure 5. For calculating the density values shown in Figure 5, all the defects (defects with a voxel size of 1 or more) was considered. The density values are approximations of the true density and a relative assessment of part quality due to the voxel size detection limit. To visualize the locations of the defects, an orthographic projection along the build plate (XY) plane of all the porous defect space for each sample is shown in Figure 6.

![Image](image.png)

**Figure 6.** An orthographic projection of the porous defect (above 4 voxels) space along the build plate (XY) plane from the XCT data of the six sample types.

In Figure 5 and Figure 6, it can be observed that sample A has few irregular defects; these are lack of fusion defects, which are attributed to the slightly lower melt pool depths as noted in Table 2 [39]. There are numerous causes for lack of fusion defects in conduction mode LPBF such as, but not limited to incomplete melting of powder particles within one-layer, incomplete
re-melting of material ejecta from previous layers or from neighbouring scan tracks, or incomplete re-melting of irregular surface topographies from previous layers. Such defects can propagate across subsequent layers, resulting in irregularly shaped lack-of-fusion defects [58] that can be observed in Figure 5 and Figure 6.

The defect population in Samples C and D spans both irregularly shaped and rounded (near spherical) defects as seen in Figure 5 and Figure 6. Although the average melt pool depth obtained for these samples is slightly higher than sample A, the higher standard deviation in melt pool depths observed for samples C and D (Table 2) can lead to random regions where under-melting may occur if the melt pool is too shallow leading to irregularly shaped lack of fusion defects, or random regions where the process transitions into the keyhole melting mode leading to keyhole defects (Figure 3D). The average melt pool widths of samples C and D are also lower when compared to samples A and B, leading the possibility of lack of fusion defects caused by insufficient stitching of melt pools between hatches (hatch distance 100 μm) in a given layer. Additionally, as per Figure 4, it is predicted that surface vaporization has taken place in samples C, D, E, and F leading to the possibility of defects related to the melt pool instabilities during transition melting mode in LPBF caused by the interplay between the drag force induced by the melt flow, the thermo-capillary force cause by the surface temperature gradients, and the recoil pressure introduced by the onset of material vaporization [59]–[61]. The defects obtained due to material vaporization are known to have both rounded and irregularly-shaped morphologies [23], [62].

In a comparative study between the laser welding of an aluminium alloy and a ferrous alloy, the higher frequency of vaporized region collapse for aluminium alloys has been associated to the lower surface tension and viscosity of molten aluminium along with the presence of volatile magnesium which vaporizes a temperature much lower than that of aluminium [63]. These observations are in line with the hypothesis proposed by Tenbrock et al. [31], wherein the importance of keyholes as a quasi-black body might be more pronounced for materials with higher reflectivity (e.g. aluminium and copper alloys) when compared to titanium, ferrous, and nickel alloys. Melt pool splatter caused by the breaking up of elongated molten pool regions
near the side and real walls of the vaporized region in transition and keyhole melting modes also contribute towards defect formation in samples C, D, E, and F by adding to the roughness of a given layer [59]. Additionally, powder denudation [64], [65] and powder ejecta [66] contribute towards the surface roughness of a given layer, thereby deteriorating the wetting behaviour of the following layers causing melt pool instabilities and increasing the possibility of irregularly shaped defects as observed in the defect space visualization for samples C, D, E, and F [67].

Samples E and F are predicted to lie in the keyhole melting mode by the processing diagram shown in Figure 4. In the keyhole melting mode, vaporization related instabilities inside the melt pool would be expected to play a dominant role in the formation of rounded porous defects, as observed in Figure 5 and Figure 6. To confirm this prediction, Figure 7 shows plots of defect aspect ratios of the defects versus frequency and percentage of defect volume. The defect aspect ratio versus frequency plot in Figure 7 shows some indication to the preference of rounded defects in samples E and F, but the aspect ratio versus percentage of defect volume plot provides a better understanding of such behaviour. Since the curves for samples E and F lean towards a higher aspect ratio in the plot against percentage of defect volume in Figure 7, it implies that most of the defects in samples E and F have a rounded morphology especially when compared to the other four samples. The rare occurrence of irregularly shaped defects can be associated to the higher average melt pool depths reported in Table 2, which are above

![Figure 7. Aspect ratio versus frequency (left), and aspect ratio versus percentage of defect volume (right) from the XCT data for the six samples.](image-url)
two times the powder layer thickness used (30 μm). Typically, a melt pool depth of about 2 times (or more) the layer thickness is targeted in LPBF to avoid the possibility of lack of fusion defects [39].

Figure 6 shows a higher concentration of rounded porous defects near the side walls of the cylinders for samples E and F that can be related to the rapid formation and collapse of deep vaporized regions due to the laser beam velocity at the turn points which occurs at the edges of a given layer in LPBF, thereby trapping the atmospheric gas in the solidified part [61]. A plot of defect volume versus frequency in Figure 8 reveals that most of the defects in samples E and F still belong to the lower volume regions of below 0.0001 mm³. However, defects closer to the side wall of the cylinders would still be expected to impact its fatigue life, since the larger subsurface defects are the biggest factor impacting a parts fatigue life. The largest defect in sample F has a volume of 0.0053 mm³. If this largest defect is assumed to be spherical, we would get a defect diameter of 0.22 mm which is close to the defects diameter of ≈0.2 mm that led to the fatigue crack initiation in AlSi10Mg as shown by Plessis et al. [37].

Sample A and B which are predicted to lie below the surface vaporization threshold seem to have almost no subsurface defects with the sporadic occurrence of small defects typical to LPBF caused by random process factors or systematic machine biases [23]. Since no particular pattern is observed in the defect space for samples A and B from Figure 3, Figure 5, Figure 6, Figure 7, and Figure 8, the process parameter combinations involving the use of a divergent
beam for samples A and B might be best suited for near fully dense LPBF AlSi10Mg components, particularly for systems with a lower beam spot radius at the focal point.

Overall, when observing the melt pool morphology and porous defect characteristics across the melting modes in conduction (samples A and B), transition (samples C and D) and keyhole (samples E and F), the benefit of deploying divergent beams in materials with a high reflectivity and high thermal conductivity becomes apparent. For these material systems, the use of a divergent beam as an energy source results in a more stable melt pool morphology, lower occurrence of porous defects in the core and sub-surface regions, and overall lower porous defect volumes. Furthermore, for such approaches, the hatch spacing, and power levels can be further optimized to minimize lack-of fusion random defects. For this class of material systems, the use of a focused beam makes it challenging to find a process parameter combination that would result in stable melt pool morphologies, resulting in a higher occurrence of defects.

4. Conclusions

The potential of divergent laser beams to achieve stable conduction mode microstructure during the laser powder bed fusion of a high reflectivity aluminium alloy (AlSi10Mg) is investigated in this work. The key findings are summarized below:

1. Divergent beams help in avoiding keyhole defects by reducing the effective beam power density as the melt pool formation progresses, thereby leading to parts with densities of over 99.98%, with close to no porous defects in the subsurface regions.
2. A melt pool aspect ratio (ratio of depth to width) of ≈0.4 is observed to be the threshold between conduction and transition/keyhole mode melt pools in AlSi10Mg, which differs from the conventionally assumed melt pool aspect ratio of 0.5.
3. The inferred absorptivity values for conduction mode melt pools are between 0.15-2 while absorptivities of 0.4-0.7 are inferred for transition and keyhole mode melt pools, pointing to the significant differences in laser absorptivity following the onset of surface vaporization in aluminium alloys, when compared to titanium, nickel, and ferrous alloys, due to its higher reflectivity.
4. In general, a higher standard deviation was observed in the melt pool depths for the transition mode (10-15 μm) and keyhole mode (4-21 μm) melt pools, when compared to the conduction mode (6-8 μm) melt pools. The predicted absence of vaporization in conduction mode melt pools could be the cause for the relatively stable melting behaviour, when compared to transition and keyhole mode melt pools.

The application of the methods proposed in this work can help to quickly identify stable conduction mode LPBF processing parameters for high-reflectivity aluminium alloys. The presence of close to no defects even near the boundaries of LPBF components helps increase the confidence of the process for load bearing and mission critical applications in particular.

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