Influence of scan strategy and molten pool configuration on microstructures and tensile properties of selective laser melting additive manufactured aluminum based parts

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Abstract
Selective laser melting additive manufacturing of the AlSi12 material parts through the re-melting of the previously solidified layer using the continuous two layers 90° rotate scan strategy was conducted. The influence of the re-melting behavior and scan strategy on the formation of the “track-track” and “layer-layer” molten pool boundaries (MPBs), dimensional accuracy, microstructure feature, tensile properties, microscopic sliding behavior and the fracture mechanism as loaded a tensile force has been studied. It showed that the defects, such as the part distortion, delamination and cracks, were significantly eliminated with the deformation rate less than 1%. The microstructure of a homogeneous distribution of the Si phase, no apparent grain orientation on both sides of the MPBs, was produced in the as-fabricated part, promoting the efficient transition of the load stress. Cracks preferentially initiate at the “track-track” MPBs when the tensile stress increases to a certain value, resulting in the formation of the cleavage steps along the tensile loading direction. The cracks propagate along the “layer-layer” MPBs, generating the fine dimples. The mechanical behavior of the SLM-processed AlSi12 parts can be significantly enhanced with the ultimate tensile strength, yield strength and elongation of 476.3 MPa, 315.5 MPa and 6.7%, respectively.

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1. Introduction

Selective laser melting (SLM), as a typical powder bed melting process of the additive manufacturing (AM) process, is increasingly paid attention to both in the academic and industrial fields from the perspective of the flexibility of fabricating near fully dense intricate metal or metal matrix parts in a fairly high degree of accuracy and a cost effective manner [1–3]. During the SLM process, metal powders are heated and subsequently melted by a fast moving high energy laser beam and then the molten pool solidifies rapidly associated with a series of mass, heat and momentum transition in the non-equilibrium metallurgical process, leading to the significant differences in mechanical properties of the SLM-processed parts compared with the casting and forged parts [4–6].

It is obvious that the characteristics of the SLM-processed parts possess the higher tensile strength combined with the undesired low dimensional accuracy and the ductility performance [7]. Prashanth et al. found that the yield and tensile strength of the SLM-processed AlSi12 specimens with 260 MPa and 380 MPa were four and two times higher than the corresponding properties of the samples processed by the casting, while the elongation was merely 3%, which was significantly decreased with respect to the casting sample [8]. The yield stresses of a series of materials (316 stainless steel, CoCr and In625) fabricated by SLM process were significantly increased by ~50% while the elongation was unfortunately decreased with 20–30% of the forging specimens, which have been systematically concluded by Yadroitsev et al. [9]. In the case of the AlSi12 the variation of the mechanical properties is not sensitive to the building direction angles to the substrate, however, a significant dependence of the mechanical properties on the application of the scan strategy has been studied and concluded by Geiger [10]. Generally, the parts fabricated by SLM process are periodically completed with the feature of the overlapping of
multi-tracks and multi-layers [11]. The molten pool boundaries (MPBs) will be accordingly obtained with an arc-shaped configuration connected with the neighboring tracks or the surfaces of the previously solidified layers. The MPBs produced through the laser melting process for the various materials have been studied in some previous researches. Thijs et al. has found that the dark band regions, indicating the MPBs in the top view of the SLM-processed AlSi10Mg part using the long bidirectional scanning tracks, were visible due to the distinction of the three different zones and the formation of different microstructures caused by the various solidification modes [12]. The same opinion was supported by Zheng et al. who proposed that the visible dark bands indicating MPBs were typical microstructures derived from the various heat effects and the resultant various crystal growth modes in the SLM-processed parts [13]. Two types of the MPBs in the SLM-processed 316 L stainless steel parts were defined by Wen et al. and then, the effects of the MPBs on the tensile properties fabricated along different directions and the fracture mechanism have been studied [14]. The microstructure of the MPBs differs from that of the other regions caused by the lower solidification rate and higher temperature gradient obtained within the molten pool, which was highly sensitive to the SLM processing parameters and the material properties [15]. Moreover, a complexly spatial topological structure in the connecting regions of the multi-track and multi-layer MPBs plays a crucial role in the performance of the SLM-processed part. Therefore, the microstructure of AlSi12 material fabricated by SLM process modified by the scan strategy is an alternative way to achieve the synergy ascension of the tensile strength and the elongation.

The objective of this study is to analyze the effect of the combination of the re-melting process and the continuous two layers 90° rotate scan strategy on the SLM process stability, dimensional accuracy, microstructure feature and the tensile properties. The temperature contour and the velocity field predicted by the finite volume method with the optimized processing parameters to investigate the re-melting process and the densification behavior are conducted. The influence of the continuous two layers 90° rotate scan strategy on microscopic sliding behavior, macroscopic ductility and fracture mechanism of the as-fabricated samples as loaded a tensile force is analyzed and assessed through experiments. The correlation of the enhanced tensile properties combined with the characteristics of the microstructure and the sliding behavior has been elucidated. This study may provide an alternative for the optimization and fabrication of the high performance of SLM-processed AlSi12 parts.

2. Experimental methods

2.1. Powder materials and SLM process

Bulk parts were produced by SLM process from the commercial and spherically shaped powder material prepared by the gas atomization and the particle size is in the range of 20–45 μm (Fig. 1a). The SLM apparatus is independently developed by Nanjing University of Aeronautics and Astronautics (NUAA), mainly equipped with the YLR-500 Ytterbium fiber laser (with a maximum laser power 500 W, a spot size of 70 μm and a continuous wave-length of 1070 ± 10 nm), an automatic powder deposition device, an inert argon gas protection system and the process control system. In order to find the optimal processing parameters for the laser-material interaction, the re-melting process and densification behavior of the previously solidified layer, four cubic parts with the dimensions of 10 mm × 8 mm × 5 mm were fabricated. The single line laser scan strategy was applied and the processing parameters were set as follows: laser power 400 W, layer thickness 30 μm, hatch spacing 60 μm and the scan speeds 4000 mm/s–1000 mm/s. In order to investigate the laser energy input to the powder layer being processed, the integrated parameter, laser volume energy density η, was defined:

$$\eta = \frac{P}{h \cdot l}$$  \hspace{1cm} (1)

and therefore, four laser volume densities of 55 J/mm³, 70 J/mm³, 110 J/mm³ and 220 J/mm³ were selected to study the complexly metallurgical behavior and, as the optimized processing parameters have been verified, the tensile testing parts, with a flat dog-bone shape as proposed by GB/T 228.1-2010, were constructed.

2.2. Microstructural characterization and mechanical tests

After the SLM process, samples were cut perpendicular to the laser beam incident direction, ground and then polished following classical procedures and, the cross-sections of the individual parts and the tensile fracture samples were characterized using a PMG3 optical microscopy (Olympus Corporation, Japan). Specimens for metallographic examinations were prepared according to the standard procedures, and then etched with a solution composing HF (2 ml), HCl (3 ml), HNO3 (5 ml) and distilled water (190 ml) for 10 s. Microstructure observations of the SLM-processed specimens and the fracture surface morphology were detected by a Zeiss Sigma 04-95 field emission scanning electron microscope (FESEM). The chemical compositions were tested using a S-4800 field emission scanning electron microscope (FESEM) (Hitachi, Japan) equipped with an EDAX energy dispersive X-ray spectroscope (EDX). Tensile tests were carried out at room temperature using a CMT5205 testing machine (MTS Industrial Systems, China) with a cross head velocity fixed at 2 mm/min.

3. Finite volume simulation

Finite volume simulation using the commercial computational fluid dynamics software FLUENT was conducted to study the thermal evolution behavior (focusing on the temperature field in the neighboring region located in the “track-track” boundary to study the re-melting behavior) and the melt mass transfer behavior (focusing on the melt flow characteristics in the neighboring region of “track-track” boundary to investigate the densification behavior) during the SLM process. The as-used AlSi12 thermo-physical properties are depicted in paper [16,17]. The Gaussian laser source is moved along the X-axis direction with various scan speeds and, the laser heat source is mathematically defined as a heat flux
which is inserted in the source term for the governing equations. The model established in this study has been described in our previous works [18–20]. Based on the mass, momentum and energy equation, the governing equations in the Cartesian coordinate system are summarized as follows [18]:

Mass conservation equation

$$\frac{\partial \rho}{\partial t} + \nabla (\rho \vec{V}) = M_i$$

(2)

Momentum conservation equation

$$\rho \left( \frac{\partial \vec{V}}{\partial t} + \vec{V} \cdot \nabla \vec{V} \right) = -\nabla p + \rho \vec{g} + \frac{\partial \tau}{\partial \xi} + \rho \vec{F}$$

(3)

Energy conservation equation

$$\frac{\partial (\rho T)}{\partial t} + \frac{\partial (\rho u T)}{\partial x} + \frac{\partial (\rho v T)}{\partial y} + \frac{\partial (\rho w T)}{\partial z}$$

$$= \frac{\partial}{\partial \xi} \left( \kappa \frac{\partial T}{\partial \xi} \right) + \frac{\partial}{\partial y} \left( \kappa \frac{\partial T}{\partial y} \right) + \frac{\partial}{\partial z} \left( \kappa \frac{\partial T}{\partial z} \right) + S_h$$

(4)

where $\rho$, $\kappa$, $\mu$, and $p$ are the density, thermal conductivity, viscosity and pressure, respectively, $M_i$ is a mass source. $\vec{V}$ is the molten metal velocity. (see Table 1)

4. Results and discussion

4.1. Re-melting process of the previous layer and densification behavior

Optical microscopes (OM) of the cross-sectional macrostructures of the SLM-fabricated different tracks and layers using various processing parameters are given in Fig. 2. Upon etching, the layer-wise feature of the additive manufacturing process and the single track molten pool boundary were apparent. The solidified molten pools were typically shown in the arc-shaped configuration, which was mainly caused by the application of the Gaussian energy distribution of the laser beam. It was obvious that the overlapping quality of the adjacent molten pool boundaries that were parallel or perpendicular to the building direction is highly dependent on the processing parameters. The “layer-layer” MPBs and “track-track” MPBs were generated through the layer-by-layer overlapping perpendicular to the building direction and the track-by-track overlapping parallel to the building direction, respectively. As the applied $\eta$ was 55 J/mm², the MPBs were significantly disordered within the solidified part due to the formation of the irregular pores with the mean size of 130 $\mu$m, which was crucially detrimental to the relative density enhancement (Fig. 2a). The irregular pore pattern depicted is typically ascribed to the low energy input and the resultant insufficient melting and spreading of the melt to the previously solidified track or layer [21]. As the applied $\eta$ increased to 70 J/mm², the contour of the MPBs was shown in a more disciplined pattern combined with the residual spherical pores of the decreased size (30 $\mu$m) remaining in the neighboring area of the MPBs (Fig. 2b). As the applied $\eta$ increased to 110 J/mm², the metallurgical bonding quality of the MPBs was significantly enhanced with the disappearance of the metallurgical defects located in the “layer-layer” MPBs. Meanwhile, it seemed that the residual pores were obviously shifted upward along the building direction, with the location higher than those of the “layer-layer” MPBs (Fig. 2c).

Therefore, it can be concluded that the densification rate of the molten pool is significantly enhanced with the entrapped gas shifting upward and the subsequent escaping behavior [22,23]. For the applied $\eta$ of 220 J/mm², a good metallurgical bonding ability of the MPBs was successfully obtained with the formation of the denser part free of the residual pores, which was caused by the sufficient re-melting process and the resultant densification behavior of the upper region within the previously solidified part (Fig. 2d). The metallurgical bonding ability of the MPBs was highly dependent on the processing parameters and the resultant re-melting behavior of the previously deposited layers.

In order to have a better understanding of the re-melting behavior during the SLM process, the temperature contour and the velocity distribution for $\eta = 220$ J/mm² are demonstrated in Fig. 3. The temperature contour from the top view was distributed following the Gaussian function and the temperature isotherm lines in the neighboring center region of the molten pool were more closer with the maximum operating temperature as high as 1100 K, which was above the melting temperature of AlSi12 material (Fig. 3a and c), implying the generation of the complete melting of the powder material in the irradiated region. Therefore, the velocity vector of the thermo-capillary convection in the free surface of the molten pool was typically shown in the radially outward pattern with the maximum velocity of 1.2 m/s caused by the temperature gradient and the resultant surface tension (Fig. 3c). From the front view, it was obvious that the depth of the molten pool was larger than the initial depth of the powder layer (Fig. 3b), indicating the formation of the re-melting of the previously solidified layer. The melt with the relative low operating temperature migrated from the bottom to the top region and then, the melt carried the energy from the laser beam to the bottom area along the molten pool boundary, generating a circulating flow and re-melting of the beneath solidified layer. The entrapped gas will float up and escape from the molten pool driven by the thermo-capillary convection. Meanwhile, it was interesting to find that the velocity vector within the molten pool was typically in the upward pattern with the velocity increasing from 0.1 m/s to 1.2 m/s (Fig. 3d) and, it seemed that the material evaporation was inevitably avoided because of the employment of the high laser energy. As studied by Yu et al. [24], the residual pores gathered in the upper region within the previously solidified layer would be reasonably eliminated due to the escape of the entrapped gas forced by the thermo-capillary convection during the re-melting process, leading to the significant enhancement of the relative density.

4.2. Scan strategy and dimensional accuracy

The schematic of the continuous two layers 90° rotate scan strategy is depicted in Fig. 4a. The real time interaction of the laser beam and powder material is depicted in Fig. 4b. The spark formation in the irradiated region within the laser scan direction along X-axis direction and Y-axis direction was bright, implying the efficient interaction of the laser beam and the powder material. The

<table>
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<td>Thermo-physical properties of AlSi12 applied in this study (temperature, in K).</td>
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<td>Quantity</td>
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<td>Thermal conductivity</td>
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<td>Density, $\rho$</td>
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<td>Heat capacity, C</td>
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Fig. 2. Optical microscope (OM) of the cross-sectional macrostructures of the SLM-processed parts ($P = 400$ W, $l = 30$ μm and $h = 60$ μm): (a) $v = 4000$ mm/s, $\eta = 55$ J/mm$^3$; (b) $v = 3000$ mm/s, $\eta = 70$ J/mm$^3$; (c) $v = 2000$ mm/s, $\eta = 110$ J/mm$^3$ and (d) $v = 1000$ mm/s, $\eta = 220$ J/mm$^3$.

Fig. 3. Temperature contour and velocity distribution during the SLM process using the processing parameters, $P = 400$ W, $l = 30$ μm, $h = 60$ μm, $v = 1000$ mm/s, $\eta = 220$ J/mm$^3$: (a) top view, (b) cross-section view, (c) position along X direction for $Y = 0$ and $Z = 0$, (d) position along Z direction for $Y = 0$ and $X = 0$. 
steady SLM process with a small spark was typically obtained as the laser beam scanned along Y-axis, which was reasonably caused by the heat effect introduced by the shorter raster scan path and the subsequent heat accumulation. The influence of the scan strategy on the dimensional accuracy of the as-fabricated part is shown in Fig. 4c. It was apparent that the defects, such as part distortion, delamination and cracks derived from the process-induced residual stress, were significantly eliminated with the deformation rate less than 1%. Meanwhile, it was found that a flat and dense top surface with the formation of the coherent bonding ability was successfully obtained by the application of the continuous two layers 90° rotate scan strategy (Fig. 4d), as a reduced edge height of the build part along building direction was generated under the combined effect of the raster scanning and the 90° rotate scan strategy. It implied that the adhesion behavior of the initial powder particles to the edge of the as-fabricated part and the scattering preference of the metallic balls to the side of the scan track was significantly restricted, which promoted the formation of the reduced side roughness and the generation of the high accuracy of the SLM-processed part. Some small balls with the size approximately equal to the diameter of the laser beam were fragmentary adhered on the smooth scan tracks, which could be ascribed to the comprehensive influence of the temperature gradient, the surface tension and the resultant Rayleigh instability within the molten pool [25].

4.3. MPBs within the as-fabricated parts

Optical microscope (OM) of the microstructures obtained in the SLM-processed parts using the continuous two layers 90° rotate scan strategy with the optimized processing parameter is depicted in Fig. 5a and b. Upon etching, the long scan vectors and the half-cylindrical shape of the molten pool were visible in the side views (Fig. 5a). Due to the partial re-melting of the previously solidified layers and the disturbance of the height of the molten pool, the longitudinal section of the molten pools from different layers can be visible in the top view (Fig. 5b). The apparent boundaries within the neighboring scan tracks were typically produced due to the heat affected zone around the molten pool. As a laser energy distribution following the Gaussian function is applied, a steep temperature gradient from the center region to the edge of the molten pool will be attendant generated and as a result, the thermo-capillary convection exhibiting a radially outward flow pattern on the free surface of the molten pool is typically formed. The residual gas entrapped in the upper region of the previously deposited layer will be subsequently floated up and escaped from the free surface of the molten pool, completing the re-melting and re-densification process (Fig. 5c). Therefore, a dense part with the perfect metallurgical bonding ability was generally produced (Fig. 5a and b). The schematic of the “layer-layer” MPBs and the “track-track” MPBs is depicted in Fig. 5d. It was obvious that the “layer-layer” MPBs had an almost parallel relationship and the length of the “layer-layer” MPBs was much longer compared with that of the “track-track” MPBs, implying that the “layer-layer” MPBs have more efficient bearing capacity to suffer from tensile force loading due to the lower force acting within the per unit area.

4.4. Microstructure of the as-fabricated parts

The typical microstructure of the SLM-processed part using the optimized processing parameters is shown in Fig. 6, where the brighter area is Si element and the grey area is Al element. No apparent cracks and porosities can be found, implying a good processability and the attendant dense part (Fig. 6a). The Al and Si elements were detected with the Al:Si atomic ratio of about 88:12 (Fig. 6b), which was unchanged before and after the SLM process. It was obvious that the much smaller grain size (400 nm) obtained in the SLM-processed AlSi12 using the continuous two layers 90° rotate scan strategy (Fig. 6c) than those obtained in the previously reported SLM-processed AlSi12 alloy (10 μm) [26]. The precipitation of Si phase was randomly oriented which was reasonably confirmed by the EDS mapping results (Fig. 6d-f). It was interesting to
find that the homogeneous distribution of the Si phase and no apparent grain orientation on both sides of the MPBs (“layer-layer” and “track-track”) were produced in the as-fabricated part (Fig. 6a-b), which was different to the epitaxial precipitation and growth of the Si phase and strong texture caused by the directional heat flux and thermal gradient during the SLM process. For the grain orientation consistent through the MPBs, the force will be uniform in the neighboring area of the MPBs during the tensile loading test. Otherwise, the force will not be delivered through the MPBs in an effective way, resulting in the formation of initial cracks and the early failure. The homogeneous dispersion of the refined Si particles embedded in the neighboring region of the MPBs with the average size of 400 nm was beneficial to the efficient transmission of the force through the MPBs and meanwhile, the refined and dispersive Si particles played a crucial role in the force delivering between the Si particles and the Al matrix, which is called “grain boundary strengthening” through the Hall-Petch relation.

4.5. Tensile properties of the as-fabricated part

The non-standard tensile sample was applied as shown in Fig. 7a. The SLM-processed and tested samples are depicted in Fig. 7b. The tensile fracture occurred near the middle of the samples. The OM microscopic morphology on the longitudinal section parallel to the X-axis after tensile fracture is represented in Fig. 7c. Some micro-pores with the large size could be found in the “track-track” MPBs and, it could be reasonably concluded that the “track-track” MPBs were the main issue that gave rise to the formation of initial cracks and the early failure. The homogeneous dispersion of the refined Si particles embedded in the neighboring region of the MPBs with the average size of 400 nm was beneficial to the efficient transmission of the force through the MPBs and meanwhile, the refined and dispersive Si particles played a crucial role in the force delivering between the Si particles and the Al matrix, which is called “grain boundary strengthening” through the Hall-Petch relation.

4.6. Effect of MPBs and grain boundary on the tensile properties

Generally, the specimen ductile deformation fabricated by the traditional casting and forging methods is typically ascribed to the grain slipping. However, for the SLM-fabricated specimens, an additional influencing factor, MPBs, has to be taken into account during the ductile deformation loaded by the tensile force. In the ductile deformation of the as-fabricated part, the sliding tends to occur along the MPBs, caused by the relative weak bonding ability compared with the grain boundaries. The schematic analysis of the sliding caused by the interaction of the shear stress and the “layer-layer”/“track-track” MPBs using the continuous two layers 90° rotate scan strategy is demonstrated in Fig. 9. The shear stress $\tau$ along the sliding direction can be expressed as [14]:

$$\tau = \frac{F \cos \theta}{A \cos \lambda} = \frac{F \cos \theta \cos \lambda}{A}$$

where $A$ is the area of the cross section, $F$ is the tensile loading force, $\lambda$ is the angle between the tensile force $F$ and the normal direction of the sliding surfaces, $\theta$ is the angle between the tensile force $F$ and the sliding direction. As the tensile loading force increases, the
tensile stress \( (F/A) \) will attendant get close to the yield point \( (\sigma_s) \) and as a result, a critical value of the shear stress, \( \tau_c \), will be obtained on the sliding surface, resulting in the propagation of the sliding. The critical shear stress, \( \tau_c \), can be expressed:

\[
\tau_c = \sigma_s \cos \theta \cos \lambda
\]  

and therefore, the yield limit, \( \sigma_y \):

\[
\sigma_y = \frac{\tau_c}{\cos \theta \cos \lambda}
\]

It can be seen that the critical shear stress has nothing to do with the tensile loading force and, the yield limit, \( \sigma_y \), is highly dependent on the angle between the tensile force \( F \) and the sliding direction. The yield limit reaches infinity as \( \theta = 0^\circ \) or \( 90^\circ \), which means that the tensile loading force is parallel or perpendicular to the sliding surface and prevents the occurrence of the sliding surface. While as \( \theta = 45^\circ \) and \( \sigma_y = 2\tau_c \), the formation of the sliding surface is easy due to the formation of the minimum value of \( \sigma_y \). For the deposited layers fabricated by the SLM process along the Y direction, the sliding surfaces of the “layer-layer” MPBs are generally parallel to the tensile loading force \( (\theta_L = 0^\circ) \) and therefore, the sliding surface along the “layer-layer” MPBs is impossible. Consequently, the ductile deformation is mainly dependent on the sliding surfaces along the “track-track” MPBs. For the deposited layers fabricated along the X direction, the ductile deformation is consisting of the combination of the sliding surfaces of the “layer-layer” and “track-track” MPBs and, the angle between the tensile loading force and the sliding surface changes with the application of the scan strategy. In theory, the sliding surfaces within the two types of the MPBs are difficult as \( \theta_L = 90^\circ \) or \( \theta_T = 0^\circ \). However, it is well known that the two types of the MPBs are not in the severe planar pattern caused by the disturbance of the dimensions of the molten pool, leading to the formation of the
ductility deformation. The excellent ductility will be reasonably obtained at $\theta_T > 0$ and $\theta_L < 90^\circ$, due to the generation of easy sliding surfaces. The distance of the adjacent “track-track” and “layer-layer” MPBs sliding surfaces are respectively equivalent to the hatch spacing (60 $\mu$m) and the layer thickness (30 $\mu$m) as the tensile force is loaded. It is apparent that a combined advance of the “track-track” and “layer-layer” MPBs sliding surfaces is realized using the continuous two layers 90$^\circ$ rotate scan strategy, leading to the formation of the excellent tensile strength and ductility of the SLM-processed samples. According to the Hall-Petch relationship, the contribution of the grain boundary strengthening mechanism ($\Delta\sigma_{\text{Hall-Petch}}$) can be expressed as below [26–28]:

$$\Delta\sigma_{\text{Hall-Petch}} = kd^{1/2}$$

where $k$ is a constant and $d$ is the grain size. As reported by the researcher, $k$ is generally set to 50 MPa $\sqrt{\mu m}$. Therefore, the calculated strength increment by the grain boundary strengthening mechanism is nearly 50 MPa.

Fig. 10 shows the fracture surface morphologies of the as-fabricated sample after room temperature test. It was obvious that a step-wise morphology and the cleavage surfaces could be found on the tensile fractured surface (Fig. 10a). It seemed that the fracture tended to propagate in the “track-track” MPBs with a low elongation feature, where the heat affected zone and the limited ductile deformation was produced (Figs. 5b and 8). Meanwhile, the ductility arose from the “layer-layer” MPBs and the Al-rich z-Al phase and thus, the initiation of the cracks was either at the “track-track” MPBs or at the cleavage of the Si phase. A large
amount of fine dimples with the dimension of ~300 nm on the fractures surface were produced, revealing the ductile fracture feature and perfect ductility during the tensile test (Fig. 10c). The notably uniform dimple size was similar to the grains of the SLM processed part due to the complete melting of powder particles. It was interesting to find that a sliding direction of the extension dimples, corresponding to the “sliding or in-plane shear mode”, was typically obtained, implying that the direction of the tensile force was almost parallel to the planar pores or cracks (Fig. 10d). Cracks preferentially initiated at the “track-track” MPBs when the tensile stress increased to a certain value, resulting in the formation of the cleavage steps along the tensile loading direction. The cracks propagated along the “layer-layer” MPBs, generating the fine dimples.

5. Conclusions

In this study, the fully dense and crack-free parts were fabricated using the continuous two layers 90° rotate scan strategy with the optimized processing parameters \((P = 400 \, \text{W}, \, v = 1000 \, \text{mm/s}, \, l = 30 \, \mu\text{m}, \, h = 60 \, \mu\text{m}, \, \eta = 220 \, \text{J/mm}^3)\), allowing to draw the following conclusions:

1. Re-melting of the previously solidified layer was proposed as an efficient method to eliminate the residual pores during the SLM of AlSi12 material and, as a higher laser volume energy density (e.g. \(\eta = 220 \, \text{J/mm}^3\)) was applied, the efficient escape of the entrapped gas located in the upper region of the previously solidified layer would be produced, promoting the formation of high densification.
2. The samples fabricated by the SLM process using the continuous two layers 90° rotate scan strategy had a high dimensional accuracy. The homogeneous dispersion of the refined Si particles embedded in the neighboring region of the MPBs was obtained and in contrast to the previously reported results of the SLM-processed AlSi12 part, the textureless microstructure of the SLM-processed AlSi12 material was obtained with the significantly refined grain size (~400 nm).
3. The “layer-layer” MPBs and the “grain boundary strengthening” were the main issues that significantly enhanced the tensile properties. The fracture propagation in the “track-track” MPBs was considerably restricted by the application of the continuous two layers 90° rotate scan strategy using the optimized processing parameters. The cracks propagated along the “layer-layer” MPBs, generating the fine dimples. The calculated contribution of the grain boundary strengthening mechanism was nearly 50 MPa.
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References