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The mechanism of substructure formation and grain growth 316L stainless steel by selective laser melting

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Abstract

This paper aims to investigate the unique substructure and grain growth of 316L stainless steel processed by selective laser melting (SLM) and clarify the mechanism. Results showed that the grain orientation on the x-z plane parallel to the build direction was the same type as the y-z plane, which all grow along the heat flow direction to form elongated columnar grains passing through the multi-layer fusion line. The epitaxial growth direction of grains changed abruptly with the vary of temperature gradient direction. Moreover, to expound the substructure formation based on solidification theory, a finite element model was established to obtain the distribution map of cooling rate G × R and solidified morphology G/R.

1. Introduction

Selective Laser Melting (SLM) is a typical 3D additive manufacturing technology representing a realistic alternative to many conventional manufacturing techniques. And to achieve single-piece small-batch rapid manufacturing of high value-added parts with complex integrated structures in multiple fields [1, 2]. Stainless steels are currently one of the most widely studied materials in metallic additive manufacturing, especially 316L stainless steels widely used in various fields due to their high strength and excellent ductility [3]. In addition, the austenitic 316L stainless steel is a significant engineering material that is broadly used in the oil and nuclear industries and the biomedical field due to its high corrosion and oxidation resistance. Due to its broad applications and excellent weld-ability has become a material-of-choice for many laser-based additive manufacturing process studies [4].

According to Yin et al [5], affecting factors of yield strength are mainly attributed to cellular substructures surrounded by dislocation walls formed during SLM. Such a high dislocation density (1.14 × 1015 m−2) contributes to yield strength different from conventionally 316L stainless steels. Prior to this, widespread reports have also described this phenomenon of high yield strength [6–9]. Furthermore, high uniform elongation is associated with microstructures whose length scales span multiple orders of magnitude, and the solute segregation within the material can pin dislocations and promote twins. Although the unique microstructure and high properties make important contributions to practical applications, such as linking scanning speed to cell spacing [10, 11] or cooling rate during solidification [12, 13], the relationship between selective laser melting process parameters and solidification microstructure is still a lack of in-depth understanding. Moreover, the interaction area of the laser and the material is only the size of the laser spot irradiation area. In such a small area, the material melts from powder to liquid and then solidifies into the solid state. Therefore, traditional experimental methods are very difficult to study the selective laser melting process. Numerical simulation has become an effective research method. Numerical simulation has become an effective research method. Numerical simulation can better understand the relationship among process parameters, solidification microstructures and mechanical properties. For example, Gusarov et al [14, 15] analyzed the temperature field...
distribution of SLM 316L stainless steel using the coupled model of heat radiation and heat conduction. Khairallah et al. [16] combined thermal diffusion with hydrodynamics and took into account temperature-dependent material properties, surface tension, as well as random particle distribution, and simulated the selective laser melting process at a three-dimensional mesoscopic micrometer scale. Körner et al. [17] adopted the Lattice Boltzmann method model and utilized numerical simulation to track the size and direction of the temperature gradient to study the characteristics of grain growth with different parameters. It is found that the solidification dominated by the building direction provokes columnar microstructures whereas equiaxed grains result if constantly varying solidification direction between successive layers. Markl et al. [18] further studied the growth process of grains during solidification by using the coupling of the CA model and the LB model. They believe that the formation of stray grains was due to insufficient energy input caused by incompletely melted particles and the layer defects at the boundary.

The distribution characteristics of the temperature field and flow field collectively determine the interface temperature gradient G and the solidification rate R of the melt, thereby confirming the final microstructure. This study mainly investigated microstructure orientation and micro-substructure, etc. Combining with the temperature gradient G and solidification rate R in the finite element model, which expound the formation of substructures and research the solidification growth direction of grain according to the temperature field characteristics.

2. Experiments and simulations

The gas-atomized 316L stainless steel powders (JINGYE GROUP) with spherical are used for SLM manufacturing. The powder morphology and particle size distribution are shown in figure 1. It can be seen that the particle size conforms to the normal distribution, and the sphericity is 0.851. The chemical composition of the powder is Fe-0.03C-1.0Si-0.6Mn-16.7Cr-11.9Ni-2.5Mo in wt%. The EOS M290 selective laser melting device equipped with a 400W Yb-fiber fiber laser to fabricated samples with the size of 15 mm × 15 mm × 50 mm in a protective atmosphere. The SLM processing parameters include laser power 135 W, scanning speed 750 mm/s, scanning hatch 0.09 mm, and layer thickness 20 μm. Adopt a cross-hatching scanning strategy, and each layer should be at 67° rotation to the previous layer. The oxygen content during SLM is less than 135 ppm.

Microstructures were analyzed by QUANTAFEG450 field emission scanning electron microscope (SEM) and electron backscatter diffraction (EBSD), the SEM equipped with the Auger electron spectrum (AES) was used for line scan analysis. The Bruker D8 x-ray diffractometer (XRD) was used to determine the phase of samples. In the experiment, the Cu target was selected with a voltage of 40 kV and a current of 40 mA, and the XRD results were analyzed using MDI jade6.0 software. After grinding and polishing, the prepared samples were electrolytically etched in a 10% oxalic acid solution for 50 s for SEM observation. The EBSD samples were electropolished in the mixed solution of 10% glycerol + 20% perchloric acid + 70% alcohol at a voltage of 12 V for the 40 s, keeping the current stable at about 0.7 A.

The three-dimensional mathematical model was established by using finite element simulation software. The high-energy laser beams irradiate to the powder bed surface with constant power, and the metal powders melt into the liquid to form melt pools. When the lasers moved at a certain speed, parts of the energy reflected back in thermal convection or radiation. The other parts were absorbed by the powders and substrate directly.
and transfer heat to the interior through heat conduction, and solidify into a solid as the heat dissipates. In order to study the formation mechanism of substructures by temperature gradient G and solidification rate R in the SLM process, this paper selected the same process parameters as the experiment to simulate the thermal field and flow field of the material being heated by a laser. The formed specimen and the schematic diagram of the SLM process model in the same parameter are shown in figure 2. The density measured by the Archimedes method is 7.925 g cm\(^{-3}\), reaching 99.3% density.

3. Results

3.1. Substructure
The general cooling rate during solidification is \(10^{-2} \sim 10^{2}\) K s\(^{-1}\) [19]. Under the conditions of SLM forming, the melt pool is only a few hundredths of cubic millimeters in size and a few micrograms in mass. For such a small melt pool, the base metal is a giant heat sink, and the cooling rate of molten metal can reach \(10^6 \sim 10^8\) K s\(^{-1}\). The unique structure formed after SLM is attributed to local areas undergoing two extreme processes in a very short time. Figure 3 shows the microstructure of conventional and SLM 316L samples. Conventional 316L structure is equiaxed single-phase austenite with uniform grain size and no substructures. The XRD results showed that SLM 316L was also a fully austenitic single-phase structure. It could be distinctly observed that there were nearly hexagonal and elongated hexagonal cell and columnar substructures uniformly distributed inside the grains according to SEM images, and no twins were found compared with conventional. The local magnification in figures 3(c)–(d) shows that the size of the cell substructure was 1 \(\mu\)m, and the width between the columnar substructures was 400–500 nm.

The line scan results of the cell structure using Auger electron spectrometer (AES) as shown in figure 3(e), the elements Mo, Cr, and Ni segregate in the cell wall. The cell structures formed by solidification were not equivalent to the traditional high-density dislocation wall but only similar in morphology.
Figure 3. Microstructure of conventional and SLM 316L samples, (a) OM and (b) SEM images of conventional 316L, (c) cell and (d) columnar structure of SLM 316L, (e) AES mapping of Ni, Mo, and Cr in the cell structure.

Figure 4. SEM images of the x-y plane at (a) low and (b) high magnification, SEM images of the y-z plane at (c) low and (d) high magnification. (b) and (d) corresponding to the yellow areas shown in (a) and (c), respectively.
Figure 4 shows substructures of the x-y plane and the y-z plane and the morphological transition between the substructures. The fine cell and columnar structures have changed to various extents to the morphology when they pass through the fusion line and the high angle grain boundary. The substructures continuously and smoothly cross the fusion line, while slight local changes are required to adjust the structural differences when crossing the high angle grain boundaries. As shown in figure 4(b), the distortion phenomenon occurs in the transition area from cellular structure to columnar, which indicates that the formation of the substructure may have a specific relationship with melt capillary flow.

3.2. Grain growth
The forming process of SLM involves the powders and part of the matrix being melted under the action of laser to form micro-melt pools, which are stacked one by one and finally solidified into a block. SEM micrographs and EBSD map parallel to the build direction are shown in figure 5. The columnar crystals continued to grow across the fusion line, and their growth direction changed after passing the fusion line, as shown in figure 5(b). Grains can even be epitaxial over ten layers, as shown by the black arrow in figure 5(c). The fusion line referred to here is the final form of the solid/liquid interface. From the perspective of metal physics, the solid/liquid two phases are in thermodynamic equilibrium at the interface. This liquid phase atomic free energy is a very thin layer equivalent to the solid phase atomic free energy. There must be a liquid phase boundary layer and a solid phase boundary layer on both sides of it. Because the arrangement of solid metal atoms is long-range ordered, the liquid metal atoms are long-range disordered. Several layers of atoms on the solid/liquid two-phase interface appear irregularly, forming fusion lines that are essentially not equivalent to the melting or the semi-melting areas. Under erosion conditions, many local galvanic cells are formed and eroded into grooves. In addition, the EBSD image showed that the directionally grown columnar grains were parallel to each other and parallel to the build direction. The maximum width of columnar grains did not exceed the laser spot diameter ($10^5 \mu m$), indicating that there was no lateral growth but only epitaxial growth along the build direction. Rotating 67° layer by layer made the shape of grains extremely irregular during the solidification process, and some relatively small grains with random orientation were easy to form. It could be ascribed to the angle difference between two adjacent layers that broke the original columnar grain growth conditions.

3.3. The analysis of orientation
Usually, the diffraction intensity of the phase to be measured is directly proportional to the content. On the diffraction pattern, the difference concerning intensity of different diffraction peaks not only correlates with the diffraction ability of the corresponding crystal planes, but also relates to the number of crystal grains paralleling to the surface of the sample sheet, which is named after texture [20]. The peak intensity has a definite relationship with the number of the exponential surface in some direction. Although it is not the only decisive factor, it is the main factor.
Because additive manufacturing is a layer-by-layer manufacturing method, the microstructure growth of SLM 316L is different from traditional processing methods. X-ray diffraction (XRD) was used to analyze the bulk sample phase formed by selective laser melting. Phase detection was performed on the cross-sections in the building direction (BD), scanning direction (SD), and transverse direction (TD), respectively, and the results are shown in figure 6. The microstructure formed after selective laser melting is still austenite but has a certain relationship between the grain orientation on different cross-sections. The results of the BD direction are apparent differences from the other two directions. The scanning strategy adopted in this experiment is to rotate 67° between layers, therefore no difference between SD and TD, both of which are marked as TD directions in the figure. It can also find from the experimental results that the x-z plane and y-z plane have similar XRD patterns. The x-y plane shows an opposite crystal plane intensity of the other two sides. Therefore, the specimens formed by SLM have an apparent orientation relationship. That is, there is a texture relationship.

In order to further explain the grain growth orientation relationship, EBSD analysis was performed on the x-y plane and y-z plane in figure 7 to obtain grains orientation information. As shown in figure 7, EBSD maps received the same test results as XRD. The grains are preferably distributed along with the (110) orientation on the x-y plane. The y-z plane comprises major (111) oriented grains and minor (001) oriented grains. In addition, except for the difference in grain orientation and various morphologies. The grains on the x-y plane are banded and arranged in order as the result of laser reciprocating scanning. Since the laser spot diameter determines the width of the melt pool, the grains are primarily polygonal or nearly circular, and most of the size does not exceed 110 μm. Noteworthily, the epitaxial growth of grains generally grows along the build direction. Fined grains will mostly appear at the overlap between the bands. On the y-z plane formed the elongated columnar grains.

Figure 6. XRD pattern of SLM 316L cross-section in different directions.

Figure 7. Grain orientation relationship of different cross-sections.
4. Discussion

4.1. Mechanism of substructure formation

The temperature gradient at the front of the solid-liquid interface and the pushing speed of the solid-liquid interface are two crucial parameters that affect the solidification behavior during solidification. The temperature gradient $G$ and the solidification rate $R$ are important factors affecting the structure and size of the metal after solidification. $G/R$ determines the morphology after solidification, while $G \times R$ confirms the size of the solidified microstructure, which is also described in literature [21]. A $G/R$ value corresponds to a size of invariant microstructure, and correspondingly, $G \times R$ corresponds to a constant microstructure, such as planar, cellular, dendritic, or equiaxed grain. When $G \times R$ remains constant, as $G/R$ decreases, the solidification interface changes from planar growth to cellular structure, then dendrite, and finally to equiaxed grain. The refine microstructure with the increase of cooling rate ($G \times R$) at the same solidified morphology ($G/R$).

$G$ is defined as the normal temperature gradient of the solid-liquid interface in the solidification front, and $R$ is defined as the normal advancing speed of the solid-liquid interface in the solidification front,

\[ G = \nabla T \cdot n^* \]
\[ R = v \cdot i \cdot n^* \]

Where $n^*$ is the normal vector, $v$ is the scanning speed, $i$ is the unit vector in the scanning direction.

According to the simulation calculation results acquired the distribution maps of $G$ and $R$ in the melt pool and further calculated $G \times R$ and $G/R$ to get the corresponding distribution map. The calculation results are shown in figure 8. It can be found that the distribution of temperature gradient and solidification rate inside the melt pool is in connection with the flow field and temperature field. The strong outward Marangoni convection resulted in a shallow and wide melt pool morphology.

It can be seen from figure 8(a) that the maximum temperature gradient at the edge of the melt pool can reach $3.1 \times 10^7$ K m$^{-1}$, while the temperature gradient at the top of the melt pool is only one-third of the top about $8.26 \times 10^8$ K/m. The distribution diagram of the solidification rate in figure 8(b) shows that the maximum solidification rate at the top of the melt pool is 0.98 m s$^{-1}$, about 50 times that at the edge. Therefore, the variation of solidification characteristic parameters is mainly controlled by the solidification rate. Figure 8(c) shows that the area with the lowest cooling rate is located at the edge of the melt pool and the largest near the heating surface. In addition, the laser spot center is not the place with the largest cooling rate but at the edge. The melt flows from the center of the melt pool to the edge, and the melt with the low temperature at the edge flows to the center to form a shallow and wide melt pool morphology. The edge near the laser heating center produces the greatest cooling rate due to incorporating low-temperature melt. Besides, although the melt pool boundary has the lowest temperature and the largest temperature gradient, it is the area with the lowest cooling rate and the slowest flow rate within the melt pool range. This results in the weakest heat and mass transfer at the melt pool edge, and the cooling rate correlates positively with the solidification rate. The basically identical morphology parameters in the entire melt pool explain the formation of fine cellular or columnar substructures.

The Columnar to equiaxed transition (CET) exists in the process of substructure formation. In general, when the undercooling at the front of the solidification interface is greater than the undercooling for nucleation,
equiaxed grain-growth may occur due to nucleation. Columnar grain growth will wrap equiaxed grain with a small volume fraction. It will block the process of columnar grain growth if the volume fraction of equiaxed grain is greater to a certain extent. According to Hunt [22], complete equiaxed grain growth requires an equiaxed crystal with a volume fraction greater than 0.49, and on the contrary, it is a columnar crystal. Gaumann et al [23] proposed an improved model based on the Hunt model, combined with other models that took into account the non-equilibrium effect of a high solidification rate. However, the Gaumann model was initially used in binary alloy systems, and it can be used to determine the CET transformation conditions of multi-component alloys after some improvements.

The dendrite tip radius is shown in Formula (3) at a specific growth rate according to the interface stability criterion. For columnar dendrite growth, the temperature gradient \( G_T = G_{L} / 2 \). \( G_{L} \) is the temperature gradient in the liquid phase,

\[
G_{C_i} = \frac{(1 - k_{L_i})\nu C_i^{q_i}}{D_{L_i}}
\]

where \( k_{L_i} \) is the equilibrium distribution coefficient of component \( i \) in the liquid phase, \( D_{L_i} \) is the diffusion coefficient of component \( i \) in the liquid phase, \( C_i^{q_i} \) is the concentration of a component in the liquid phase at the tip of the dendrite, \( \nu \) is the dendrite growth rate.

Each parameter in the above formula has the following relationship,

\[
m_{L_i} = m_i \left( 1 + \frac{k_i}{k_{L_i}} \ln \left( \frac{k_{L_i}}{k_i} \right) \right)
\]

where \( k_i \) is the balanced distribution coefficient, \( m_i \) is the liquidus slope.

\[
\xi_c(Pe_i) = 1 - \frac{2k_{L_i}}{[1 + (1/\sigma^*Pe_i)^{1/2} - 1 + 2k_{L_i}]}
\]

where \( \sigma^* \) is the surface tension.

\[
Pe = \frac{\nu r}{\alpha}
\]

where \( \alpha \) is the thermal diffusivity.

\[
k_{L_i} = \frac{k_i + av/D_i}{1 + av/D_i}
\]

where \( a \) is the interatomic distance, \( D_i \) is the atom exchange coefficient between the solid and liquid phases.

\[
C_i^{L*} = \frac{C_{oi} - I(Pe_i)}{1 - (1 - k_{L_i})I(Pe_i)}
\]

where \( I(Pe_i) \) is the Ivantson function.

\[
I(P) = Pe_i \exp(Pe_i) E_i(Pe_i)
\]

where \( E_i \) is the exponential integral function.

\[
E_i(Pe_i) = \int_{Pe_i}^{\infty} \frac{\exp(-z)}{z} dz = -E_i(-Pe_i)
\]

According to the Gibbs-Thomson temperature equation, the dendrite tip temperature \( T_i \) is

\[
T_i = T_m + \sum_{i=1}^{n-1} m_{L_i} C_i^{L*} - \frac{2T_i}{R} - \frac{\nu}{\mu_k}
\]

where \( T_m \) is the melting point of the pure component, \( \mu_k \) is the linear kinetic coefficient.

Grain growth during solidification requires dynamic undercooling (\( \Delta T_{k} \)). When calculating the dendrite tip undercooling, it is necessary to take into account temperature undercooling, composition undercooling, tip curvature, and other factors at the dendrite tip during the dendrite growth. The undercooling caused by the interface curvature is called curvature undercooling (\( \Delta T_{r} \)). Thermal undercooling (\( \Delta T_{T} \)) is caused by the temperature gradient at the front of the interface. Solute undercooling (\( \Delta T_{C} \)) is caused by the concentration of solute in the front of the interface. Therefore, the total undercooling at the solid/liquid interface is the sum of these undercoolings.
\[ \Delta T = \Delta T_K + \Delta T_T + \Delta T_r + \Delta T_C \]  

(13)

The very small solidification kinetic undercooling on account of the relatively low melting entropy of metals or alloys can be ignored. Therefore, the undercooling caused by curvature \( \kappa \) can be calculated using:

\[ \Delta T_T = 2\kappa \Gamma^* T_m \]  

(14)

where \( \Gamma^* = \gamma / \Delta h_m \), \( \gamma \) is the solid/liquid interface energy, \( \Delta h_m \) is the melting enthalpy per unit volume.

Solute undercooling can be calculated using:

\[ \Delta T_C = -m_1 C_0 \Omega(1 - k_0) \]  

(15)

where \( \Omega \) is the solute supersaturation.

Thermal undercooling can be calculated using:

\[ \Delta T_T = -\Omega_T \frac{\Delta h_m}{c_p} \]  

(16)

where \( \Omega_T \) is the thermal supersaturation, \( c_p \) is the liquid phase hot melt.

Therefore, the equiaxed grain growth rate can be determined according to the total undercooling and temperature gradient at the dendrite tip and the Ivantsov function. The maximum radius of the equiaxed grain can be obtained using:

\[ r_e = \int_0^{Z_n} \frac{v_e[z]}{v} \]  

(17)

where \( v_e \) is the equiaxed crystal growth rate, \( Z_n \) is the distance between the position where the liquid phase undercooling at the front of the interface is equal to the nucleation undercooling and the solid-liquid interface.

The Avarami equation can be used to determine the actual volume fraction of equiaxed crystals assuming that equiaxed crystals nucleate randomly.

\[ \phi = 1 - \exp[-\phi_e] \]  

(18)

Where \( \phi \) is the actual volume fraction, \( \phi_e \) is the expanded volume fraction.

It is assumed that the equiaxed grain grows spherical and the total nucleation point will quickly reach a consistent nucleation number, in consequence, the expanded volume fraction can also be expressed using:

\[ \phi_e = \frac{4\pi r_e^3 N_0}{3} \]  

(19)

where \( N_0 \) is the total number of heterogeneous nucleation per unit volume.

### 4.2. Epitaxial growth of grain

The solidification of molten metal still includes two stages of nucleation and grain growth despite the fast solidification rate of the micro-melt pool. Since the metal powder and the matrix have the same composition and the excellent wettability between the two leads to a wetting angle of nearly 0°, and there is almost no potential barrier for nucleation. If the melt pool temperature is slightly below the equilibrium melting point temperature, it will nucleate on the base metal. No new grain nuclei are produced due to the extremely fast cooling rate and the large undercooling, and there is almost no independent nucleation. The continuously growing grains and the base metal grains on which they grow have the same crystal structure and crystallographic orientation so that the grains can grow epitaxially at the metal liquid-solid interface. As shown in figure 5(d), no obvious fusion line was observed in the EBSD results, indicating that the crystallographic orientation on both sides of the fusion line did not vary significantly.

The large temperature gradient in the micro-melt pool made the columnar grains fully grow. There should be a layer of planar crystals growing on the first layer in the scan and the substrate, then transforming into columnar crystal growth, similar to the solidification of ingots. In the initial stage of solidification, some grains grow faster and coarser and will prevent other grains when it was more conducive to growth, resulting in columnar crystals with a specific orientation. As shown in figure 5(c), the grain in the red area was blocked and stop growing due to the neighboring grains that had more growth advantages. The growth direction of the columnar crystals is not always perpendicular to the isotherm but has a certain angle with the isotherm according to the schematic diagram given in figure 9. The appearance of such an included angle provides the crystal with growth a specific orientation and matches the crystal growth speed with the scanning speed. In addition, there are also many complex forces in the melt pool, and the presence will affect the crystal growth.

The crystal growth speed after the metal powder melts must match the scanning speed to ensure the continuity of the SLM process. As shown in figure 9(a), the local growth speed of crystal \( v_w \) and the heat source moving speed \( v \) must meet equation (20) if the crystal growth is isotropic.

\[ v_w = \]
\[ w = v \cos \theta \]

Where \( \theta \) is the angle between the local growth direction and the scanning direction.

Figure 9 shows an anisotropic crystal growth direction, which is not perpendicular to the isotherm but with an angle of \( \theta' \). At this time, the crystal growth speed \( v_w \) and the heat source moving speed \( v \) must meet equation (21),

\[ v_w = \frac{v_n}{\cos(\theta' - \theta)} = \frac{v \cos \theta}{\cos(\theta' - \theta)} \]  

Where \( v_n \) is the growth speed perpendicular to the isotherm.

The cooling rate in the micro-melt pool is rapid, and the temperature gradients inside the melt pool are also different. Because the heat source keeps moving forward leads to the direction of the temperature gradient also keeps changing. Figure 10 shows a schematic diagram of columnar crystal growth parallel to the temperature gradient and the principal axis deviates from the temperature gradient. For isotropic crystals, the growth direction of columnar crystals usually changes abruptly in order to keep the maximum growth rate direction consistent with the temperature gradient direction and to maintain their inherent growth direction. The black line shows the A-B-C-D-E partial grain boundary of columnar grain in figure 5(b). When the columnar grain crosses the fusion line at points B and D, the growth direction of the columnar crystal changes and is perpendicular to the fusion line, similar to figure 10(a). Sudden changes in growth direction at points C and E in order to meet the requirements of preferential growth and temperature gradient direction, as shown in figure 10(b).

5. Conclusions

(1) The microstructure of 316L stainless steel formed by selective laser melting has an apparent orientation relationship on different sections. The adopted cross-hatching scanning strategy makes the growth relationship parallel to the build direction basically the same, consisting of primary \((111)\) oriented grains and secondary \((001)\) oriented grains. On the contrary, perpendicular to the build direction showing the strong \((110)\) orientation.
(2) Using the solidification theory to explain the formation of cellular substructures. Due to the higher temperature gradient and solidification rate in the solidification front of the melt pool during the selective laser melting process, the cellular or columnar substructures are formed. The fine cellular and columnar substructures in SLM 316L are all below 1 μm in size. AES results have been presented to show that the substructure wall constituted by the enrichment of elements is not simply a dislocation wall formed by dislocations, morphologically similar only.

(3) The grain growth has obvious directionality and grows layer by layer along the direction of heat flow. The epitaxial growth mechanism results in columnar grains with the same orientation and passing through multiple layers of fusion lines. When the temperature gradient direction deviates from the principal axis, the grain growth direction suddenly changes as it crosses the fusion line but remains perpendicular to the fusion line.

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Data availability statement

The data that support the findings of this study are available upon reasonable request from the authors.

Declaration of competing interest

The authors confirm that there is no conflict of interest regarding the work presented in this paper.

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