Macro-micro coupled simulation for the morphological evolution of the solidification structures in the entire weld

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Abstract: The heat and mass transfer and the morphological evolution of the various weld solidification structures for the entire weld were simulated by the macro-micro coupled model which combined the welding process and the solidification structures in the molten pool. The weld profile were calculated for different welding conditions, and the effects of the welding parameters on the solidification conditions were analyzed along the trailing pool boundary. The formation mechanism of the axial structure and the equiaxed grains in the weld were studied by the developed model. The results indicate that the temperature gradient G , solidification rate R_w , and the ratio G/R_w along the trailing pool boundary are related with the location in the weld and the welding parameters. When the weld solidification structure is only consisted with the epitaxial columnar grains from the fusion boundary, the axial structure forms in the central region of the weld at low welding speeds owing to the anisotropic kinetics of the columnar grains. When the heterogeneous nucleation occurs in the molten pool, the weld structure is directly determined by the competitive growth between the columnar and new nucleated grains. At low welding speeds, the preferential orientations of the survival grains get closer to the welding direction with the decrease of the distance to the weld center.

With the increase of the welding speed, more grains form in the molten pool, and the new nucleated grains also become more competitive. The columnar grains from the fusion line are blocked by the equiaxed grains. The continuous equiaxed grain zone forms in the weld, and the columnar to equiaxed transition (CET) occurs.

Key words: weld pool, solidification conditions, weld grain structure, cellular automata

1. Introduction

The fusion welding is one of the most important materials processing technique and applied in many fields of manufacturing industry. The solidification behavior of the molten pool controls the grain structure and the ultimate mechanical properties of the weld [1]. The morphological evolution of the grain structure in the weld is directly determined by the competitive growth of grains with different preferential orientations under the complicated solidification conditions in the entire molten pool. The heat and mass transfer in the molten pool exerts an important influence on the grain structure evolution through changing the transient thermal conditions during the solidification [2]. Therefore, the investigation on the solidification behavior of the molten pool which considers the complicated heat and mass transfer process during welding can provide useful information for the understanding to the formation mechanism of the various weld grain structures. However, the welding process is quite complicated. It is difficult to observe the dynamic solidification process of the molten pool directly through the conventional method.

With the development of the computational power and the numerical model, the morphological evolution of the solidification structure could be simulated. Some investigations have been performed to simulate the morphologies of the solidification

structures in the weld pool. Yin [3] adopted a two-dimensional CA model to simulate the dendrite growth in the molten pool of the laser-engineered net shaping process. Zhan [4, 5] and Pavlyk [6] simulated the dendrite morphology during the gas tungsten arc welding (GTAW) by the CA method. Tan [7-9] developed a two-dimensional CA-PF model to simulate the dendrite growth, and a further study on the growth of grains and sub-grain dendrites was also performed during the laser keyhole welding. Chen[10] proposed a coupled CA-FE model to simulate the grain structure during GTAW. Farzadi[11] and Fallah [12] calculated the dendrite growth in the molten pool using the PF method. Considering the transient solidification conditions during welding, Zheng [13, 14] developed a PF model to simulate the microstructure evolution near the fusion line. A review of the exiting studies show that most of the investigations focused on the morphological evolution of the dendrite for different solidification conditions in the local region of the molten pool. A few studies also simulated the meso-scale grain structure of the weld. However, the systematic investigation on the morphological evolution of the various solidification structures within the entire weld has not been reported, and the relationship between the welding parameters and the actual weld solidification structures has not been established. As mentioned above, the weld grain structure evolution is mainly controlled by the competitive growth of grains with different preferential orientations within the entire weld. To clearly reveal the formation mechanism of the various weld solidification structures, it is necessary to accurately simulate the heat and mass transfer behavior and the competitive growth of grains with different preferential orientations in the molten pool.

In the present paper, the three dimensional weld pool model and the cellular automata

(CA) model were developed to simulate the heat and mass transfer behavior and the solidification structure evolution in the weld pool. The transient thermal conditions used in the CA model are obtained from the three dimensional weld pool model through the data interpolation method. The weld profile for different welding conditions are calculated. The temperature gradient G , solidification rate R_w , and the ratio G/R_w along the trailing pool boundary are extracted to analyze the effect of the welding parameters on the transient solidification conditions. The morphological evolution of the weld solidification structures for the entire weld are simulated under different welding conditions, and the formation mechanism of the various weld solidification structures are discussed.

2. Numerical model description

2.1 The three dimensional weld pool model

The heat and mass transfer in the molten pool dominate the weld profile and the solidification conditions along the trailing pool boundary which can affect the morphological evolution of the weld solidification structures directly. To accurately simulate the weld solidification structure, it is necessary to develop the three dimensional model for the heat and mass transfer during welding. In the weld pool model, several basic assumptions have been made to simplify the calculation, including the heat and mass transfer behavior is symmetric in the longitude plane along the welding direction, the fluid flow in the weld pool is laminar and incompressible, and the deformation of the free surface during welding is ignored. The governing equations have been solved to obtain the velocity and temperature fields during welding. The conservation equations in three-dimensional Cartesian coordinate system are detailed described in our previously published work[15].

In general, the temperature conditions during the solidification change within the three dimensional space. As a result, the morphological evolution of the weld solidification structures also occurs in the three dimensional molten pool. In order to simplify the solidification conditions during welding and easily reveal the formation mechanism of the various solidification structures, a thin plate-shaped workpiece and the full penetration welding process are applied in present work. Under such simplified conditions, the solidification conditions and the weld structures will not change obviously along the thickness direction and can be approximately considered as two dimensional problems on the upper surface of the workpiece. This method is also applied in many previously experimental works for the weld solidification structures [16-19]. The workpiece used in the calculation is an aluminum alloy plate with the length of 100 mm, width of 50 mm, and thickness of 1.5 mm. The thermal properties of the workpiece used in present simulation are shown in Tab.1 [20, 21].

2.2 The CA model for weld solidification structure

The temperature and the solute concentration are the necessary conditions for the dendrite growth. In the present CA model, the solute concentration within the entire calculation domain is obtained by solving the equation as follows[22]:

$$\frac{\partial C_i}{\partial t} = D_i \cdot \nabla^2 C_i + C_i \cdot (1 - k_0) \frac{\partial f_s}{\partial t} \quad (1)$$

where C is the composition, D is the solute diffusivity, f_s is the solid fraction, k_0 is the equilibrium partition coefficient, and the subscript i indicates liquid or solid. Moreover, the transient temperature field used in the CA model is obtained from the three dimensional weld

pool model. Owing to the mesh size for the weld pool model and the CA model are quite different. To obtain the thermal conditions of each CA mesh for simulating the dendrite growth, the spatial bilinear interpolation method is applied in present work, as shown in Fig.1[23]. Using the values of points A, B, C and D as shown in Fig.1 (a), the value of each CA mesh within this macro mesh can be calculated by the spatial bilinear interpolation method, as shown in Fig.1 (b). Assuming that the values of points A (x_1, y_1), B(x_2, y_2), C(x_1, y_2) and D(x_2, y_1) are $f(a)$, $f(b)$, $f(c)$ and $f(d)$, respectively. The value of an arbitrary point O(x, y) can be obtained through two steps. Firstly, the interpolation along x direction should be operated to calculate the values of points $P_1(x, y_1)$ and $P_2(x, y_2)$ by Eq.(2) and Eq.(3). And then the value of point O(x, y) can be calculated from the values of points $P_1(x, y_1)$ and $P_2(x, y_2)$ by Eq.(4).

$$f(P_1) = \frac{x_2 - x}{x_2 - x_1} f(c) + \frac{x - x_1}{x_2 - x_1} f(d) \quad (2)$$

$$f(P_2) = \frac{x_2 - x}{x_2 - x_1} f(a) + \frac{x - x_1}{x_2 - x_1} f(b) \quad (3)$$

$$f(O) = \frac{y_2 - y}{y_2 - y_1} f(P_1) + \frac{y - y_1}{y_2 - y_1} f(P_2) \quad (4)$$

Besides the temperature and the solute concentration, the growth kinetics of the dendrite is also calculated by the CA model. The local curvature of the interface cells can be calculated by the equation as follows [20]:

$$K = \left[\left(\frac{\partial f_s}{\partial x} \right)^2 + \left(\frac{\partial f_s}{\partial y} \right)^2 \right]^{-3/2} \times \left[2 \frac{\partial f_s}{\partial x} \frac{\partial f_s}{\partial y} \frac{\partial^2 f_s}{\partial x \partial y} - \left(\frac{\partial f_s}{\partial x} \right)^2 \frac{\partial^2 f_s}{\partial y^2} - \left(\frac{\partial f_s}{\partial y} \right)^2 \frac{\partial^2 f_s}{\partial x^2} \right] \quad (5)$$

Considering the effects of the thermal conditions and the local curvature at the S/L interface, the interface equilibrium composition C_l^* can be calculated by [20]

$$C_l^* = C_0 + \frac{T^* - T^{eq}}{m_l} + \frac{\Gamma K f(\varphi, \theta)}{m_l} \quad (6)$$

where C_0 is the initial solute concentration, T^* is the interface temperature obtained by the interpolation, T^{eq} is the equilibrium liquidus at the initial solute concentration, m_l is the liquidus slope, Γ is the Gibbs-Thomson coefficient, K is the interface curvature calculated by Eq. (5), and $f(\varphi, \theta)$ is a function accounting for the anisotropy. The increase of the solid fraction in the interface cells within each time step can be calculated by[22]

$$\Delta f_s = (C_l^* - C_l) / (C_l^* \cdot (1 - k)) \quad (7)$$

To simulate the formation of sidebranches, a random fluctuation has been added to the solid fraction increment for the interface cells during each time step, the random fluctuation F_{fs} can be calculated by[24]

$$F_{fs} = A_{fs} \cdot R_{fs} \quad (8)$$

where A_{fs} is the fluctuation coefficient, and R_{fs} is the random number in the range of $[-1, 1]$. In the calculation, the A_{fs} is set to 0.01. The amplitude of the fluctuation can be controlled through the fluctuation coefficient A_{fs} . The accurate simulation of the dendrite with random preferential orientations is the basic requirement for the following investigation of the competitive grain growth. The capturing algorithm [7, 25] applied in the present paper is shown in Fig. 2. At initial time, a square-shaped solid seed is formed at the center of a cell. With the increase of the solid fraction in the central cell, the seed grows along its diagonals. The relationship between the solid fraction of the cell and the diagonal length of the seed can be obtained by

$$L = \sqrt{2} \cdot f_s \cdot \Delta x \quad (9)$$

where L is the length of the diagonal, and Δx is the cell size. When the corners of the square seed reach any liquid cell of the eight neighbors, the neighboring liquid cell will be changed to an interface cell. Then a new square seed with the same preferential orientation as the original seed is formed in the new interface cell. The center of the new square seed is placed at the corner of the original square seed. The new square seed starts to grow according to the increase of the solid fraction in the new interface cell.

Owing to the solidification rate in the molten pool is higher than that in the casting process. As a result, the partition coefficient and the liquidus slope will deviate from its balance value. In the present model, the non-equilibrium partition coefficient k_p for various solidification rate can be calculated by[26]:

$$k_p = \frac{k_0 + (\alpha_0 V / D_l)}{1 + (\alpha_0 V / D_l)} \quad (10)$$

where k_0 is the equilibrium partition coefficient, α_0 is a constant. In addition, the apparent liquidus slope m_p can be obtained by[21]:

$$m_p = m_l (1 + f(k)) \quad (11)$$

where the m_l is the equilibrium liquidus slope, and the $f(k)$ can be calculated by[21]:

$$f(k) = \frac{k_0 - k_p \left[1 - \ln(k_p / k_0) \right]}{(1 - k_0)} \quad (12)$$

3. Results and discussions

3.1 The weld pool shape and the solidification conditions

The heat and mass transfer in the molten pool determines the weld profile, which is related with the solidification conditions along the trailing pool boundary and affecting the weld solidification structures. The investigation for the effect of the welding parameters on the weld profile and the transient solidification conditions in the molten pool is useful to understand the formation mechanism of the various weld solidification structures. During the arc welding, the weld pool size is determined by the welding line energy and the welding speed. To better study the morphological evolution of the weld solidification structures, the weld width should keep constant under different welding conditions. According to the previous experimental works [17, 18], the higher welding currents are applied with the increasing welding speed. In the present paper, the welding speeds used in the calculation are 2, 4, and 10 mm/s, respectively. In addition, the welding current also correspondingly changes from the 70 to 150 A. The simulated results of the weld pool shape for various welding conditions are shown in Fig.3. It can be observed that with the increase of the welding speed, the weld pool was elongated along the welding direction, and the weld pool shape changes gradually from the elliptical to the tear-drop shaped. The solidification conditions along the trailing pool boundary is related with the weld pool shape. To analyze the effect of the welding conditions on the solidification conditions which can directly dominate the resulting weld solidification structure, the temperature gradient G , the solidification rate R_w , and the ratio G/R_w along the trailing pool boundary are extracted from the welding temperature field.

Fig.4 shows the temperature gradients for various welding conditions, the x -axis indicates the distance to the weld center. It can be seen that the temperature gradient increases gradually along the trailing pool boundary with the decrease of the distance to the weld center.

During the welding, the moving weld pool can affect the heat diffusion process within the workpiece. With the increase of the welding speed, the temperature gradient along the trailing pool boundary decreases near the weld center and increases near the fusion line. It is because that with the higher welding speed, the heat transfer process behind the weld pool will be elongated in the welding direction and compressed in the weld piece width direction, and the effects of these two factors will reach the balance at the location shown in the Fig.4.

Besides the temperature gradient, the normal solidification rate also exerts important influence on the solidification behavior. In the molten pool, the solidification rate at the trailing pool boundary changes gradually from the fusion line to the weld center. The solidification rate R_w can be calculated by[27]:

$$R_w = V_{welding} \cdot \cos(\alpha) \quad (13)$$

Where $V_{welding}$ is the welding speed, α is the angel between the welding direction and the normal to trailing pool boundary. The relationship between the welding speed and the solidification rate is shown in Fig.5. The angle α can be obtained from the simulated weld pool shape for different welding speeds.

Fig.6 illustrates the angle α along the trailing pool boundary, the x -axis is the distance to the weld center. It can be observed that the angle α along the trailing pool boundary increases from fusion line to weld center. At low welling speeds, the trailing pool boundary is a continuously curved line, and the increasing rate of the angle α is near constant. At high welding speeds, the angle α does not change obviously in the middle region of the trailing pool boundary. It is because that with the increase of the welding speed, the elliptical weld pool will change to tear-drop shaped, and the trailing pool boundary also become straight, as

shown in Fig.3. According to the Eq.13, the normal solidification rate along the trailing pool boundary can be calculated and shown in the Fig.7.

During the solidification of the molten pool, the ratio G/R_w along the trailing pool boundary has an important influence on the weld solidification structure morphologies [28-31]. With the decrease of the ratio G/R_w , the constitutional undercooling in front of the trailing pool boundary will increase, and the more equiaxed grains can form in the molten pool and compete with the epitaxial columnar grains [21]. The ratio G/R_w along the trailing pool boundary from the fusion line to the weld center for various welding speeds are shown in Fig.8. It can be seen that the ratio G/R_w will become smaller with the decrease of the distance to the weld center. In addition, the ratio G/R_w also decreases with the increase of the welding speed. The results indicate that with the higher welding speed, the new nucleated equiaxed grains in front of the trailing pool boundary will increase and become more competitive during the solidification of the molten pool.

3.2 The morphological evolution of the solidification structures in the entire weld

The accuracy of the developed CA model is important for the simulation results of the morphological evolution of the various weld solidification structures. To verify the present CA model the grow kinetics of the solidification structures were calculated and compared with the analytical model and other numerical models. Fig.9 shows the variation of the tip velocity with the solidification time for different undercoolings. It can be seen that in the initial unstable growth stage, the tip velocity is larger than the steady value owing to the unsatisfied solute balance condition at the S/L interface. At this stage, the solute rejected by the growing crystal is more than that can be transferred away from the S/L interface by solute

diffusion[22]. With the development of solidification, the solute concentration in front of the S/L interface increases gradually due to the solute accumulation. The tip velocity of the dendrite decreases with the increasing solidification time. When the dendrite growth comes to the steady-state stage, the solute rejected by the growing crystal is equal to that which can be transferred away by solute diffusion. Through the comparison between the results for different undercoolings of 3 K and 5 K, the solidification time for reaching the steady-state stage is shorter with the higher undercooling. Fig.10 shows the results of the steady tip velocities and equilibrium concentrations for various undercoolings. It can be seen that the simulated results by the present CA model are consistent with the results from the LGK analytical model and the other recent numerical models. Beside the growth kinetics, the actual growth directions of the simulated solidification structures were also verified in our previous work[23].

The weld solidification structure morphology is directly determined by the competitive growth of grains with different preferential orientations in the molten pool. The solidification conditions dominate the competition behavior of grains and the resulting different weld solidification structures. Based on the discussions in section 3.1, the morphological evolution of the weld solidification structures for different welding parameters within the entire weld have been simulated in this section. When there is no heterogeneous nucleation phenomenon in the weld pool, the morphology of the weld solidification structure is completely determined by the competitive growth between the epitaxial columnar grains. Fig.11 shows the morphological evolution of the weld grain structure without the heterogeneous nucleation. The welding speed used in the calculation is 2 mm/s, and the transient solidification conditions are obtained from the simulated results of the welding temperature field through

the interpolation method. It can be observed that with the movement of the molten pool, the columnar grains begin to grow from the fusion line to the weld center. At the micro-scale, the columnar grains are consisted with the dendritic arrays with the same preferential orientation. Owing to the large curvature of the dendritic structure, the anisotropic interface energy can obviously affect the growth kinetics of the columnar grains. As a result, the growth direction of the columnar grain will not affected by the change of the maximum temperature gradient direction during solidification. With the decrease of the distance to the weld center, the angle between the normal to trailing pool boundary and the welding direction decreases gradually owing to the continuously curved trailing pool boundary. The angel between the growth directions of the columnar grains and the maximum temperature gradient direction increases with the development of the solidification. In this case, the columnar grains from the fusion line become less competitive near the weld center. Therefore, the axial grains will block the columnar grains through the competitive growth, and then the continuous axial structure forms in the center of the weld. Fig.12 shows the experimental result of the axial structure in the actual aluminum weld [16]. The width of the axial structure zone is determined by the result of the competitive growth between the columnar grains from fusion line and the axial grains near the weld center. With the increase of the welding speed, the weld pool changes to tear-drop shaped, and the angle between the welding speed and the normal to the trailing pool boundary also increases gradually. As a result, the axial grains become less competitive and eventually disappear in the weld. The micro-mechanism for the formation of the axial structure and its morphological evolution with various welding speeds have been detailed described in our previously work[24].

For some alloys, there exists the heterogeneous particles by artificial adding of in-situ generation in the molten pool during welding. These particles can provide the available surface for the nucleation in the constitutional undercooled zone near the trailing pool boundary. As a result, the competitive growth between the epitaxial columnar grains and the new nucleated grains in the weld pool occurs, and the weld solidification structure is consisted with both the columnar and new nucleated grains. With the change of the welding conditions, the morphologies of the weld solidification structure will be more complicated. Fig.13 illustrates the experimental results of the weld solidification structures for different welding conditions. To clearly reveal the formation mechanism of the various weld solidification structures, the competitive growth between the columnar and new nucleated grains under different welding conditions have been simulated. In the calculation, a continuous Gaussian nucleation distribution is adopted to relate the nuclear density increase to the increased effective undercooling. The total number of the nuclei for a given ΔT can be obtained by [32-34]:

$$n(\Delta T) = \frac{N_{\max}}{\sqrt{2\Delta T_{\sigma}}} \int_0^{\Delta T} \exp\left[-\frac{(\Delta T - \Delta T_N)^2}{2(\Delta T_{\sigma})^2}\right] d(\Delta T) \quad (14)$$

where ΔT_N is the mean nucleation undercooling, ΔT_{σ} is the standard deviation of the distribution and $N_{\max}$ is the maximum nucleation density. The above nucleation parameters control the number and the location of the equiaxed grains in the weld during solidification. In the present simulation, the $N_{\max}$ is $3.0 \times 10^7 \text{ m}^{-2}$, the ΔT_N is 6 K, and the ΔT_{σ} is 0.5 K[34].

Fig.14 shows the morphological evolution of the weld solidification structure with the welding speed of 2 mm/s. It can be observed that the columnar grains grow toward the weld

center with the development of the solidification. Owing to the higher temperature gradient and the lower solidification rate near the fusion line, it is difficult for the new grains to form in the molten pool. In this region, the solidification structure is mainly consisted with the columnar grains from the fusion line. With the decrease of the distance to the weld center, the solidification rate increases and the temperature gradient decreases gradually along the trailing pool boundary. In this case, more nuclei can grow up in front of the trailing pool boundary. Moreover, the preferential orientations of these nuclei are selected randomly. The competitive growth between the columnar and new nucleated grains with different preferential orientations will occur during the solidification. The solidification conditions play an important role in this competition. Owing to the trailing pool boundary curves gradually from the fusion line to the weld center at low welding speeds, the preferential orientation of the survival grains near the weld center will be closer to the welding direction. On the other hand, the new nucleated grains which can grow up and block the columnar grains under this welding conditions are not enough. Therefore, the survival new nucleated grains will elongate along its solidification direction in the weld, and the weld grain structures will curve toward the welding direction at the macro-scale.

With the increase of the welding speed, the weld pool shape and the solidification conditions along the trailing pool boundary will change correspondingly. Fig.15 illustrates the morphological evolution of the weld solidification structures with the welding speed of 4 mm/s. It can be found that the number of nuclei increases obviously owing to the higher solidification rate and the smaller temperature gradient at the trailing pool boundary. Moreover, since the undercooled zone width becomes larger, the new grains will form in front

of the trailing pool boundary and become more competitive compared with the columnar grains. Therefore, the columnar grains from the fusion line are blocked by the new nucleated grains, and the columnar to equiaxed transition (CET) occurs in the weld. The grain density in the weld increases gradually from the fusion line to the weld center. In addition, the grain morphology changes gradually from the columnar to equiaxed from the fusion line to the weld center. Between the columnar and the equiaxed grains, there exists an elongated grain region in the weld.

When the welding speed increases to 10 mm/s, the weld pool become tear-drop shaped and the trailing pool boundary is nearly straight. Under this condition, the morphological evolution of the weld solidification structures is shown in Fig.16. As discussed above, the ratio G/R will decreases with the increase of the welding speed. Thus, the constitutional undercooling in the molten pool also increases. According to the Eq.14, the heterogeneous nucleation becomes easier. Compared with the result shown in Fig.14, the equiaxed grain zone enlarges toward the fusion line. Moreover, the finer equiaxed grains are obtained at this higher welding speed. It should be noticed that the equiaxed grain size will keep constant in the central region of the weld. Because the number of the heterogeneous particles is limited in the molten pool. And the grain density will not increase when the undercooling in the weld pool reaches the critical value. According to the simulated results above, the higher welding speed and welding current can obviously increase the CET tendency in welding.

4. Conclusions

A macro-micro coupled model was developed which combined the welding process and the solidification structure in the molten pool. The weld pool morphologies and the

solidification conditions along the trailing pool boundary were analyzed together with the morphological evolution of the various weld solidification structures for different welding conditions. The main conclusions of the present paper can be surmised as follows:

1. The solidification conditions are related with the location in the weld and the welding parameters. With the decrease of the distance to weld center, the temperature gradient G decreases and the solidification rate R_w increases, and the ratio G/R_w becomes smaller gradually. With the increase of the welding speed, the temperature gradient decreases near the weld center and increases near the fusion line. The solidification rate along the trailing pool boundary becomes larger, and the ratio G/R_w will decrease at the same time.

2. When the heterogeneous nucleation does not occur in the molten pool, the weld solidification structure is only consisted with the epitaxial columnar grains from the fusion boundary. Owing to the anisotropic kinetics of the columnar grains, the axial structure forms in the central region of the weld at low welding speeds. With the increase of the welding speed, the axial structure region becomes smaller and eventually disappears in the weld.

3. The heterogeneous nucleation can affect the weld solidification structure obviously. The weld grain structure is directly determined by the competitive growth between the columnar and new nucleated grains. At low welding speeds, the preferential orientations of the survival grains get closer to the welding direction with the decrease of the distance to the weld center. The solidification structures in the weld curve toward the welding direction at the macro-scale. With the increase of the welding speed, more grains form in the molten pool, and the new nucleated grains also become more competitive. The columnar grains from the fusion line are blocked by the new nucleated grains, and the continuous equiaxed grain zone forms in

the weld.

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Figure captions

Fig.1 The schematic diagram of spatial bilinear interpolation method. (a) The macro-scale mesh. (b) The micro-scale cell.

Fig. 2 Schematic diagram of capturing algorithm used in present CA model. (a) A square-shaped nucleus grows with the defined preferential crystal orientation. (b) The central cell captures new interface cells from its neighboring cells. (c) The new interface cells begin to grow.

Fig.3 The simulation results of the weld pool shape for different welding parameters (a) The welding speed is 2mm/s. (b) The welding speed is 4mm/s. (c) The welding speed is 10mm/s.

Fig.4 The simulation results of the temperature gradient along the trailing pool boundary for different welding parameters

Fig.5 The schematic diagram for the relationship between the welding speed $V_{welding}$ and the solidification rate R_w

Fig.6 The simulation results of the angle between the normal to trailing pool boundary and the welding direction for different welding parameters

Fig.7 The simulation results of the normal solidification rate along the trailing pool boundary for different welding parameters

Fig.8 The results of G/R_w along the trailing pool boundary for different welding parameters

Fig.9 The various tip velocities of the Al-4%Cu equiaxed dendrite. (a) The undercooling is 3K. (b)The undercooling is 5 K.

Fig.10 The verification of the growth kinetics for the Al-4%Cu equiaxed dendrite. (a)The steady tip velocities for different undercoolings. (b) The equilibrium solute concentrations for different undercoolings.

Fig.11 The morphological evolution of the columnar grain structure in the weld. Figs. (a)、(b)、(c)、(d) shows the results for 2、4、6、12 s, and the welding speed is 2 mm/s.

Fig.12 The axial structure in the gas tungsten arc (GTA) weld of the aluminum alloy [16] The welds shown in this figure are full penetrated in order to obtain the two-dimensional solidification structures.

Fig.13 The weld solidification structures for different welding conditions. (a)The welding speed is 5 mm/s[35]. (b) The welding speed is 12.7 mm/s [36].

Fig.14 The morphological evolution of weld solidification structure with the welding speed of 2 mm/s. Figs. (a)、(b)、(c)、(d) shows the results for 2、4、6、12 s.

Fig. 15 The morphological evolution of weld solidification structure with the welding speed of 4 mm/s. Figs. (a)、(b)、(c)、(d) shows the results for 1.2、2.5、4.1、8 s.

Fig. 16 The morphological evolution of weld solidification structure with welding speed of 10 mm/s. Figs. (a)、(b)、(c)、(d)、(e)shows the results for 0.9、1.3、1.9、2.3、6 s.

Tables

Table 1 Physical properties of the Al-Cu alloy [20, 21]

Physical property	Al-Cu
Liquidus temperature (K)	921
Thermal conductivity ($\text{W m}^{-1} \text{K}^{-1}$)	120
Specific heat ($\text{J kg}^{-1} \text{K}^{-1}$)	1070
Magnetic permeability ($\text{H} \cdot \text{m}^{-1}$)	1.26×10^{-6}
Thermal expansion coefficient (K^{-1})	1×10^{-4}
Viscosity ($\text{kg m}^{-1} \text{s}^{-1}$)	0.006
Density (kg m^{-3})	2600
Latent heat (J kg^{-1})	3.97×10^5
Alloy initial composition (wt pct)	4.0
Liquidus slope (K wt pct^{-1})	-2.6
Equilibrium partition coefficient	0.17
Solute diffusion coefficient in liquid ($\text{m}^2 \text{s}^{-1}$)	3×10^{-9}
Solute diffusion coefficient in solid ($\text{m}^2 \text{s}^{-1}$)	3×10^{-13}
Gibbs-Thomson coefficient (K m)	2.4×10^{-7}
Anisotropy coefficient	0.159

Figures

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