Simulation study on temperature field and microstructure of Ti-6Al-4V alloy round ingot during EBCHM

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Abstract
The solidification structure of Ti-6Al-4V round ingot during the electron beam cold hearth melting (EBCHM) directly determines the quality of the ingot and the performance of the subsequent rolled coil. In this paper, the Cellular Automaton Finite Element (CAFE) method is used to numerically simulate the solidification structure of Ti-6Al-4V ingot. Firstly, the mathematical model is established with a numerical solution. Secondly, effects of process parameters including the pouring temperature and pulling speed on the solidification structure are revealed. The results show that the microstructures predicted by the numerical method match the experimental results. For the case of fixed pulling speed, a reduction in the pouring temperature leads to the grain refinement and the decreased volatilization of Al. With an increase of the pulling speed, the number of grains first increases and then decreases, but the average grain size first decreases and then increases. Furthermore, the maximum grain size monotonically increases with increasing the pulling speed. Thus, the fine solidified structure with fine grains can be obtained at the pouring temperature of 1700 °C and the pulling speed of 4 × 10^{-4} m s^{-1}.

1. Introduction

Titanium alloys exhibiting excellent properties including high corrosion and temperature resistance, and high specific strength have been widely used in aerospace, ocean and other fields [1–3]. Plates and strips of titanium alloys are the most used types. Normally, titanium alloy plates and strips are made from sponge titanium with adding alloy elements by smelt into ingot, with forging followed. Because titanium alloys have high chemical activity, especially in the molten state, it is easy to react with C, H and other impurities element to form brittle impurities [4]. Melting of titanium alloys should be carried out in vacuum or under the protection of inert gases such as argon [5]. Normally, titanium alloy ingots are mainly processed by the methods of the traditional vacuum arc remelting (VAR) and electron beam cold hearth melting (EBCHM) [6, 7]. Due to the EBCHM is much better than the VAR in solving the component segregation, microstructure uniformity and impurity removal. Therefore, EBCHM has become an irreplaceable advanced technology for the smelting of high-quality titanium alloy ingots [8, 9]. However, there are still some quality defects such as the cold scars and cracks on the surface of Ti-6Al-4V alloy ingots during EBCHM. Particularly, the inhomogeneity of microstructures will cause the deteriorated quality of plates and strips of titanium alloys. In the EBCHM process, the pouring temperature and speed play a critical role in the shape and depth of the molten pool [10]. The pouring temperature is determined by the power of the electron gun. Since the power of the electron gun is more complicated, in order to simplify the model, we only need to study the process parameters of the melt surface temperature in this paper. Based on the solidification theory, the shape and depth of the molten pool determine the temperature field distribution and microstructure of the ingot. In the solidification process of Ti-6Al-4V alloy round ingot
during EBCHM, the temperature field distribution and microstructure directly determine the quality of ingots [11, 12]. Reasonable temperature field distribution can reduce macro-segregation and improve the quality of ingot. The fine grain structure can improve the strength and toughness of the ingot and affect the processing performance of the subsequent products. Thus, the temperature field and microstructure were studied by numerical simulation in this paper. In order to study the influence of the main process parameters on the temperature field distribution and microstructure of Ti-6Al-4V alloy ingot, the following two cases were set: Case 1, controlling the pulling speed of the subsequent products. Thus, the temperature is 1760 °C, the pouring temperature is 1760 °C, and the pulling speed is set to 1 × 10⁻⁴ m s⁻¹, 1.66 × 10⁻⁴ m s⁻¹, 2 × 10⁻⁴ m s⁻¹, 3 × 10⁻⁴ m s⁻¹, 4 × 10⁻⁴ m s⁻¹, and 6 × 10⁻⁴ m s⁻¹, respectively.

2. Mathematical model

The heat transfer process during the solidification of titanium alloy ingot, is represented by Fourier-Kirchhoff equation [13]:

\[ \rho C \frac{\partial T}{\partial t} = \nabla \cdot (k \nabla T) + Q \]  \tag{1}

where \( C \), \( \rho \) and \( k \) are the specific heat capacity, the density and the thermal conductivity of the material as the function of temperature \( T \), respectively. \( \nabla \) = \( \frac{\partial}{\partial x} i + \frac{\partial}{\partial y} j + \frac{\partial}{\partial z} k \). \( i, j, k \) represent the unit vectors of \( x, y, z \) axes, respectively. \( Q \) is the heat source intensity. When the temperature is higher than the liquidus temperature or lower than the solidus temperature \( Q = 0 \), and between them [14]:

\[ Q = \rho L \frac{\partial F_s}{\partial t} \]  \tag{2}

where \( L \) is the solidification latent heat, \( F_s \) is the solid fraction. The enthalpy of Ti-6Al-4V alloy ingot during solidification satisfies the following equation [15]:

\[ H(T) = \int_0^T C dt + L(1 - F_s) \]  \tag{3}

Taking the derivative of temperature \( T \) on both sides of this equation, we have

\[ \frac{\partial H}{\partial T} = \rho L \frac{\partial F_s}{\partial T} \]  \tag{4}

Equation (2) and equation (4) can be substituted into equation (1), then

\[ \rho \frac{\partial H}{\partial T} = \nabla \cdot (k \nabla T) \]  \tag{5}

In the melting process of Ti-6Al-4V titanium alloy ingots, besides the heat transfer of the alloy itself, the crystalizer itself is accompanied by complex heat transfer. Its heat transfer process is expressed by the following equation:

\[ \rho_M C_M \frac{\partial T_M}{\partial t} = \nabla \cdot (k_M \nabla T_M) \]  \tag{6}

where, each symbol in the equation (6) has the same meaning as equation (1), the subscript \( M \) represents the crystalizer, \( C_M \), \( \rho_M \) and \( k_M \) are functions of temperature \( T_M \).

Based on the solidification theory, crystal nucleation can be divided into homogeneous nucleation and heterogeneous nucleation. In actual solidification process, nucleation is mainly achieved by heterogeneous nucleation, with external particles or walls as the substrates. In the solidification simulation, heterogeneous nucleation can be divided into instantaneous nucleation and continuous nucleation. In order to make the simulation process closer to the actual nucleation situation, the continuous nucleation model proposed by Rappaz et al [16] is used, which is obtained by statistical methods.

Within a certain period of time, the decrease in temperature leads to an increase in undercooling, and the change in grain nucleation density can be expressed by the equation:

\[ \delta_n = n[\Delta T + \delta(\Delta T)] - n(\Delta T) = \int_0^{\Delta T + \delta(\Delta T)} \frac{dn}{d(\Delta T)} d(\Delta T) \]  \tag{7}
where $dn/d(\Delta T)$ is corresponding Gaussian nucleation site distribution in the following equation:

$$
\frac{dn}{d(\Delta T)} = \frac{n_{\text{max}}}{\sqrt{2\pi} \Delta T_\sigma} \exp \left[ -\frac{1}{2} \left( \frac{\Delta T - \Delta T_{\text{max}}}{\Delta T_\sigma} \right)^2 \right]
$$

(8)

where $n$ is the total density of grains, $\Delta T_{\text{max}}$ is the mean undercooling, $\Delta T_\sigma$ is the standard deviation, and $n_{\text{max}}$ is the maximum density of nuclei obtained by a continuous nucleation distribution.

In the actual solidification of the Ti-6Al-4V titanium alloy, the degree of undercooling has an effect on crystal growth. The total degree of undercooling of the dendrite tip is composed of the following four parts:

$$
\Delta T = \Delta T_c + \Delta T_T + \Delta T_r + \Delta T_k
$$

(9)

where $\Delta T_c$ is the constitutional undercooling, $\Delta T_T$ is the thermodynamic undercooling, $\Delta T_r$ is the curvature undercooling caused by solid-liquid interface, $\Delta T_k$ is the kinetic undercooling. For most metals, except $\Delta T_c$, the other three undercooling effects are very small and can be ignored. It can be approximated as $\Delta T = \Delta T_c$. Thus $\Delta T$ can be determined by the follow equation:

$$
\Delta T = \Delta T_c = (C_0 - C^*_l) m
$$

(10)

where $C_0$ is the initial concentration of alloy, $C^*_l$ is the concentration at the tip of dendrite, $m$ is the liquidus slope.

Kurz et al [17] established the KGT model based on the theory of interface stability. When using the KGT model to calculate the growth rate of the dendrite tip, in order to accelerate the calculation, the KGT model is coupled, and the growth rate can be solved only by knowing the undercooling of the dendrite tip. After the KGT model is simplified, the relationship between the growth rate of the dendrite tip and the undercooling as follow equation [18]:

$$
\nu(\Delta T) = \alpha \Delta T^2 + \beta \Delta T^3
$$

(11)

where $\alpha$ and $\beta$ are the dynamic coefficient of dendrites tip growth.

3. Geometry model and boundaries conditions

Figure 1 (a) is the schematic diagram of Ti-6Al-4V alloy round ingots during EBCHM. During the smelting process the raw materials enter the cold bed, followed by melting and refining in the cold bed. The materials then flow into the crystallizer with solidification, and finally pulled out by the pull spindle (made from copper) to get Ti-6Al-4V round ingot. This paper mainly studies the temperature field distribution and solidification structure of Ti-6Al-4V titanium alloy round ingot. Therefore, the final solidification stage, where the round ingot solidifies in the crystallizer, is used to establish the geometric model. This paper mainly studies the changing law of the influence of pouring temperature and drawing speed on solidification structure when solidification reaches the steady stat. For ease of calculation, we use the boundary movement method (The ingot does not move and the crystallizer moves upward at the pulling speed, so as to realize the process of pulling the ingot downwards relative to the crystallizer.) for numerical simulation.
Figure 1(b) shows the built geometric model with diameter of 220 mm, length of 900 mm. The height of the crystallizer is 180 mm, the wall thickness is 10 mm; the pull spindle diameter is 220 mm, the length is 100 mm. The geometric model is meshed by the Mesh module in ProCAST. After meshing, the total mesh number is 987792, with the surface mesh number of 26812 and the volume mesh number of 960980. Selecting the microstructure calculation area from the round ingot, and the selected position is 250 mm from the bottom of the ingot, as shown in figure 1(b). The initial position of the crystallizer is at the bottom of the ingot; V is the upward movement speed of the crystallizer, which is equal to the pulling speed.

When calculating the temperature field and microstructure of the Ti-6Al-4V alloy round ingot, the thermal parameters, dynamic growth coefficient and nucleation parameters need to be determined. The thermal parameters are calculated by the ProCAST thermodynamic database and corrected by experiments. The liquidus of the Ti-6Al-4V alloy is 1650 °C and the solidus is 1600 °C. The Relationship of the density, heat enthalpy and heat conductivity of Ti-6Al-4V alloys with temperature are shown in figure 2. The crystallizer material is copper, with density of $\rho = 8360.5 \text{ kg m}^{-3}$. The relationship of the conductivity and latent heat with temperature is shown in figure 3. The kinetic growth coefficient and nucleation growth parameters of the Ti-6Al-4V alloy are $\alpha = 5.58 \times 10^{-6}$ and $\beta = 0$, respectively. Other parameters include the body nucleation density $n_{b, \text{max}} = 2 \times 10^{-9} \text{ m}^{-3}$, surface nucleation density $n_{s, \text{max}} = 5 \times 10^7 \text{ m}^{-2}$, body nucleation undercooling $\Delta T_{b, \text{max}} = 7 \text{ K}$, surface nucleation undercooling $\Delta T_{s, \text{max}} = 0.5 \text{ K}$, standard variance of body undercooling $\Delta T_{b, \sigma} = 0.5 \text{ K}$, standard deviation of surface undercooling $\Delta T_{s, \sigma} = 0.5 \text{ K}$. The heat transfer coefficient of the interface between the ingot and the crystallizer, the ingot and the pull spindle are set to 1000 $W/ (\text{m}^2 \cdot \text{K})$, and the heat transfer coefficient of the outer surface of the mold and the pull spindle is set to 5000 $W/ (\text{m}^2 \cdot \text{K})$. The movement speed of the crystallizer is controlled by the C code.

Based on above parameters, we will obtain the numerical solution of the mathematical model. For the reliability of the model, it was verified in our previous work [20].

4. Results and discussion

4.1. Effect of process parameters on temperature field distribution

Figure 4 shows the steady-state temperature field distribution of the Ti-6Al-4V alloy round ingot under the pouring temperature of 1700 °C and the pulling speed of $1.66 \times 10^{-4} \text{ m s}^{-1}$. As can be seen, the different colors
represent different isotherms, and the liquidus and solidus are listed in the partial magnification. When the temperature field reaches the steady state where the temperature field distribution does not change with the time, the liquidus depth \((h_1)\), solidus depth \((h_2)\) and the width of mushy zone \((h_2-h_1)\) can be directly measured from the temperature field distribution.

4.1.1. Temperature field distribution under different pouring temperatures

Figure 5 is the temperature field distribution of the Ti-6Al-4V alloy under the cases 1. It can be seen that the molten pool gradually deepens, and the mushy zone gradually narrows with increasing the pouring temperature. In order to quantitatively analyze the relationship of liquidus depth \((h_1)\), solidus depth \((h_2)\) and the width of mushy zone \((h_2-h_1)\) with the pouring temperature, the morphology of the molten pool is measured and drawn in figure 6. It can be visually seen that when the pouring temperature increase from 1700 °C to 1960 °C, the \(h_1\) increases from 3.20 cm to 7.20 cm, the \(h_2\) increases from 4.48 cm to 8.08 cm, and the \(h_2-h_1\) decreases from 1.28 cm to 0.88 cm. When the pouring temperature increases the undercooling region of the solidification front narrows, the temperature gradient increases, the melt nucleation rate decreases, the growth of columnar crystals is promoted, and the development of equiaxed crystal regions is hindered. The Ti-6Al-4V alloy contains volatile element Al. When the pouring temperature is too high, the volatilization of the Al element in the alloy increases, and the insufficient Al element content leads to insufficient quality of the ingot. When the pouring temperature is too low, the fluidity of the solution will be insufficient, and quality defects such as shrinkage will be formed inside the ingot. Therefore, a reasonable pouring temperature can suppress the growth of columnar crystals, reducing the segregation of the ingot composition, and promoting the uniformity of the alloy composition.

4.1.2. Temperature field distribution under different pulling speed

Figure 7 shows the temperature field distribution of Ti-6Al-4V alloy ingot under the case 2. It can be seen from the figure that the molten pool is gradually deepened and the mushy zone becomes wider. Figure 8 shows the relationship of \(h_1\), \(h_2\) and \(h_2-h_1\) with the pulling speed. It can be seen from figure 8 that when the pulling speed increases from \(1 \times 10^{-4}\) m s\(^{-1}\) to \(6 \times 10^{-4}\) m s\(^{-1}\), the \(h_1\) increases from 1.72 cm to 16.51 cm, and the \(h_2\) increases from 2.80 cm to 20.60 cm, the \(h_2-h_1\) increases from 1.10 cm to 4.08 cm. In this simulated condition, when the pulling speed is \(5.4 \times 10^{-4}\) m s\(^{-1}\), the depth of the solidus interface reaches 180 mm. The actual crystallizer height is 180 mm in this paper. When the depth of the molten pool exceeds the depth of the crystallizer, it is prone to leakage and safety accidents. Leakage refers to an accident in which the solidified shell
Figure 6. Relationship of $h_1$, $h_2$ and $h_2-h_1$ with pouring temperature.

Figure 7. The temperature distribution of Ti-6Al-4V alloy round ingot with different pulling speed: (a) $1 \times 10^{-4}$ m s$^{-1}$; (b) $1.66 \times 10^{-4}$ m s$^{-1}$; (c) $2 \times 10^{-4}$ m s$^{-1}$; (d) $3 \times 10^{-4}$ m s$^{-1}$; (e) $4 \times 10^{-4}$ m s$^{-1}$; (f) $6 \times 10^{-4}$ m s$^{-1}$.

Figure 8. Relationship of $h_1$, $h_2$ and $h_2-h_1$ with pulling speed.
cannot resist the hydrostatic pressure of the metal after exiting the crystallizer during continuous casting, causing the molten metal to flow out of the weak part of the shell. So, in order to prevent the occurrence of safety accidents, the pulling speed should be controlled within $5.4 \times 10^{-4} \text{ m s}^{-1}$ under this simulation condition. It is worth noting that for other crystallizer sizes, the maximum drawing speed depends on the actual crystallizer size, cooling water conditions and other conditions. For other conditions, please refer to the [21]. Comparing figure 5, it is found that the influence of the pulling speed change on the temperature field of the round ingot is more significant than that of the pouring temperature change. Therefore, in practical production, controlling the pulling speed is more important than controlling the pouring temperature.

### 4.2. Effect of process parameters on microstructure

Figure 9 shows the simulation result of microstructure evolution of the Ti-6Al-4V alloy round ingot under the pouring temperature of 1700 °C and pulling speed of $1.66 \times 10^{-4} \text{ m s}^{-1}$: (a) cross section; (b) longitudinal section. The selected position of cross section is 250 mm away from the bottom of the ingot. The longitudinal section starts from the bottom and goes through the center of the circle. Different colors represent crystallographic orientations, and the gray area in the center is the un-solidified liquid phase. It can be seen from figure 10(a), after solidification starts, the grains first nucleate in the surface of the ingot. This is because the surface of the ingot is in contact with the wall of the crystallizer, producing a greater undercooling part. And the wall of the crystallizer provides a substrate for nucleation, resulting in a large number of grains forming on the surface of the ingot. At this time, the formed grains are fine with distributed crystallographic orientations. The columnar crystal area with relatively coarse grains then starts to form immediately adjacent to the fine crystal area, and finally the fine equiaxed grain area with grains is formed in the central area.

It can be seen from figure 10(b), in the longitudinal section of the ingot, during the grain at the position less than 180 mm from the bottom of the ingot, it grows from outside to inside with a relatively fine size. The grains at the bottom of the ingot tend to grow parallel to the ingot. The grains at the side of the ingot tend to grow perpendicular to the ingot. During the grain at the position more than 180 mm from the bottom of the ingot, it
begins to grow obliquely from bottom to top. This is because the bottom of the ingot is in contact with the pull spindle at the beginning of crystallization, which produces the largest undercooling, and the formed crystal grains are relatively fine. It gradually becomes smaller and the solidified structure of the ingot tends to be stable. This is because at the beginning of crystallization, due to the contact between the bottom of the ingot and the ingot rod, the maximum undercooling is generated and the grains formed are relatively small. With the ingot pulling out continuously, the influence of the pull spindle on the cooling rate of the ingot gradually decreases and the microstructure of the ingot tends to be stable. It can be seen from the figure that the grain growth tends to be stable when the distance from the bottom of the ingot is greater than 180 mm. Therefore, when studying the grain change on the cross section, the selected cross section position should be greater than 180 mm. The cross sections selected in this paper are all 250 mm away from the bottom of the ingot.

4.2.1. Effect of pouring temperature on microstructure of Ti-6Al-4V alloy round ingot

Figure 10 shows the microstructure of the Ti-6Al-4V alloy ingot at the same cross-section under case I. An increasing of the pouring temperature results in a decrease of proportion of surface fine grain area and an increase of the proportion of columnar crystal area. The change of central equiaxed crystal area is not obvious. In order to intuitively observe the change rule of grain number and average grain radius with pouring temperature, the relationship between the change of grain number and average grain radius with pouring temperature at the cross section is calculated and plotted by ProCAST, as shown in figure 11. An increase of the pouring temperature leads to a decrease of the number of grains and an increase of the average grain radius. This is because when the pulling speed is fixed, the cooling capacity of the crystallizer is fixed. As the pouring temperature increases, the undercooling of the melt gradually decreases, the nucleation rate decreases, and the columnar crystal growth is promoted. Therefore, the number of grains decreased and the average grain radius increased. Drop the pouring temperature can refine grains.

4.2.2. Effect of pulling speed on microstructure of Ti-6Al-4V alloy round ingot

Figure 12 shows the microstructure of the Ti-6Al-4V titanium alloy round ingot at the same cross section under case II. When the pulling speed is $1 \times 10^{-4}$ m s$^{-1}$, equiaxed grains can be observed with a large average size. When the pulling speed is above $1.66 \times 10^{-4}$ m s$^{-1}$, the typical ingot three areas begin to form on the cross section, the grains in the surface fine grain zone become finer, and the columnar crystal zone becomes more and more distinct with increasing the pulling speed. This is because when the pulling speed is $1 \times 10^{-4}$ m s$^{-1}$, the undercooling in the melt increases sharply due to the slow pulling speed, and a large number of crystal nuclei are formed in the melt at the same time, almost all of them form equiaxed crystals with large average grain area. With increasing pulling speed, the undercooling of the melt gradually decreases and the fine grain area begins to form on the surface of the ingot. The columnar crystal begins to form inside, in the meantime the fine grain area is getting finer and the columnar crystal is becoming more and more distinct.

In order to intuitively observe the change of grain number and grain size with the pulling speed, the relationship between the grain number and average grain radius with the ingot speed is analyzed from extract relevant data at the same cross section, as shown in figure 13. The average grain radius first decreases and then increases, and the number of grains first increases and then decreases with increasing the pulling speed. In order
to better respond to the change of grain, other conditions of grain are counted, as shown in Table 1. It can be seen from Table 1, an increasing of the pulling speed results in a decrease of maximum grain area followed by a subsequent increase. Also, the average grain area decreases first and then slightly increases. This is because although the pulling speed is increased to promote the growth of columnar crystals, the growth rate of columnar crystals cannot keep up with the degree of surface grain refinement within a certain range of pulling speeds so there is an increase in the number of grains and a decrease in the average grain radius. When the pulling speed is greater than $4 \times 10^{-4} \text{ m s}^{-1}$, melt subcooling drops further, and the columnar crystal area formed is more and more obvious, the number of grains begins to decline, and the average grain radius increases. In a word, the influence of pulling speed on the microstructure of round ingot is greater than that of pouring temperature.

When the pulling speed is set to $2 \times 10^{-4} \text{ m/s}$ under the simulation conditions, can obtain the microstructure with the largest grain area and the smallest.

![Figure 12. Grain structures within the identical cross-section with different pulling speed: (a) $1 \times 10^{-4} \text{ m s}^{-1}$; (b) $1.66 \times 10^{-4} \text{ m s}^{-1}$; (c) $2 \times 10^{-4} \text{ m s}^{-1}$; (d) $3 \times 10^{-4} \text{ m s}^{-1}$; (e) $4 \times 10^{-4} \text{ m s}^{-1}$; (f) $6 \times 10^{-4} \text{ m s}^{-1}$.]

![Figure 13. Relationship of the mean radius and number of grains with the pouring speed.]

Table 1. Results of the identical cross-section grain with different pulling speed.

| Pulling speed($10^{-4} \text{ m s}^{-1}$) | 1   | 1.66 | 2   | 3   | 4   | 6   |
|----------------------------------------|-----|------|-----|-----|-----|-----|
| Maximum grain area(cm$^2$)             | 0.416 | 0.322 | 0.267 | 0.371 | 0.430 | 0.464 |
| Average grain area($10^{-2} \text{cm}^2$)| 2.345 | 1.463 | 0.898 | 0.636 | 0.592 | 0.598 |
5. Conclusions

In this paper, the finite element software ProCAST was used to simulate the temperature field and microstructure of the Ti-6Al-4V alloy ingot. The main result is as follows.

1. When simulating the temperature field of the Ti-6Al-4V alloy ingot, it was found that as the pouring temperature and the pulling speed increased, the molten pool became deeper and the influence of the pulling speed on the shape of the molten pool was more significant than the that of the pouring temperature. Therefore, it is necessary to strictly control the pulling speed in practical production.

2. When the pulling speed was constant and the temperature increased from 1700 °C to 1960 °C, the depth of the solidus increased from 4.48 cm to 8.08 cm, the depth of the liquidus increased from 3.20 cm to 7.20 cm, and the width of the mushy zone decreased from 1.28 cm to 0.88 cm. Choosing the proper pouring temperature in practical production not only could suppress the growth of columnar crystals with reduced segregation of ingot composition, but also ensure the uniformity of alloy ingot composition.

3. When the pouring temperature was constant, the pulling speed increased from $1 \times 10^{-4}$ m s$^{-1}$ to $6 \times 10^{-4}$ m s$^{-1}$, the depth of the solidus increased from 2.80 cm to 20.60 cm, the depth of the liquidus increased from 1.72 cm to 16.51 cm, and the width of the mushy zone increased from 1.1 cm to 4.08 cm. In order to prevent the danger of drawing leakage in the simulation condition, the pulling speed should be controlled within $5.4 \times 10^{-4}$ m s$^{-1}$.

4. After simulating the microstructure of Ti-6Al-4V alloy round ingot, it was found that the experimental results were consistent with the simulation results. The number of grains increased and the average grain area decreased with increasing the pouring temperature. With increasing the pulling speed, the grain number first increased and then decreased, the average grain radius first decreased and then increased, and the maximum grain area first decreased and then increased. Under the simulated conditions, the maximum grain area of solidification structure was obtained when the ingot pulling speed was $2 \times 10^{-4}$ m s$^{-1}$.

The results obtained in this paper have practical guiding significance for the production of ingots by the directional solidification.

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Data availability statement

The data generated and/or analysed during the current study are not publicly available for legal/ethical reasons but are available from the corresponding author on reasonable request.

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