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Study of the precipitation, nucleation, and grain growth of phases in the ZL201 cast aluminum alloy by integrated calculations

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Abstract
The purpose of this study is to investigate the influence of alloy composition on the evolvement rule of microstructure for ZL201 cast aluminum alloy. The ZL201 casting samples with different components were prepared by gravity diecasting process, the phase composition was characterized and calculated by x-ray diffraction (XRD) and thermo-calc (TC) software respectively, the evolvement rule of microstructures was observed by metallographic microscope (OM) and scanning electron microscope (SEM), and simulated by ProCAST software. The results show that the Al2Cu(θ) formed by Cu and Al matrix, which increased the free energy and reduced the precipitation temperature for the alloy system. For the microstructure of ZL201, the nucleation direction of Al-Cu was along the positive direction of z-axis, the morphology was a regular ribbed morphology; Al-Mn and Al-Ti were along the negative directions of x- and y-axis and morphologies were columnar. Besides, the nucleation mode of casting center was dominated by the growth pattern of Al-Cu, and that of casting surface was dominated by Al-Ti and Al-Mn.

1. Introduction

Recently, the development of lightweight automobiles has received increasing attention. Because of the rapid growth of car ownership, lightweight automobiles provide an effective measure to mitigate the associated energy and environmental problems [1]. The advantages of the ZL201 aluminum alloy include low density, good formability, and excellent corrosion resistance [2]. Consequently, ZL201 has become an important material for realizing lightweight automobiles. The ZL201 aluminum alloy has been widely used in the automobile body, chassis, and powertrain. The Al-Cu matrix plays an important role in the ZL201 alloy [3]. However, the hot cracking tendency of Al-Cu alloys is serious [4]. As a result, the casting performance is poor, and its application in industry is greatly limited. The factors affecting ZL201 casting defects and mechanical properties are multiscale and multifaceted. They primarily include alloy composition, casting properties, casting conditions, and grain size [5]. In particular, the alloy composition has a significant effect on the microstructure during material processing and preparation. The primary method to improve the casting performance of the ZL201 alloy and avoid thermal cracking in the forming process is to add trace alloying elements [6]. The most commonly used elements are Cu, Mg, Zn, Mn, Ti, V, and Re [7]. The θ strengthening phase is formed by Cu and α-Al, and can significantly improve the casting strength [8]. Mn and the Al-Cu binary alloy form an Al-Cu-Mn phase, which could significantly improve the room temperature and high temperature mechanical properties [9]. Ti and α-Al form the Al3Ti phase, which can refine grains [10]. Zr reacts with Al to form Al2Zr in a peritectic manner. Because the melting point of Al3Zr is high, it refines the grain structure by acting as an external nucleus during solidification [11]. Per unit volume, increasing the number of grains decreases the grain size. In general, microstructure refinement can enhance the alloy properties. Different nucleation sites have a significant effect

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on the growth patterns of grain, and the growth patterns of grains determine the refinement for the alloy structure. M. Jin [12] studied deformation-induced grain growth during nanoindentation of ultrafine-grained Al at room temperature, and the results suggest that grain growth and coalescence are important modes of response in the deformation of ultrafine and nanograin materials. W. T. Kim [13] studied the nucleation and growth of \( \text{Al}_2\text{CuMg} \) in \( \text{AlLiCuMg} \) and \( \text{AlCuMg} \) alloys, and the results show that the primary nucleation mechanisms of \( \text{S}^+ \) in \( \text{Al-Cu-Mg} \) alloys are associated with the clustering of Mg and Cu at vacancy loops and helixes, and heterogeneous nucleation at subgrain boundaries the dominant nucleation sites are the \( \gamma'/\text{aluminum matrix} \) interphase boundaries and dislocations in the Li-containing alloy. At present, there is an abundance of experimental reports on the effect of grain growth patterns and nucleation sites on alloy microstructure, but there is still a lack of theoretical information. Because of the randomness of grain growth, which makes it impossible to achieve the desired pattern solely through experiments, integrated computing has become a new research trend to solve this problem computationally. First, the simulation model is established by theoretical calculations. Second, the different process parameters, such as alloy composition, casting properties, and casting conditions, are adjusted by model optimization. Finally, the optimum process parameters are used in experiments and production.

The complicated relationship between the growth patterns of grains and nucleation sites makes the theoretical study and mathematical analysis of the solidified microstructure difficult. In recent years, with the development of experimental technology, a simulation model has been established that significantly shortens the processing time and production cycle [14]. In the microscopic simulation, the grain size and morphology are predicted by establishing a dynamic model of the nucleation and growth process [15]. Therefore, it is not only necessary to calculate the macroscopic and microscopic fields, but also simulate the grain nucleation and growth process [16]. In the study of microstructure simulation, the influence of alloy composition on the solidification process was established and quantitatively calculated by the cellular automation (CA) method [17]. It can describe the formation process of free dendrites and columnar crystals, as well as the transformation process of columnar and equiaxial crystals [18]. In addition, thermodynamic theory can calculate complex homogeneous and heterogeneous phase equilibria, and then the results can be plotted as property diagrams and phase diagrams by computer [19]. These thermodynamic models and databases can be used to make calculations on a large variety of materials, such as steels, alloys, slags, salts, ceramics, solders, polymers, subcritical aqueous solutions, supercritical electrolyte solutions, non-ideal gases and hydrothermal fluids, and organic substances [20]. The relationship among the technology, microstructure, and properties for cast aluminum alloys is described quantitatively by integrated computing [21]. It allows casting processes to be designed that avoid defects in ZL201 aluminum alloy castings and improve the comprehensive performance [22].

In this work, the microstructures of Al-X binary alloys, Al-X-X ternary alloys, and the ZL201 alloys were simulated by the CA method. Alloy phase diagrams and cooling solidification curves were simulated by thermodynamic theory. To verify the effect of different elements on the ZL201 microstructure, the experimental results were characterized by x-ray diffraction (XRD), optical microscopy (OM), and scanning electron microscopy (SEM). Some consistency of the results between theory and experiment is evident. The nucleation mode was dominated by the growth pattern of Al-Cu, the nucleation site was near the center of the sample, and the microstructure had a regular ribbed morphology. The nucleation mode was dominated by the growth patterns of Al-Ti and Al-Mn, the nucleation site was on the surface of the samples, and the microstructure also had a columnar morphology.
2. Establishment of the model and determination of the parameters

2.1. Meshing models
The casting of ZL201 was prepared according to figure 1. The detailed parameters are as follows: \( R \geq 15 \text{ mm} \), \( P \approx 60 \text{ mm} \), \( L = 50 \text{ mm} \), and \( D = 10 \text{ mm} \).

In the simulation process, the surface mesh and volume mesh were divided to establish the microstructure simulation (CAFE). As shown in figure 2, equilateral triangles were used to divide the grid cells. The grid size of the ZL201 casting simulation was 1.2 mm, and the sampling casting was 0.3 mm. To thoroughly analyze how the microstructure is affected by different alloy compositions, the nucleation site was sampled and analyzed. The sample was a cylinder with a diameter of 15 mm and a height of 10 mm. To accurately simulate the microstructure during grain growth, the grid needs to be continuously optimized. The nucleation particle grew in a subdivided mesh. The mesh division should not only ensure the number of nucleated particles in the mesh, but also ensure that the growth of grains cannot exceed the range of a cell.

2.2. Thermophysical parameters
The composition of alloying elements is shown in table 2. The microstructure models were simulated by the CA method. The temperature of the molding was set to 250 °C and the casting was set to 750 °C. In the microstructure fields, the accuracy of thermophysical parameters directly affects the results of the simulation. In the ZL201 casting model, the solid phase fraction was calculated by the Scheil equation for intracellular grains [23].

\[
C_i = k_e C_0 (1 - f_s) k_e^{k_e - 1}
\]  

(1)

\( k_e \) is the effective solute solidification coefficient; \( C_0 \) is the initial composition; \( f_s \) is the frequency; and \( C_s \) is the solute segregation of the solid phase.

In the microstructure simulation, the growth coefficient directly determines the growth rate of dendrites. The coefficient of growth can be expressed as follows [24]:

\[
\alpha = \left[ \frac{-\rho}{(m \cdot c_0)(1 - k)^2 \cdot 2 \cdot \Gamma \cdot k} + \frac{m \cdot c_0}{(m \cdot c_0)(1 - k) \cdot D} \right]
\]  

(2)
Table 1. Simulation parameters for CAFE fields.

| Parameter                                      | Unit         | Numerical value |
|------------------------------------------------|--------------|-----------------|
| Coefficient of heat conduction [25]            | W/(m²·K)     | 1050            |
| Heat [26]                                      |              | FilmCo = 10, T = 25 °C |
| Gate velocity                                  | cm s⁻¹       | 0.24            |
| Coefficient of growth (α, β)                   |              | 4.45 × 10⁻⁷ (α), 3.74 × 10⁻⁸ (β) |
| Maximum nucleation density (n_max) [27]        | m⁻³          | 1 × 10⁸         |
| Average nucleation and sub-cooling (ΔT_n) [27]  | K            | 0.4             |
| Standard curvature sub-cooling (ΔT∂) [28]       | K            | 0.1             |

Table 2. Elemental composition of the ZL201 auxiliary frame.

| Elementary composition | Cu    | Mn    | Ti    | Al    |
|------------------------|-------|-------|-------|-------|
| Binary alloy           | 5.0 wt% | \   | 0.8 wt% | \   | Margin |
|                        | \    | \    | 0.3 wt% | \   |       |
| Ternary alloys         | 5.0 wt% | 0.8 wt% | \   | 0.3 wt% | Margin |
|                        | 5.0 wt% | \    | 0.8 wt% | 0.3 wt% | Margin |

\[
\beta = \frac{D^2}{\pi \Gamma} \cdot \frac{1}{(m \cdot c_0)(1 - k)^2 \cdot D}
\]

\( m \) is the liquid phase slope; \( c_0 \) is the alloy composition; \( \Gamma \) is the Gibbs-Thompson parameter; \( D \) is the liquid diffusion coefficient; \( k \) is the equilibrium solute partition coefficient; \( \alpha \) is the first level growth coefficient; and \( \beta \) is the second level growth coefficient. The values of the thermophysical parameters and growth coefficients required for the microstructure simulation are listed in Table 1.

The cooling solidification curves and phase diagrams of the ZL201 aluminum alloy were analyzed by thermodynamic theory. The simultaneous solution of the above equations was calculated to obtain a series of simultaneous equations. From these equations, the component with the lowest total Gibbs free energy was obtained. According to the results, the cooling and solidification curves and the phase diagrams of the ZL201 aluminum alloy system were drawn by thermodynamic calculation [29].
3. Experimental preparation

As shown in figure 3, the ZL201 aluminum alloy was cast by the gravity diecasting process. The composition of the different samples prepared for this study are shown in table 2. In addition to these components, impurity elements, including Fe, Si, and B, were present in concentrations of less than 0.01 wt%. The calcination
temperature was 700 °C. The metal mold was made from carbon steel C45. The casting time was 3.5 s. The cooling method was air cooling \( \text{FilmCo} = 10, T = 25 ^\circ C \). Figure 4 shows the sampling sections that were analyzed by OM, SEM, and XRD.

The microstructural sampling process of the ZL201 aluminum alloy is as follows:

1. As shown in figure 4, samples were taken from the aluminum alloy castings. The sampling position of the casting samples was inside the label and were incised by a metallographic sample cutting machine.

2. The casting samples were ground with 180, 400, 800, and 1200 grit sandpaper. After each grind, the casting samples were rotated by 90°.

3. The grinding samples were polished on the polishing machine at a speed of approximately 1000 rpm.

4. The polished samples were cleaned with anhydrous ethanol (99.99 mol%). The sample surface was corroded with hydrofluoric acid (10 mol%) for approximately 5–10 s. The observation position of the microstructure is shown in figure 4.

4. Results

4.1. Casting process

The figure 5 shows that the ZL201 casting was cast by the gravity diecasting process, the initial temperature field was obtained by steady-state analysis. The initial temperature of the casting was 700 °C. Contacting between castings and molds results in a strong heat exchange. Section 1 was the first to reach the liquidus temperature at 646.3 °C. Then, the temperature of the most parts of casting begin to drop. The most parts of liquid aluminum alloy begin to solidify, which accords with the basic physical process of heat conduction. When the process continued to 10 min, the sections 2 and 3 pouring gate were still the liquid, and other positions had dropped below the temperature of the solid phase line. In addition, the heat dissipation conditions of the casting edge part are supreme. The transformation from liquid state to solid state about 11.783 s, and the phase transition zone began to spread to the surface of the casting. When solidification reached 1206 s, the temperature of the casting drops to 20 ~ 66 °C. Due to internal and external temperature difference is less than 15 °C, leading to the rate of descent of the casting temperature becomes slow. The heat exchange process is basically complete until it cools to room temperature.

4.2. Al-Cu, Al-Mn, and Al-Ti binary alloys

The binary phase diagrams of Al-5.0 wt%Cu, Al-0.8 wt%Mn, and Al-0.3 wt%Ti were calculated by thermodynamics and are shown in figure 6. As shown in figure 6(a), the Al-Cu had a eutectic reaction at 548 °C, and the solubility of Cu in Al was 5.65 wt%. As the temperature decreased, the solid solubility also decreased, and the Cu solid solubility was only 0.1% at room temperature. Therefore, as the temperature decreased, Al₆Cu (θ phase) subcrystals would precipitate from the solid solution. The subcrystalline precipitation process is also called the desolventizing process \[30\]. The enhancement of aluminum alloys through heat treatment is based on the dissolution of the solid solution. If the dispersity and solid solubility of the \( \theta \) phase are properly controlled in

![Figure 7. Calculated average grain size of the binary aluminum alloys.](image-url)
the desolventizing process, it can significantly improve the strength of ZL201 [31]. Figure 6(b) and figure 6(c) show that the alloying content of Mn and Ti were very low. It was difficult to form an intermediate phase with Al and Cu. During the cooling and solidification process, Al and Ti formed Al₃Ti, and Al and Mn formed Al₆Mn. When the Ti content was 0.3 wt%, the precipitation temperature of Al₃Ti was 742.8 °C. When the Mn content was 0.8 wt%, the precipitation temperature of Al₆Mn was 663.4 °C.

To analyze the influence of different elements on the grain growth of ZL201, metallographic samples with a diameter of 15 mm were taken at 25 °C. The corresponding finite element simulation and CAFE prediction analysis were carried out. Different colored grains represent different growth orientations [32]. It can be seen from the comparison between the macro structure and the CAFE fields in the corrosion state that the Al-5.0 wt% Cu sample had a rosiness grain structure, and the Al-0.8 wt%Mn and Al-0.3 wt%Ti samples had dendritic grain structures.

Using the simulated CAFE fields, the average grain size was calculated. Figure 7 shows that Al-0.3 wt%Ti had the minimum grain size (36.7 μm), while Al-0.8 wt%Mn had the maximum grain size (48.2 μm). The addition of Ti to the aluminum alloy refined the α-Al [33]. In addition, a small amount of Ti could be added to the α-Al. Ti and Al formed many Al₃Ti solid phase particles in the aluminum solution. Before the α-Al began to solidify, these tiny dispersed particles were separated. Because the lattice constant of Al₃Ti is close to that of α-Al, the α-Al could nucleate by attaching to the Al₃Ti particles. Therefore, the addition of Ti resulted in the α-Al forming small heterogeneous nuclei under a low subcooling. In addition, the low subcooling reduced the grain growth rate, and the grain size was refined. Because Al₆Mn and Al have significantly different lattice constants [34], the influence of Mn in the Al-Mn alloy was almost negligible.

By comparing the grain growth process of the Al-Cu, Al-Mn, and Al-Ti binary alloys, it is possible to determine how the ratio of different components would affect its nucleation. The OM observation location is
shown in the figure 8, and a quarter of the viewing surface was tailored according to figure 4. According to figure 8 (a) Al-5.0 wt%Cu, (b) Al-0.3 wt%Ti, and (c) Al-0.8 wt%Mn, the nucleation sites of the Al-Cu alloy were at the \( z = 0 \) surface in the CAFE model, while the nucleation sites of the Al-Mn and Al-Ti alloys were in the plane of rotation centered on the \( z \)-axis in the CAFE model. The nucleation direction of the Al-Cu alloy was along the positive \( z \)-axis direction in the CAFE model, and the nucleation directions of the Al-Mn and Al-Ti alloys were along the negative \( x \)- and \( y \)-axis directions. This is because the grain orientation was affected by the solid solubility and the degree of supercooling. The precipitation temperatures were 742.8 °C for \( \text{Al}_3\text{Ti} \), 548.6 °C for the \( \theta \) phase, and 663.4 °C for \( \text{Al}_5\text{Mn} \) in the binary alloy. The precipitation temperature of the \( \theta \) phase was the lowest, and figure 10 shows that the thermal conductivity of the lateral surface (section 2) was higher than the

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Figure 10. Variation of the temperature of different surfaces over time during casting.

Figure 11. Cooling and solidification curves of ternary aluminum alloys.
bottom surface (section 1) of the casting samples. Therefore, the nucleation sites with a fast cooling rate (section 1) were more suitable for the Al-Cu alloy, while the nucleation sites with a slow cooling rate (B) were more suitable for the Al-Mn and Al-Ti alloys [35].

During the solidification process, the morphology of the primary phase was markedly affected by the cooling rate of the aluminum alloy. The cooling rate was proportional to the degree of undercooling. As shown in figure 9(a), when the cooling rate was high, the \( \theta \) phase grew in a flat plane, and it formed a regular ribbed morphology. As shown in figure 9(b), because of the hindrance of the Al matrix and low cooling rate in section 2, the growth directions of the Al-Mn and Al-Ti alloys were only dependent on the same horizontal line, and they formed columnar morphologies. The supersaturation and the degree of supercooling were also different at different nucleation positions, which resulted in different nucleation tendencies.

4.3. Al-Cu-Mn, Al-Cu-Ti, and Al-Mn-Ti ternary alloys

Figure 11 shows the Al-5.0 wt%Cu-0.8 wt%Mn, Al-5.0 wt%Cu-0.3 wt%Ti, and Al-0.8 wt%Mn-0.3 wt%Ti cooling and solidification curves that were calculated by the thermodynamic method. As shown in figure 11(a),

![Graph of Gibbs free energy vs. temperature for different aluminum alloy components.](image)

**Figure 12.** Variation in the Gibbs free energy with temperature for different aluminum alloy components.

![Macroscopic corrosion diagrams of ternary aluminum alloys with corresponding CAFE fields.](image)

**Figure 13.** Macroscopic corrosion diagrams of ternary aluminum alloys with corresponding CAFE fields.
the precipitation temperature of the $\theta$ phase was 548.6 °C, and for Al$_{23}$Cu$_2$Mn it was 622.5 °C. When the alloy only contained Al and Mn, it formed the Al$_{23}$Cu$_2$Mn phase rather than Al$_6$Mn phase. This might be because the amount of Mn was very low, and Mn forms the Al$_{23}$Cu$_2$Mn intermetallic compound more easily with an Al-Cu alloy. Al$_{23}$Cu$_2$Mn can prevent the recrystallization of an aluminum alloy, raise the recrystallization temperature, and remarkably refine recrystallization grains. Figure 11(b) indicates that Ti formed Al$_3$Ti (only bonding with Al) in the Al-5.0 wt%Cu-0.3 wt%Ti ternary series alloys. The precipitation temperature of the $\theta$ phase was 548.6 °C, and for Al$_3$Ti it was 652.4 °C. From the above results, it can be concluded that the addition of Ti and Mn did not affect the precipitation temperature of the $\theta$ phase. The Cu and Mn significantly decreased the precipitation temperature of Al$_3$Ti, and the difference was approximately 90 °C. According to figure 12, the free energies of the Al-Mn and Al-Ti binary systems were smaller than those of the Al-Cu-Mn, Al-Cu-Ti, and Al-Mn-Ti ternary systems. A portion of the Cu was dissolved in the $\alpha$-Al, and the solid solution phase increased the free energy and reduced the precipitation temperature for the alloy system [36]. The lower free energy made the maximum useful work of the system on the environment higher, causing the system to become more stable and making the spontaneous reaction easier.

Figure 12 shows macroscopic corrosion diagrams and CAFE fields for Al-5.0 wt%Cu-0.8 wt%Mn, Al-5.0 wt%Cu-0.3 wt%Ti, and Al-0.8 wt%Mn-0.3 wt%Ti at different solidification stages. Different colors represent grains with different orientations in the CAFE fields. From figures 13(a) and (b), it can be seen that the $z=0$ surface in the CAFE model was dominated by the growth pattern of Al-Cu, and the microstructure had a regular ribbed morphology. The growth patterns for Al-Ti and Al-Mn dominated the position near the outside of the sample, and the microstructure had a columnar morphology. According to figure 13(c), because the growth directions of Al-Mn and Al-Ti were only dependent on the same horizontal line, the microstructure only had a columnar morphology. Given the consistency of the macro corrosion diagram and CAFE model results, the values of the simulation parameters are in line with the actual standards.

In addition, as the grains were gradually growing, the number of grains gradually decreased, and the grain orientations were close to each other. As a result, the competitive growth became slow among grains, and the rate at which the number of grains decreased was slower [37]. It can be seen from the growth modes of Al-Cu-Mn and Al-Cu-Ti that as the solidification front began to advance, there were small equiaxed crystal regions with random orientations distributed in the ternary aluminum alloy, and the grains began to gradually spread as the solidification time increased. Different methods of grain growth created competition, which caused the Al-Ti,
Al-Mn, and Al-Cu systems to inhibit growth of one another at the grain boundary. The polarized images are shown in figure 14. The metallographic microstructure of Al-5.0 wt%Cu-0.8 wt%Mn, Al-5.0 wt%Cu-0.3 wt%Ti and Al-0.8 wt%Mn-0.3 wt%Ti are shown in (a), (b), and (c), respectively. The viewing positions are shown in the figure 14, and a quarter of the viewing surface was tailored according to figure 4. At the $z=0$ surface, the growth patterns were dominated by Al-Cu, and the microstructure had a regular ribbed morphology. At the plane of rotation centered on the $z$ axis, the growth patterns were dominated by Al-Ti and Al-Mn, and the microstructure also had a columnar morphology. Different nucleation patterns were described in detail for the binary systems. Given the consistency of the metallographic structure and CAFE model results, the values of simulation parameters were in line with the actual standards [38].

Figure 15. XRD pattern of the ZL201 aluminum alloy.

Figure 16. Phase diagram of ZL201 that were calculated by thermodynamic theory.
4.4. ZL201 cast aluminum alloy

The XRD pattern of ZL201 aluminum alloy is shown in figure 15. The composition of the sample was Al-5.0 wt%Cu-0.8 wt%Mn-0.3 wt%Ti, and the impurity elements included Si, Fe, and B. The impurity concentrations were all less than 0.01 wt%. The $\theta$ and diffraction peaks of AlFcc were 38.2° (200), 45.4° (211), 65.1° (200), 77.8° (220), 82.7° (311), and 98.4° (222), respectively. For the $\theta$ phase, they were 21.7° (111), 45.4° (200), 65.1° (200), 47.8° (330), and 98.9° (600), respectively. For the Al$_3$Ti phase, they were 20.8° (002), 42.5° (004), and 47.1° (200), respectively. For the Al$_{20}$Cu$_4$Mn$_9$ phase, they were 23.4° (111), respectively. In addition, the addition of Mn induced Al-Cu to form the $\theta$ phases, while Mn formed the Al$_{20}$Cu$_4$Mn$_9$ intermetallic compound easier with the Al-Cu alloy. The Ti formed Al$_3$Ti (only bonding with Al), and Cu formed Al$_2$Cu ($\theta$ phase) with $\alpha$-Al. As shown in figure 16, the phase composition of the ZL201 aluminum alloy was calculated by thermodynamic theory. ZL201 was primarily composed of Al$_3$Ti, $\theta$, and Al$_{20}$Cu$_4$Mn$_9$ phases. The presence of other impurity phases was considered, such as Al$_2$Cu$_2$Fe. Al$_{20}$Cu$_4$Mn$_9$ was calculated by thermodynamics in the ZL201 alloy, while the detected XRD pattern and PDF-2009 was Al$_{20}$Cu$_4$Mn$_9$. Because of the randomness in the distribution of atoms and the uncertainty in the data, some differences were evident between them. However, the experimental results were consistent with the thermodynamic phase diagrams.

In the process of crystal growth, changes in the environment and composition are important factors. In addition, the crystal orientation might result in subtle changes in the physical properties and composition at different times. Therefore, different shapes and structures can often be seen in the cross-section of a crystal [39]. Figure 17(a) shows the ZL201 metallographic microstructure. The orientation of the grains was more inclined toward dendritic growth. This is because when the Cu content was 5.0 wt%, it could not form more phases with Al to stabilize energy. In addition, the degree of supercooling for Al-Mn and Al-Ti was larger than Al-Cu, and the rapid cooling rate promoted grains to form a strip structure. Figure 17(b) shows the SEM images of the ZL201 aluminum alloy; the $\theta$ indicates the $\theta$ phase, which was distributed by dispersive particles. Because the content of Mn and Ti was very low, Al$_3$Ti and Al$_{20}$Cu$_4$Mn$_9$ were evident after screening a large number of SEM images.

5. Conclusion

A portion of the Cu was dissolved in $\alpha$-Al, and the solid solution phase increased the free energy and reduced the precipitation temperature for the alloy system. This is the cause of the different precipitation temperatures of the Al$_3$Ti phase in the Al-Ti binary and Al-Cu-Ti ternary alloys.

At the $z = 0$ surface in the CAFE model, the nucleation mode was dominated by the growth pattern of Al-Cu in the Al-Cu-Mn and Al-Cu-Ti ternary alloys. At the surface of rotation centered on the $z$ axis, the nucleation mode was dominated by the growth pattern of Al-Ti and Al-Mn in the Al-Cu-Mn, Al-Cu-Ti, and Al-Mn-Ti ternary alloys. This is because the nucleation sites with high heat transfer efficiency (section 1) were more suitable for Al-Cu, and the nucleation sites with low heat transfer efficiency (section 2) were more suitable for Al-Mn and Al-Ti.

The nucleation direction of Al-Cu was along the positive $z$-axis direction with the CAFE model, and it formed a regular ribbed morphology. The nucleation directions of Al-Mn and Al-Ti were along the negative $x$-

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**Figure 17.** Microstructure of ZL201: (a) metallographic organization diagram; (b) scanning electron microscopy images.
and y-axis directions, and they formed columnar morphologies. This is because grain growth was affected by different cooling rates.

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Data availability statement

All data that support the findings of this study are included within the article (and any supplementary files).

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