Restrictions of Physical Properties on Solidification Microstructures of Al-based Binary Alloys by Cellular Automaton

Hongwei ZHANG,1) Keiji NAKAJIMA,2) Hong LEI1) and Jicheng HE1)

1) Key Laboratory of Electromagnetic Processing of Materials, Ministry of Education, Northeastern University, Wenhua Road 3-11, Shenyang 110004 China. E-mail: hongweizhang@epm.neu.edu.cn 2) Division of Applied Process Metallurgy, Department of Materials Science and Engineering, Royal Institute of Technology (KTH), SE-100 44 Stockholm, Sweden.

(Received on May 28, 2010; accepted on August 20, 2010)

The solidification microstructure evolution and the Columnar to Equiaxed Transition (CET) during Al–Si and Al–Cu binary alloy solidification processes are analyzed by the help of Cellular Automaton-Finite Difference (CA-FD) model. The effects of the physical properties, except the effects of the nucleation parameters and the operation parameters, on the cooling curves, the dendrite growth, the solidification morphologies and the CET of the Al–Si and Al–Cu binary alloys are emphatically discussed. Results show that the solidification morphologies are internally influenced by the physical properties related to the dendrite tip growth kinetics. Besides the solute diffusivity in liquid $D_s$, and the growth restriction parameter $Q = mC^0(k_{SL}^0 - 1)$, the Gibbs–Thomson coefficient $\Gamma$ shows a great effect on dendrite tip growth rate. Their effects on the dendrite tip kinetics are ordered as $\eta(k_{SL}^0 - 1) > \Gamma > D_s$. The growth rate can be predicted efficiently by the present simplified expression based on GGAN model combining several physical parameters and the local undercooling, which clearly shows the physical meaning of the constant coefficients in the simplified expression based on KGT model. The equiaxed ratio related to the solidification morphologies can also be evaluated as a function of those physical properties.

KEY WORDS: physical parameter; growth restriction; solidification; CET; equiaxed ratio; cellular automaton.

1. Introduction

Generally, the solidification morphologies of the alloys are controlled by several aspects such as the physical properties of an alloy, the nucleation parameters and the operation parameters, etc. In the previous work of the authors,1,2) the effects of the nucleation parameters (including the nucleation density, the nucleation undercooling in bulk and at surface), and the operation parameters (such as the initial concentration of an alloy, the cooling rate, the flow velocity, etc.) on the solidification morphologies are deeply analyzed. While it is observed in the experiments as well as in the simulations that although adopting the same nucleation parameters and the same operation parameters, the different solidification morphologies take on for varied alloy systems, that is, the physical properties of an alloy also play an important role on the solidification microstructure evolution. So, Tarshis et al.3) proposed the concept of constitutional supercooling parameter, $P = mC^0(k_{SL}^0 - 1)/k_{SL}^0$, and considered the effects of $P$ on the growth restriction of the dendrites. Spittle and Sadli4) explained their experimental work in terms of $P$ on the inoculated binary alloy melt within a wide range of solute types and amounts and found the grain size is related closely to parameter $P$. Corresponding to the maximum solutal undercooling, $P$ characterizes the largest growth restriction for a given concentration. Afterwards, Maxwell and Hellawell5) proposed the growth restriction parameter $Q = mC^0(k_{SL}^0 - 1)$ which takes an inversely proportional effect on the growth kinetics. Greer et al.6) interpreted Spittle and Sadli’s experimental data on grain size in terms of $Q$ and suggested that $Q$ is a better parameter than $P$ for quantifying the degree of growth restriction at a small undercooling. Hodaj and Durand7) also introduced a new growth-restriction parameter as $U = Q/D_s$.

In this paper, the evolutions of the solidification morphologies of Al–Si and Al–Cu binary alloys are modeled by the Finite Difference (FD)-Cellular Automaton (CA) model. Considering the key effects of the dendrite tip growth kinetics on the consequent solidification morphologies, the influences of the related physical properties on grain growth procedure and the Columnar to Equiaxed Transition (CET) of Al–Si and Al–Cu alloys are mainly analyzed and the empirical relations of the growth rate and the equiaxed ratio versus these parameters are deduced. The Gibbs–Thomson coefficient besides the growth restriction parameter mentioned in the above literature is proposed for its obvious effects on the growth grain.

2. Model Description

2.1. Heat Transfer

An unsteady two dimensional heat transfer combined with the solute redistribution at solid/liquid (S/L) interface8) during Al–Si and Al–Cu alloy solidification processes is considered. The temperature $T_{n+1}$, solid fraction $f_{S,n+1}$ and concentration $C_{s,n+1}$ can be determined by iterative solution
of the unsteady heat transfer, energy balance and Kobayashi’s microsegregation equations. Details can be found in Ref. 1). The obtained temperature at FD nodes is interpolated into CA cells.

2.2. Nucleation and Growth

The CA model of Gandin and Rappaz\(^9\) is adopted to predict the nucleation and grain growth during the solidification procedure. Both the locations and the crystallographic orientations of the new nucleus are chosen randomly. The heterogeneous nucleation distribution is described by two types of Gaussian distributions at the mold wall (the maximum nucleation density, \(n^*\), the maximum undercooling, \(\Delta T_{so}\), and the standard deviation, \(\Delta T_{s,ap}\), of the nucleation distribution) and in the bulk of liquid (\(n^*_v\), \(\Delta T_{v}\), \(\Delta T_{c}\)), respectively.

The grain growth is described by the decoated square growth algorithm.\(^9\) In the algorithm, each solid cell grows respectively. dendrite tip radius the effect of the thermal gradients on each side of the S/L interface, dendrite tip radius growth according to a certain rule. After all the neighbors of the solid grain become solid, it stops growth.

2.3. Dendrite Tip Growth Kinetics

For most metallic alloys, the contributions of the thermal diffusion and the kinetics on the local undercooling are small and can be omitted, so the dendrite tip growth kinetics is a function of the constitutional and curvature undercooling, which can be described by GGAN model and KGT model.

2.3.1. GGAN Model Considering Constitutional and Curvature Undercooling

Assuming the total undercooling \(\Delta T\) at the dendrite tip is the contributions of the solute diffusion, \(\Delta T_{c}\), and S/L interface curvature, \(\Delta T_{r}\), as given by Gandin et al.\(^10\)

\[
\Delta T = mC^0(1 - \frac{1}{1 - (1 - k_{SL})\Omega}) + \frac{2\Gamma}{r} \quad \text{(1)}
\]

With no fluid flow, the supersaturation is expressed as the Ivantsov function,

\[
\Omega = Iv(P_v) = P_v \exp(P_v)E_v(P_v) \quad \text{(2)}
\]

\[
P_v = \frac{rv_{tip}}{2D_v} \quad \text{(3)}
\]

Applying the marginal stability criterion\(^11\) with ignoring the effect of the thermal gradients on each side of the S/L interface, dendrite tip radius \(r\) is written as a function of growth Peclet number \(P_v\)

\[
r = \frac{\Gamma}{2\sigma^*mC^*(k_{SL}-1)P_v} \quad (\sigma^* = 1/(4\pi^2)) \quad \text{(4)}
\]

where \(C^*\) is the liquid concentration at S/L interface,

\[
C^* = \frac{C^0}{1 - (1 - k_{SL})\Omega} \quad \text{(5)}
\]

And \(v_{tip}\) is expressed as,

\[
v_{tip} = \frac{D_v mC^0(k_{SL}-1)}{\pi\Gamma} \left[ \frac{P_v}{1 - (1 - k_{SL})Iv(P_v)} \right] \quad \text{(6)}
\]

Given the local undercooling, \(\Delta T\), the dendrite tip growth rate can be obtained by iterative solution of Eqs. (1)–(5).\(^2\)

The dendrite tip growth rates predicted by the GGAN model are illustrated as the solid lines in Fig. 1.

2.3.2. KGT Model

Another dendrite tip growth kinetics model, called KGT model,\(^13\) only considers the contribution of the solute diffusion to the total undercooling. The relation between the dendrite tip growth rate and the undercooling \(\Delta T\) can also be obtained by the solution of Eqs. (1)–(5) while omitting the curvature undercooling, i.e. \(\Delta T_r = 2\Gamma/r = 0\), in Eq. (1).

2.3.3. Simplified Expressions Base on KGT Model

As shown in the above models, the dendrite tip growth kinetics is dependent on the local undercooling and some physical properties. The simplified growth kinetics expression follows this idea and expresses \(v_{tip}\) as a certain function, e.g. power or polynomial expression, of the local undercooling with keeping consistent with the dendrite tip growth kinetics model. The one adopted here writes \(v_{tip}\) as a direct proportion to the power of the local undercooling. The proportional constant, i.e. the combination of some physical properties, varies corresponding to the concentration and the kinds of alloys.

The simplified expressions based on KGT model for Al–Si alloys come from the best fitting of the KGT model. They are \(v_{tip} = 2.9 \times 10^{-6} \times (\Delta T)^2\) m/s for Al–7mass%Si and \(v_{tip} = 1.7 \times 10^{-7} \times (\Delta T)^2\) m/s for Al–3mass%Si binary alloy,\(^14\) \(v_{tip} = 5.75 \times 10^{-6} \times (\Delta T)^2\) m/s for Al–5mass%Si binary alloy according to the growth rate curves of Al–5mass%Si and Al–7mass%Si alloy in Ref. 15).

The dendrite tip growth rates predicted by the simplified expressions based on KGT model are illustrated as the dotted lines in Fig. 1. The contributions of the physical properties to the values of these coefficients are discussed in Sec. 3.2.

3. Results and Discussion

The solidification evolution of Al–Si and Al–Cu binary alloy ingot with configuration 0.03 m x 0.03 m is modeled
by CA-FD coupling model. The related physical properties for Al–3mass%Si and Al–3mass%Cu alloy are shown in Tables 1 and 2, others can be found in Ref. 1). A seen, the latent heat of the solidification is regarded as the functions of the solute concentration in order to fit with Kobayashi’s model.8)

The cooling rate in intermediate position  is set as 0.7, 2.3 and 5.0 K/s. The heat transfer coefficient  corresponding to each alloy and  is obtained beforehand through several attempts. Since this work mainly discusses the influences of the physical properties on the solidification morphologies, most of the nucleation parameters are fixed, such as , , except the nucleation undercooling in the bulk of liquid  and the nucleation density at the surface. The former one is changed for the determination of the critical range for the CET, and the latter one is determined by the primary arm spacing, i.e. , which is related to the cooling rate at the surface, as seen in Table 1.

The heat transfer equation combined with microsegregation model and energy balance is adopted to obtain the temperature field. The temperature is then interpolated into CA

### Table 1. Thermophysical properties of Al–3mass%Si and Al–3mass%Cu alloy.

<table>
<thead>
<tr>
<th></th>
<th>Al–3mass%Si</th>
<th>Al–3mass%Cu</th>
</tr>
</thead>
<tbody>
<tr>
<td>Initial solute concentration in liquid,</td>
<td>0.03</td>
<td>0.03</td>
</tr>
<tr>
<td>melting point of pure Al,</td>
<td>660.35 [14]</td>
<td>660.35 [18,19]</td>
</tr>
<tr>
<td>Initial temperature,</td>
<td>800</td>
<td>800</td>
</tr>
<tr>
<td>Primary arm spacing,</td>
<td>220×10^{-6} × (T_{surf})^{0.55} [20]</td>
<td>250×10^{-6} × (T_{surf})^{0.55} [18,19]</td>
</tr>
<tr>
<td>Secondary arm spacing,</td>
<td>9.89×10^{-6} × (T_{sg}^{0.59}) [20]</td>
<td>11.07×10^{-6} × (T_{sg}^{0.59}) [21]</td>
</tr>
<tr>
<td>Diffusion coefficient of Si in solid,</td>
<td>2.0×10^{-9} [22]</td>
<td>6.47×10^{-9} [23]</td>
</tr>
<tr>
<td>Diffusion constant in expression of</td>
<td>1.335×10^{12} [22]</td>
<td>1.349×10^{12} [23]</td>
</tr>
<tr>
<td>Specific heat,</td>
<td>2.6×10^{0} [24]</td>
<td>3.0×10^{0} [18,19,25]</td>
</tr>
<tr>
<td>Thermal conductivity,</td>
<td>3.65×10^{5} [29]</td>
<td>77 [18,19]</td>
</tr>
<tr>
<td>Thermal conductivity in liquid,</td>
<td>253−0.110 T [26]</td>
<td>153 [18,19]</td>
</tr>
<tr>
<td>Average density of liquid and solid,</td>
<td>2452.51 [14]</td>
<td>2531.55 [18,19]</td>
</tr>
<tr>
<td>Latent heat,</td>
<td>397.0−473.68C_{60}−2004.5C_{60}^{2}−272.44−37.08C_{60}−128.71C_{60}^{2}</td>
<td>272.44−37.08C_{60}−128.71C_{60}^{2}</td>
</tr>
<tr>
<td>Eutectic solute concentration in liquid,</td>
<td>0.126</td>
<td>0.332 [27]</td>
</tr>
<tr>
<td>Eutectic point,</td>
<td>577</td>
<td>548 [26,27]</td>
</tr>
</tbody>
</table>

### Table 2. Growth restriction parameters for Al–Si and Al–Cu alloy (data without citation is from Ref. 1).

<table>
<thead>
<tr>
<th></th>
<th>Al–7mass%Si</th>
<th>Al–5mass%Si</th>
<th>Al–3mass%Si</th>
<th>Al–3mass%Cu</th>
</tr>
</thead>
<tbody>
<tr>
<td>Initial solute concentration in liquid,</td>
<td>0.07</td>
<td>0.05</td>
<td>0.03</td>
<td>0.03</td>
</tr>
<tr>
<td>Slope of liquudus,</td>
<td>-650</td>
<td>-625</td>
<td>-600 [26]</td>
<td>-337 [26]</td>
</tr>
<tr>
<td>Partition coefficient,</td>
<td>0.13</td>
<td>0.125</td>
<td>0.12</td>
<td>0.17 [18,19,26,27]</td>
</tr>
<tr>
<td>Diffusion coefficient of solute in liquid,</td>
<td>6.45×10^{-9}</td>
<td>6.725×10^{-9}</td>
<td>7.0×10^{-9} [26]</td>
<td>5.0×10^{-9} [26]</td>
</tr>
<tr>
<td>Gibbs-Thomson coefficient,</td>
<td>1.96×10^{7}</td>
<td>1.96×10^{7}</td>
<td>1.96×10^{7} [26]</td>
<td>2.41×10^{-7} [25]</td>
</tr>
<tr>
<td>Dynamic viscosity,</td>
<td>138×10^{0}</td>
<td>138×10^{0}</td>
<td>138×10^{0} [10]</td>
<td>1.2×10^{0} [28]</td>
</tr>
<tr>
<td>Liquidus temperature ,</td>
<td>618</td>
<td>632</td>
<td>642.816</td>
<td>650 [26]</td>
</tr>
<tr>
<td>Schmidt number,</td>
<td>90.6</td>
<td>86.58</td>
<td>83.2</td>
<td>98.0</td>
</tr>
<tr>
<td></td>
<td>304.5</td>
<td>218.75</td>
<td>132</td>
<td>49.36</td>
</tr>
<tr>
<td></td>
<td>39.59</td>
<td>27.34</td>
<td>15.84</td>
<td>8.39</td>
</tr>
<tr>
<td></td>
<td>6.14×10^{0}</td>
<td>4.07×10^{0}</td>
<td>2.26×10^{0}</td>
<td>1.68×10^{0}</td>
</tr>
</tbody>
</table>
3.1. Physical Properties Influencing Heat Transfer

Heat transfer is the basic stage for the solidification and evolution of the morphologies. Figure 2 plots the cooling curves for Al–3mass%Si and Al–3mass%Cu alloy in the same cooling rate in intermediate position $T_\text{c}$. The curves show the same tendency that the slope of the curve becomes smaller temporarily due to the appearance of the primary phase and the eutectic phase. After the completion of the solidification, the slope returns to the original state. Here the cooling rates at the three positions (intermediate, center and surface) are defined by the ratio of the temperature decrease to the solidification time.

With different solute in Al melt, the solidification interval is different due to their different slope of the liquidus and different eutectic temperature. As illustrated in the cooling curves in Fig. 2, since the slope of the liquidus for Al–3mass%Si alloy is larger than that for Al–3mass%Cu alloy, with the same solute concentration, the liquidus temperature for Al–3mass%Si alloy is lower. And the eutectic temperature of Al–3mass%Si alloy (577°C) is higher than that of Al–3mass%Cu alloy (548°C). So the solidification temperature for Al–3mass%Si alloy is lower. And the eutectic temperature of Al–3mass%Si alloy is larger than that for Al–3mass%Cu alloy.

3.2. Physical Properties Influencing Grain Growth

It is inclined to nucleate in the undercooled melt, and then the nuclei begin to grow up with a certain growth rate during the solidification process.

There are several physical properties related to the dendrite tip growth kinetics such as the partition ratio $k_{\text{SL}}$, the slope of the liquidus $m$, the Gibbs–Thomson coefficient $\Gamma$ and the diffusion coefficient in liquid $D_L$, etc. Several works have discussed their effects on the dendrite growth and proposed the constitutional supercooling parameter, $P=mC^0(k_{\text{SL}}-1)/k_{\text{SL}}$, the growth-restriction parameter $Q=mC^0(k_{\text{SL}}-1)$, since Maxwell and Hellawell found that $v_{\text{tip}} \propto D_L/Q$. Hodaj and Durand further expressed the growth-restriction parameter as $U=Q/D_L$.7

Through Eq. (6), we have gotten the exact relations of $v_{\text{tip}}$ versus these restriction parameters and other physical properties such as Gibbs-Thomson coefficient $\Gamma$. To clearly show their relations, Fig. 3 plots the evolution of dendrite tip growth rate under the local undercooling at 1 K, 5 K and 10 K and the physical properties of Al–Cu and Al–Si alloys as listed in Tables 1 and 2. As can be seen, the growth rate $v_{\text{tip}}$ increases obviously with the decrease of $mC^0(k_{\text{SL}}-1)$, with the increase of $D_L$ and with the decrease of $\Gamma$. That is, besides the parameters $mC^0(k_{\text{SL}}-1)$ and $D_L$ as mentioned in Refs. 3), 5), 7), the Gibbs–Thomson coefficient $\Gamma$ also shows great effects on the dendrite tip growth rate. Their effects on the dendrite tip growth rate are in the order of $mC^0(k_{\text{SL}}-1) > \Gamma > D_L$.

With the concentration decrease in Al–Si alloy and with the alloy changing from Al–Si alloy to Al–Cu alloy, as seen in Table 2, it results in the decrease of $mC^0(k_{\text{SL}}-1)$, increase or decrease of $D_L$ and increase of $\Gamma$. Since $mC^0(k_{\text{SL}}-1)$ dominates the change of $v_{\text{tip}}$, the value of dendrite tip growth rate under a certain undercooling is ordered as Al–3mass%Cu>Al–3mass%Si>Al–5mass%Si>Al–7mass%Si alloy, as presented in Fig. 1.

Considering the dendrite tip growth kinetics expression by GGAN model in Eq. (6), the variance of $v_{\text{tip}}$ with the physical properties in Fig. 3, the growth restriction parameters in Refs. 3), 5), 7), as well as the equation...
which is derived from Eq. (15) in Ref. 1) in the case of \( r/\Omega \sim \propto \) (i.e. flat S/L interface), the following simplified expression based on GGAN model in the case of \( r/\Omega \) limited value is assumed,

\[
\frac{v_{tip}}{v_0} = \frac{D_L}{\Gamma} \frac{\Delta T^2}{mC\left(k^{S/L} - 1\right)\Gamma} \tag{7}
\]

Keeping \( D = 2.7 \) as adopted for Al–Si alloys in Sec. 2.3.3 and also for Al–Cu alloy, best fit to GGAN model through Levenberg–Marquardt method16) for Al–3mass%Si, Al–5mass%Si, Al–7mass%Si and Al–3mass%Cu alloy with local undercooling in the range of 1–10 K, the coefficients are obtained

\[
A = 1.03 \tag{8a}
\]

\[
B = 1.85 \tag{8b}
\]

\[
C = 0.88 \tag{8c}
\]

\[
D = 2.7 \tag{8d}
\]

The simplified expression based on GGAN model (Eq. (8)) indicates the contributions of the physical properties of the alloys to the coefficients in the simplified expressions based on KGT model. Correspondingly, the values of \( D_L/(\left[mC\left(k^{S/L} - 1\right)\Gamma\right]^{0.88}) \) were roughly equal to \( 2.9 \times 10^{-6} \text{ m/s} \) for Al–7mass%Si, \( 5.75 \times 10^{-6} \text{ m/s} \) for Al–5mass%Si, \( 1.7 \times 10^{-5} \text{ m/s} \) for Al–3mass%Si, and \( 3.35 \times 10^{-5} \text{ m/s} \) for Al–3mass%Cu alloy.

Figure 1 compares the dendrite tip growth kinetics by the GGAN model, the simplified expressions based on KGT model and the present simplified expression based on GGAN model. The deviation in the faster evolution region of the dendrite tip growth rate in Fig. 1 is just due to the simple power function style based on the GGAN model. The GGAN model considers only the contribution of mass transfer as the same as the KGT model. As illustrated, the results by the present simplified expression based on GGAN model fit well with the GGAN model where the local undercooling is less than 6 K. This range is the most common case in actual solidification process of Al alloys, since the Al alloys always include much contamination which can be regarded as the heterogeneous nucleants1) and consequently decreases the local undercooling for dendrite growth. There is some difference for Al–Cu alloy between the growth rate predicted by the present simplified expression based on GGAN model and by the GGAN model where the undercooling is greater than 6 K, although it shows a good agreement with the curves for Al–Si alloy. It indicates that for the faster evolution of the dendrite tip growth rate with the undercooling under the larger undercoolings, the present empirical expression with the unique value of power for undercooling is not enough and we might have to adopt the polynomial expression instead. However, the present simplified expression based on GGAN model is a good and quick evaluation of the growth kinetics for Al–Si and Al–Cu binary alloy and in the case of Al–Si alloy, it shows even better fitness with the curves by GGAN model than the simplified expressions based on KGT model.

### 3.3. CET and Equiaxed Ratio

The solidification morphologies for Al–Si and Al–Cu alloy under different concentration, cooling rate and nucleation undercoolings are shown in Fig. 4. The CET can be evaluated through the solidification morphologies. As seen, with faster dendrite tip growth rate, Al–Cu alloy requires a much smaller critical range of nucleation undercoolings for the CET compared with Al–Si alloy. For the case of smaller concentration in Al–Si alloy, it’s the same. Since the transition...
tion from fully columnar growth (fully C) to mixed columnar/equiaxed growth (C/H) under $T_\dot{\nu}/H_{11005}<0.7$ K/s for Al–3mass%Cu alloy and Al–3mass%Si alloy occurs at the nucleation undercooling between 3–4 K and 4–5 K, respectively, and the transition from C/H to fully equiaxed growth (fully E) occurs at the nucleation undercooling between 2–3 K, the solidification morphologies at the nucleation undercooling $D_{\dot{\nu}}/H_{11005}=3$ K are chosen as the typical case for further discussion.

The CET is also predicted by Hunt’s model for Al–3mass%Si and Al–3mass%Cu alloy with cooling rate $T_\dot{\nu}/H_{11005}=0.7, 2.3$ and 5.0 K/s and nucleation undercooling $D_{\dot{\nu}}/H_{11005}=3$ K. As shown, the equiaxed ratio decreases obviously with the increase of $PP$ (i.e. increase of $v_{tip}$) under the same cooling rate and decreases a little with the increase of the cooling rate for the same alloy ($PP$). Now, the change of solute concentration and alloy kind are summarized into the change of their physical properties. And the difference of the solidification behaviors such as the CET of the alloys is interpreted as the results of the difference of their physical properties.

Polynomial fit of this relation under $T_\dot{\nu}/H_{11005}=0.7$ K/s obtains a parabolic function as illustrated the solid line in the figure,

$$r_{\text{Equiaxed}} = 0.95 - 6.38 \times 10^3 \cdot PP - 5.79 \times 10^5 \cdot PP^2$$

**4. Conclusions**

Besides the nucleation parameters and the operation parameters, the solidification morphologies are internally affected by the thermophysical properties of the alloys. The
results through the modeling of the solidification morphologies of Al–Si alloy and Al–Cu alloy are as follows:

1. The solidification interval is determined by the decrease of the temperature and the solidification time. For Al-based binary alloys, the larger slope of the liquidus and the higher eutectic temperature result in the smaller solidification interval as the case of Al–3mass%Si alloy compared to that of Al–3mass%Cu alloy.

2. The simplified expression based on GGAN model indicates the empirical relation between them under the cooling rate $T \approx 0.7 \text{K/s}$ and nucleation undercooling $\Delta T_c = 3 \text{K}$ is approximated by a parabola, $r_{e\text{equiaxed}} = 0.95 - 6.38 \times 10^{-3} \cdot PP - 5.79 \times 10^{-2} \cdot PP^2$, $PP = D_L^{1.03} / \left( \left[ mC(k_{\text{LS}} - 1) \right]^2 T_{\text{CET}} \right)$. The effects of these parameters on the growth rate are in the order of $m(k_{\text{LS}} - 1) > T > D_L$. The Gibbs–Thomson coefficient $\Gamma$, which is firstly proposed in this work, shows greater effect on the dendrite tip growth rate than the diffusion coefficient in liquid, $D_L$.

3. The solidification morphologies and the CET are internally affected by the physical properties of alloys. The equiaxed ratio decreases clearly with the increase of the combination of the physical properties $D_L^{1.03} / \left( \left[ mC(k_{\text{LS}} - 1) \right]^2 T_{\text{CET}} \right)$.

Acknowledgements

This work is supported by 111 Project [B07015], Natural Science Foundations of Liaoning Province [20031010] and [20072033] of China, the Fundamental Research Funds for the Central Universities [N100409004] National High-tech R&D Program of China [2009AA03Z530], National Natural Science Foundation of China and Shanghai Baosteel [50834010], the Key Project of Chinese Ministry of Education [108036], SRF for ROCS, SEM [20071108-2] and NEU, Science and Technology Planning Project of Liaoning Province [2009221007].

REFERENCES