A New Equiaxed Solidification Predictor from a Model of Columnar Growth

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A new indicator of the potential for the formation of an equiaxed zone during alloy solidification is proposed. The indicator, or equiaxed index, is calculated from the predictions of a numerical model of non-equilibrium columnar solidification. This model uses a front-tracking approach to simulate the nucleation and growth of an undercooled columnar dendritic front into the liquid phase in a 2D casting process. The algorithm for the advancing front is based on expressions developed from considerations of dendrite tip growth. A comparison is made with models in which growth of individual, and competing, columnar crystals are simulated. The equiaxed index is based on numerical integration of an undercooled ravine in front of the advancing columnar front, and changes with time. This proposed metric is a predictor of the relative tendency to form an equiaxed zone. Study of the peak values confirm that equiaxed solidification is more likely in concentrated alloys, and also where the rate of heat extraction to the mould is low. This is in agreement with experimental data from the literature.

KEY WORDS: alloy solidification; columnar growth; equiaxed microstructure, front-tracking model.

1. Introduction

The grain structure in an as-cast alloy component results from the conditions in which it solidified. The importance of this macrostructure and its effects on service properties have long been appreciated. For example, in 1856, Robert Mallet reported his observations, via macrophotographs, of as-cast grain structure in cast iron cannon. He noted that “...the principal axes of the crystals will always be found arranged in lines perpendicular to the bounding planes of the mass; that is to say, in the lines of direction in which the wave of heat has passed onwards from the mass in the act of consolidation...”. These crystals formed either columnar or elongated equiaxed grains. Mallet also proposed plausible “laws” on the effects of mould material, pour temperature, and cooling rate on the grain size and type, and showed a high level of understanding of the formation of shrinkage porosity in castings. In the intervening years, significant advances in the understanding of microstructural evolution during alloy solidification have been made by generations of researchers.

It is the competition between columnar and equiaxed solidification that largely determines the grain structure in metallic alloy castings. Often both types of crystals grow: the transition from the outer columnar zone to the inner equiaxed zone occurs at the columnar-to-equiaxed transition, or CET. The origin of the equiaxed zone has been studied for many decades, and recently numerical mesoscale models of the nucleation and growth of crystals during alloy solidification have been employed to assist in the prediction of the grain structure of castings. Perhaps the main computational method used in such work has been that of Cellular Automata. In such models, the nucleation and growth of each crystal in the domain of interest is directly simulated, yielding a computed macrostructure. At smaller scales, the growth of individual dendrites has been studied, either by sharp interface tracking, or by phase field methods. It is beyond the scope of this paper to outline the attributes of these models, but such an appraisal has been completed.

The present work is based on a front-tracking model of columnar solidification. This model is unique in that it adapts the methodology previously used only for tracking interfaces at a microscopic level to the meso-scale of grain growth, which is of direct and immediate industrial interest.

2. Front Tracking Model of Columnar Solidification

A new model of alloy solidification has been developed, in which a front-tracking technique is used, on a fixed grid, to visualise the growth of dendritic solid into undercooled liquid. A detailed description of the model have previously been published, and only a summary account of its salient features is presented here.

The 2D model is of an alloy solidifying in a mould. The initial condition was of an isothermal superheated liquid. A control volume finite difference formulation of heat transfer, with latent heat source terms, was employed. The source terms were computed from the results of the front-tracking predictions of solid growth, i.e. there is full coupling between the heat transfer model and the tracking of the front. Following nucleation, solid growth was modelled...
via consideration of non-equilibrium dendrite tip kinetics. The algorithm used was based on expressions developed from analysis of dendrite tip growth where the primary spacing is relatively large, implying that most of the solute is rejected interdendritically. In this case the additional treatment of diffusion of solute through a continuous layer ahead of the growing dendritic front was not necessary. The solidification fronts were represented as boundaries joining discrete computational markers which moved, according to their temperature, in a direction normal to the front. The front could be considered as a line separating liquid from a zone of partially solid alloy. This front was not the liquidus isotherm, but rather was undercooled according to the local thermal and solidification conditions. The kinetics of dendrite tip advance were used to determine the velocity of any computational marker \( V_m \) according to its undercooling \( \Delta T_m \), as follows:

\[
|V_m| = C_1(\Delta T_m)^2 \quad \text{(1a)}
\]

where

\[
\Delta T_m = T_L - T_m \quad \text{(1b)}
\]

and

\[
C_1 = \frac{-D}{8m(1-k_p)C_0} \theta \quad \text{(1c)}
\]

The advance of marker particles into the liquid.

Fig. 1. The advance of marker particles into the liquid.

The Cellular Automata-Finite Element (CA-FE) model of Rappaz et al.\(^6\) provided information on the crystallographic orientation of grains, and so the competitive growth of columnar grains from a chill, in which only those grains with the \(\langle 100 \rangle\) direction closely aligned with the direction of heat flow survive, could be simulated. This aspect of the CA-FE model is particularly important in the DS or single crystal dendritic alloy production for aerospace applications. But for more general castings, this detail may not be of high importance—rather it is grain size and the extent of the equiaxed zone that is of interest. If, in columnar growth, one grain has a substantial misorientation from the preferential growth direction, it will not survive due to being impeded and overgrown by the neighbouring crystals with better alignment. A columnar zone consisting of three grains, the central one of which is misoriented at an angle \(\theta\) to the direction of heat flow in directional solidification, is illustrated in Fig. 2. The outlying grains, in contrast, are growing with the primary arms of their dendrites parallel to the direction of maximum thermal gradient. As such their growth velocity \( V_0 \) is the same as the velocity of the liquidus isotherm. The dendrites of the central grain extend at a higher normal velocity \( V_\theta \) where

\[
V_\theta = \frac{V_0}{\cos \theta} \quad \text{(3a)}
\]

i.e. the vertical component of the growth rate is the same for both grains. Assuming that, in general, the dendritic primary extension rate

\[
V_\text{eq} = (\Delta T)^n \quad \text{(3b)}
\]

and \(n > 1\), then the dendrite tips of the misoriented grain, which need to grow faster, do so at a higher undercooling. The tips of the central dendrites are a greater distance behind the liquidus isotherm than those of the perfectly
aligned outer grains, i.e.

\[ d_\theta > d_0 \] ..................................(3c)

Tip undercooling is a function of velocity, such that

\[ \Delta T_0 = f(V_0) \] ..................................(3d)

and

\[ \Delta T_\theta = f(V_\theta) \] ..................................(3e)

So, for a linear thermal gradient in the vertical direction

\[ \frac{d_\theta}{d_0} = \frac{\Delta T_\theta}{\Delta T_0} = \left( \frac{V_\theta}{V_0} \right)^{\frac{1}{n}} \cos \theta = f_1(\theta) \] ............(3f)

If we calculate, for \( n=2 \), the variation of \( f_1(\theta) \) with \( \theta \) (Table 1) we get a picture of the ratio of the distance of the misorientated grain behind the liquidus isotherm to that of the grain with \( \theta=0 \).

So, for crystallographic misalignment of up to 30° from the heat flow direction, the percentage difference in distance behind the liquidus isotherm from the perfectly aligned crystal is less than 7.5. And the distance \( d_0 \) itself is very small—the columnar growth undercooling predicted in this work being of the order of 1°C (see below)—relative to the scale of the casting. So, for the purposes of prediction and visualisation of generic columnar front growth, the effects of such misorientation can be safely ignored, as is done in our front-tracking model. As \( \theta \) increases beyond 30° the discrepancy becomes more marked, but here the grain selection mechanism and competition will mean that such crystals will not emerge from the outer equiaxed zone.

The fact that this model does not treat such crystallographic orientation is not, in the light of the above analysis, a serious shortcoming for the prediction of as-cast structure in many shape castings. The model presented here predicts the development of columnar grains and equiaxed grains\(^ {\dagger}\) and will in future predict the evolution of and competition between both forms of dendritic solidification.

<table>
<thead>
<tr>
<th>( \theta ) (degrees)</th>
<th>5</th>
<th>10</th>
<th>15</th>
<th>20</th>
<th>25</th>
<th>30</th>
</tr>
</thead>
<tbody>
<tr>
<td>( f_1(\theta) )</td>
<td>1.002</td>
<td>1.008</td>
<td>1.017</td>
<td>1.032</td>
<td>1.050</td>
<td>1.074</td>
</tr>
</tbody>
</table>

Table 1. Variation of relative grain distance behind liquidus isotherm with misorientation of crystal with reference to the direction of heat flow.

Table 2. Alloy thermophysical data used in the model.

<table>
<thead>
<tr>
<th>Description</th>
<th>Symbol</th>
<th>Value</th>
<th>Units</th>
</tr>
</thead>
<tbody>
<tr>
<td>Latent heat of melting</td>
<td>( L )</td>
<td>10(^0)</td>
<td>J/m(^3)</td>
</tr>
<tr>
<td>Thermal conductivity</td>
<td>( k )</td>
<td>100</td>
<td>W/(mK)</td>
</tr>
<tr>
<td>Specific heat</td>
<td>( \rho C_p )</td>
<td>3.0 \times 10(^6)</td>
<td>J/(m(^3)K)</td>
</tr>
</tbody>
</table>

3. Test Case

The model was used to study the solidification of Al–Cu binary alloys in a square mould (18×18 cm). Since only diffusive transport was considered in the model, the experiment was represented by just one quadrant—the lower left one. In the computational domain, the corner was the origin \((0,0)\) and \((x,y)\) represented the horizontal and vertical coordinates. The mould temperature was fixed at 400°C, the initial liquid alloy was at 700°C, and the columnar nucleation undercooling at the mould wall was set to 2°C. The thermophysical properties of the alloy were considered invariant, see Table 2. Calculations were performed on a 100×100 control volume (c.v.) grid. For the Al–Cu system, the constant in Eq. (1) becomes\(^ {21,14}\) in cm/(K\(^2\)s)

\[ C_1 = \frac{0.16}{C_0} \] ..................................(4)

where \( C_0 \) in this case is in wt%Cu. So, for an alloy of Al–2wt%Cu, \( C_1 = 0.08 \text{ cm/(K}^2\text{s)} \) in Eq. (1).

4. Results

The model was run for an Al–2 wt%Cu alloy, for which the liquidus and solidus temperatures are, respectively, \( T_l = 655°C \) and \( T_s = 614°C \). The alloy/mould interface heat transfer coefficient \( h = 3000 \text{ W/(m}^2\text{K)} \). Following nucleation at the mould walls, solid growth proceeded into the domain.\(^ {15} \) Figure 3 shows the position of the columnar front (from the marker particle positions) superimposed (in thick line) on the thermal contour map, at time \( t=60 \text{s} \). This is not an isotherm, although the temperature along the front does not vary appreciably. For example, the difference between the average and the maximum front temperature at 60 s is less than 0.04°C. With the average front undercooling of 0.83°C, this represents a deviation of 4.8%. When expressed in terms of the front temperature (in °C), the deviation is only 6×10\(^{-3}\)%.

So, at the scale of the casting and for the purposes of visualisation one would expect the average front temperature isotherm and the tracked front to be practically coincident. A plot of both contours at the scale
of Fig. 3 would confirm this. Even at higher resolution (Fig. 4) the two curves are not easily distinguishable. Here the line closest to the lower left corner is the front. Although not used for computational purposes, it is therefore proposed that the average front temperature isotherm be used to represent the front visually. This is adopted for practical reasons—the isotherm contour plotting software can readily shade areas of superheated liquid, undercooled bulk liquid, mushy zone and solid. For such a picture, the bounding isotherms are those of the liquidus, “front” and solidus, respectively. In Fig. 5, the white area is solid, the dark grey is the columnar mushy zone, medium grey is undercooled bulk liquid, and light grey is superheated liquid. This illustration shows these zones every 30 s during the process.

Figure 6 shows the partial temperature profile along the half-diagonal from the mould corner to its centre, at various times, for an Al–2 wt%Cu alloy with $h=3\,000\,\text{W/(m}^2\text{K)}$, in which the front position is also marked. It is very clear that the thermal gradient reduces ahead of the front, and becomes very flat towards the end of solidification.

The area of undercooled bulk liquid is clearly visible in Fig. 5, but this does not show the extent of undercooling. This is shown in the 3D wireframe picture of undercooling.
in Fig. 7. The near corner is the centre of the casting. The computational grid is shown, and the vertical elevation is the magnitude of bulk liquid undercooling. Near the casting centre is superheated liquid, behind the undercooled ridge (or ravine) is the columnar mushy zone, and solid. The ridge is travelling towards the corner, and its position is shown at 90 s. A c.v.\((i, j)\) is checked for bulk liquid undercooling \(U_b(i, j)\) if it is liquid, or less than half covered by semi-solid (as, for example, the c.v. bounded by markers \(m+2\) and \(m+3\) in Fig. 1), \(i.e.\) the nodal point should still be liquid. It can be seen that the plotted undercooling drops sharply to zero at the columnar mushy zone front. Of course in reality the liquid within this columnar zone is also undercooled, but for reasons outlined below, this is ignored in the present study.

5. New Predictor for Equiaxed Solidification

In the absence of equiaxed solidification in the undercooled liquid ahead of it, the columnar front could proceed unhindered to the centre of the casting. It is the intention of the author to directly simulate the nucleation and growth of equiaxed crystals in the bulk undercooled liquid, \(via\) a similar front-tracking approach. However, the representation of the extent of the undercooled zone as the ridge depicted in Fig. 7, has led to the proposal of a new metric for the relative tendency to form an equiaxed zone between one process and another. This metric can then be employed, for example, to tailor experimental variables to minimise (or maximise) the chances of equiaxed solidification, whilst only actually modelling columnar solidification.

Equiaxed nucleation and growth can occur under the ridge. Embryonic crystals may be advected into this area from the columnar mushy zone behind, \(via\) dendrite detachment mechanisms, or may be present due to the “big bang” effect. Alternatively, heterogeneous nucleation may occur here due to the use of grain refiners.

Nucleation events can also occur within the undercooled interdendritic liquid in the columnar mushy zone, leading to the growth of equiaxed grains behind the columnar dendrite tips. Such grains become embedded in the columnar zone, resulting in a mixed columnar and equiaxed grain structure. However, as the columnar dendrites have themselves emerged \(via\) competitive growth from the outer chill zone near the mould wall, their crystalline orientation is aligned for maximum growth rate. So the equiaxed crystals which emerge in columnar interdendritic liquid cannot overtake the columnar front to block columnar growth. Certainly this is true under the assumed zero convection conditions. The advance of the columnar front can only be stalled by a high volume fraction of equiaxed grains which have nucleated and grown ahead, in the bulk undercooled liquid. Because of this, the undercooled liquid in the columnar mushy zone is not considered here; the following analysis relates only to the formation of a fully equiaxed zone.

Within the bulk undercooled liquid, the closer any nucleation site is to the front the greater is the probability of nucleation. However, in this case the time for growth is limited as the site will be quickly enveloped by the columnar dendrites. The converse is true of sites close to the leading edge of the undercooled ridge. Indeed, Fig. 7 could be considered as a map related to the probability of equiaxed nucleation. And it is proposed here, therefore, that the volume under the ridge is an indication of the potential for equiaxed growth (preceded by nucleation). Thus is because the amount of growth depends on both undercooling and the time available for growth. The greater the area of the base of the ridge, the greater is its dwell time over any particular point, or possible nucleus, in the casting. The value of the maximum undercooling of the bulk liquid alone is not a measure of such potential—it accounts only for nucleation and, as this maximum occurs at the columnar front, there is no time for growth. The undercooled ridge of Fig. 7 is a snapshot at a particular time. If we need to follow the process with time, we can calculate this volume at regular time intervals. We can define an equiaxed index \(I_{ex}(t)\):

\[
I_{ex}(t) = \sum_{i=1}^{nrows} \sum_{j=1}^{ncols} U_b(i, j) \Delta x \Delta y |_{\Delta T = \text{const}} \quad \text{...........(5)}
\]

where \(nrows\) and \(ncols\) are the number of rows and columns of c.v.s in the computational grid, respectively, and the bulk liquid undercooling

\[
U_b(i, j) = \begin{cases} 
0 & \text{if } T(i, j) > T_L \\
T_L - T(i, j) & \text{if } T(i, j) \leq T_L \text{ and } d(i, j) < 0.5 \\
0 & \text{if } T(i, j) \leq T_L \text{ and } d(i, j) \geq 0.5 
\end{cases} \quad \text{...........(6)}
\]
where $d(i, j)$ is the fraction of any c.v. which is covered by the columnar mushy zone. So, the bulk undercooling is calculated, as described, only in undercooled liquid c.v.s—elsewhere $U_b(i, j) = 0$. For the conditions under discussion, this index is plotted in Fig. 8. The peak is of relevance (5.9 cm$^2$ K), as is the time of the peak (98.55 s). If there is to be a CET, it should occur at or before this time. Note that, at the start of solidification, the equiaxed index is very small, as the extent of the undercooled region ahead of the front is small. The index increases with time, but eventually starts to decrease as columnar solidification converges at the casting centre and the undercooled region again shrinks. This is illustrated in Fig. 5, in which the temporal advance (Figs. 5(a) and 5(b), widening (Fig. 5(c)) and disappearance (Figs. 5(d) and 5(e)) of the undercooled zone can be seen.

Equation (5) gives the volume under the undercooling ridge, such as that illustrated in Fig. 7. This graphic is produced using Surfer® 8 software from Golden Software Inc., USA. In this package the volume under this surface is calculated by numerical integration as 3.796 cm$^2$ K. The crude calculation of Eq. (5) used in the current model gives a value of 3.863 cm$^2$ K. The difference is only 1.8%, and this is quite acceptable as the index is relative.

As stated, these results are for an Al–2%Cu (wt% is inferred from here on) alloy, and the heat transfer coefficient ($h$) was 3000 W/(m$^2$ K). The effects of varying $h$ and alloy concentration $C_0$ were then investigated. A matrix of numerical experiments with three levels of $h$ (1500, 3000 and 5000 W/(m$^2$ K)) and three levels of $C_0$ (1, 2 and 4% Cu) was used. From the equilibrium phase diagram, the liquidus and solidus temperatures of these alloys are presented in Table 3, along with the value of $C_1$, from Eqs. (1) and (4). It should be noted that the alloys chosen exhibit significantly different freezing ranges.

Figure 9 shows the plots of $I_{ex}(t)$ for each case. For any particular alloy it can be seen that the peak index, and time for which levels of $I_{ex}(t)$ exceed zero, are greatest for the lowest heat transfer coefficient (h.t.c.). Also, the more concentrated alloys have higher peak values of $I_{ex}(t)$, for any given value of $h$. These peak values are given in Table 4.

The average front undercooling was calculated every 30 s for each process. The results for each alloy are sufficiently different ($h$ has a smaller effect) to enable all nine plots to be shown, and clearly distinguishable, on one graph (Fig. 10). Increasing the solute content in the alloy has a very significant effect on increasing the front undercooling. This is because most of the undercooling is solutal, an effect which is incorporated into Eq. (1). For a given alloy, raising the rate of heat extraction increases the undercooling of the

![Fig. 8. Equiaxed index as a function of time; Al–2%Cu, $h=3000$ W/m$^2$ K.](image)

<table>
<thead>
<tr>
<th>Table 3. Alloy-specific data.</th>
</tr>
</thead>
<tbody>
<tr>
<td>$C_0$ (wt.%Cu)</td>
</tr>
<tr>
<td>----------------</td>
</tr>
<tr>
<td>1.0</td>
</tr>
<tr>
<td>2.0</td>
</tr>
<tr>
<td>4.0</td>
</tr>
</tbody>
</table>

![Fig. 9. Equiaxed index as a function of time for three different alloys.](image)

<table>
<thead>
<tr>
<th>Table 4. Peak values of $I_{ex}(t)$, in cm$^2$ K, for each process.</th>
</tr>
</thead>
<tbody>
<tr>
<td>Composition (wt.%Cu)</td>
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<tr>
<td>-----------------</td>
</tr>
<tr>
<td>1.0</td>
</tr>
<tr>
<td>2.0</td>
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<tr>
<td>4.0</td>
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</table>
columnar front, but the influence is less dramatic than that on the new index.

6. Comparison of Predictions with Experiments

Use of this new metric indicates that low h.t.c.s favour equiaxed growth, and that increasing the solute content also promotes equiaxed solidification. This is in agreement with experimental evidence (see Ref. 22)), as will be described, and established theory. 23) Because the front-tracking solidification model treats diffusion and not convection, the ideal experiment would be one in which natural liquid convection is completely suppressed, for example under zero gravity conditions. Dupouy et al. 24) report on such experiments, carried out in the Spacelab microgravity laboratory. Solidification experiments on Al–4%Cu were conducted, and the resulting grain structure compared to similar experiments on earth. However, the effects of varying alloy composition and the process h.t.c. were not studied. Rather a difference in as-cast macrostructure between the samples solidified in space and on earth was recorded, attributed to the lack of convection during solidification in microgravity. For example, the sedimented equiaxed grains observed in the as-cast earth samples were not evident in the space samples. Currently a European Space Agency research project on the effects of convection and microgravity on the CET in solidification (CETSOL) is underway.

A recent experimental and theoretical study was carried out by Siqueira et al. 25) to quantitatively determine the solidification parameters which affect the CET during directional solidification (from below) of Sn–Pb alloys. In this binary system, the solute-rich interdendritic liquid has a higher density than the bulk melt, thus minimising melt convection, as the vertical thermal gradient is also stabilising. So such an experiment can be useful for validation of diffusion-only models, such as the present one under discussion. The authors had previously 26) carried out a similar study on Al–Cu alloys, which also solidify with minimum convection if chilled unidirectionally from the bottom. They carried out a series of experiments in which the alloy concentration and h.t.c. at the lower chill were varied independently. The CET was measured from the bottom of the cylindrical casting along a vertical centre section. For Pb–10%Sn, they found that the columnar length increases with h.t.c., in agreement with the new predictor. The values of h.t.c. studied were the same as those reported in Table 4. These values, determined by the authors by thermal analysis of their casting experiments, are in agreement with those calculated at University College Dublin (UCD) for solidification of aluminium alloys in permanent steel moulds. 27) using a numerical inverse technique. 28) For both Al–Cu and Pb–Sn, Siqueira et al. found that the as-cast columnar length decreased with increasing concentration of the solute. This finding is also in agreement with the new predictor. Similar bottom-cooled directional solidification experiments on Al–1%Cu and Al–10%Cu, carried out by Rerko et al. 29) showed that, in the absence of grain refiners, the dilute alloy was completely columnar, the concentrated alloy completely equiaxed. Previously, Griffiths et al. had also noted 30) that the size of the equiaxed zone increased with nominal Cu concentration in a series of Al–Cu alloy ingots solidified vertically downwards in order to promote convection in the bulk liquid. The ingot macrostructure consisted of columnar grains above a region of equiaxed grains of varying extent. Fredriksson and Olsson, in a study 31) of solidification of Fe-based alloys, also calculated that, theoretically, the length of the columnar zone increases with cooling rate. This was verified via experiments done by Ares et al. 32) on Pb–2%Sn alloys directionally solidified from below. It should be noted, however, that in this case the rejected solute is lighter than the bulk liquid and the fluid layer at the solidifying interface would be solutally unstable, but thermally stable. This relationship between columnar length and cooling rate also held true for more concentrated alloys. The fact that a higher cooling rate means that there is less time for equiaxed grains to grow is reflected both in these many experimental findings, and in the predictions of the new equiaxed index presented here.

In a series of shape casting experiments on Al–Cu alloys performed in the UCD Solidification Laboratory, increasing solute levels, up to 15% Cu (at which no columnar zone was present) was found to increase the size of the equiaxed zone. Increasing the rate of heat extraction locally (via a steel chill placed at a section of a sand casting) was found to result in a local area of columnar grains in an otherwise equiaxed macrostructure.

The undercooling at the columnar front (which is the maximum undercooling of the bulk liquid) is not as sensitive to changes in h as the new predictor. Indeed if this undercooling were used as an indicator for equiaxed solidification, the effect of increasing h would be opposite to the findings of experimental studies, some of which have been cited above. Its use as an indicator of equiaxed solidification ignores the well-documented and significant effect of thermal gradient in the bulk liquid. The information in Fig. 10 is important, however, in that some critical undercooling is required in order for equiaxed nucleation to occur at all. For example, it has been reported 33) that nucleation undercooling for certain grain-refined aluminium alloys varies from 0.2 to 1.5°C. If the nuclei are dendrite fragments,
lower growth activation undercooling values are expected. So, provided the peak undercooling in the bulk liquid is sufficient to cause nucleation, the proposed equiaxed index can be a useful predictor of the formation of an as-cast equiaxed structure.

7. Conclusions

A front-tracking model of columnar solidification has been used as a predictor of equiaxed growth. It was shown that the simplification of not treating individual columnar crystals in such a model is, for many applications, reasonable. The extent (magnitude plus area covered) of the bulk undercooled liquid zone was presented as a 3D wireframe plot: the undercooled “ridge”. It is postulated that the volume under this ridge, which varies with time \( t \), is an indication of the potential for the formation of a completely equiaxed zone, and was therefore denoted as the equiaxed index \( I_{eq}(t) \). This index grows, peaks and then drops throughout columnar solidification. Study of the peak value suggests that equiaxed solidification is more likely in alloys with a high solute content and where the rate of heat extraction to the mould is low. This is in agreement with the published experimental observations of numerous research groups. The level of solute in the alloy was found to have a significant effect on the front undercooling, with more pure alloys growing at higher temperature. Increasing the h.t.c. was shown to lower the front temperature, but only very slightly. Simply using the columnar front undercooling as an indicator of the tendency for equiaxed solidification indicates that pure alloys solidifying with low rates of heat extraction would be less likely to form an equiaxed zone, but this metric is not as sensitive to alterations in the heat transfer coefficient as is \( I_{eq}(t) \). The new indicator provides an improved criterion on which to base predictions of equiaxed zone formation.

It should be noted that the best way to get information on the formation of the equiaxed zone is to actually model the nucleation and unconstrained growth of solid, and it is planned to carry out this work in future. However, the presented technique is a method of ranking processes with regard to their equiaxed potential, without having to consider individual nucleation events, growth and impingement. It is true that the effects of interdendritic and bulk liquid flow may be significant, and work is currently under way to take account of this. Consideration of eutectic solidification is also needed.

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