Grain Refinement during Melt-spinning of Dilute Cu-base and Ni-base Alloys

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(Received on December 17, 1996; accepted in final form on March 7, 1997)

In order to obtain a fine and equiaxed grain structure in a casting, grain formation has been examined in past work in terms of undercooling of the melt, promotion of heterogeneous nucleation and remelting or fragmentation of dendrites. Some techniques to obtain grain refinement are commercially used, but the mechanism is not fully understood. The addition of a second element to a pure metal is one of the possible ways to obtain grain refinement in bulk samples. In the present work, this approach was taken with a melt-spinning method, which allows for an extremely directional and rapid heat extraction.

Dilute Cu-base and Ni-base alloys with selected solute additions were studied to examine the operating mechanism of the grain refinement in melt-spun alloys. In melt-spun Cu-base and Ni-base binary alloys, it was found that the type of solidification structure (columnar or equiaxed) can be predicted by the constitutional undercooling parameter, P , in most alloy systems as well as the case of bulk samples. There were some cases in which the melt-spun grain structure was not consistent with the P parameter prediction. In all cases, however, the final solidification temperatures of the alloys appear to be correlated to the development of grain refinement. As a result, to explain all the results in the present work, a model has been proposed which considers the effects of the parameter P , dendrite fragmentation and final solidification temperature concurrently and offers useful guidance in alloy design for optimum grain refinement.

KEY WORDS: melt-spinning; grain refinement; phase diagram; dendrite fragmentation.

1. Introduction

It is important to control the microstructure of castings because of the direct relationship between microstructure and the properties of materials. Generally, a fine and equiaxed microstructure in cast alloys leads to improved strength, ductility and formability. As a result continuing efforts have been directed to obtain such grain refinement and to identify grain formation mechanisms.

In general, the methods which have been developed to obtain grain refinement may be classified into intrinsic and extrinsic types. Grain refinement induced by a large amount of undercooling,¹⁾ *i.e.* by suppressing the onset of heterogeneous nucleation, belongs to the intrinsic type. The introduction of inoculant particles to obtain grain refinement by promoting heterogeneous nucleation²⁾ is an extrinsic method. Other methods such as the addition of vibration to the mold³⁾ may be either intrinsic if natural nuclei are activated or extrinsic if artificial nuclei are introduced but the effects of vibration are usually not defined clearly. Some methods are commercially used, but the mechanisms of grain refinement are not fully understood.

It is beyond the scope of the present work to review all of the past work in this area. However, from the

existing literature, a number of solidification processing variables that appear to influence grain refinement may be identified including melt superheat, alloy composition, cooling rate, melt undercooling, mold character (*e.g.* mold material and surface roughness), fluid flow, vibration and inoculation practice.

In terms of alloy composition which is the focus of the current work, Tarshis *et al.*⁴⁾ have shown that grain refinement tendency in various Ni-base and Al-base alloy ingots can be explained very well by the constitutional undercooling parameter, P , which is described as:

$$P = -\frac{m_L(1-k)C_0}{k} \dots\dots\dots (1)$$

where m_L is the liquidus slope, k is the solute partition ratio, and C_0 is the nominal solute concentration. On the other hand, Southin *et al.*⁵⁾ have reported that the solute partition ratio k is more reliable than P in predicting the grain refinement in some Cu-base alloy ingots. These studies suggest that the mechanism of grain refinement caused by the addition of an alloying solute element is strongly related to parameters determined by phase diagrams.

Recently, associated with the development of rapid solidification processing (RSP), the manipulation of the cast structure of rapidly solidified alloys has become

important. Melt-spinning is one of the RSP methods. Highly directional heat extraction from the melt is a typical feature of melt-spinning. Metallic glasses or metastable phases are commonly produced by melt-spinning, therefore the studies on melt-spun alloys have focused upon novel structures. However, in terms of grain formation of equilibrium phases in melt-spun alloys, not much work has been reported aside from the common observation of a columnar grain morphology in single phase alloys.

In some recent works, however, a large undercooling in the melt in contact with the substrate has been reported,⁶⁾ and the influence of solute concentration on grain structure has been shown to be significant.^{7,8)} In bulk samples such a large undercooling and the role of solute concentration are important to induce grain refinement. Therefore, it seems quite probable to create conditions for grain refinement during melt-spinning.

In the present work a systematic study of grain structure development in dilute Cu-base and Ni-base alloys was conducted during melt-spinning. The specific solute additions were considered in terms of the existing models of grain refinement. As a result solute additions were examined which yielded large variations in the P and k parameters to allow for a determination of the controlling mechanism of grain structure development.

2. Experimental Procedure

As parent metals, Cu with the purity of 99.999 mass% and Ni with the purity of 99.99 mass% were used.

A free jet melt-spinner was used in the present work with the operational specifications shown in **Table 1**. The chamber was evacuated down to 40 Pa, and then backfilled with Ar. After the sample melted completely, Ar pressure was added through the tube connected to the top of the crucible. The melt ejected from the crucible through the hole solidified rapidly on the wheel and in most cases formed ribbons. Most experiments were conducted with a tangential wheel speed of 23 m/s. In addition, Cu-0.1mass%Bi alloy was melt-spun with various wheel speeds to examine the effect of wheel speed on melt-spun grain structures. Also, Cu-1mass%Pb alloy was cast on the Cu wheel without any rotation to make an ingot and the microstructure was compared with melt-spun ribbons.

During a melt-spinning run, powder of the alloy was often produced in addition to ribbons. The microstructure of such powder was examined for the analysis of solidification phenomena.

Grain size of melt-spun ribbons was measured manually from photographs of ribbon cross sections for each alloy system with respect to length (vertical to the substrate) and width (parallel to the substrate). The

maximum values were taken for each direction. In case that a columnar grain represents the ribbon thickness, the thickness was taken as the length. The figures of longitudinal cross sections of melt-spun ribbons were used for the measurement. For each alloy system, 50–80 grains in the regions with a representative grain structure were measured for grain structure determination.

3. Experimental Results

The binary alloy systems examined in the present work are listed in **Table 2**. In Table 2, the final solidification temperature is defined by the lowest liquidus temperature in a phase diagram. Thicknesses of ribbons melt-spun with a tangential wheel speed of 23 m/s were generally 40–70 μm and the widths were about 5 mm. The pure metal melt-spun ribbons were continuous with a coarse structure as shown in **Fig. 1** for Cu where the length of a grain often represents the thickness of the ribbon.

Figure 2 shows the microstructure of a melt-spun Cu-0.5mass%Pb ribbon. By the addition of Pb, the as-cast grain structure becomes finer and equiaxed. According to the back-scattered electron image shown in **Fig. 3**, the lead-rich phase exists mainly at grain boundaries, but it is sometimes found also within grains.

By the addition of Pb, melt-spun ribbons become less continuous. In the melt-spinning of Cu-1mass%Pb, the longest length of the ribbons obtained was as short as 20 mm. Clearly, the ribbon length is related to the accumulation of the lead-rich phase at grain boundaries.

Table 2. Alloy systems melt-spun in the present work.

Alloy system	P (at 1mass%)	k	Final solidification temperature (°C)
Cu-Pb	∞	≈ 0	326
Cu-Bi	∞	≈ 0	271
Cu-B	1 411	0.02	1 013
Cu-Zn	4	0.87	(902)
Cu-Mn	3	0.75	871
Cu-Ag	13	0.24	780
Ni-Pb	11	0.25	324
Ni-Bi	∞	≈ 0	270
Ni-Ag	4	0.46	960

Fig. 1. Microstructure of melt-spun pure Cu ribbon.

Table 1. Specification of melt-spinner.

Type	Free jet melt-spinner
Substrate	Cu
Wheel diameter	250 mm
Tangential wheel speed	23 m/s (max)
Atmosphere	Ar

Fig. 2. Microstructure of melt-spun Cu-0.5mass%Pb ribbon.

Fig. 3. Backscattered electron image of melt-spun Cu-1mass%Pb ribbon.

If the lead-rich phase freezes after the ribbon has left the wheel, long continuous ribbon production would be difficult.

Unlike Pb, Zn has a large solubility in solid Cu and a large k value. Therefore, according to the constitutional undercooling parameter P , the cast structure should be coarse. Figure 4 shows the microstructure of a melt-spun Cu-1mass%Zn ribbon with a fully columnar grain structure. The width of these grains is about 10 μm , which is comparable to the grain size in Cu-Pb alloys. Moreover, many of the columnar grains appear to extend completely across the thickness of the ribbon. Figure 5 shows the transition of average grain size of melt-spun Cu-base alloy ribbons with the addition of Pb or Zn. The grain refinement effect induced by Pb is obvious.

The experimental results are summarized in Figs. 6–8 where the potential grain refinement parameters P , k , and final solidification temperatures (T_f) are related to the observed grain structure. The grain refinement can be predicted by P or k for most alloy systems. In general, the trends indicate that high P , low k , and low T_f favor a fine equiaxed grain structure. However, there are some

Fig. 4. Microstructure of melt-spun Cu-1mass%Zn ribbon.

Fig. 5. Transition of grain size of melt-spun Cu-Pb and Cu-Zn ribbons.

 Fig. 6. Relation of constitutional undercooling parameter P and grain structure of melt-spun ribbons.

Fig. 7. Relation of solute partition ratio k and grain structure of melt-spun ribbons.

Fig. 8. Relation of final solidification temperature T_f and grain structure of melt-spun ribbons.

exceptions. For example, the results for Ni-Ag and Ni-Pb alloys are not perfectly explained by P . Also, the grain structure in Cu-B alloys appears to contradict the overall trend. Only T_f is totally consistent with the trend in grain morphology, therefore it should be taken into account when the mechanism of grain refinement is examined.

4. Discussion

4.1. Melt-spinning Process and Grain Structure

One of the advantages of solidification processing by melt-spinning is the strong directionality of heat flow which together with the positive temperature gradient tends to promote a columnar grain structure in alloys. The development of an equiaxed grain structure requires an alloy solidification response that can overcome the directional heat flow effect.

At the same time there have been numerous reports of metastable phases and amorphous structures produced by melt-spinning which clearly demonstrate that the melt

Fig. 9. Constitutional undercooling in front of solid/liquid interface.

can develop substantial thermal undercooling in contact with the wheel. While some level of thermal undercooling may be expected in most (if not all) cases of melt-spinning, the actual measurement of the amount of undercooling is difficult to perform in a reliable manner. In the present experiments the grain structures were produced under conditions of constant melt superheat, uniform wheel speed and only small changes in alloy composition so that the structures may be compared under constant thermal undercooling values. In fact, most analyses of grain structure development in single phase alloys (including those with minor amounts of second phase) consider the observations in terms of constitutional undercooling and dendrite fragmentation. For melt-spun ribbon these considerations are also appropriate for the understanding of grain structure formation.

4.2. Calculation of Constitutional Undercooling

The work by Tarshis *et al.*⁴⁾ has shown that the constitutional undercooling parameter P explains grain refinement effects very well in the case of Ni-base and Al-base bulk samples. One possible interpretation of their experimental results is that constitutionally undercooled liquid ahead of the solid/liquid interface promotes the nucleation and growth of equiaxed grains.⁹⁾ The concept of constitutional undercooling is schematically shown in Fig. 9.

When the solid/liquid interface is assumed to be planar, the effect of solidification conditions on constitutional undercooling can be estimated by simple calculations. Assuming a single phase binary system with the liquidus slope (m_L) of -10 K/%, which is a representative value for Cu-base alloys tested in the present work, the liquid composition ahead of the solid/liquid interface, C_L , is given by:¹⁰⁾

$$C_L = C_0 \left(1 + \frac{1-k}{k} e^{-(V/D_L)x'} \right) \dots \dots \dots (2)$$

where V is solidification rate, D_L is diffusion coefficient in the liquid, and x' is the distance from the solid/liquid interface. Then, equilibrium liquidus temperature, T_L , can be obtained immediately by:

$$T_L = T_m + m_L \cdot C_L \dots \dots \dots (3)$$

These values were calculated for different k 's. Other parameters used in the calculation are shown in Table 3.

If a value of 10^3 K/mm for G_L , the temperature gradient

Table 3. Parameters used in calculation of constitutional undercooling and dendrite tip radius.

Parameters		Used values
Melting point at $C_0=0$	T_m	1 356 K
Liquidus slope	m_L	-10 K/%
Diffusion coefficient in liquid	D_L	$2.0 \times 10^{-3} \text{ mm}^2/\text{s}$
Temperature gradient in liquid	G_L	1 000 K/mm
Gibbs-Thomson coefficient	Γ	$1.5 \times 10^{-4} \text{ K mm}$
Solidification rate	V	10, 100 mm/s

Fig. 10. Distribution of ΔT_{CU} when V is 10 mm/s. (a) $C_0 = 0.1 \text{ mass\%}$ (b) $C_0 = 1 \text{ mass\%}$

in the liquid, is applied to one of the calculated results above, the amount of constitutional undercooling (ΔT_{CU}) can be defined by the difference of the equilibrium liquidus temperature (T_L) and the actual temperature. **Figures 10 and 11** show the plot of ΔT_{UC} with the distance from the solid/liquid interface. A smaller partition ratio k ($0 < k < 1$) or higher solute concentration promotes an increase in the constitutional undercooling. The tendency of grain refinement observed in the present work is almost consistent with these calculated results. In melt-spinning it is not adequate to assume that the interface is planar, but the possibility of heterogeneous nucleation in the melt puddle should be taken into account when the mechanism of grain refinement is examined.

4.3. Dendrite Fragmentation

Another possible interpretation of the work by Tarshis *et al.*⁴⁾ is that the fragmentation of dendrites is the mechanism for the grain refinement effect.

The fact that the dendrite tip radius tends to become smaller with an increase in solute concentration may support this hypothesis. The dendrite tip radius is assumed to be correlated to the size of equiaxed grains if such grains are produced by the fragmentation of

Fig. 11. Distribution of ΔT_{CU} when V is 100 mm/s. (a) $C_0 = 0.1 \text{ mass\%}$ (b) $C_0 = 1 \text{ mass\%}$

Fig. 12. Calculated dendrite tip radius for various k 's. (a) $V = 10 \text{ mm/s}$ (b) $V = 100 \text{ mm/s}$

dendrites. For dendritic solidification, the dendrite tip radius (R_s) can be determined by:¹¹⁾

$$R_s = 2\pi(\delta_c \cdot s)^{1/2} \approx 2\pi\{(D_L/V)(\Gamma/P)\}^{1/2} \dots\dots\dots(4)$$

where δ_c is the geometric mean of a diffusion length, s is a capillary length, Γ is the Gibbs-Thomson coefficient. As noted by Kurz and Fisher¹¹⁾ Eq. (4) applies when $(1-k)P_c < 1$ where $P_c = R_s V / 2D_L$ is the solutal Peclet number. However, Kurz *et al.*¹²⁾ have also noted that

Eq. (4) is a reasonable approximation at high P_c values for the range of interface velocities considered in the current work. **Figure 12** shows the result of the calculation of dendrite tip radius based on Eq. (4) and given the solidification conditions listed in Table 3. In the present work, a higher alloy concentration ends up with grains with reduced width in any alloy system, which is consistent with the tendency shown by the calculation.

4.4. Dendrite Size in Small Ingots

There is another experimental result which supports the hypothesis of the fragmentation of dendrites. **Figure 13(a)** shows the microstructure of the Cu-1mass%Pb ingot which was cast on the Cu wheel without any rotation. The thickness of the ingot was about 5 mm. In this case, there may not be significant fragmentation of dendrites, therefore some columnar structures were expected to be found near the chill surface. In fact, the grains of this ingot were very coarse compared to those of melt-spun ribbons, but the network of Pb-rich phase, which represents the structure at solidification, can be seen inside grains. The width of columnar grains seen in Fig. 13(a), which is about $10\ \mu\text{m}$, is consistent with the grain size of melt-spun Cu-1mass%Pb ribbons (**Fig. 13(b)**).

4.5. Dendrite Remelting in Powder Samples

The transition of microstructure from dendritic to equiaxed in a droplet sample has been reported in some papers. For example, Cochrane *et al.*¹³⁾ have shown such a transitional structure in a Fe-35%Ni sample which

was produced by drop-tube processing. In melt-spun ribbons, it is difficult to recognize such a clear transition, although a mixture of columnar and equiaxed grains is observed sometimes. However, in powders which are produced during a melt-spinning run, sometimes transitional structures (columnar to equiaxed) are observed (**Fig. 14**). Such powders experienced neither a strong directional heat extraction nor a large scale convection, but still the transition can occur with a dilute solute concentration.

4.6. Effect of Fluid Flow

When the fragmentation of dendrites is examined as the mechanism of grain refinement, fluid flow in the melt should be taken into account. For example, an equiaxed microstructure with no sign of dendrites (*i.e.* no directionality) shown in Fig. 13(b) can be obtained with the aid of some kind of fluid flow which breaks up dendrites. Unlike the case of bulk samples, there is little space for natural convection to develop in the melt-spinning process. Therefore, it is not adequate to consider that natural convection causes the fragmentation of dendrites. However, it should be noted that there exists strong fluid flow in the melt pool where ribbons are formed from the liquid metal. Takeshita *et al.*¹⁴⁾ have analyzed the melt flow within the puddle formed by the single roller chill block casting technique. According to their calculated results, there is no melt flow due to natural convection, but a drastic change of flow direction due to the impingement of the melt onto the substrate. It is understood that there exists a shear stress along the

Fig. 13. Cast structure of Cu-1mass%Pb alloy.
(a) 5 mm thick ingot (b) Melt-spun ribbon

Fig. 14. Interior section of Cu-0.1mass%Bi powders produced during a melt-spinning run.

solid/liquid interface. It seems reasonable to consider that such a shear stress can promote the fragmentation of dendrites or even the separation of crystals from the substrate, which is the mechanism proposed by Ohno¹⁵⁾ for grain refinement in bulk samples. Regarding the fact that even with such a shear stress grain refinement does not occur in pure metals and some dilute alloys in the present work, it may be necessary to consider also the role of the final solidification temperature which will be discussed later.

4.7. Wheel Speed and Microstructure

Some melt-spinning runs with lower wheel speeds were carried out to investigate the grain structure in thicker ribbons. As shown in Fig. 15, the grains in melt-spun Cu-0.1mass%Bi ribbons become longer toward the free surface with a decrease in wheel speed. From the position of the fragmentation theory, it may be explained that the fragmentation of dendrites is not promoted in this case because melt flow in the puddle is milder with a lower wheel speed and reduced solidification rate. For the heterogeneous nucleation theory, however, it is difficult to explain this trend. As the thickness of a melt-spun ribbon increases, there will be more chance of heterogeneous nucleation in the melt puddle and equiaxed grains will be formed. Such equiaxed grains ahead of the solid/liquid interface will block the growth of columnar grains rather than promote them, as suggested by Jang and Hellawell.¹⁶⁾

4.8. Final Solidification Temperature

As mentioned above, the parameter P can not explain all the results because there is the exception of the results in the Cu-B system. However, the observations of low final solidification temperatures (T_f) are totally consistent with the trend of grain refinement. It is worthwhile to consider the significance of this observation.

Comparing the phase diagrams, the T_f for Ni-Ag (960°C) or Cu-B (1 013°C) is higher than that of Cu-Pb (326°C) or Ni-Pb (324°C). If the melt which remained in the interdendritic region solidifies before the fragmentation of dendrites occurs, grain refinement will not

be observed. In other words, if T_f is high, there is more chance for dendrites to survive without fragmentation. On the other hand, heterogeneous nucleation will not be affected much by the difference in T_f .

4.9. Grain Refinement Model

It is reported that at very high solidification rates the solute partition ratio becomes unity out of equilibrium.¹⁷⁾ It suggests that the difference in the parameters k and P , which are the values under the equilibrium state, should not strongly affect the grain structures of melt-spun alloys. Despite that, from the results of the present systematic study of grain refinement in melt-spun ribbon, it is evident that the parameters k and P seem to have certain significance and play a role in the development of fine, equiaxed grains as well as T_f . Thus all three parameters appear to be coordinated in achieving optimum grain refinement and should be considered together.

A recent model for grain refinement in thermally undercooled melts by Schwarz *et al.*¹⁸⁾ provides a means of treating the combined effect of P , k and T_f . In this model dendrite breakup is due to a morphological instability driven by the solid/liquid interface energy and described by a characteristic time, Δt_{BU} , which depends on undercooling. The actual breakup process is treated as developing during the time when the liquid-solid mixture is solidifying under external heat flow control (*i.e.* post-recalcence). When the time for breakup is shorter than the freezing time for the mixture a grain refined equiaxed structure develops with a size scale related to the dendrite size. Under solute diffusion control, the model yields an estimate for the breakup time as;

$$\Delta t_{BU}(\Delta T) \approx \frac{3}{2} \frac{R(\Delta T)^3}{d_0 D_L} \left| \frac{m_L C_0 (1-k)}{\Delta H_f / C_p} \right| \dots \dots \dots (5)$$

where $R(\Delta T)$ is the dendrite trunk radius which is proportional to the tip radius, *i.e.* $R(\Delta T) = Z R_s$ with $Z = \text{constant}$, and $d_0 = \Gamma C_p / \Delta H_f$, with C_p , the specific heat and ΔH_f the heat of fusion. This relation can be expressed in terms of the parameter P by substitution of Eqs. (1) and (4) to yield;

$$\Delta t_{BU}(\Delta T) \approx 12 Z^3 \pi^3 k \left(\frac{D \Gamma}{P V^3} \right)^{1/2} \dots \dots \dots (6)$$

It is now evident that a short $\Delta t_{BU}(\Delta T)$ is strongly favored by a high value of P and V and a low value of k . Moreover, for a given value of $\Delta t_{BU}(\Delta T)$ a low T_f will increase the overall freezing time which will promote grain refinement. This model has been shown to provide a good accounting for the quantitative evaluation of grain refinement in undercooled Cu-Ni alloys.

The application of the model to melt-spun ribbon allows for a coordination of all of the variables observed to influence grain refinement. For example, the plots in Figs. 16 and 17 for solidification rates of 10 and 100 mm/s reveal a consistent grouping for systems that exhibit grain refinement and an equiaxed grain morphology. The systems with grain refinement during melt-spinning also exhibit low values of $\Delta t_{BU}(\Delta T)$ which is consistent

Fig. 15. Transition of grain size of melt-spun Cu-0.1mass%Bi ribbons due to wheel speed.

Fig. 16. Relation of time for dendrite break up and grain size of melt-spun ribbons; $V = 10$ mm/s.

Fig. 17. Relation of time for dendrite break up and grain size of melt-spun ribbons; $V = 100$ mm/s.

with the dendrite breakup mechanism.¹⁸⁾ At the same time, it can be noted that the Cu-B system has a low value of $\Delta t_{BU}(\Delta T)$, but exhibits a columnar grain structure. Similarly, the Cu-Ag alloy has a $\Delta t_{BU}(\Delta T)$ value close to that for Ni-Pb, but Cu-Ag develops a columnar grain structure while Ni-Pb reveals an equiaxed grain structure.

The complete understanding of the behavior in Figs. 16 and 17 also requires a consideration of the final solidification temperature for the systems. Both Cu-B and Cu-Ag have higher T_f values than Ni-Pb and also smaller freezing ranges, $\Delta T_e = T_L - T_f$ where T_L is the initial liquidus temperature. In fact, the effective cooling rate to bypass the time needed for dendrite breakup, which can be estimated as $T_B \approx T_e / \Delta t_{BU}(\Delta T)$, is about 8×10^4 K/s for Cu-B and 3×10^3 for Cu-Ag at $V = 100$ mm/s. These cooling rates are exceeded at high T_f values during melt-spinning. For Ni-Pb with a low T_f value it is difficult to estimate the cooling rate at T_f during melt-spinning since the ribbon probably reaches T_f after

it has lost contact with the wheel. In any case this comparison again emphasizes the importance of a low T_f in allowing for the operation of dendrite breakup and effective grain refinement. Furthermore, the relationship expressed in Eq. (6) offers useful guidance in alloy design for optimizing grain refinement. The main parameters influencing $\Delta t_{BU}(\Delta T)$ may be judged from available phase diagram information which can be used in the selection of grain-refining alloying additions.

5. Summary

Several kinds of Cu-base and Ni-base alloys were melt-spun to investigate grain refinement effects. It was found that the addition of a small amount of Pb or Bi produces a very fine and equiaxed grain structure. On the other hand, the addition of Zn or Mn to Cu did not cause such grain refinement.

To explain the grain refinement observed in the present work, heterogeneous nucleation in the melt puddle and the fragmentation of dendrites were examined as the possible mechanism. There exist some situations in which the heterogeneous nucleation theory can not explain the observed results and only the fragmentation theory can be applied. In all cases, however, the final solidification temperature appears to be correlated with the operation of grain refinement. In order to incorporate the concurrent operation of all of the factors observed to influence grain refinement, a model for the morphological instability of dendrites has been applied to develop a relation for the time to cause dendrite breakup. In accordance with the observed behavior, the model indicates that large P values, rapid solidification rate and small k values favor fine dendrites and short breakup times that develop within the overall freezing time which is determined by the final solidification temperature. With the model a comprehensive accounting of grain refinement is developed which also offers guidance in alloy and process design for the most effective refinement.

Acknowledgments

The support of a graduate fellowship from Nippon Metal Industry Co., Ltd. and the U.S. Army Research Office (DAAH04-93-G-0296) are gratefully acknowledged.

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