

The Morphology of Gradient and Growing Rate on the Solidification of TiAl Intermetallic Compounds

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Abstract

In the metallic solidification the morphology of gradient and growing rate has been narrated. According to the critical velocity $V_{cr}=G \times C^{-1}$ the solidification velocity will be attained, here G is gradient and C is the solute concentration. With increasing gradient the velocity will increase and with increasing C the velocity will decrease. Meantime according to $\Delta L_s = \Delta T \times C^{-1}$ the solidification length can be presented, ΔT is the temperature variation. On the other hand, the homogeneous nucleation and heterogeneous one dynamics has been narrated as well. As for the former the dynamics will describe as $\Delta G_{hom} = 4/3 \times \pi r^3 \Delta G_v / V_s + 4\pi r^2 \sigma_{is}$ while the later will be as $\Delta G_{het} = [\Delta G_{vol} + G_{sur}] \times \text{Func}(\cos\theta)$. We shall consider the former firstly and then the later.

Keywords: Metallic solidification; Gradient; Rate; Morphology; Growth; Solute concentration

Introduction

The solidified morphology plays an important role in metallic solidification course. Therein, the research on it will be dominant priority one. In the metal solidification the growth rate will dominate the main performance according to gradient and solute concentration. It can wield function in directional solidification even single crystals particularly. In this paper the detailed narrative has been proceeded about it in order to search for the intrinsic relationships between them. Only if the one is clarified further the controlling will become feasible through its gradient and solute concentration etc. The theoretical dynamics of generating nucleation can control the nucleation generation in the initial stage particularly, so many researches are proceeding on it now. More research for this will proceed further study on the theoretical dynamics to form the complete solidification specimen. Furthermore the homogeneous nucleation and heterogeneous one has been narrated with their generating

dynamics to judge the difference between them. In general, the former has been considered firstly then the later has been done [1]. In short, the solidification morphology of gradient and growing rate has been studied in order to catch up this blank. Looking for the advantage and disadvantage to distinguish from their dynamics behaviors is the main destination in this study. The solidified length variation is considered with temperature as well. The directional nucleated solidification has been narrated as well like TiAl intermetallic compounds on the basis of Hunter equation [2].

Discussions

Gradient and growing rate

It should be discussed that how to promote thermal dynamics on super cooling so that rapid solidified alloys were formed in solidification. The cooling is difference between the nucleation beginning virtual temperature T_v and liquid line temperature T_l in equilibrium [3-4]. The cooling gradient will depend on initial

viscosity, it promoted velocity and energy difference, which would depend on temperature between cooling liquid and solid, besides interface energy G_I , volume and density, non-homogeneous and cooling velocity. [5] Actually the factors to obtain super cooling will reduce impurities as possible when the pure liquid alloys were cooled to below liquid line. Turnbull who was theory founder indicated if we eliminated pure liquid metal and nucleation factor for example intrinsic contamination and contacted among melt and vessel walls. The melted nucleation should be fully derived from energy and construction fluctuating. The crystalline behaviour will transfer from non-homogeneous nucleation to homogeneous, which make to increase cooling meantime limiting cooling is 0.2 melting point. The crystalline nucleation will still be formed as non-spontaneous type i.e. mainly that of melting body and vessel. However limited nucleation has obtained the higher cooling gradient in different alloys. Have made four deduced sections on alloys, which included thermal dynamics cooling ΔT_1 , constitutional cooling and curvature of dendrite growing tip to cause Gibbs-thomson cooling and melting dynamical cooling ΔT_4 . Here $\Delta T = \Delta T_1 + \dots + \Delta T_4$. In terms of the formula it has shown that rapid solidification may increase dynamics cooling so that it needs to be increased melt gross cooling. It has important effects as, 1) in low speed range, the change of cell to dendrite happened to while dendrite to cell was evolved. 2) At enough high gradient the G limits branch no more to change on inter cell and dendrite morphology. 3) These changes to the low speed side V_{cr} has taken a reliant relation in $G \cdot C^{-1}$ (gradient G & solute concentration C). In the meantime V_{cr} was reduced with the high G and the low C . In the low speed range, the rate v raises to make the plane interface unstable while in the high rate v it will make it absolutely stable evolving. That is on account of increasing v to promote constitutional cooling effect. Basis conditions are mono thermal flow i.e. a dimensional construction. However due to complexed thermal conduct the ideal types are very difficult. Through the solidified range the high forming defect probabilities are raised while solidified length $\Delta L_s = \Delta T \cdot C^{-1}$, so the high G is a main matter in the ΔS (directional solidification). T is temperature. The thermal dynamic of super α cooling has avoided or eliminate heterogeneous nucleation role, promote G_{cr} and hold back homogeneous nucleates to make alloys or metal difficult to arrive cooling on the general status. Super cooling method had changed thermal dynamic to obtain high cooling.

Herlach theory etc.

Herlach had demonstrated that that super cooling melt, rapid cooling, liquid alloys and metal has same mechanism being rapid solidification. The solute at the S/Liquid interface is distributed, in the local zone of the secondary dendrite arm spacing with

0.01~0.1mm by diffusion or convection [6]. It showed that the effect of coarsening can be accounted for a conventional segregation model on a back-difficult term. That results in a net diffusion process. It was known that they are for ideal state in terms of $\Delta G = 4/3 \times \pi r^3 DG_v / V_s + 4\pi r^2 \sigma_{ls}$. Here ΔG is the change of free energy. R is critical of crystal. $DG\#$ is forming work for homogeneous nucleation. The course from liquid to solid has become the same to that from liquid to growth grain through nuclear. The $r\#$ and $DG\#$ are critical radius and forming energy respectively. It was from equation of $\Delta G_{het} = [\Delta G_{vol} + G_{sur}] \times \text{Func}(\cos\theta) = 1/4 \times (2 - 3\cos\theta + \cos^3\theta) \Delta G_{hom}$. As for heterogeneous nucleation to be know, if the ΔG_{het} to be the general change of free energy was lower than the ΔG_{hom} which is easy to happen. Here q is the contact angle to be zero and no need to cool. With increasing q the cooling raised. The nuclear of heterogeneities had been formed generally through particles and surface to be distributive in liquid. They actually made much nucleation in the condition of cooling to be much smaller than the homogeneity. The embryos were nucleated through interface [7,8]. It would reach the critical nucleated radius if the volume was smaller than the homogeneity. The cooling to compare with the homogeneity had high cooling. It was possible that the single crystals should be formed with small grains of metal when temperature passed the solidified point. The crystalline had been started at boundaries through a certain time if the interface temperature would be decreased below eutectic temperature. The final structures had columnar with relation to non-direction in surface [9]. The dislocation would promote nucleation. The resolved dislocations will be good to form the nucleation in condition for the ordered HCP due to the low interface energy.

Hunt's analytical model for DS

In this study it is thought to be a positive liquid thermal gradient. The model parameters and formula are used as follow. Comparing with Hunt's analytical model, a Model presently may be written as $G_I = 0.67n0.5\Delta T_t [1 - (\Delta T / \Delta T_t)^3]$. Here, T_t is undercooling of the dendrite tips relative to the liquid temperature of alloy composition. N is the density of equated grains per unit area. A numerical simulation of TiAl has involved in the calculation dendrite volume and surface. The liquid temperature is a head of the growth front in liquid. When total under cooling is in a cell $T(y) > DT_n$, the nucleation growth will happen in c cell. Here, DT_n is nucleation undercooling. The tip under cooling is thought to be a function of the velocity. The number density of equated grains is taken to be equal to the seed density. Exquiated growth occurs if the thermal gradient is less than the value given by above equation. Here, the tip under cooling is thought to be a function of applied pulling velocity [11]. The basis of Solid-liquid interface stabilities assumed that the dendrite tip should be oval. The value

between cell/dendrite Ttip and non-equilibrium liquid T is ΔT . The shape can be described as the secondary tip position as a function of the distance from the primary tip. Measurements of the secondary tip positions have been performed on the images to obtain at different times for various super cooling [12,13]. For a comparison of data from different super cooling, it was of use the tip radius R. The growth model proposed that describing structural growth will be based on the continuous theory on the same to thermal and solute diffusion. In the case of dendrite alloys, due to F it will drive solidified course. Under a certain cooling gradient T there is only one growth velocity and radius R. The certain angles between the growth direction and lamellar orientation has been arranged along the GD(growth direction) in terms of the primary phases, they may be formed on parallel rule.

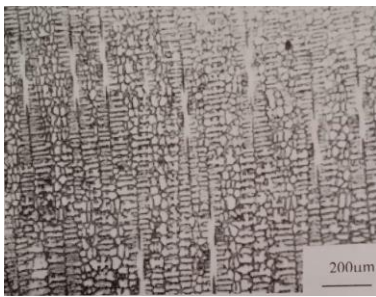


Figure 1: Graph of quenched morphology at 180mm/h [14].

Liquidus, °C	Reference, °C	Literature ^[17] , °C	Al, at %
1540		1557	44
1520	1535	1531	46
1510		1503	48
1485	1480	DLTA G	50
1480	1470	DLTA Ls	52

Table 2: Cooling for liquid in front of Solid-liquid interface

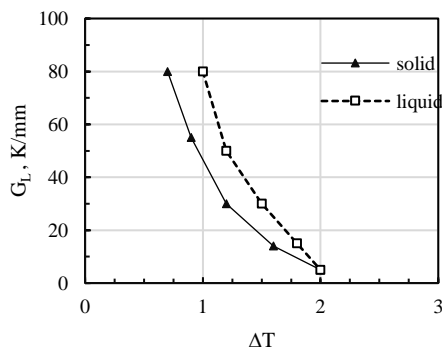


Figure 2: Trend of G_L and ΔT at $v=0.25\text{mm/s}$ for Ti-44at.%Al. [10]

Control of the Growth Velocity

As for the preferred growth direction (PGD) it is confirmed through the bcolumnar growth simultaneously. At high gradient FZ (Floating zone) as possible, the [001] β preferred growth velocity is confirmed. Henry reports that in Fcc Al alloys in DS (Directional solidification), the solidified velocity and gradient will raise. Therefore, the greatly dendrite direction {001} to {110} has been changed. [001] has been prohibited and [110] has raised. Some researcher reported that as for the Bcc Cr when the rate is raised the growth direction of [001] had taken changing to [111] for determining preferred growth. The dendrite space size could be caused by gradient though increasing cooling field. The solute liquid diffusion among inter dendrites would result in space to be decreased. In final directional solidification's solute did not take out forming segregation to be castings'. In Bridgman method in DS the G was very high so that it affects macro segregation. [16] In the condition of 25K/mm due to the C to be vG, strain energy anisotropic one could be caused by the preferred growth with [111]. As the velocity rate was 0.25K/s, the [001] direction would be caused by boundary energy which had no report on this relation of rate and preferred growth direction so far. With the rate of 0.75K/s the directional of [111] would be taken to think. If the rate was low it had caused [001] preferred growth direction. The primary β , on the contrary, if the rate became high it cause [111] preferred growth. The high G had resulted the [001] preferred growth which should be caused by primary b. as for Fcc Al [001] to [110], the velocity is so high that G had resulted in great high. In the case of 0.25K/s under the velocity of 45mm/h, b dendrite will grow along [001]. In the case of 0.75K/s it might be [111] under the 135mm/h, which are considered. Specimens with Bcc generally evolves at low velocity and [001] is preferred growth direction due to boundary energy. [15] The strain energy attributing to heterogeneities with raising temperature that will transfer to [111]. Therefore for b dendrite parallel to growth direction, high temperature gradients alloys could be chosen and appropriate growth rate will be essential. At the parallel direction with growth direction two alloys produce dendrite growth. On the {112} had been perpendicular to boundary plans by observations. As a speed of 30mm/h with parallel to growth direction it might cause to be 10% RT (Room temperature) elongation, the same as that at 800°C. In the case of 2mm/h, plane boundary had been grown and made seeding effect. Up to 90mm/h the direction of [111] became to be wanted structure. Hence the [001] is proper condition. 1) Temperature gradient and solidified velocity i.e. crystal growth velocity; 2) Phase transformation studied at the interface of S/Liquid [16]. The decreased trend will be observed. The total value is about 140~40mm/s, which might be caused by low velocity. For example, a certain cooling rate of 0.5K/s and low value of 0.3K/s approximately. As shown in Table1, the value about 20K would be fitting to 48at.%Al in terms of phase

diagram for references. The phase at interface in the inter dendrite region is to form during the quenching cooling to RT as seen in Figure 1, which need to be considered. The G_l has a negative proportion to ΔT as seen in Figure 2. It is 5K/mm of intersection between two curves. The G has higher than that in the solid, while T is higher than liquid. The value smaller than 30K/mm fits to nucleation to happen as ΔT is 1K. If more than 2K no equiaxed would happen due to no slope of DT a long distance. It was found that the coarse grain will be gained in low velocity. This is a thermal flow. The growth grain has been right part. The left takes role of seed effect, the better state is 10mm/s. The final freezing temperature might be reduced by the length of the dendrite, which was measured at the solid. Liquid interface and the temperature gradient. The final freezing temperature could be estimated to be 1172°C in the simple equation. If one consider that dendrite length is 21mm, the temperature gradient is 8.8K/mm in the mushy zone, and dendrite tip temperature is 1492°C. Usually the first rate was demanded lowly to make sure to grow with morphology of plane and cell. Heating temperature was 1492°C taken on the T_{tip} . According to $G = (T_{tip} - T_{base})/L$, G is 40~33mm which was the length of the dendrite. The final freezing temperature estimated by the length of the dendrite showed that some deviation from the results had been by temperature profile and DTA. There might be some error in estimating the freezing dendrite region.

Conclusion

The critical velocity has equation as $V_{cr} = G \times C^{-1}$ which is main factor in metallic solidification. With increasing gradient and decreasing solute concentration the value increases. Meantime, the solidified length variation $D_{Ls} = DT \times C^{-1}$ means that the temperature variation and solute concentration has the similar function as well. The priority analysis of homogeneous nucleation has been taken into account and then the heterogeneous one in general. Furthermore the Hunt's analytical model may describe the directional solidification like TiAl intermetallic compounds in nucleation generations too. Meantime, the GL will become low if DT increases. The liquid GL has been bigger than solid one.

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