

HOW DOES COALESCENCE OF DENDRITE ARMS OR GRAINS INFLUENCE HOT TEARING ?

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Abstract

Hot tearing, a severe defect occurring during solidification, is the conjunction of tensile stresses which are transmitted to the mushy zone by the coherent solid underneath and of an insufficient liquid feeding to compensate for the volumetric change. In most recent hot tearing criteria, one of the critical issues is the definition of a coherency point which, in low-concentration alloys, corresponds to the bridging or coalescence of the primary phase. A coalescence model has been developed recently using the concept of the disruptive pressure in thin liquid films.^[1] It has been shown that large-misorientation grain boundaries, which are characterized by an interfacial energy, γ_{gb} , larger than twice the solid-liquid interfacial energy, γ_{sl} , solidify at an undercooling $\Delta T_b = (\gamma_{gb} - 2\gamma_{sl})/(\Delta s_f \delta)$, where Δs_f is the entropy of fusion and δ the thickness of the diffuse interface. When $\gamma_{gb} < 2\gamma_{sl}$ (e.g., low-angle grain boundaries), dendrite arms coalesce as soon as they impinge on each other. Using such concepts and a back-diffusion model, the percolation of equiaxed, randomly oriented grains has been studied in 2D : it is shown that the grain structure gradually evolves from isolated grains separated by a continuous interdendritic liquid film, to a fully coherent solid with a few remaining wet boundaries. The implication of such findings for the hot cracking tendency of aluminum alloys are discussed.

1. Introduction

Hot cracking is a severe defect occurring during solidification of low concentration alloys. Arising at fairly high volume fraction of solid, it involves deformation of the solid skeleton and flow of the interdendritic liquid.^[2] Indeed, three regions can be distinguished in the mushy zone of a single dendritic grain, as shown in Fig. 1 for a succinonitrile-acetone alloy. At low volume fraction of solid (region 1), the permeability of the mushy zone is fairly high and the interdendritic liquid is continuous. Solidification shrinkage and deformation of the solid can therefore be easily compensated by fluid flow in this region. At volume fraction of solid close to unity (region 3), dendrite arms have bridged together and the remaining liquid is in the form of small droplets. This region behaves therefore as a continuous viscoplastic solid. In between (region 2), the dendrites are densely packed (i.e., low permeability of the mushy zone) but not yet coalesced. Thus, the interdendritic liquid is a continuous film which cannot sustain tensile stresses, i.e., it behaves as a brittle phase. Any opening of the mushy zone in this region cannot be compensated by liquid flow, thus inducing a hot crack.

Recognizing the importance of hot cracking, Flemings and his group devised at the end of the sixties an interesting shearing apparatus for testing the mechanical resistance of aluminum alloys in the semi-solid state.^[3,4] They explained the influence of grain size, shearing rate, microstructure morphology, volume fraction of solid in terms of particles bonding/debonding, grain sliding and rearrangement. One of the first conclusions outlined in the second

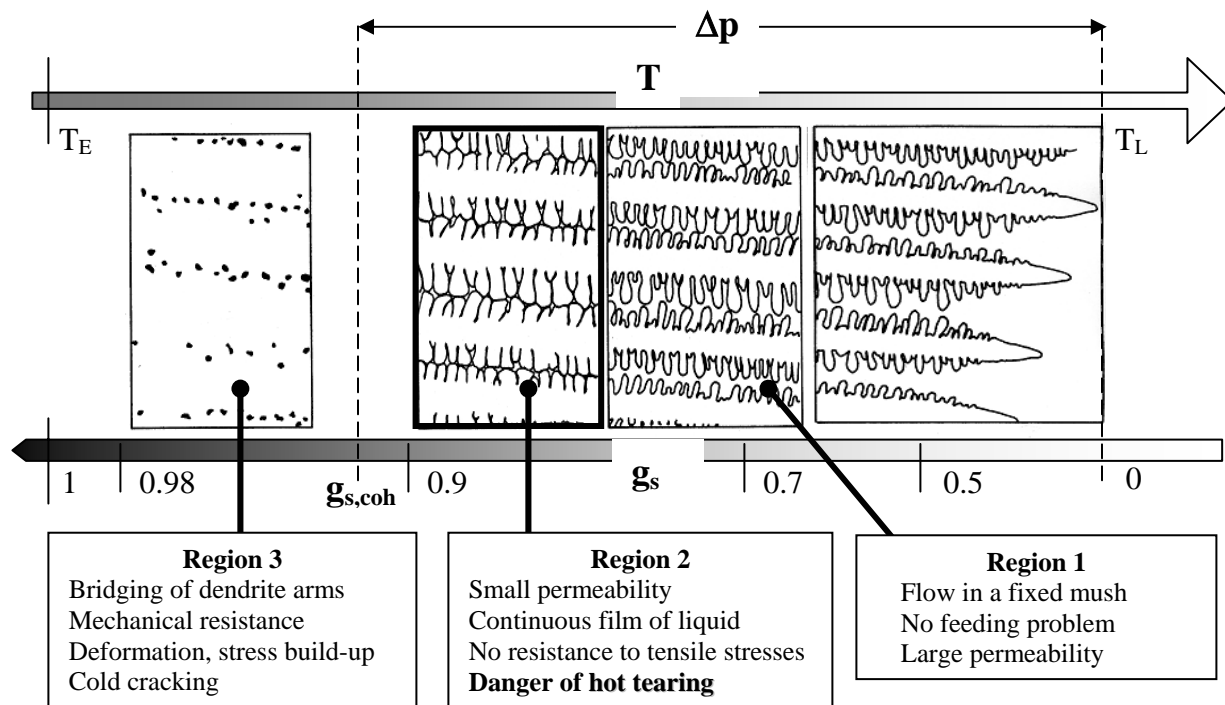


Figure 1 : Solidification sequences in a transparent succinonitrile-acetone alloy solidified under directional conditions in between two glass plates. The pictures, redrawn from the original micrographs to enhance contrast, have been placed on a schematic temperature and solid fraction scale. The transition at which the continuous liquid film is transformed into a continuous solid corresponds to $g_{s,coh}$.^[5]

contribution^[4] was that processing of metallic alloys in the semi-solid state should be feasible with various advantages. As Flemings' group focused its attention toward semi-solid processing, prediction of hot tearing did not progress much. For many years, the main criterion applied to characterize the Hot Cracking Sensitivity (HCS) of an alloy was based on the solidification interval:^[2] the HCS increases with the width of the mushy zone. Feurer^[6], maybe inspired by the work of Flemings and Piwonka on porosity^[7], tried to derive a criterion based on the pressure drop of the interdendritic liquid, driven by solidification shrinkage. Clyne and Davies^[8], and much earlier Pellini^[9], recognized the fact that hot tearing occurs in the critical region 2 of the mushy zone. They derived a HCS criterion based on a critical time spent by the mushy zone in this region.

More recently, Rappaz, Drezet and Gremaud^[10] derived a very simple model for HCS, similar to the Niyama criterion developed for porosity formation^[11], but which accounts also for strain-induced flow in the mushy zone. This so-called RDG model is briefly summarized in Section 2. One of the critical parameter of this model is the fraction of solid, $g_{s,coh}$, at which the continuous liquid film transforms into a continuous viscoplastic solid (see Fig. 1). In section 3, it is shown that this parameter is directly linked to the coalescence behavior of the dendrites/grains. A coalescence model based on thermodynamic considerations allows to define this coherency point as a function of involved interfacial energies and to explain why hot cracking occurs at grain boundaries. As an application, percolation of equiaxed grains is calculated. Finally, some experimental observations are presented in Section 4, while future directions of research are outlined in the Conclusion.

2. RDG Hot Cracking Criterion

Consider the situation shown in Fig. 1 of columnar dendrites growing at steady state in a constant thermal gradient, G , and at a velocity, v_T , equal to the speed of the liquidus isotherm.

For the sake of simplicity, a single grain has been considered. In the absence of solid deformation and solid or liquid contraction, solidification shrinkage has to be compensated by interdendritic liquid flow in order to avoid porosity formation. In the well-known Niyama's criterion, the pressure drop, Δp_{sh} , associated with this flow can be easily calculated considering the average mass conservation equation and Darcy's equation. The RDG criterion simply extended this calculation to encompass a tensile deformation rate, $\dot{\epsilon}_{pT}$, perpendicular (transverse) to the growth direction. Therefore, as introduced in Ref. [10], a hot tear forms if the critical pressure, p_{min} , at the roots of the dendrites falls below a certain cavitation pressure, p_c :

$$p_{min} = p_m - \Delta p_\epsilon - \Delta p_{sh} = p_c \quad (1)$$

where p_m is the metallostatic pressure at the dendrite tips, Δp_ϵ and Δp_{sh} are the pressure drop contributions associated with deformation and shrinkage, respectively (taken as positive values). These two contributions can be obtained from the average mass balance. Assuming that the specific masses of the solid and liquid, ρ_s and ρ_l , are constant, that the liquid flows along the x-axis of the thermal gradient while deformation of the solid is in the perpendicular direction, one has :

$$\begin{aligned} \frac{\partial(\rho_l g_l v_{l,x})}{\partial x} + \frac{\partial(\rho_s g_s v_{s,y})}{\partial y} + \frac{\partial(\rho_s g_s v_{s,z})}{\partial z} - v_T \left[\frac{\partial(\rho_s g_s)}{\partial x} + \frac{\partial(\rho_l g_l)}{\partial x} \right] = \\ = \frac{d(g_l v_{l,x})}{dx} + (1+\beta) g_s \dot{\epsilon}_{p\perp} - v_T \beta \frac{dg_s}{dx} = 0 \end{aligned} \quad (2)$$

$\beta = \frac{\rho_s}{\rho_l} - 1$ ($\beta > 0$) is the solidification shrinkage factor and $\dot{\epsilon}_{p\perp} = \frac{\partial v_{s,y}}{\partial y} + \frac{\partial v_{s,z}}{\partial z}$ is the deformation rate of the solid in the direction perpendicular to the thermal gradient. Please note that the x-component of the plastic deformation of the solid is not considered. In the fully coherent solid, the incompressibility condition states that $(\epsilon_{p_{xx}} + \epsilon_{p_{yy}} + \epsilon_{p_{zz}}) = 0$. However, the non-coalesced dendrites do not need to satisfy this relationship. Furthermore, the x-component of the solid deformation (i.e., along the thermal gradient) is considered to be much smaller than the transverse components. After integration of Eq. (2), one gets :

$$g_l v_{l,x} = - (1+\beta) E(x) - v_T \beta g_l \quad (3)$$

where $E(x)$ is the cumulated deformation rate defined as :

$$E(x) = \int g_s \dot{\epsilon}_{p\perp} dx \quad (4)$$

In the absence of deformation ($E = 0$), the velocity of the fluid, $v_{l,x}$, is constant and equal to $-v_T \beta$ at any point of the mushy zone. This was already established by Niyama for the formation of porosity.^[11] The LHS term of Eq. (3) can be related to the pressure gradient in the liquid via Darcy's equation:^[7,10,11]

$$g_l v_{l,x} = - \frac{K}{\mu} \frac{dp}{dx} \quad (5)$$

where K is the permeability of the mushy zone, μ is the viscosity of the liquid and the contribution of gravity has been neglected. Combining Eqs. (3) and (5) and integrating over the whole length of the mushy zone finally gives the pressure drop between the tips and roots of the dendrites :

$$\Delta p_{max} = \Delta p_\epsilon + \Delta p_{sh} = (1+\beta)\mu \int_0^L \frac{E}{K} dx + v_T \beta \mu \int_0^L \frac{g_l}{K} dx \quad (6)$$

At this stage, Eq. (6) is quite general and can be easily extended to non steady-state situations. Setting up a cavitation pressure (Eq. (1)), it gives the maximum strain rate (or value of E) that the mushy zone can sustain before a hot tear nucleates. It should be pointed out that, under steady-state conditions, the integration over x can be replaced by an integration over temperature, thus introducing the thermal gradient and the solidification interval of the alloy.^[10] Assuming that the permeability is given by Carman-Kozeny's relationship as employed in previous works,^[7,10,11] one gets :

$$\Delta p_{\max} = \frac{180}{\lambda_2} \frac{(1+\beta)\mu}{G} F(\dot{\epsilon}_{p\perp}) + \frac{180}{\lambda_2} \beta \frac{v_T}{G} H \quad (7)$$

where λ_2 is the secondary dendrite arm spacing, F and H are two contributions given by :

$$F(\dot{\epsilon}_{pT}) = \int_{T_S(g_{s,\text{coh}})}^{T_L} \frac{E(T) g_s(T)^2}{(1 - g_s(T))^3} dT \quad H = \int_{T_S(g_{s,\text{coh}})}^{T_L} \frac{g_s(T)^2}{(1 - g_s(T))^2} dT \quad (8)$$

Setting a value to the cavitation pressure p_c and assuming that $\dot{\epsilon}_{p\perp}$ is uniform, Eq. (8) allows then to determine the maximum strain rate, $\dot{\epsilon}_{p,\max}$, that the mushy zone can sustain before a hot crack nucleates. Therefore, a fairly good HCS estimator is given by $(\dot{\epsilon}_{p,\max})^{-1}$.

Besides a microsegregation path giving $g_s(T)$, the RDG model requires to specify the lower bound of integration, i.e., the fraction of solid at which coherency is reached. If the alloy exhibits a substantial amount of eutectic, $g_{s,\text{coh}}$ can be set to $g_s(T_E)$. However, in low concentration alloys, $g_{s,\text{coh}}$ corresponds to the transition between a continuous liquid to a continuous solid. Such a transition will differ very strongly, depending whether one considers dendrites belonging to the same grain (Fig. 1) or dendrites separated by a grain boundary. A calculation of $g_{s,\text{coh}}$ for low-concentration alloys therefore requires to address the question of primary phase coalescence.

3. Coalescence Model

Coalescence of two solid grains/dendrites depends strongly on the grain boundary energy. Consider two solids of a pure phase but not necessarily of the same orientation in contact with the liquid just below the melting point (Fig. 2).^[11] As the thickness, h, of the liquid film separating them tends towards zero, the overall excess free energy of the system (per unit surface) should go from $2\gamma_{sl}$, when the two solid-liquid interfaces are far apart, to γ_{gb} when $h = 0$. The excess free energy of the atoms located in a thickness δ of the diffuse interface and integrated over the normal to the interface defines the interfacial energy in a macroscopic (or multi-phase field) approach. When these two solid parts have the same orientation (i.e., $\gamma_{gb} = 0$), there is a clear energy gain for the system to solidify the last interdendritic liquid. This situation is "attractive", i.e., two solid-liquid interfaces have a tendency to attract each other once they get close enough. It is an unstable situation since the excess surface energy decreases with the spacing h : a perturbation between two flat solid-liquid interfaces will naturally develop. This situation is typical of two dendrite arms belonging to the same grain.

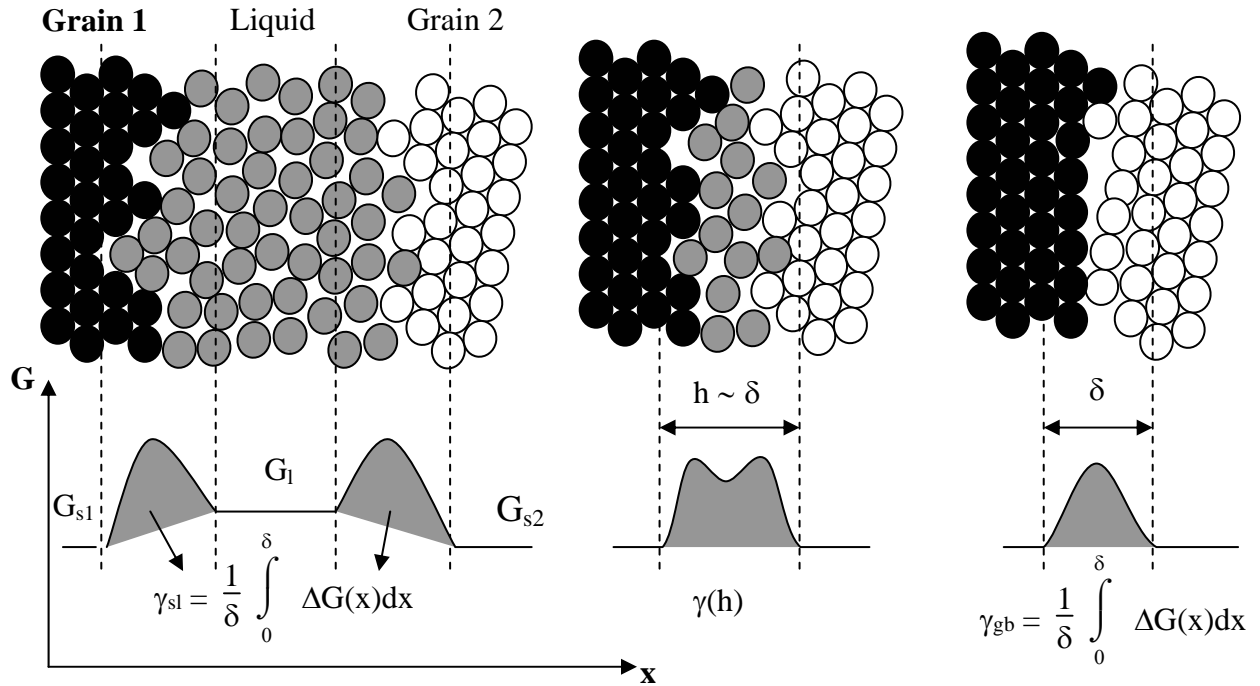


Figure 2 : Schematic atomistic view of a closing liquid interface in between two grains of different orientations (top) and schematics of the Gibbs free energy at a temperature T below the melting point (case where $\gamma_{gb} > 2\gamma_{sl}$) to be used in a multi-phase field approach (bottom).

For two grains or dendrite arms which are misoriented, however, the interfacial energy, γ_{gb} , is larger than twice the interfacial solid-liquid energy, γ_{sl} .^[12,13] This "repulsive" situation is such that the intergranular liquid film may remain stable in a pure substance, even below the melting point, T_m . At a given undercooling, ΔT , the energy balance per unit area of the film is given by :

$$\Delta G_s(h) = (G_l - G_s)h + \gamma(h) = \frac{L \Delta T}{T_m} h + \gamma(h) \quad (9)$$

where L is the latent heat of fusion per unit volume (i.e., $L = \Delta_s T_m$). The thickness dependence of the interfacial energy can be taken as:^[1]

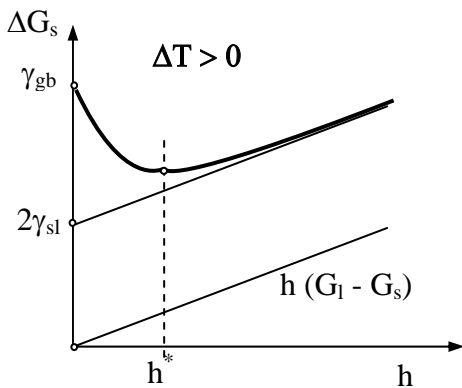


Figure 3 : Overall excess free energy $\Delta G_s(h)$ for $\gamma_{gb} > 2\gamma_{sl}$ and $\Delta T > 0$.

$$\gamma(h) = 2\gamma_{sl} + (\gamma_{gb} - 2\gamma_{sl}) \exp\left[\frac{-h}{\delta}\right] \quad (10)$$

When h is infinite, the interfacial energy of the two solid-liquid interfaces is equal to $2\gamma_{sl}$, while when $h = 0$, the grain boundary has an energy γ_{gb} . The excess free energy, ΔG_s , is shown in Fig. 3. Indeed, the minimum energy is obtained for a certain film thickness, h^* . It can be shown that $h^* = 0$ (i.e., disappearance of the liquid film) is obtained in a pure substance at a *bridging/coalescence* undercooling, ΔT_b , given by :

$$\Delta T_b = \Delta T_{h^*=0} = \frac{(\gamma_{gb} - 2\gamma_{sl}) T_m}{L} \frac{1}{\delta} \quad (11)$$

In alloys, the situation is made more complex by solute diffusion. While the interdendritic liquid film remains in between two grains, back-diffusion has a tendency to pump solute from the liquid and thus to solidify it. Introducing Eq. (10) into a standard 1-dimensional microsegregation model based on a Landau transformation for tracking of the interface^[14] produces the result shown in Fig. 4. The parameters are given in the figure caption with : D_s the solid diffusion coefficient, λ the secondary dendrite arm spacing, ρc_p the volumetric specific heat, \dot{H} the heat extraction rate, k the partition coefficient, m the slope of the liquidus and c_0 the nominal concentration. The average composition of the solid starts at kc_0 and increases progressively towards c_0 as solidification progresses, but not along the solidus. Indeed, this situation is closer to Scheil's than lever rule. Although diffusion in the liquid is also calculated, the composition of the liquid is nearly uniform but has nevertheless been represented with the average in Fig. 4. It starts with the value c_0 and continuously increases along the liquidus line as T decreases. However, when the remaining liquid has a thickness h close to δ , the two interfaces start to "feel" each other and remain wet. As back-diffusion pumps solute, the concentration in the liquid decreases and cooling proceeds until the *coalescence line* is reached. This line is parallel to the liquidus, shifted by the undercooling ΔT_b given by Eq. (11).

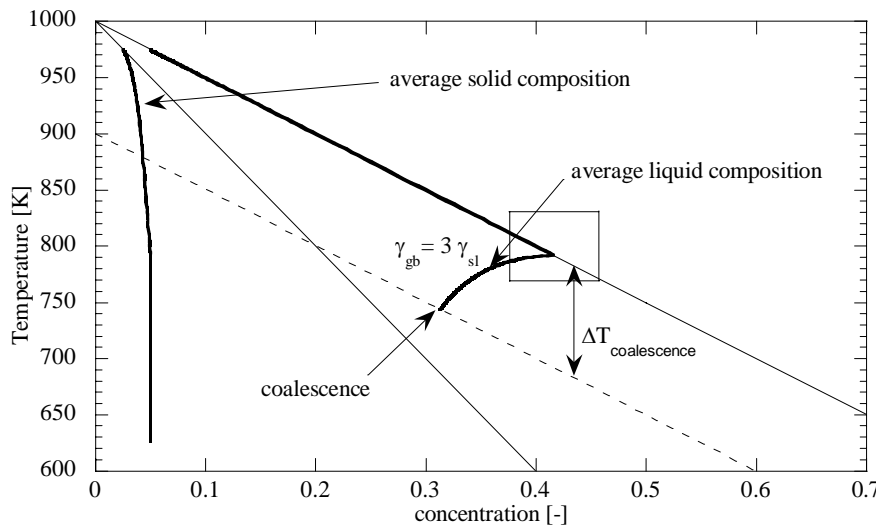


Figure 4 : Evolution of the average composition of solid and liquid as a function of temperature as predicted with a 1D sharp interface model.

$$\begin{aligned}
 D_s &= 10^{-13} \text{ m}^2/\text{s} & T_m &= 1000 \text{ K} \\
 \gamma_{sl} &= 0.1 \text{ J/m}^2 & m &= -500 \text{ K} \\
 \gamma_{gb} &= 0.3 \text{ J/m}^2 & k &= 0.5 \\
 \lambda/2 &= 10 \text{ } \mu\text{m} & c_0 &= 0.05 \\
 \delta &= 1 \text{ nm} \\
 L &= 10^9 \text{ Jm}^{-3} \\
 \rho c_p &= 2.5 \cdot 10^6 \text{ Jm}^{-3}\text{K}^{-1} \\
 \dot{H} &= -1.25 \cdot 10^8 \text{ Wm}^{-3}
 \end{aligned}$$

This simple coalescence model explains why hot tearing is located at grain boundaries. In a columnar structure (Fig. 5), one can define two coherency temperatures : one for dendrites belonging to the same grains, T_{cd} , and another one for dendrites separated by a grain boundary, T_{cg} . Please note that this last entity depends on the type and misorientation of the grain boundary. For low angle grain boundaries, γ_{gb} is smaller than $2\gamma_{sl}$ and the dendrites behave as those belonging to a single grain. For high angle grain boundaries, however, γ_{gb} is larger than $2\gamma_{sl}$ and $\Delta T_b > 0$. Coalescence in this case is delayed and a continuous liquid film can remain. The presence of such liquid films has two effects : i) it increases the fraction of solid, $g_{s,coh}$, up to which feeding must be performed if any opening of the mushy zone occurs (see Eq. (8)). This obviously increases the pressure drop for a given strain rate or, equivalently, decreases the maximum strain rate that the mushy zone can sustain before cavitation occurs in the liquid film ; ii) it localizes the deformation by a factor D/λ_1 with respect to a single columnar grain, where λ_1 is the primary dendrite arm spacing and D the average transverse distance separating repulsive-type boundaries.

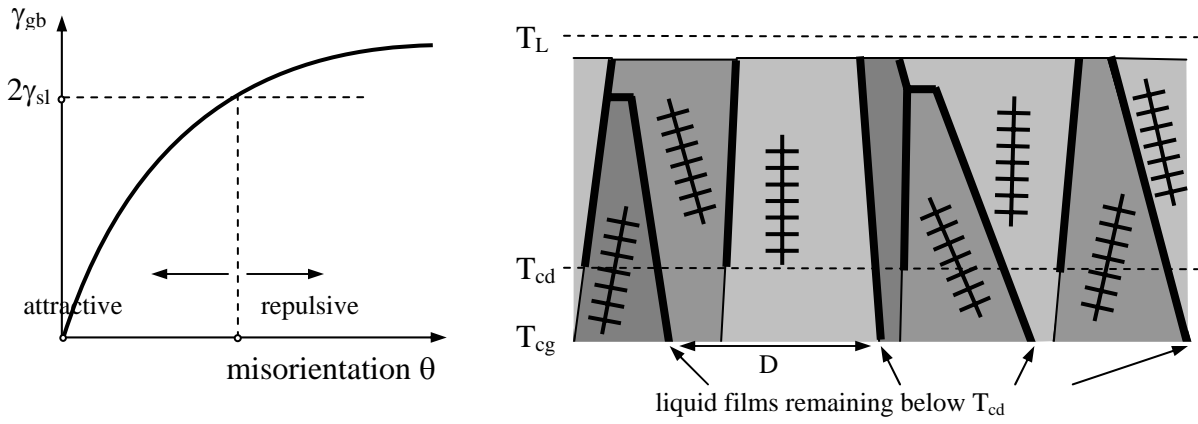


Figure 5 : (left) Schematics of γ_{gb} as a function of the misorientation between two grains. The critical misorientation at which $\gamma_{gb} = 2\gamma_{sl}$ defines the transition between attractive and repulsive grain boundaries; (right) Schematics of columnar grains growing in a vertical thermal gradient. Attractive grain boundaries coalesce at T_{cd} whereas at repulsive ones, liquid films remain up to a temperature T_{cg} function of the misorientation. This localizes thermal strains at these liquid films.

The 1D microsegregation sharp-interface model cannot account for diffusion of solute in the liquid parallel to the impinging interfaces and for geometrical aspects. For that reason, a multi-phase field approach of coalescence has also been developed.^[1] It will not be reported here. Instead, a recent extension of this 1D model to a collection of equiaxed grains is presented.^[15] As opposed to the very computation intensive phase-field approach, this model allows to describe a fairly large population of grains. Assuming that all the grains nucleate at the same time in a 2D domain and that the grains are fully globulitic, the final grain boundaries are straight segments perpendicular to the lines linking closest nucleation centers (Fig. 6).^[16] This defines a Voronoï tessellation of the domain. During growth, the solid-liquid interface of a given grain is gradually changing from a nearly spherical shape at the beginning of solidification to the final polyhedral shape outlined by the grains boundaries at the end. It is approximated during growth by a set of linear segments : each segment is parallel to the final grain boundary edge within the triangle linking this edge and the nucleation center (see Fig. 6). Considering then each triangle, the solid and liquid regions can be enmeshed with a set of nodes aligned on a perpendicular to the interface. A Landau transformation allows to follow the interface position and the solute balance accounts for diffusion in the liquid and solid domains, as well as for the change in the interface length. This segregation model is similar to that of Ohnaka^[17] for a hexagonal arrangement of primary dendrite trunks, but near the end of solidification, when two solid-liquid interfaces impinge at a grain boundary, the sharp interface coalescence model is considered. The undercooling of the liquid, ΔT_b , is a function of the misorientation of the two neighbor grains, their orientations being set randomly at the beginning of the calculation. Therefore, in this 2D situation, the misorientation of the boundary is given by $\Delta\theta = (\theta_i - \theta_j)$, where θ_i and θ_j are the angles characterizing the orientation of two neighbor grains with respect to the horizontal axis. The function $\gamma_{gb}(\Delta\theta)$ was prescribed from the shape given in Fig. 5 (left).^[13]

In the result shown in Fig. 6, only the remaining wet grain boundaries have been drawn with a thick line. Simulations have been done for an Al-1wt% Cu alloy and the results are shown for four different temperatures. At 600 °C, most of the boundaries are wet and the macrostructure is therefore still made of a population of disjointed grains which could certainly not sustain tensile stresses. As the temperature is lowered, the structure evolves from a continuous interdendritic liquid to a continuous solid, i.e., the grain structure percolates. Please note that this occurs at a temperature of around 500 °C (Fig. 6c), i.e., at a temperature substantially lower than the

eutectic temperature. This value is of course strongly dependent of the selected parameters, in particular the thickness of the diffuse interface, δ , and of the approximations made for the grain boundary energy function. Using phase field simulations, it has been shown that the geometry of the impinging dendrite arms (or grains) has a strong influence on their coalescence, at least within a single grain.^[18]

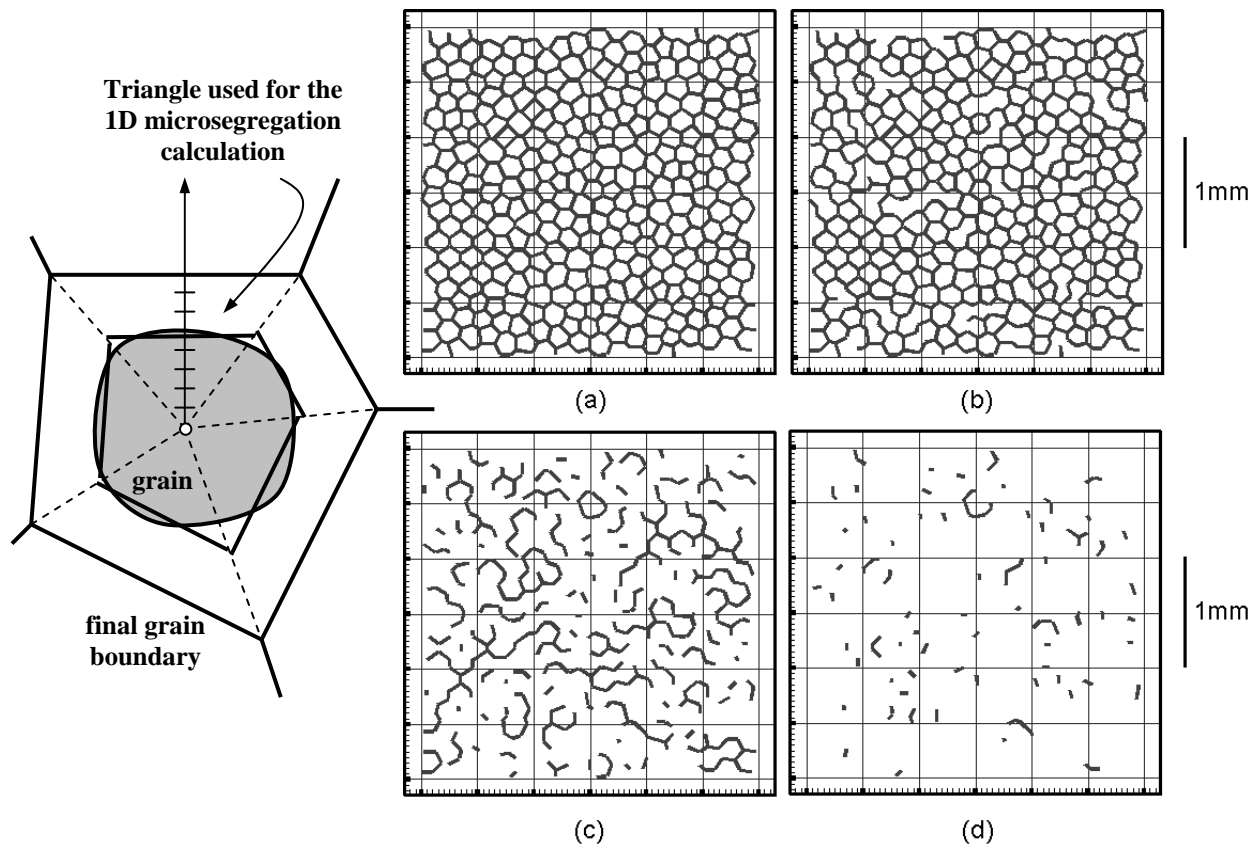


Figure 6 : Percolation of equiaxed grains in Al-1wt%Cu. The distribution of the remaining wet grain boundaries are represented with thick lines as a function of temperature : (a) 600°C (b) 550°C (c) 500°C (d) 490°C.

4. Experimental Observations

The transition between the non-coalesced and coalesced regions of the mushy zone is not abrupt. As already mentioned, it first occurs within the grains and later at grain boundaries. It depends on the grain boundary energy, but also on the configuration of the impinging dendrite arms. Observing by SEM the grain boundary surfaces after a hot crack has formed at this location is particularly enlightening (Fig. 7).^[19] The hot crack surface is very smooth, with small undulations corresponding precisely to secondary dendrite arms (Fig. 7a). This clearly shows that hot cracking occurred while the grain boundary was still wet, i.e., was covered by a thin liquid film. At many locations, drape-like spikes such as those shown in Fig. 7a can be observed on both sides of the hot cracked surface. They seem to have formed by elongation of a liquid region with simultaneous oxide formation at the surface, but without appreciable deformation. In a few more rare cases, spikes such as that observed in Fig. 7b are seen. They clearly exhibit a torn surface indicating that they have probably been formed by deformation and necking of a solid bridge extending across the grain boundary. These deformed spikes can also exhibit a drape-like appearance at the roots, meaning that some liquid was still present. Both mechanisms of spike formation (i.e., stretching of liquid menisci and of solid bridges) have been corroborated by in-situ observations of hot cracking formation in organic alloys.^[19]

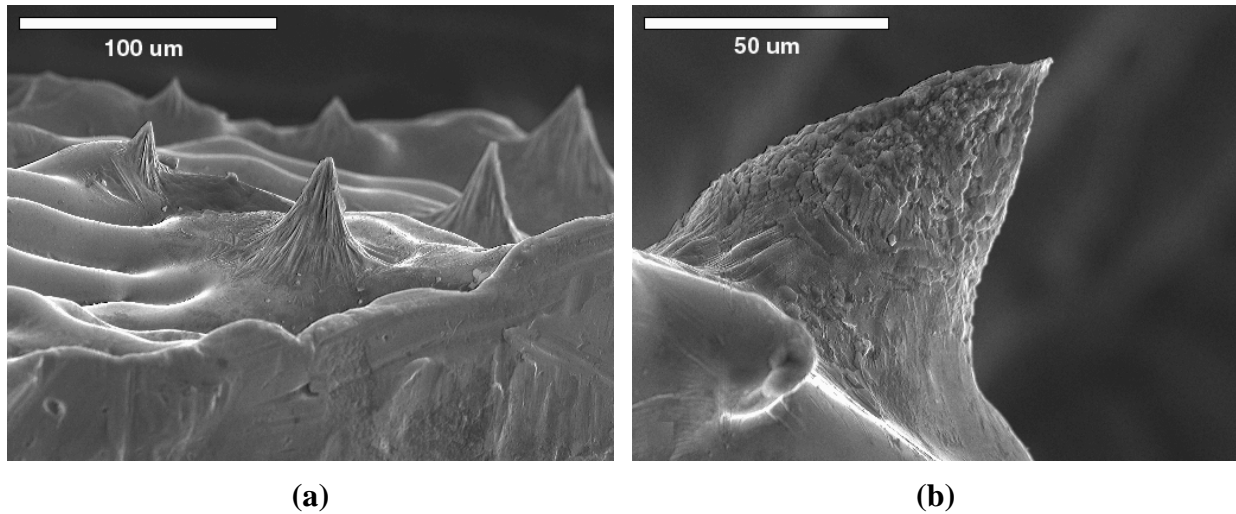


Figure 7 : SEM observation in Al-3wt% Cu of a drape-like spike probably formed by the last solidification of interdendritic liquid (a), and of a deformed spike probably formed by necking of a solid bridge (b).

Observations of organic solidification have also clearly revealed the gradual coalescence occurring within a grain (Fig. 8a) and at grain boundaries (Fig. 8b). Within a grain, bridging of secondary dendrite arms attached to the same primary trunk seem to occur fairly early during solidification (attractive boundaries), leaving small pockets of liquid near their roots. Bridging of arms belonging to two dendrite trunks is slightly delayed. This can be due to the positive curvature of the impinging interfaces, to the microsegregation profiles and/or to the small misorientations between two primary trunks. At a grain boundary, however, coalescence does occur much later : Fig. 8b has been taken much deeper in the mushy zone, where it seems that everything is solid. At large magnification, a separation between the two grains is still visible until it disappears. This contrast line probably indicates the presence of a thin liquid layer.

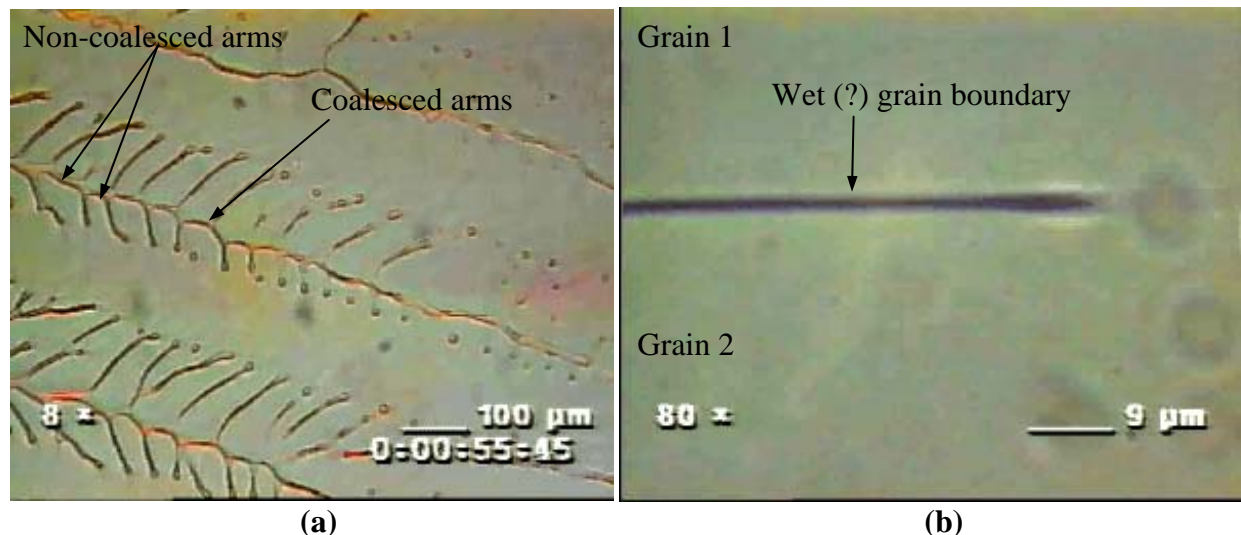


Figure 8 : Coalescence with a single dendritic grain (a) and at a grain boundary (b) of a succinonitrile-acetone alloy which is directionally solidified. Figure (b) is taken much deeper in the mushy zone, where the alloy seems fully solid, but a large magnification seems to indicate still the presence of a wet (?) grain boundary, suddenly disappearing.

Wang et al.^[20] have tried to correlate the hot cracking tendency with the coalescence behavior during welding of bi-crystals of superalloys. They showed that, when the misorientation of the two crystals is small (typically less than 15 deg.), no crack develops at the grain boundary in

this hot cracking sensitive alloy. Increasing the misorientation beyond this value led to a crack at the weld centerline coinciding with the grain boundary.

Grasso et al.^[21] have devised a quite interesting hot tearing test in which the tensile strength of the mushy zone can be measured without having any interaction with a free (oxidized) surface. In this experiment, an Al-Cu alloy is solidified in a small vertical thermal gradient within a two-part cylindrical mold which has a neck at mid-height (Fig. 9). A thermocouple is inserted near the neck and the two parts of the mold can be pulled apart in a tensile-test machine. The vertical stress measured by a load cell can be then reported as a function of the temperature at which pulling occurred. Fig. 9 shows the shape of the ruptured surface, whereas in Fig. 10, the maximum stress measured during the hot tearing experiments is reported as a function of the temperature, together with the curve $g_s(T)$ measured by Single Pan Thermal Analysis (SPTA). As can be seen, at 640 °C, the mushy zone does not offer much resistance and a thin liquid needle in between the two parts of the casting indicates that feeding is still fairly easy at this temperature. This temperature corresponds to a volume fraction of solid of about 90%, as measured by thermal analysis (Fig. 10).^[22] At 620 °C, the ruptured surface is slightly conical indicating that the crack tries to find the weakest path in the slightly more liquid, upper part of the casting (see schematic shape indicated in Fig. 9, middle). At 590 °C, the surface of the crack is flatter but starts also in the upper part of the neck, where the volume fraction of solid is slightly lower and coalescence not as advanced.

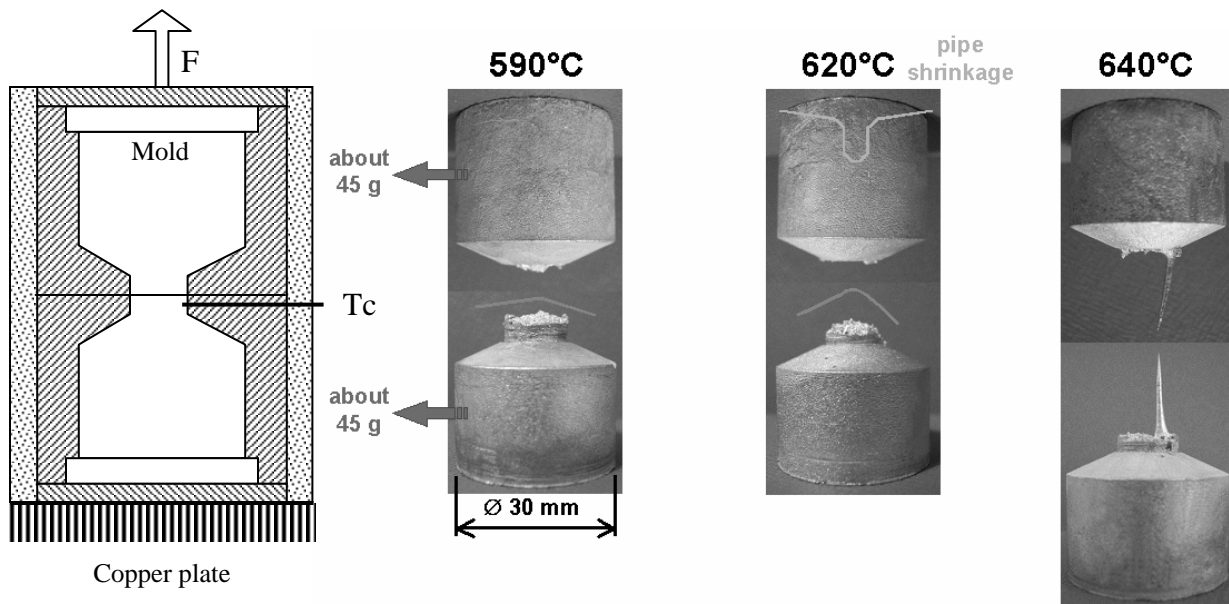


Figure 9 : Schematics of the hot tearing device using a necked cylindrical mold (left) and hot cracks produced at 640, 620 and 590 °C in Al-1wt% Cu.

As can be seen in Fig. 10, the alloy offers little resistance to tensile stresses just below the liquidus. The resistance slightly increases between 650 and 630 °C, as solidification proceeds. The alloy was inoculated with titanium diboride and exhibits a globulo-dendritic structure. The increase of resistance can be attributed to the reduced ability of the grains to move as they start to impinge on each other at around 0.3-0.5 volume fraction of solid. The increased viscosity of the liquid may also be responsible for this overall resistance increase. At a volume fraction of solid of around 0.95, i.e., between 630 and 620 °C, the resistance of the mushy zone exhibits a fairly sharp increase, from 1.25 to 2.7 MPa. This could be due to the first coalescence events of low angle grain boundaries. There is then a small plateau followed by a gradual increase of the resistance up to 590 °C. At this temperature, high angle grain boundaries are certainly still wet

and full coherency is not yet reached. Unfortunately, the load cell did not allow to measure the strength of the mushy zone below this temperature.

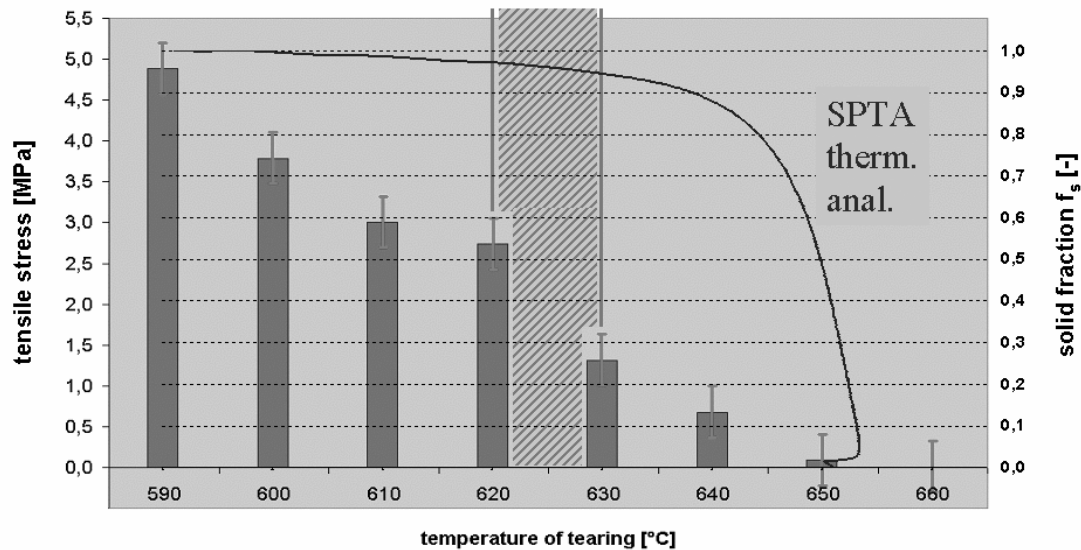


Figure 10 : Maximum stress measured in the hot tearing experiment of Fig. 9 as a function of the tearing temperature for an Al-1wt% Cu alloy. The error bar corresponds to the deviation measured for each temperature between three trials. The solid fraction – temperature curve superimposed to these measurements has been obtained using a Single Pan Thermal Analysis (SPTA) ^[22].

5. Conclusion

Hot cracking is strongly influenced by the bridging/coalescence of dendrite arms. Bridging occurs later at grain boundaries, as compared with dendrites belonging to the same grain, thus explaining why thermally-induced strains and hot cracking are localized in these regions. While hot cracking criteria based on physical mechanisms start to emerge, a theoretical framework for describing bridging of dendrites in low concentration alloys has been proposed. Although still in a fairly early stage of development, such coalescence model allows to understand the gradual increase of coherency of the mushy zone as the temperature is lowered. Well controlled tearing tests and other experiments, such as in situ observations of organic alloy solidification, counting of spike density on torn surfaces or welding of bi-crystals, should allow to calibrate the parameters of the coalescence model, in particular the thickness δ of the diffuse interfaces. Once this is done, the next topics to be addressed will be the localization of strains and feeding induced by the presence of remaining liquid films, as well as the percolation of equiaxed grain structures.

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