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Susceptibility of ternary aluminum alloys to cracking during solidification

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A B S T R A C T

The crack susceptibility map of a ternary Al alloy system provides useful information about which alloy compositions are most susceptible to cracking and thus should be avoided by using a filler metal with a significantly different composition. In the present study the crack susceptibility maps of ternary Al alloy systems were calculated based on the maximum $|dT/d(f_S)^{1/2}|$ as an index for the crack susceptibility, where T is temperature and f_S fraction solid. Due to the complexity associated with ternary alloy solidification, commercial thermodynamic software Pandat and Al database PanAluminum, instead of analytical equations, were used to calculate f_S as a function of T and hence the maximum $|dT/d(f_S)^{1/2}|$ for ternary Al-Mg-Si, Al-Cu-Mg and Al-Cu-Si alloy systems. A crack susceptibility map covering 121 alloy compositions was constructed for each of the three ternary alloy systems at each of the following three levels of back diffusion: no back diffusion, back diffusion under a 100 °C/s cooling rate, and back diffusion under 20° C/s. The location of the region of high crack susceptibility, which is the most important part of the map, was shown in each of the nine calculated maps. These locations were compared with those observed in crack susceptibility tests by previous investigators. With back diffusion considered, either under 20 or 100 °C/s, the agreement between the calculated and observed maps was good especially for Al-Mg-Si and Al-Cu-Mg. Thus, the maximum $|dT/d(f_S)^{1/2}|$ can be used as a crack susceptibility index to construct crack susceptibility maps for ternary Al alloys and to evaluate the effect of back diffusion on their crack susceptibility.

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1. Introduction

Cracking during solidification of an alloy is a serious defect, called solidification cracking in welding [1] and hot tearing in casting [2,3]. During welding of an alloy, a semisolid region, called the mushy zone, exists between the weld pool and the completely solidified weld metal. In the mushy zone the solid phase usually exists in the form of columnar dendritic grains, separated by the liquid phase along grain boundaries. The fraction solid f_S continues to increase as temperature T decreases toward the end of the mushy zone. Cracking usually occurs during the terminal stage of solidification (near $f_S = 1$) along grain boundaries, where a small amount of liquid can still exist to form thin liquid films between grains to keep them from bonding together to resist cracking. The prominent RDG model of Rappaz, Drezet and Gremaud [4] was the first hot

tearing model with a physically sound basis. However, the grain boundary, where cracking occurs, was not yet taken into account.

Kou [5] considered three factors at the grain boundary. Suppose two columnar dendritic grains are growing side by side in their axial direction in the mushy zone. The first factor is the lateral separation of grains from each other under tensile strain to cause cracking. The second factor is the lateral growth of grains toward each other to bond together (called bridging) to resist cracking. The third factor is the liquid feeding along the grain boundary to resist cracking. Tension is induced because the mushy zone cannot contract freely under solidification shrinkage and thermal contraction especially when the workpiece is rigid or clamped down tightly. Kou [5] showed during terminal solidification the lateral growth rate of a columnar dendritic grain is proportional to $|d(f_S)^{1/2}/dT|$. Based on the consideration of the space in a volume element positioned between the two grains near $(f_S)^{1/2} = 1$, an equation was derived to serve as the criterion for cracking to occur during solidification. According to the criterion, if the rate of space increase due to grain separation minus the rate of space decrease due to

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lateral grain growth exceeds the rate of space decrease due to liquid feeding, a void can form in the volume element, that is, a crack can form at the grain boundary. For a crack to actually form, however, crack initiation sites are needed, such as folded oxide films, micropores or the external surface of the weld or casting [3,6,7].

Kou [5] further showed that in a plot of T vs. $(f_S)^{1/2}$, the steepness $|dT/d(f_S)^{1/2}|$ near $(f_S)^{1/2} = 1$ can be considered as an index for the crack susceptibility. Since the slope $dT/d(f_S)^{1/2}$ is negative, it is more convenient to deal with its absolute value, that is, the steepness $|dT/d(f_S)^{1/2}|$. The reasons $|dT/d(f_S)^{1/2}|$ near $(f_S)^{1/2} = 1$ can be used as the crack susceptibility index are as follows. First, the higher $|dT/d(f_S)^{1/2}|$ is, the slower the lateral growth rate becomes, that is, the slower the two neighboring columnar grains grow toward each other to bond together and resist cracking. Second, with slow lateral growth the columnar grains can grow very long without bridging. This means the intergranular liquid channel can be very long and hence difficult for liquid to flow through it (due to viscosity of liquid [8]) to feed shrinkage and resist cracking. Third, a long intergranular liquid channel can act as a long sharp notch to promote crack initiation. Thus, $|dT/d(f_S)^{1/2}|$ near $(f_S)^{1/2} = 1$ can be used as an index for the susceptibility to cracking during solidification. To calculate the index, a short interval of $(f_S)^{1/2}$ near $(f_S)^{1/2} = 1$ can be taken as $\Delta(f_S)^{1/2}$ and the corresponding temperature interval can be taken as ΔT to calculate the $|\Delta T/\Delta(f_S)^{1/2}|$ near $(f_S)^{1/2} = 1$ [5]. An alternative is to use the maximum steepness $|dT/d(f_S)^{1/2}|$ if it occurs near $(f_S)^{1/2} = 1$. The curves of T vs. $(f_S)^{1/2}$ for binary Al alloys and commercial wrought Al alloys showed that the maximum steepness $|dT/d(f_S)^{1/2}|$ occurs near $(f_S)^{1/2} = 1$ [5]. Thus, Kou [9] also proposed to use the maximum $|dT/d(f_S)^{1/2}|$ as the crack susceptibility index.

The T-$(f_S)^{1/2}$ curve shows $(f_S)^{1/2}$ continues to increase as T decreases during solidification. Thus, near the end of solidification, there exists a point where liquid still remains but is no longer enough in quantity to form continuous liquid films to separate the grains. At this point, extensive bonding between grains, i.e., extensive bridging, can occur to end the crack susceptibility. Let f_{SB} be the fraction solid at which extensive bridging occurs to end the crack susceptibility. Kou [9] assumed that $(f_{SB})^{1/2} = 0.99$, that is, $f_{SB} = 0.98$. It has been shown that $(f_{SB})^{1/2} = 0.99$ works well for Al alloys [9]. Since the crack susceptibility ends at f_{SB}, if $|dT/d(f_S)^{1/2}|$ continues to increase beyond $(f_S)^{1/2} = 0.99$, the maximum $|dT/d(f_S)^{1/2}|$ before or at $(f_S)^{1/2} = 0.99$ should be used. The RDG model also assumed $f_{SB} = 0.98$ [4].

Kou [9] verified the validity of using the maximum $|dT/d(f_S)^{1/2}|$ as the crack susceptibility index. He showed that a filler metal that is known to reduce the crack susceptibility of an alloy (e.g., filler metal 4145 Al for welding 2014 Al and filler metal 4043 Al for welding 6061 Al and 7075 Al) reduces the maximum $|dT/d(f_S)^{1/2}|$ of the alloy. He also showed that the maximum $|dT/d(f_S)^{1/2}|$ increases in the order of 2219, 2014, 2024, 7075 and 6061, consistent with their ranking in crack susceptibility tests of welds. In addition, he showed that Al casting alloy A206 is higher in the maximum $|dT/d(f_S)^{1/2}|$ than A356, consistent with the well-known higher crack susceptibility of A206 than A356 [5]. Thermodynamic software Pandat [10] and Al database PanAluminum [11] were used to calculate the maximum $|dT/d(f_S)^{1/2}|$, assuming negligible back diffusion of the solute from the interdendritic liquid to the Al-rich dendrites, that is, the Scheil solidification model, which is also called the Gulliver-Scheil solidification model [12,13].

Binary Al-Si, Al-Cu and Al-Mg alloy systems were also analyzed based on the maximum $|dT/d(f_S)^{1/2}|$ as the crack susceptibility index [9]. When the maximum $|dT/d(f_S)^{1/2}|$ is plotted against the solute content, a λ-shaped crack-susceptibility curve with a peak was obtained, consistent with the λ-shaped curves observed in crack-susceptibility tests of binary Al alloys. The location of the peak is the most important information provided by the crack

susceptibility curve because it indicates the solute content most susceptible to cracking. A filler metal with a very different solute content can thus be used to change the weld metal composition and move it away from the peak to reduce the crack susceptibility. The height of the peak, on the other hand, shows the level of the highest crack susceptibility.

Subsequently, Liu and Kou [14] considered back diffusion during solidification by using the following equation of Kurtz and Fisher [15] to calculate f_S as a function of temperature T:

$$f_S = \frac{1}{1 - 2\alpha'k}\left[1 - \left(\frac{T_m - T}{T_m - T_L}\right)^{\frac{1-2\alpha'k}{k-1}}\right] \tag{1}$$

where

$$\alpha' = \alpha\left[1 - \exp\left(-\frac{1}{\alpha}\right)\right] - \frac{1}{2}\exp\left(-\frac{1}{2\alpha}\right) \tag{2}$$

$$\alpha = \frac{4D_S t_f}{\lambda_2^2} \tag{3}$$

k is equilibrium segregation coefficient, T_m melting point of pure Al, T_L liquidus temperature, D_S diffusion coefficient of solute in solid dendrites, t_f local freezing (solidification) time, and λ_2 secondary dendrite arm spacing. In Eq. (1) the solidus and liquidus lines are assumed straight lines, i.e., k is constant. Without diffusion, $\alpha = \alpha' = 0$ and Eq. (1) reduces to the simple Scheil equation [12,13]. Liu and Kou [16] derived simple analytical equations to calculate the locations and heights of the peaks in the crack susceptibility curves of binary alloys. To deal with binary phase diagrams with curved solidus and liquidus lines such as Al-Zn and Al-Sn, Liu and Kou [17] calculated f_S as a function of T by numerical integration.

Liu and Kou [18] quenched the mushy zones and measured the cooling curves during gas-tungsten arc welding of Al alloys. Microstructural examination and composition analysis by EPMA (electron probe microanalysis) showed that significant back diffusion occurred in both 2014 Al (~Al-4.4Cu) and 5086 Al (~Al-4.0 Mg) during solidification.

2. Calculation of T vs. $(f_S)^{1/2}$

Ternary alloy solidification is significantly more complicated than binary alloy solidification. Consequently, simple analytical equations can no longer be used, such as the Scheil equation (no diffusion) [12,13] or the equation of Kurtz and Fisher (back diffusion) [15]. Instead, commercial thermodynamic software Pandat [10] and Al-alloy database PanAluminum [11] were used to calculate T vs. f_S for ternary Al alloys with and without back diffusion. The diffusion coefficients were already provided in PanAluminum based on the data in the handbook of Neumann and Tuijin [19]. In addition to the alloy composition, the cooling rate (through the freezing temperature range ΔT_f) and the secondary dendrite arm spacing λ_2 also need to be specified when using Pandat to calculate T vs. f_S involving back diffusion. Liu and Kou [18] measured the cooling rate during gas-tungsten arc welding (GTAW) of 2014 Al. The workpiece was 102 mm by 102 mm by 2.0 mm, and the welding conditions were direct current electrode negative (DCEN), 100 A welding current, 13 V welding voltage, 4.25 mm/s torch travel speed, and high-purity Ar shielding (i.e., standard welding grade Ar). The cooling rate $\Delta T_f/\Delta t$ was essentially constant at 95 °C/s. They also measured the cooling rate during GTAW of 5086 Al. The workpiece was 102 mm by 102 mm by 1.6 mm, and the welding conditions were identical to those for 2014 Al except the welding current was reduced to 80 A in view of the smaller workpiece