Acta Materialia 122 (2017) 47–59

Contents lists available at ScienceDirect

Acta Materialia

journal homepage: www.elsevier.com/locate/actamat

Full length article

Evidence of back diffusion reducing cracking during solidification

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ARTICLE INFO

Article history:
Received 1 August 2016
Received in revised form
20 September 2016
Accepted 22 September 2016
Available online 30 September 2016

Keywords:
Welding
Solidification
Solidification cracking
Hot tearing
Quenching
Diffusion
Aluminum alloys
Heat transfer
Mass transfer
Fluid flow

ABSTRACT

Al-Mg alloys, despite their wide freezing temperature range ΔT_f, can have good resistance to cracking during solidification. To help understand why, the mushy zone of 5086 Al (~Al-4.0 Mg) was quenched during arc welding and the cooling curve measured to locate the beginning of the original mushy zone (liquidus temperature T_L) and the end (eutectic temperature T_E). Since little eutectic was visible just slightly behind the beginning of the quenched mushy zone, little liquid was here in the original mushy zone, i.e., solidification already ended well above T_E. Since no dendrites were visible, either, and since the highest Mg content measured was well below the maximum solubility in solid Al, C_{SM} (17.5 wt% Mg), microsegregation was very mild here in the original mushy zone. These results suggest significant Mg back diffusion occurred during solidification (because of very high C_{SM}), causing: 1. fraction solid f_S to increase much faster with decreasing temperature T, 2. ΔT_f to narrow down, and 3. dendritic grains to bond together extensively ($f_S \approx 1$) to resist intergranular cracking earlier (well above T_E). Since $|d(f_S)/dT|$ increased, $|dT/d(f_S)^{1/2}|$ decreased to decrease the crack susceptibility index, i.e., the maximum $|dT/d(f_S)^{1/2}|$. All these changes reduce the crack susceptibility. For comparison, 2014 Al (~Al-4.4Cu) was also quenched during arc welding. At the end of the quenched 2014 Al mushy zone, continuous eutectic, dendrites and microsegregation were all very clear. Thus, solidification ended at T_E and thin liquid films still separated grains at the end of the original mushy zone to allow intergranular cracking. Calculated T-$(f_S)^{1/2}$ curves showed the index is reduced significantly by back diffusion in Al-4.0 Mg (~5086 Al) but not in Al-4.4Cu (~2014 Al).

1. Introduction

Cracking of an alloy during solidification is a serious defect. In welding it is called solidification cracking [1]. In casting it is called hot tearing [2,3]. Al alloys are known to be susceptible to cracking during solidification in welding and casting [1–3]. It has long been a puzzle why Al-Mg alloys can have good weldability despite their very wide freezing temperature range. Because a liquid alloy freezes over a temperature range, a weak semisolid region, called the mushy zone, exists between the liquid pool and the completely solidified alloy. For most Al alloys the freezing temperature range is $\Delta T_f = T_L - T_E$, where T_L is the liquidus temperature and T_E the eutectic temperature. An alloy with a wider ΔT_f can be expected to have a wider weak mushy zone and hence a higher susceptibility to solidification cracking. Cracking occurs near the end of the mushy zone, where a small amount of liquid still can exist as thin liquid films along grain boundaries to keep grains from bonding together firmly to resist intergranular cracking under tension. Tension is induced when free contraction due to solidification shrinkage and thermal contraction is obstructed, e.g., by rigid workpiece in welding or by rigid mold walls in casting.

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] developed a model focusing on the grain boundary. Consider two columnar dendritic grains growing side by side in the same axial growth direction. The following three factors can be relevant: 1. lateral separation of the grains from each other under tension to cause cracking, 2. lateral growth of the grains toward each other to bond together to resist cracking, and 3. liquid feeding along the grain boundary in the opposite direction of axial growth to resist cracking.

Consider a volume element Ω between the two grains near the end of the mushy zone, where cracking occurs. The net volume expansion rate of Ω is that due to lateral grain separation minus

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http://dx.doi.org/10.1016/j.actamat.2016.09.037

that due to lateral grain growth. The net volume flow rate of liquid entering Ω, on the other hand, is the volume flow rate of liquid entering Ω minus that leaving Ω. When the net volume expansion rate of Ω exceeds the net volume flow rate of liquid entering Ω, a void (i.e., crack) can occur in Ω [5] if crack initiation sites are available, such as folded oxide films, micropores or the external surface of the weld or casting [3,6,7].

Kou [5] showed that near $f_S = 1$ the lateral growth rate of grains toward each other to bond together to resist cracking is proportional to $|d(f_S)^{1/2}/dT|$, where T is temperature and f_S fraction of solid. Kou [5] pointed out that $|dT/d(f_S)^{1/2}|$ near $(f_S)^{1/2} = 1$ can be considered as an index for the crack susceptibility for the following three reasons. First, the higher $|dT/d(f_S)^{1/2}|$ is, the slower the two neighboring columnar grains can grow toward each other to bridge 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. For binary Al alloys and commercial wrought Al alloys, the maximum $|dT/d(f_S)^{1/2}|$ occurs near $(f_S)^{1/2} = 1$ [5]. Thus, the maximum $|dT/d(f_S)^{1/2}|$ can also be used as the crack susceptibility index [9].

The validity of using the maximum $|dT/d(f_S)^{1/2}|$ as the index for the crack susceptibility was verified [9]. The curves of T vs. $(f_S)^{1/2}$ can be calculated based on the composition of an alloy or a weld using Pandat [10], PanAluminum [11] and the Scheil solidification model (no diffusion), which is also called the Gulliver-Scheil solidification model [12,13]. First, a filler metal that reduced the maximum $|dT/d(f_S)^{1/2}|$ of an alloy was shown to actually reduces its solidification cracking in welding, e.g., filler metal 4145 Al for welding 2014 Al and filler metal 4043 Al for welding 6061 Al and 7075 Al. Second, the maximum $|dT/d(f_S)^{1/2}|$ increased in the order of 2219, 2014, 2024, 6061 and 7075 Al, consistent with their ranking in crack susceptibility tests.

However, the predicted crack susceptibility of Al-Mg alloys was too high, higher than that of Al-Cu alloys. This is because the maximum freezing temperature range ($\Delta T_f = T_m - T_E$, where T_m is the melting point of Al 660 °C) is much wider for Al-Mg (210 °C) than for Al-Cu (112 °C) and this significantly increases the maximum $|dT/d(f_S)^{1/2}|$ of Al-Mg alloys. However, Cross et al. [14] and Rosenberg et al. [15] showed less cracking during solidification for Al-Mg alloys than Al-Cu alloys, even though Al-Mg alloys are expected to be more crack susceptible because of their much wider ΔT_f. Instead of using the Scheil model and assuming no back diffusion, the following equation of Kurtz and Fisher [16] was used to include back diffusion:

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

where

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

$$\alpha = \frac{4D_S t_f}{\lambda_2^2} \quad (3)$$

α' is the diffusion parameter, k 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].

Curves of T vs. $(f_S)^{1/2}$ were calculated using Eq. (1). It was found that the maximum $|dT/d(f_S)^{1/2}|$, that is, the crack susceptibility index, is reduced by back diffusion. The higher the diffusion parameter α' is, the greater the reduction [17]. The crack susceptibility curve for a binary alloy system is a curve of the crack susceptibility vs. the solute content. The curve is λ-shaped, that is, with a peak at an intermediate solute content [1]. When the maximum $|dT/d(f_S)^{1/2}|$ is plotted against the solute content of a binary Al alloy, a λ-shaped crack-susceptibility curve was obtained, consistent with the λ-shaped curves observed in crack-susceptibility tests of binary Al alloys. The peak of the crack-susceptibility curve shows the solute content most susceptible to cracking and the level of the crack susceptibility. With the back diffusion parameter α' raised from 0 (no diffusion) to 0.025, 0.050 and 0.075, for instance, the peak of the crack-susceptibility curve decreased in magnitude and shifted to a higher solute content. The decrease was much greater with Al-Mg alloys than Al-Cu alloys.

The objectives of the present study were to: 1. develop an experimental procedure to quench the mushy zone during welding, and 2. use the quenched-in microstructure and microsegregation of the mushy zone to explain why Al-Mg alloys, despite their much wider freezing temperature range ΔT_f, can be less susceptible to solidification cracking than Al-Cu alloys.

2. Experimental procedure

2.1. Welding

2000-series and 5000-series Al alloys are widely used commercial wrought Al alloys. The former are essentially binary Al-Cu alloys with about 2–6 wt% Cu and the latter essentially binary Al-Mg alloys with 1–5 wt% Mg. 2014 Al (~Al-4.4Cu) was selected as a representative alloy for Al-Cu alloys, so was 5086 Al (~Al-4.0 Mg) for Al-Mg alloys. The chemical compositions of the two alloys used are shown in Table 1. The workpiece was in the form of a thin sheet, 102 mm by 102 mm by 2 mm in the case of 2014 Al and 102 mm by 102 mm by 1.6 mm in the case of 5086 Al. Before welding, the workpiece surface was cleaned with a stainless steel brush to remove oxide films and then rinsed with acetone.

Bead-on-plate welding was conducted without a filler metal using gas-tungsten arc welding (GTAW), which is also called tungsten-inert gas (TIG) welding. The workpiece was welded along its centerline, starting from about 10 mm away from its leading edge and proceeding inward. The welding conditions were as follows: direct current electrode negative (DCEN), 4.25 mm/s torch travel speed, 13 V welding voltage, and pure Ar gas shielding. The welding current was 100 A for 2014 Al and 80 A for 5086 Al. Since 5086 Al (1.6 mm thick) was thinner than 2014 Al (2 mm thick), the welding current was reduced from 100 A to 80 A in order to keep the weld width close to that of 2014 Al.

In GTAW, weld penetration is deeper with the DCEN polarity [1], which is desirable in thick-plate welding. Although the present study involved thin-sheet welding, DCEN was still selected because

Table 1
Compositions of workpiece in wt%.

	Cu	Mg	Si	Mn	Cr	Fe	Al
2014 Al	4.40	0.56	0.83	0.77	0.01	0.26	balance
5086 Al	0.04	4.00	0.22	0.26	0.11	0.35	balance