

Applying Solidification Theory to Aluminum Weldability and Consumable Development

A review of weldability studies performed for aluminum alloy development, including theoretical aspects for grain refinement and solidification crack growth

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Abstract

One defect encountered in the fusion zone when welding aluminum alloys involves solidification cracking (i.e., the tearing apart of grain boundary liquid films at the trailing edge of the weld pool). This problem can often be mitigated by the proper selection of filler metal. Two key engineering examples, one aerospace and one maritime, where this has occurred were examined in terms of alloy development to achieve optimum mechanical properties while maintaining weldability. Specifically, base metal/filler metal systems susceptible to cracking were examined in terms of filler metal dilution. A mechanism for crack growth was presented based upon critical strain rate. Conditions needed for improved weldability through grain refinement were defined based upon the columnar-to-equiaxed solidification theory.

Keywords

- Aluminum
- Alloy Development
- Weldability
- Solidification Cracking
- Grain Refinement

Introduction

Classical solidification theory has traditionally been developed and applied to the casting of metal alloys. Building upon this knowledge and applying it to weld solidification to explain observed microstructures and weldability has been the goal of this work. Welds experience unique solidification boundary conditions that differ from castings, resulting in characteristic behavior. One of these weld characteristics is that the solidification rate is linked to weld travel speed, whereas in castings, it is linked to the rate of heat removal. In addition, welds typically experience higher temperature gradients and growth rates, resulting in finer microstructures.

The addition of consumable filler metals during welding provides the opportunity to vary weld metal composition from the base metal composition, a method that improves weld strength and weldability. Wrought alloys are seldom designed for good weldability. Thus, it is incumbent upon the welding engineer to pair the base metal alloy with the best possible filler alloy that will provide the desired outcome, be it weldability, strength, or corrosion resistance.

Examples will be provided in this paper where weldability testing was used in alloy development to evaluate and improve the weldability and mechanical behavior of high-performance aluminum alloy joints. Understanding the critical conditions needed for solidification cracking and improved weldability with grain refinement will help avoid problems when welding crack-susceptible alloys.

Background

The 1980s saw a keen interest in the development of lithium-containing aluminum alloys to achieve improved strength/weight ratios for aerospace usage. There were two

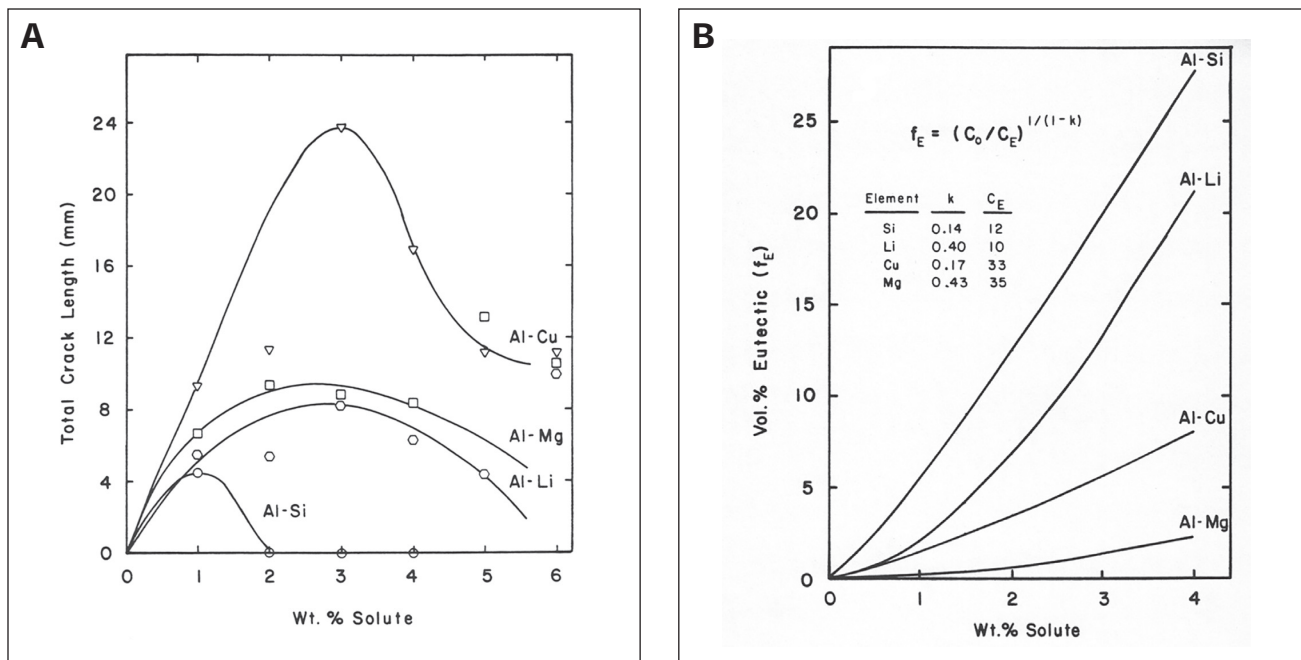


Fig. 1 — A — Vareststraint weldability comparison of cast aluminum binary alloys (0.5% augmented strain); B — associated amounts of interdendritic eutectic calculated from the Scheil equation (Ref. 6).

Table 1 — Comparison of Cryo-Tank Aluminum Alloys. Nominal Major Alloying Elements in wt-% (Refs. 8, 9)

	Cu	Li	Mg	Ag	Zr	Ti	V
Al 2219	5.8–6.8	0	0.02	0	0.10–0.25	0.02–0.10	0.05–0.15
Weldalite™ 049	5.4	1.3	0.4	0.4	0.14	0	0

different approaches taken to achieve this: 1) low Cu-Mg and high Li, to achieve low density (e.g., Alloys O1420 and 8090) vs. 2) high Cu-Mg and low Li to achieve high strength (e.g., Alloys 2090, 2094, 2095, and 2195).

To lay the groundwork, the Vareststraint test (Ref. 1) was used to compare different binary aluminum systems. It was encouraging that test results showed that Al-Li alloys displayed a favorable behavior (Fig. 1A), markedly better than Al-Cu and similar to Al-Mg (Ref. 2). This observed peak in cracking susceptibility is typical of what has been found in other weldability (Ref. 3) and castability studies (Refs. 4, 5). To explain this variation in behavior with alloy addition, conventional thought has been that at low alloy additions, as solute is increased, continuous liquid films begin to form interdendritically and along grain boundaries. As these thin films prevent bonding between dendrites, the film experiences the strain due to welding. At some point, however, as the amount of eutectic increases with alloy content (Fig. 1B),

it becomes possible to more easily backfill any cracks that may form (Ref. 7).

Al-Si is found to be particularly resistant to cracking because its characteristic flowability (low viscosity) and large quantity of eutectic (Fig. 1B) provide an ability to backfill cracks. This has proven useful as filler metal (e.g., Al 4047 for brazing and Al 4043 for welding difficult alloys). Unfortunately, its lack of solid solution and precipitate strengthening result in relatively low strength while coarse Si particles result in low ductility.

High Strength Aluminum Aerospace Alloy

The external tank of NASA's space shuttle was a candidate for reduced weight due to the need to propel into space vast amounts of material for constructing Spacelab. This external fuel tank, consisting of two stacked cryogenic tanks (liquid

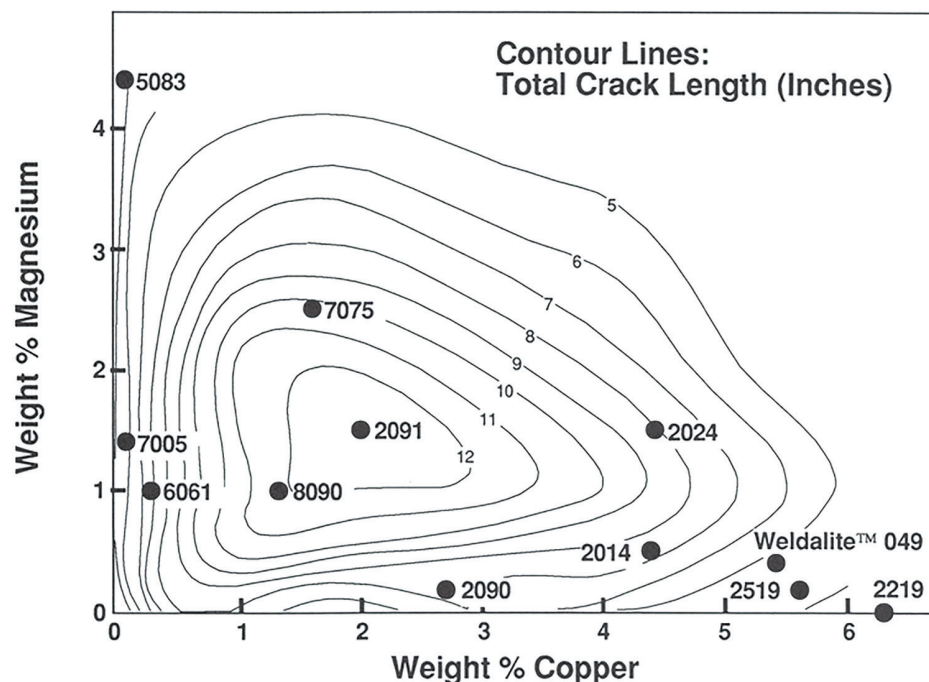


Fig. 2 — Ring casting castability profile for Al-Mg-Cu ternary alloys with contour lines showing crack length in inches (Ref. 4). Superimposed are modern wrought alloys comparing potential weldability (Ref. 10).

Table 2 — Nominal Primary Alloy Compositions for Low-Cu Werdalite Family of Alloys in wt-% (Ref. 15)

	Cu	Li	Mg	Ag	Zr
Al 2094	4.7	1.3	0.4	0.4	0.14
Al 2095	4.3	1.3	0.4	0.4	0.14
Al 2195	4.0	1.0	0.4	0.4	0.14

O₂ and liquid H₂), was initially constructed by arc welding barrel sections and end caps made from Alloy 2219. Alloy 2219 is basically an Al-6 wt-% Cu binary alloy specifically developed by NASA for exceptional weldability related to its high copper content (recall Fig. 1). The contender alloy considered to replace Alloy 2219 was Werdalite™ 049, developed by J. Pickens (Ref. 8) as a high-strength alloy based upon T₁ (Al₂CuLi) precipitation (yield strength 2219-T87: 400 MPa; Werdalite™ 049-T8: 690 MPa). From Werdalite 049's composition, given in Table 1, it follows that Cu and Li are present for the T₁ phase, Mg is added for solid solution strengthening, Ag is added to enhance precipitation, and Zr is added

for grain control. It is important to note that Werdalite 049 is high in Cu, which contributes to its good weldability but also adds to its density (density 2219: 2.84 g/cm³; Werdalite 049: 2.60 g/cm³).

Werdalite 049 is compared to other Al-Cu-Mg alloys in the ring castability diagram of Fig. 2, where modern alloys have been superimposed on the crack length data from Pumphrey and Moore (Refs. 4, 10). This diagram is shown to suggest a possible comparison of weldability, assuming that Cu and Mg in combination exert a major influence over cracking susceptibility. For example, commercial Alloys 7075, 2024,

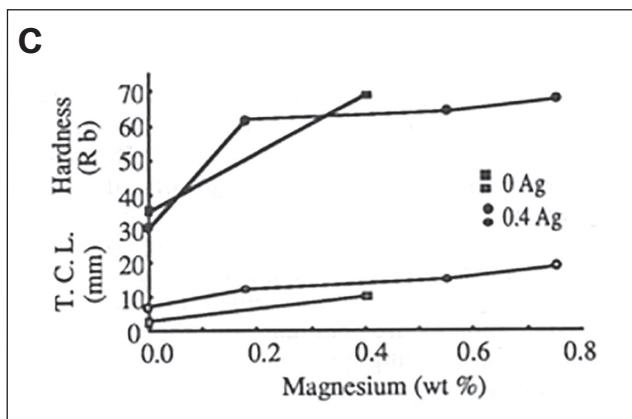
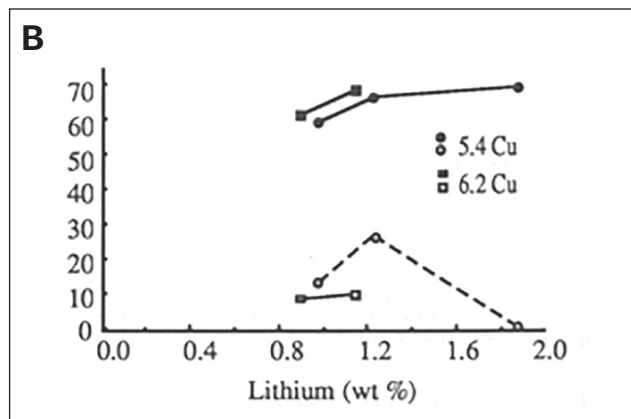
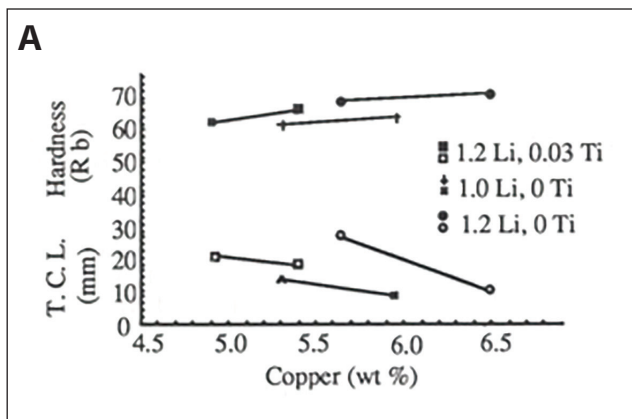


Fig. 3 — Varestraint (total crack length) and weld metal Rockwell-B comparisons of Weldalite™ O49 for small variations in Cu, Li, and Mg from the original alloy composition (Ref. 11).

and 2014 contain both Cu and Mg and are known to exhibit high susceptibility to cracking.

To determine more precisely how alloy additions affect weldability, Varestraint tests were performed on experimental variants of Weldalite O49 extruded and heat treated to peak hardness. The results, shown in Fig. 3, indicate that Cu, Li, and Mg additions all increase weld metal hardness, but in particular, Cu additions improve weldability (i.e., less cracking) while Mg additions reduce it (Ref. 11). It is interesting to note that a special high-strength Li-containing filler alloy was patented based upon these results: 4.5 Cu + 1.0 Li + 0.4 Ag wt-% (Ref. 12). However, filler Alloy 2319 was deemed more appropriate, particularly when considering the propensity for hydrogen pickup in Li-containing alloys.

As part of the Weldalite O49 weld development, evaluation of mechanical properties was pursued (Ref. 13). Figure 4A gives a comparison of hardness traverses for the processes evaluated for welding Weldalite O49, including gas metal arc (synergic and plasma), single- and double-torch gas tungsten arc (GTA) (direct current and variable polarity), variable polarity plasma arc, laser beam, and electron beam. In all cases, there was significant loss of strength in the weld metal.

Surface crack tension (SCT) tests were performed at room and cryogenic temperatures (0°, -196°, and -269°C)

on Weldalite O49 coupons joined using different welding processes with the crack centered in the weld metal (Fig. 4B). Results compared in Fig. 4C show K_{Ic} fracture toughness values that varied between 23 and 38 MN m^{-3/2} depending upon temperature, welding process, and corresponding dilution with 2319 filler metal. This represents an improvement over 2219/2319 welds (bottom line in Fig. 4C) with toughness values between 18 and 26 MN m^{-3/2}.

Nevertheless, there was a concern about this new alloy because cracks always migrated to the weld interface, propagating through a fine-grain region unique to this alloy — Fig. 5. While fine equiaxed grains at the weld interface are not a common occurrence, Gutierrez and Lippold (Ref. 14) found that they were nucleating on ZrAl₃ particles (dispersoids) released from the base metal. The high Cu content of Weldalite O49 generated a large amount of grain boundary eutectic CuAl₂ phase, which works against weld toughness. Also, the large difference in hardness between the weld metal and heat-affected zone likely resulted in deformation concentration at the weld interface. Because of this concern, variants of Weldalite O49 were made for evaluation having lower Cu content (Table 2). Mg content was kept low based upon the earlier results from Fig. 3.

Due to concern over what reduced Cu content may do to weldability, testing was performed to compare the 4.7 wt-% Cu variant of Weldalite O49 (Alloy 2094) against other common base metal/filler metal aluminum alloys. A custom Trans-Varestraint (Ref. 16) testing machine with GTA, cold wire feed capability was constructed for this task. These welds were made using a bead on plate with an appropriate filler alloy. Heat input was varied to maintain an approximately equal-sized weld pool. The augmented strain applied was 0.5%. The amount of filler dilution can be assumed to be low (approx. 30%). Results shown in Fig. 6 suggest that the low-Cu Alloy 2094 variant is not as weldable as other alloy pairs known to be weldable from general experience, in particular 2219/2319, 5086/5356, and 6061/4043 (Ref. 17).

Nevertheless, the alloy finally selected for use on the external tank, Al 2195 (4.0 wt-% Cu), was even lower in Cu than Alloy 2094. Although weldability tests do not necessarily relate directly to real-world applications, later experience during external tank weld fabrication at the NASA Michoud Assembly Facility in Louisiana confirmed that this low Cu alloy was susceptible to cracking, particularly during weld repair. This was one of the reasons arc welding was replaced

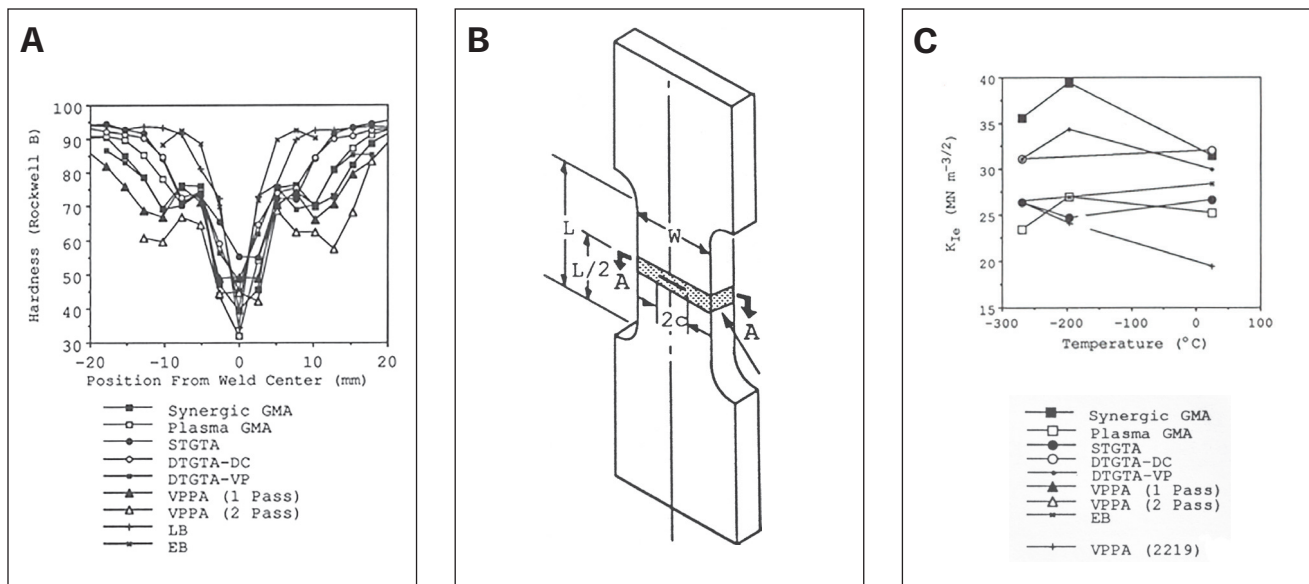


Fig. 4 — Properties of Weldalite 049/2319 weldments made using different welding methods: A — Rockwell-B hardness traverses across weldment; B — schematic of a SCT specimen showing precrack placed in weld metal (Ref. 13); C — cryogenic toughness measurements from SCT tests. The bottom line represents 2219/2319 weldment.

Table 3 — Nominal Primary Alloying Element Compositions for Ship Construction in wt-% (Ref. 9)

	Mg	Zn	Mn	Cu	Ti	Zr
Al 5083	4.0–4.9	0.25	0.4–1.0	0.10	0.15	0.00
Al 7108	0.7–1.4	4.5–5.5	0.05	0.05	0.05	0.12–0.25

by friction stir welding of vertical panel joints at the Michoud facility (Ref. 18).

High-Strength Aluminum Maritime Alloy

The use of aluminum in fast catamaran ferries that hydrofoil above the water is important for light weight, just as in aircraft. Such ferries are of importance to Norway for shuttling cars and people along its vast coastline. The commonly used aluminum alloy in seawater applications is rolled Al 5083 plate, which achieves moderate strength from cold work plus good corrosion resistance and weldability. A Norwegian consortium (SINTEF, NTNU, Hydro Aluminium, Kvaerner, and DNV) was formed in the 1990s to, among other things, make use of higher-strength extruded Alloy 7108 (Table 3). However, of concern was its weldability and joint strength, known to be weak in the weld metal. In particular, the desire was to assemble car decks for ferries by welding extruded box-frame panels of Al 7108. Since car decks are not immersed

directly in salt water, it was believed that corrosion would not be an overriding concern.

While extruded Al 7108-T6 has a higher yield strength than Al 5083-H321 (290 MPa vs. 200 MPa), the strength of its welded joints is limited by the strength of the weld metal, which is dependent upon the filler alloy selected. Al 7108 is considered weldable when using an Al-Mg filler metal (Table 4) to improve its weldability. This is demonstrated in the castability diagram of Fig. 7, where it is observed that greater dilution with Al-Mg filler alloys shifts the cracking susceptibility downhill from peak susceptibility. It is clear from this graph that high Zn content does not promote weldability, which is assumed to follow a similar trend as castability.

To characterize Al 7108 weldments made with different filler alloys, GTA welds with cold wire feed were made on 5-mm-thick extruded plate heat treated to peak (T6) hardness (Ref. 21). A butt joint configuration was used giving about 20% filler dilution. The weld metal was the weakest link in the joint as demonstrated by the hardness traverses of Fig. 8. Use of Al 4043 resulted in the weakest weld metal, whereas Al 5356 had the strongest relative hardness.

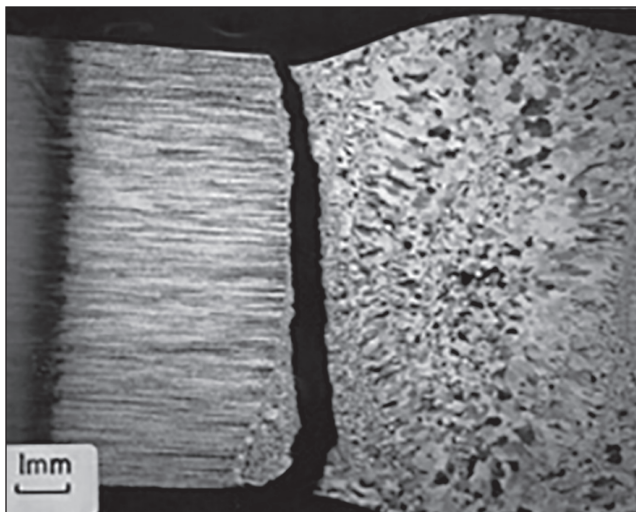


Fig. 5 — The cross section of a Weldalite 049/2319 weldment tensile specimen showing a fracture in the weld metal along the fine-grained weld interface (Ref. 13).

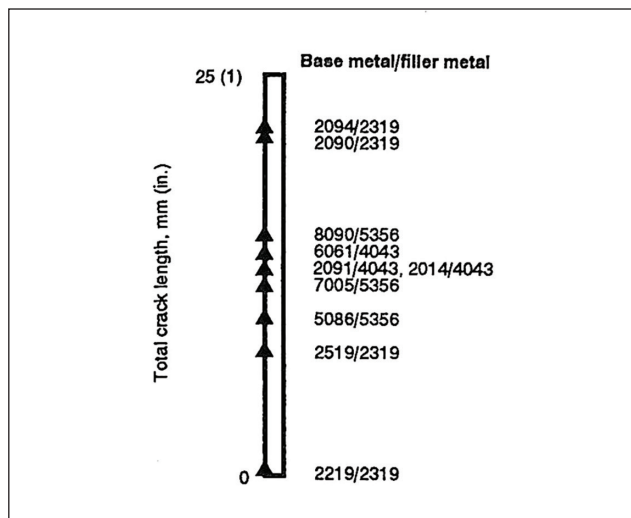


Fig. 6 — Trans-Varestraint test comparison for a variety of aluminum alloys using 0.5% augmented strain and 30% filler dilution (Ref. 17).

Weld Strength Study

A series of simulated weld metal experimental alloys were chill cast having controlled amounts of Mg and Zn to determine the optimum combination for strength (Ref. 22). The legend in Fig. 9 shows the range of compositions examined,

each belonging to one of three Zn/Mg ratios (1.5, 2.5, and 4.5). Because alloy specimens were observed to naturally age over approximately 300 h, all tensile bars were tested after natural aging for nine to 11 weeks. The 2.5 ratio represents a quasi-binary condition found in the Al-Mg-Zn ternary diagram. Alloys with a low Zn/Mg ratio were found to form a

Table 4 — Nominal Primary Alloy Composition for Filler Alloys Suitable for Welding Al 7108 in wt-% (Refs. 19, 20)

	Si	Mg	Mn	Fe	Zn	Ti	Zr
4043	4.5–6.0	0.05	0.05	0.80	0.10	0.20	—
5654	0.45**	3.1–3.9	0.01	—	0.20	0.10	—
5039	0.1	3.3–4.3	0.3–0.5	0.40	2.4–3.2	0.10	—
5183	0.38	4.3–5.2	0.5–1.0	0.40	0.25	0.15	—
5356	0.25	4.5–5.5	0.05–0.2	0.40	0.10	0.10	—
Safra 66*	0.4	5.5–6.1	0.8–1.0	0.20	0.20	0.15	0.08–0.12

*trade name; **Fe + Si

T-phase ($\text{Mg}_3\text{Zn}_3\text{Al}_2$) eutectic, replaced by an η -phase (MgZn_2) eutectic at higher Zn/Mg ratios but in smaller amounts. The η -phase in precipitate form (η') is known to be the primary strengthening phase in Al-Zn-Mg alloys.

Tensile test results are given in Fig. 9 showing the peak strength was achieved with test alloy 18 having the high Zn/Mg ratio and 8.18 Zn and 1.82 Mg, values significantly above those of the Al 7108 base alloy (Ref. 22). From this, it appears that a high-Zn/low-Mg filler alloy would be ideal (e.g., 8 wt-% Zn and 2 wt-% Mg). This composition sits near the peak of the castability curve (Fig. 7), and concern over potential weldability was addressed by examining grain refinement. It is well established that the castability of aluminum alloys can be improved by adding grain-refining elements, commonly titanium and boron for aluminum alloys.

Effect of Grain Refinement on Weldability of Al 7108

The grain-refiner Ti is found as a minor addition to most wrought aluminum base metal and filler metal alloys and is usually limited to a maximum of 0.15 wt-% in Aluminum Association (AA) standards (Refs. 9, 19). A Ti-B master alloy (Tibor®) is typically added during the melt processing stage, just before pouring ingots to avoid dissolving of nucleating particles (fading). Different ratios of Ti/B can be used, but 3/1 and 5/1 are most common. While some remains in solution, Ti is present in the master alloy as aluminide TiAl_3 particles, ideal for nucleating aluminum grains according to the peritectic reaction: $\text{TiAl}_3 + \text{Al}_L = \text{Al}_S$ (Ref. 23). Partitioning of Ti is also very effective at increasing undercooling at the liquid-solid interface, as will be demonstrated later. Boron is believed to enhance the efficiency of nucleation, possibly by forming insoluble TiB_2 particles that are then coated by TiAl_3 (duplex nucleation theory) (Ref. 24). Because boron is used in such small quantities (i.e., < 0.03 wt-%), it is not listed in AA standards even though its presence may be critically important.

Other transition elements besides Ti are also effective in providing grain refinement, including V, Zr, and Sc, all of which form aluminides at high temperatures: V and Zr by means of a peritectic reaction and Sc by forming a eutectic (Ref. 25). In wrought products, these elements also form coarse precipitates (i.e., dispersoids) that are effective in controlling grain growth during annealing, hot rolling, and extrusion.

A circular patch test (Ref. 26) was performed on cast coupons made by treating melted 7108 Alloy with varying amounts of Tibor aluminum master alloy (Al + 5 wt-% Ti + 1 wt-% B). A scandium master alloy (Al + 2 wt-% Sc) was

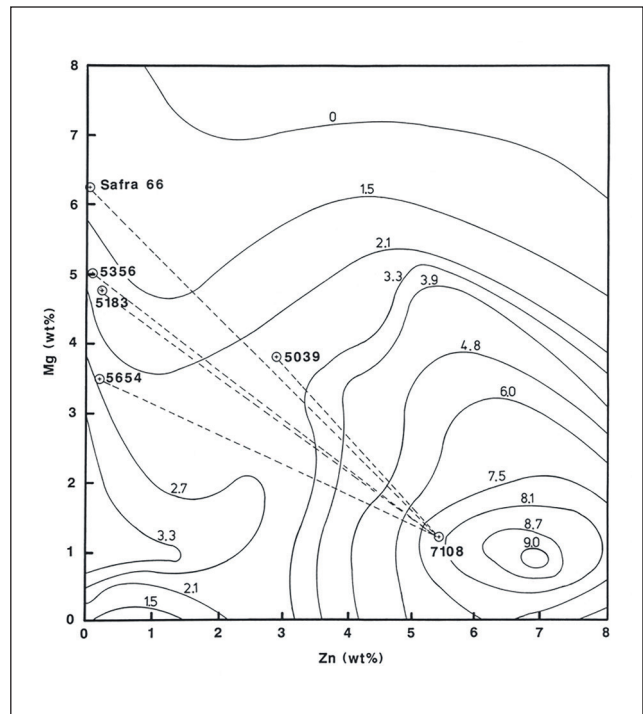


Fig. 7 — Ring castability test results showing crack length contours in inches (Ref. 5). Superimposed on this chart is the Al 7108 alloy and dilution paths for various Al-Mg filler alloys (Ref. 21).

also added to a separate series of cast coupons to allow for a comparison. This test utilized a lathe whereby a square coupon (52 × 52 × 10 mm thick) was bolted into a water-cooled copper platen attached to the lathe head stock with a bolt in each corner of the coupon (Ref. 27). A GTA torch was fixed to the tool holder, allowing a 40-mm-diameter autogenous weld to be made upon rotating the head stock. Weld parameters are given in Table 5.

Results given in Fig. 10A show a remarkable decrease in grain size with Tibor additions but less for Sc additions. For Tibor, the grain size was reduced from 300 to 30 μm at 0.1 wt-% Ti (0.02 wt-% B), as observed in the cross-sectional micrograph of Fig. 10B. No cracking was observed for any of the coupons with Tibor additions.

Significant cracking was measured in terms of both total crack length (TCL) and maximum crack length (MCL) for the Sc additions (Fig. 10C), eventually going to zero at 0.2 wt-% Sc. Figure 10D shows intergranular cracking for an Al 7108 coupon with zero grain refiner extending into the base metal.

Table 5 — Welding Parameters Used in Circular Patch Test (Ref. 27)

Polarity	Shielding Gas	Current	Travel Speed
DCEN	Helium	130 A	4 mm/s

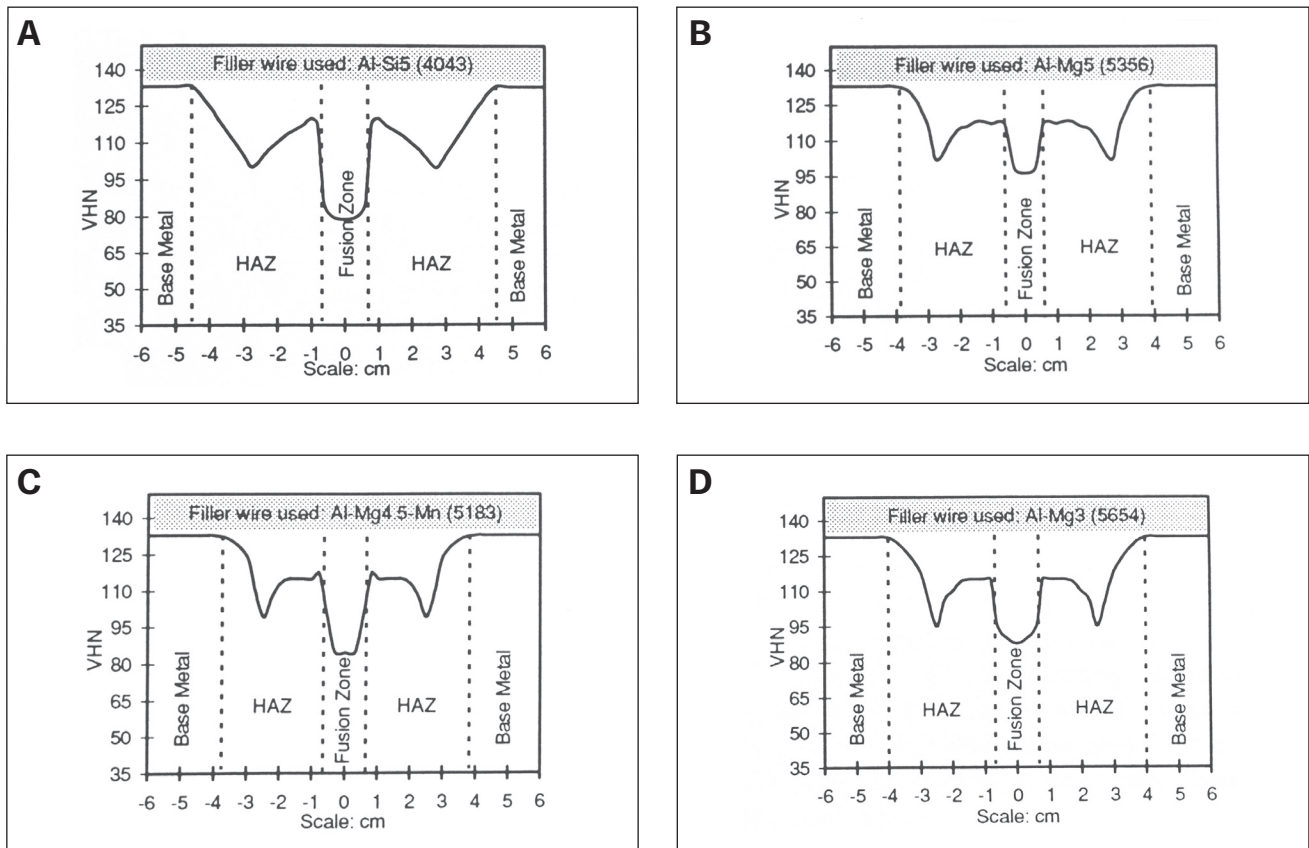


Fig. 8 — Vickers hardness traverses for Al 7108 weldments made using four different aluminum filler alloys: A — 4043; B — 5356; C — 5183; D — 5654 (Ref. 21).

It is interesting to note that most commercial base metal and filler metal alloys contain some Ti, ranging from 0 to 0.15 wt-% max (Ref. 9). Depending upon how much Ti and B are actually in the base metal and filler metal, together with the degree of weld dilution, it is uncertain if the critical amount needed for optimum refinement (e.g., 0.1 wt-% Ti + 0.02 wt-% B) can be reached. This problem is further complicated by possible variations in the number of nucleating particles associated with a given amount of added grain refiner, which can vary with processing (e.g., breaking up of particles during wire drawing as well as dissolution of particles in the weld pool) (Refs. 28, 29).

Conditions Needed for Grain Refinement

The remarkable effect of reducing grain size and improving weldability by adding Tibor was examined further to better understand the mechanisms involved.

Two conditions are required to nucleate new grains in the weld metal: suitable particles upon which to nucleate and sufficient undercooling to activate nucleation, particular to the particle. Suitable particles are those that are structurally close to the aluminum crystal and have similar atom spacing and bonding (Ref. 30). Undercooling is affected by the partitioning of alloying elements and the solidification parameters

G (thermal gradient) and R (growth rate), both of which tend to be large for weld solidification in comparison to castings.

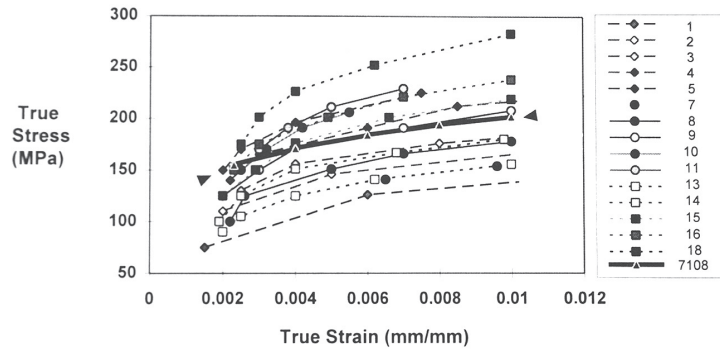
Weld Metal Undercooling

The ratio G/R has traditionally been used as an indication of undercooling in castings, where low values of G/R indicate high undercooling ahead of the solid-liquid interface. This came from defining the critical temperature gradient G^* as the tangent to the liquidus curve ahead of a planar solid/liquid interface (Ref. 31):

$$G^* = -\frac{R}{D} \frac{m_L C_o (1-k)}{k} \quad (1)$$

where D is the solute diffusivity in liquid, m_L is the liquidus slope, C_o is the solute concentration, and k is the equilibrium partition coefficient. Gradients below this critical value will experience constitutional undercooling. Dividing through by R in Equation 1 gives a ratio that is a constant dependent only upon material properties.

For weld solidification, a small G/R ratio and concurrent grain nucleation are most often observed near the weld center, where growth rate is highest and the thermal gradient is lowest, as demonstrated by Savage, Lundin, and Aronson (Ref. 32). But fine grains in some alloys can additionally be found along the weld interface, as has been reported for Zr



Alloy	1	2	3	4	5	7	8	9	10	11	12	13	14	15	16	17	18	
Mg	2.00	2.40	2.80	3.20	3.60	1.43	1.71	2.00	2.29	2.57	2.86	0.91	1.09	1.27	1.45	1.64	1.82	
Zn	3.00	3.60	4.20	4.80	5.40	3.57	4.29	5.00	5.71	6.43	7.14	4.09	4.91	5.73	6.55	7.36	8.18	
Mg+Zn	5	6	7	8	9	5	6	7	8	9	10	5	6	7	8	9	10	
Zn/Mg	1.5					2.5					4.5							
variant	low Zn/Mg					quasi-binary					high Zn/Mg							

Fig. 9 – Comparison of tensile test results for simulated weld metal that was chill cast and machined into cylindrical tensile bars: 7.5 mm dia. × 25.4 mm gauge. The legend provides nominal compositions for the 19 experimental alloys examined (Ref. 22).

Table 6 – Comparison of Phase Diagram Parameters for Aluminum Binary Alloys (Ref. 37)

Parameter	Binary System										
	Al-Si	Al-Fe	Al-Cu	Al-Mn	Al-Mg	Al-Cr	Al-Ni	Al-Zn	Al-Ti	Al-V	Al-Zr
Liquidus Slope m_L in $K \text{ wt}\%^{-1}$	-6.6	-3.0	-3.4	-1.6	-6.2	3.5	-3.3	-1.6	33.3	10	4.5
Partition Coefficient k_f	0.11	0.02	0.17	0.94	0.51	2.0	0.007	0.4	7.8	4.0	2.5
$m_L(k_f - 1)$ in $K \text{ wt}\%^{-1}$	5.9	2.9	2.8	0.1	3.0	3.5	3.3	1.0	220	30	6.8

containing aluminum-lithium (Ref. 14) and aluminum-zinc alloys (Ref. 33). It therefore appears that the application of the G/R ratio in welds to predict grain refinement is problematic.

It is appropriate to use an analytical expression for undercooling at the dendrite tip ΔT as defined by Burden and Hunt (Ref. 34):

$$\Delta T = \frac{GD}{R} + A\sqrt{R} \quad (2)$$

where A is a material constant and θ is curvature undercooling. The first term in Equation 2 originates from constitutional undercooling ahead of a flat surface, and the second term comes from kinetic undercooling associated with curvature. For most castings, the first

$$A = 2\sqrt{-2\left(\frac{\theta m_L (1 - k) C_o}{D}\right)} \quad (3)$$

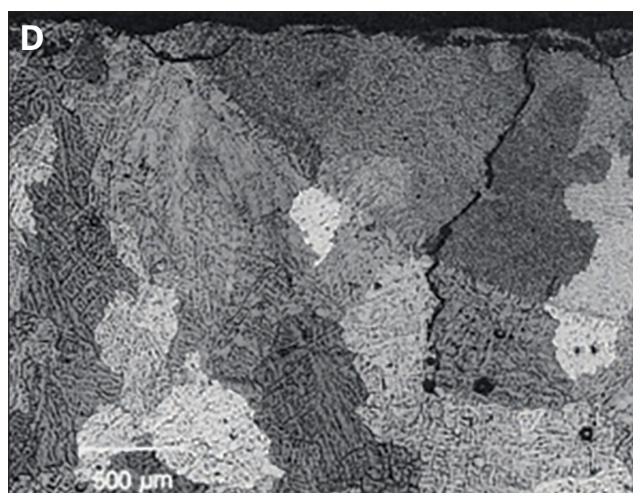
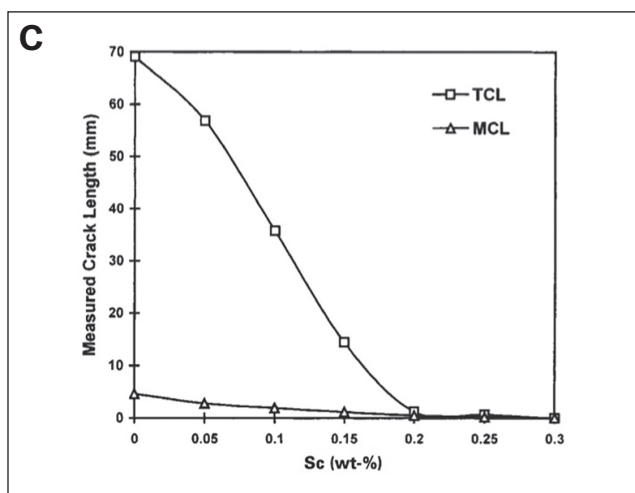
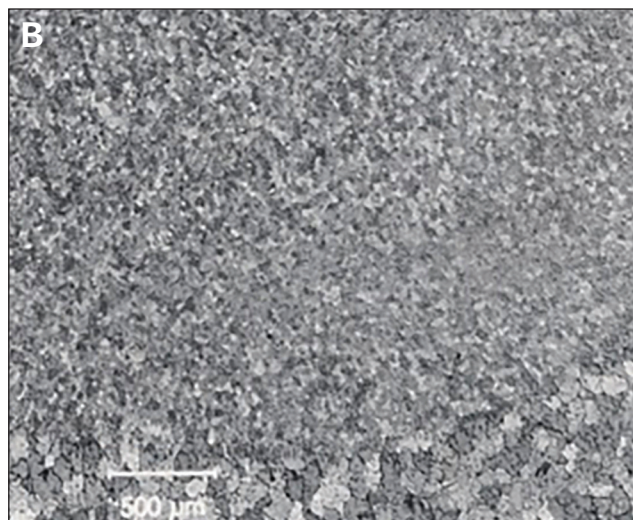
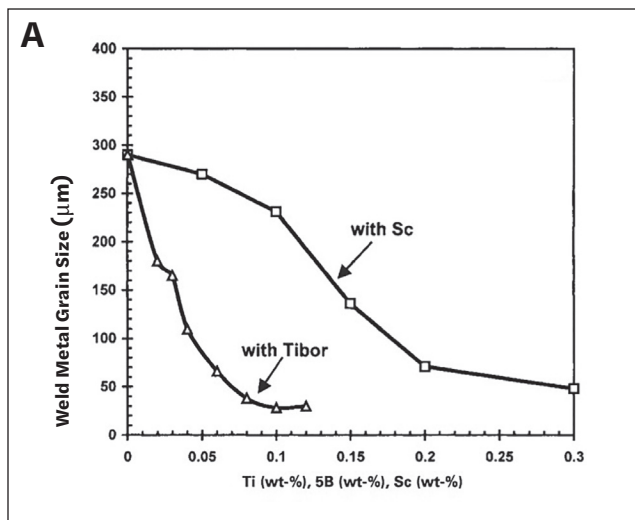


Fig. 10 — A — Grain size results from a circular patch test showing a decrease in grain diameter with increasing Tibor of Sc content; B — weld cross section for Al 7108 treated with Tibor (0.1 wt-% Ti + 0.02 wt-% B) showing grain-refined weld metal (upper left) and base metal (lower right); C — circular patch test crack lengths for Sc additions; D — weld cross section for untreated Al 7108 (Ref. 27).

term in Equation 2 is relatively small and is neglected. However, for welds along the weld interface, where R goes to zero, this term can become very large. This can explain the grain refinement observed along the weld interface for Weldalite O49 and Al 7108 alloys (Ref. 33). Like Weldalite O49, Al-Zn alloys contain $ZrAl_3$ particles in the base metal for grain control, and the particles can be released into the weld pool along the fusion boundary.

Effect of Alloying on Undercooling

It is of interest to know how alloying affects undercooling and understand how different alloys will react to grain refinement. An early approach to this for binary alloy castings involved comparing a constitutional undercooling P factor similar to the second term in Equation 1 (Ref. 35):

$$P = \frac{m_L (k - 1) C_o}{k} \quad (4)$$

Binary alloys with a high P factor will have greater undercooling and finer grains. A revised form of this is the Q growth restriction factor of Easton and StJohn (Ref. 36):

$$Q = kP = m_L (k - 1) C_o \quad (5)$$

based upon the idea that partitioning of certain elements restricts growth and requires more driving force (i.e., undercooling). For multiple solute alloys, the Q factor can be additive for each alloying element. A comparison of

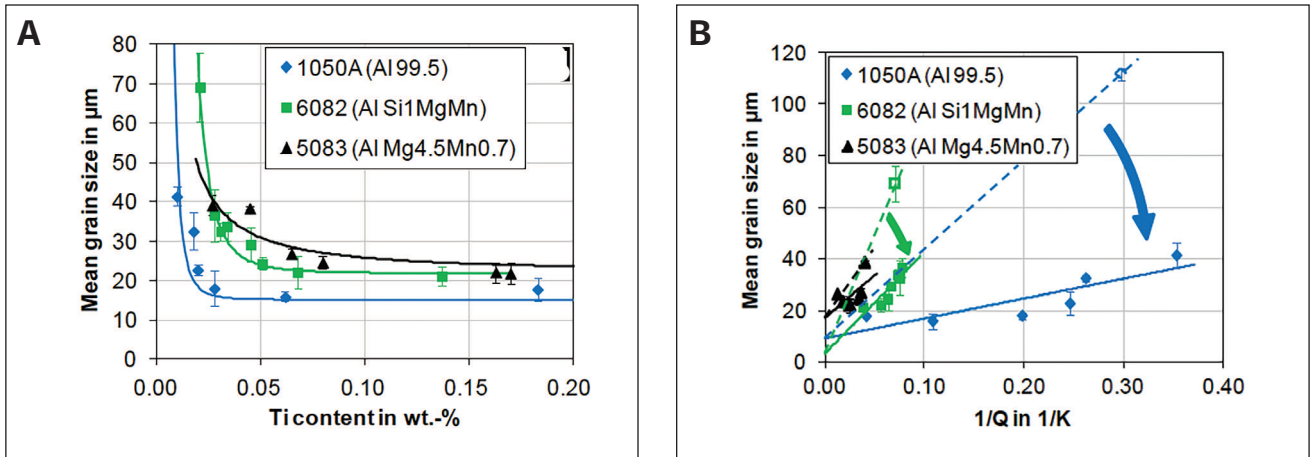


Fig. 11 – A – Weld metal grain refinement of aluminum Alloys 1050, 6082, and 5082 (composition indicated in legend) due to addition of Tibor; B – linear dependence of grain size based upon the inverse of Q factor. Here, Q factor is determined primarily by amount of dissolved titanium content (Ref. 37).

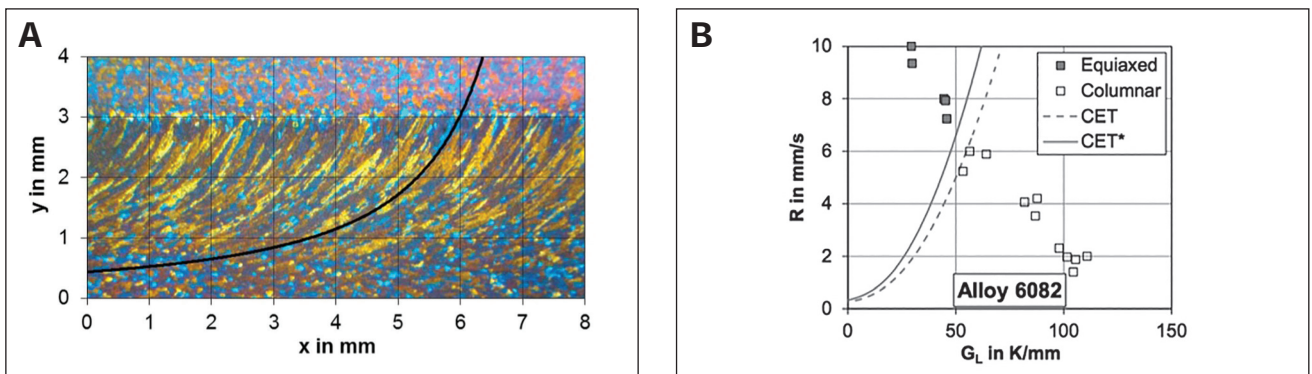


Fig. 12 – A – Top view of the weld cross section on GTA Al 6082 weld; B – comparison of the measured CET values vs. the predicted CET curve based upon Equation 6 (Ref. 37).

$m_L(k-1)$ values for different alloy additions is given in Table 6, where it is observed that titanium, in solution, should provide the greatest amount of undercooling.

To test the effect of Q on GTA weld metal, three aluminum alloys (Al 1050, 6082, and 5083) inoculated with variable amounts of Tibor were examined (Ref. 37). The associated difference in alloy elements, in addition to dissolved titanium, resulted in variable Q values. Figure 11A shows that commercially pure Al 1050, with the lowest Q value, had the greatest response to grain refinement, a result that was not expected. This may be related to a latent heat effect more prevalent in higher-alloyed aluminum. However, the grain size dependence on Q^{-1} was found to be linear, as predicted by Easton and StJohn and shown in Fig. 11B (Ref. 37). The change in slope for each alloy represents the absence of Tibor (dashed line) in comparison to the addition of Tibor (solid line) and, hence, the presence of more potent nucleating particles.

Columnar-to-Equiaxed Transition (CET)

Another approach to quantify necessary conditions for grain refinement is to apply the CET model of Hunt, originally

developed for directional solidification in castings (Ref. 38). A critical gradient is defined, below which columnar grains will transition to equiaxed grains:

$$G < 0.617N_o^{1/3} \left[1 - \left(\frac{\Delta T_N}{\Delta T_t} \right)^3 \right] \Delta T_t \quad (6)$$

where N_o is the density of nucleation sites (mm^{-3}), assumed to be the inverse cube of grain size: $1/d^3$. ΔT_N is the supercooling needed to activate the nucleation particles, and ΔT_t is supercooling at the dendrite tips as defined in Equation 2.

As part of a study to apply the CET model to welding (Ref. 39), GTA bead-on-plate welds (single pass, complete joint penetration) were made on coupons cut from 3-mm-thick Al 6082-T6 plate using the welding parameters given in Table 7. Parameters were selected to give a wide range of weld pool R and G values.

The top of each weld was ground, polished, and etched to reveal grain structure, thereby delineating the location of CET occurrence. An example of one such weld surface

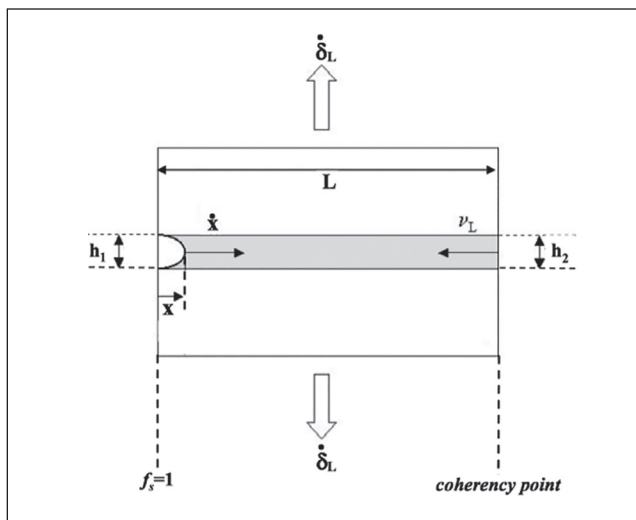


Fig. 13 — Unit cell used to establish mass balance for continuous crack growth consisting of liquid film (shaded) between two adjacent grains. Crack tip is at x and liquid was backfilled from the right (Ref. 43).

is shown in Fig. 12A, including a superimposed trace of the weld pool shape.

Columnar grains are shown originating from the weld interface, transitioning to fine equiaxed grains near the weld centerline. The solidification growth rate (R) can be determined at various positions using the weld travel speed (v) and the angle between weld direction and growth direction (α) (Ref. 40):

$$R = v \cos (\alpha) \quad (7)$$

The dendrite tip undercooling was calculated using the second term of Equation 2. Temperature gradients were determined using pre-embedded thermocouples placed mid-depth at different distances from the weld center. Values for R and G_L are plotted in Fig. 12B for all welds together with the predicted CET transition curve obtained from Equation 6. It was observed that equiaxed grains occurred at high growth

rates where temperature gradients were lower, below 50 K/mm. CET predictions were made for two cases with $A = 2.0 \text{ s}^{1/2} \text{ Kmm}^{-1/2}$ (dashed line) from the literature (Ref. 38) and $A^* = 1.7 \text{ s}^{1/2} \text{ Kmm}^{-1/2}$ (solid line), found by setting the ΔT_t in Equation 6 to the ΔT_t in Equation 2 and solving for A . A nearly identical curve was obtained in the same way for Al 1050 welds, suggesting that this CET delineation may apply to many other aluminum alloys.

Conditions Needed for Crack Growth

Solidification cracks are sometimes found to grow continuously behind the weld pool along the weld centerline, and this scenario results in long cracks rendering the weld unusable. When this happens, there exists a condition unique to welding, whereby the crack grows at the same rate as the weld travel speed, with the crack tip observed to remain in the mushy zone (Ref. 41). It is assumed that crack growth is initiated and driven by the transverse strain displacement across the weld associated with solidification shrinkage and thermal contraction from the weld thermal experience. Furthermore, it is assumed that crack growth involves the separation of a liquid film located at a grain boundary and that this film experiences most of the strain imposed.

Crack Growth Mass Balance

Following a model for the feeding of shrinkage in castings by Braccini et al. (Ref. 42), a simplified mass balance has been established fixing the rate of advancement of the crack tip to the weld travel speed ($\dot{x}$) (Ref. 43). Accordingly, a unit cell has been defined in Fig. 13 to represent a liquid film of thickness h located between two grains at the trailing edge of the mushy zone. This thickness can be related to primary dendrite spacing (λ) and solid fraction (f_s):

$$h = (1 - f_s)\lambda \quad (8)$$

The size of the cell (length: L) extends from a solid fraction

Table 7 — Weld Parameters Used in CET Study of GTA Welds Made on Al 6082-T6 (Ref. 39)

Welding Speed (mm/s)	2.0	4.2	6.0	8.0	10.0
Current (A)	170	175	181	184	190
Voltage (V)	10.7	11.2	11.2	11.2	11.2

Polarity: 50 Hz AC (80% electrode negative, 20% electrode positive)
 Shielding Gas: 50% Ar + 50% He (26 L/min)
 Arc Gap: 3 mm
 Electrode: 3.2 mm ceriated tungsten, 30-deg tip angle

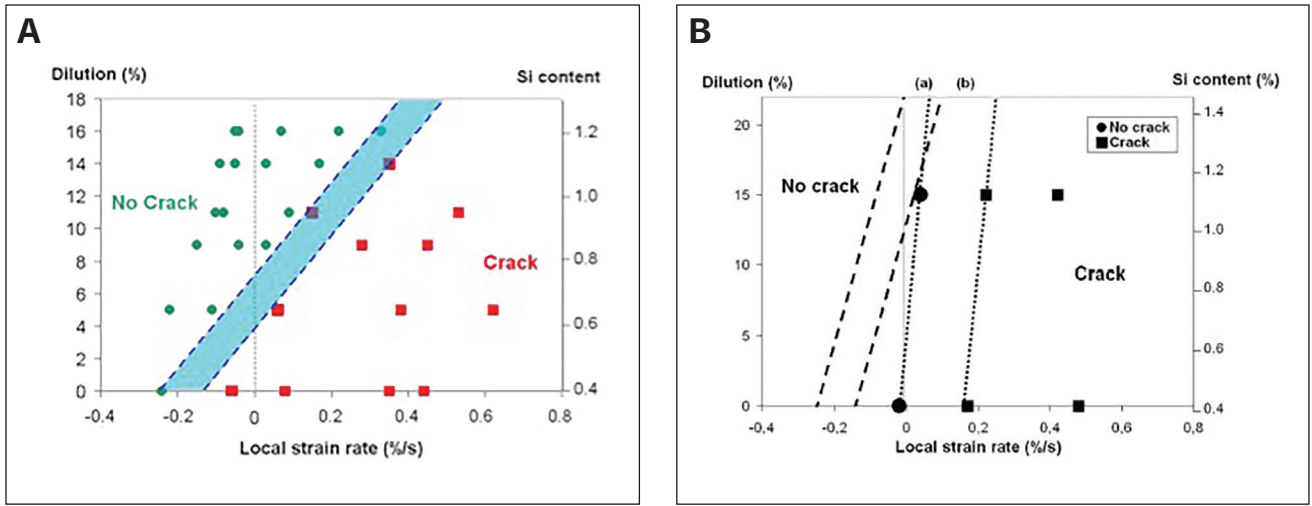


Fig. 14 — A — Results from CTW weldability tests on Al 6060-T4 with 4043 filler metal showing the critical strain rate dividing the region (blue) between no crack and crack conditions (Ref. 50); B — comparison between 6060-T6 for no grain refiner (coarse dash) versus with grain refiner (fine dash). Data points are shown only for grain refiner (Ref. 51).

of one ($f_s = 1$) to the coherency point (i.e., the point at which bonding between adjacent dendrites allows for a transverse stress to first be sustained). The crack tip is assumed to remain located at x , where the solid fraction $f_s = 0.98$. Any larger solid fraction would arguably isolate the crack from liquid feedback.

To maintain this cell constant over a differential length of time, assuming unit cell depth, the volume of backfilling needed is set equal to the volume created by grain separation less the volume of crack advancement:

$$v_L h_2 = \dot{\delta}(L - x) - \dot{x} h_1 \quad (9)$$

where v_L is the flow rate of liquid into the cell and $\dot{\delta}$ is the rate of transverse displacement. If the crack is to remain at a fixed distance into the cell, the rate of liquid backfilling must be sufficient to maintain this equality. This flow value is assumed constant and can be estimated using Darcy's Law:

$$v_L = \frac{-K}{\mu f_L} \frac{dP}{dx} \quad (10)$$

where K is permeability, P is pressure along the interdendritic channel, μ is liquid viscosity, and f_L is fraction liquid. The interdendritic pressure gradient dP/dx can be calculated with the aid of the Rappaz, Drezet, and Gremaud (RDG) model (Ref. 44). Permeability can be calculated based upon the relation from Ganesan, Chan, and Poirier (Ref. 45).

$$K = \frac{\lambda^2(1 - f_s)^2}{8\pi} \quad (11)$$

If the displacement rate is too slow, an abundance of liquid will fill the crack beyond the solidus, and the crack will stop growing. If the displacement rate is too fast, the crack will advance cyclically toward the weld pool and experience increased backfilling, returning to its equilibrium position. This suggests that there is a critical displacement rate, below which cracks cannot grow continuously.

Application of Crack Growth Model

To apply the crack growth mass balance to a practical problem, an Al 6060-T4 alloy extrusion was selected to be welded with Al 4043 filler metal (16% filler dilution). What follows is a condensed summary of the calculations in this study by Coniglio and Cross (Ref. 43):

Pressure Gradient and Backfill Flowrate. The pressure gradient (dP/dx) was calculated based upon an RDG calculation and was estimated to be 10^4 Pa/mm.

Coherency Temperature. Using a single sensor thermal technique of Alexandrov and Lippold (Ref. 46) and the thermal differential analysis of Backerud et al. (Ref. 47), the coherency temperature was determined to be 507°C with a corresponding fraction solid $f_s = 0.82$ (Ref. 48). The cell size ($L - x$) was 4.7 mm.

Critical Displacement Rate Prediction. Solving for the critical grain boundary displacement rate ($\dot{\delta}$) in Equation 9, this value was found to be $0.33 \mu\text{m/s}$.

To experimentally determine if this calculated critical grain boundary displacement rate is reasonable, it was decided to measure the displacement rate across the entire weld and then divide this by the number of grains across the weld metal. This assumes that strain is distributed in the liquid

grain boundary films and not in the grain itself and that there is equal partitioning between all films. Herein is where grain refinement can play a major role and reduce the displacement rate experienced by any grain pair. The German Federal Materials Testing Institute (BAM) in Berlin developed a test frame and methodology where transverse strain is applied at a constant rate during welding using a controlled tensile weldability (CTW) test, basically a horizontal tensile machine into which a weld coupon is clamped and pulled during welding (Ref. 49).

GTA, bead-on-plate welds were made on Al 6060-T4 coupons using different 4043 filler dilutions (Ref. 50). To find the critical strain rate needed for cracking, CTW tests were repeated at incremental applied strain rates until cracking was first observed. The transverse strain rate across the entire weld was measured during welding by placing a 10.5 mm strain gauge on the back side of the plate. The critical strain rate determined in this manner was found to be 0.42%/s for 16% filler dilution. Critical strain rates for other filler dilutions are compared in Fig. 14A, where it is observed that higher amounts of Si in the weld metal give greater critical strain rates (i.e., better weldability) (Ref. 50). There is typically an overlap between no-crack/crack conditions where intermittent cracking is observed.

The displacement rate across the Al 6060-T4 + 16% 4043 weld (i.e., over the 10.5 mm gauge) was (0.0042/s) (10.5 mm) or 44 $\mu\text{m/s}$. Dividing this by the number of grains gives the critical displacement rate across a grain boundary: (44 mm/s)/(157) or 0.28 $\mu\text{m/s}$. The difference between predicted and measured critical displacement rates (0.33 vs. 0.28 $\mu\text{m/s}$) was remarkably small considering the extensive assumptions made in the model and method of testing.

Figure 14B shows CTW data for a stronger temper 6060-T6 that appears less weldable (lower critical strain values) than the T4 temper in Fig. 14A (Ref. 51). Also shown is data for a 6060-T6 weld made with Al 6060-plus-Tibor cast inserts containing 0.8 wt-% Ti. The grain refinement markedly improved weldability (higher critical strains), particularly at low 4043 dilution. This demonstrates the ability of CTW testing to distinguish between weldability due to small differences in material properties and composition.

Summary

Practical examples have been given demonstrating how weldability testing can be used effectively in aluminum alloy development for both base metal and filler metal. It is often the case that weld properties (strength and toughness) are in conflict with weldability and may require compromise. However, the use of grain refiners in filler alloys may allow acceptable weldability for higher-strength alloys that would normally not be weldable. The AA standards for filler alloys are lax when it comes to defining grain refiner Ti content, and B is never listed even though its presence may be critical. Using a solidification crack growth model, it was demonstrated that grain refinement reduces the displacement rate experienced by grain boundary liquid films and helps to keep the displacement rate below a critical value needed for crack growth.

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