Fluid Flow and Solidification in Welding:
Three Decades of Fundamental Research at the University of Wisconsin

Fundamental understanding was gained in fluid flow in weld pool and solidification in fusion zone, in partially melted zone, and with dissimilar filler metals

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ABSTRACT

Fluid flow and solidification during welding dominate the fusion zone of the resultant weld, including its shape, microstructure, properties, and defects. Fundamental research conducted at the University of Wisconsin led to the first demonstrations of the following things in four different areas. The first area, fluid flow in welding, includes: 1) computer models capable of calculating the weld-pool shape and showing how weld-pool fluid flow affects the pool shape, including Marangoni flow driven by surface-tension gradients along the pool surface, 2) visualization of Marangoni flow in simulated weld pools, including its reversal by the surface-active agent and its oscillation, and 3) a theory on the effect of the surface-active agent on pool-surface deformation, pool-surface oscillation and ripple formation as well as weld penetration. The second area, solidification in the fusion zone, includes 1) quenching of the weld pool and its surroundings to reveal the microstructure development, microsegregation, and nucleation mechanisms during welding, e.g., with liquid Sn for stainless steels and water for Al alloys, and 2) suppression of solidification cracking with a wavy crack path, as demonstrated by transverse magnetic arc oscillation in welding of Al sheets. The third area, solidification in the partially melted zone (PMZ), includes: 1) weakening of the PMZ by severe grain-boundary segregation caused by planar solidification of the grain-boundary liquid, as demonstrated in arc welds of high-strength Al alloys, and 2) prediction and elimination of the susceptibility of Al alloys to liqation cracking, i.e., cracking in the PMZ along grain boundaries where liquid formation (i.e., liqation) occurs. The fourth area, solidification with dissimilar filler metals, includes 1) fundamental concepts in welding with dissimilar filler metals, including non iso thermal pool boundaries and quick freezing of one liquid metal in another to cause macrosegregation, and 2) macrosegregation mechanisms based on the concepts to explain the formation of beaches, peninsulas, and islands different in composition from the bulk weld metal (i.e., bulk fusion zone).

Introduction

Fluid flow and solidification are two important subjects in fusion welding. Fluid flow in the weld pool affects the shape and solidification of the weld pool, and solidification affects the microstructure, properties, and defects of the resultant fusion zone. Failures can occur if the weld-pool shape does not provide sufficient penetration into or fusion with the workpiece, if the solidification microstructure is weak and brittle, or if cracking occurs during solidification (Ref. 1). Understanding fluid flow and solidification during welding can help control the resultant weld, including its shape, microstructure, properties, and defects. Fundamental research on welding fluid flow and solidification was conducted by the author in the past 33 years, the first four years at Carnegie Mellon University and the rest at the University of Wisconsin. The research covered the following four areas: 1) fluid flow in welding, 2) solidification in the fusion zone, 3) solidification in the partially melted zone, and 4) solidification with dissimilar filler metals.

Fluid Flow in Welding

Marangoni flow, also called surface-tension-driven flow, is driven by the surface-tension gradient along the pool surface $d\gamma/d\ell$, where $\gamma$ is the surface tension and $\ell$ the distance along the pool surface. The surface tension of a pure liquid metal tends to decrease with increasing temperature, that is, $d\gamma/dT < 0$. Thus, when the welding heat source induces a temperature gradient $dT/d\ell$ along the pool surface, it also induces $d\gamma/d\ell$ because $d\gamma/d\ell = (d\gamma/dT) \times (dT/d\ell)$. Sulfur (S) is an impurity always present in steels and stainless steels, and it acts as a surface-active agent to reduce their surface tension. Experimental data show that with low S (e.g., 40 ppm) $d\gamma/dT < 0$, but with increasing S (e.g., to 150 ppm) $d\gamma/dT$ can become positive (Ref. 2). Heiple and Roper (Ref. 2) proposed a theory to explain the significant effect of the surface-active agent on weld penetration. According to the theory, when sufficient surface-active agent is present to make $d\gamma/dT > 0$, Marangoni flow is reversed and the weld pool can become significantly deeper. DebRoy and coworkers (Refs. 3, 4) developed thermodynamic models to predict $d\gamma/dT$ and its influence by the surface-active agent. For convenience of discussion here, $d\gamma/dT$ will be taken as the driving force for Marangoni flow because without it there will be no $d\gamma/d\ell$ to drive Marangoni flow.
Computer Modeling of Fluid Flow and Heat Transfer in Welding

Kou and coworkers (Refs. 5–8) developed the first computer models (2- and 3-dimensional) to calculate the unknown shape of the weld pool and demonstrate the effect of fluid flow on the pool shape, including fluid flow caused by \( \frac{d\gamma}{dT} \), the Lorentz force, and the buoyancy force. In previous computer models, the pool shape was specified first, and fluid flow and heat transfer in the pool were then calculated. Since the weld shape can be critical in welding practice, it is highly desirable to be able to calculate the shape of the weld pool.

Tsai and Kou (Refs. 9–11) further improved the computer models of Kou and coworkers (Refs. 5–8) by calculating the unknown shape of the weld pool surface, which can be deformed by fluid flow in the weld pool and volume expansion associated with melting and thermal expansion. Their results are most relevant to the following discussion, and thus, will be adopted, although other computer models have also been developed (Refs. 12–16).

By using curvilinear coordinates fitting the pool surface, the boundary conditions for fluid flow and heat transfer at the pool surface were treated precisely. After each iteration of pool-shape computation, the coordinates were calculated again and updated.

Figure 1 shows the effect of \( \frac{d\gamma}{dT} \) alone on the weld pool (Ref. 9). For the purpose of illustration, it shows the steady-state velocity and temperature fields in stationary weld pools caused by heating a workpiece of 6061 Al alloy (~Al-1Mg-0.6Si) with a laser beam of 1800-W power and 8-mm diameter at its top surface. With \( \frac{d\gamma}{dT} < 0 \) (Fig. 1A), the warmer lower-\( \gamma \) liquid at the center of the pool surface is pulled outward by the cooler higher-\( \gamma \) liquid at the pool edge. This outward surface flow carries the laser heat to the pool edge instead of the pool bottom, resulting in a shallow pool. With \( \frac{d\gamma}{dT} > 0 \) (Fig. 1B) the opposite is true. The liquid at the pool surface flows inward toward the center to receive the laser heat, turns axially downward and carries the heat to the pool bottom, resulting in a deeper pool. These results are consistent with the theory proposed by Heiple and Roper (Fig. 1C, D) (Ref. 2).

The effect of \( \frac{d\gamma}{dT} \) alone on pool-surface deformation is also demonstrated. With \( \frac{d\gamma}{dT} < 0 \) (Fig. 1A), the fast outward surface flow (~1 m/s) is suddenly decelerated near the pool edge. Bernoulli’s principle (Ref. 17), although not involved in actual computer modeling, can be used here to simplify the explanation. The deceleration, according to the principle, causes the liquid pressure to rise and push the pool surface upward near the edge, resulting in a concave pool surface. Likewise, with \( \frac{d\gamma}{dT} > 0 \) (Fig. 1B), the fast inward flow is suddenly decelerated near the pool center. The pool surface is raised by the pressure increase near the center and becomes convex.

In fact, volume expansion caused by melting (and, to a much smaller extent, thermal expansion) is another factor to consider. The liquid density is lower than the solid density, e.g., by about 6% for aluminum (Ref. 18). Volume expansion makes the pool surface convex as shown by computer modeling (Ref. 11) (and subsequently by Fig. 5D). However, even with volume expansion, a strong Marangoni
flow induced by $\frac{dy}{dT} < 0$ can still make the pool surface concave by making the pool surface higher near the edge (as shown subsequently by Fig. 5A).

In addition to the weld pool, Tsai and Kou (Ref. 19) also calculated the temperature and velocity fields in gas tungsten welding arcs. Previous arc computer modeling targeted at long arc plasmas unrelated to welding.

Visualization of Marangoni Flow in Simulated Weld Pools

Kou and coworkers (Refs. 20–23) developed a flow-visualization technique to 1) reveal Marangoni flow in a transparent simulated weld pool; 2) verify the effect of $\frac{dy}{dT}$ on Marangoni flow proposed by Heiple and Roper (Ref. 2), and 3) demonstrate oscillatory Marangoni flow. The most direct way to verify the theory of Heiple and Roper (Ref. 2) is by direct observation of Marangoni flow inside the weld pool, which is difficult because metal is opaque.

The flow-visualization technique, as illustrated in Fig. 2, consists of 1) a transparent hemispherical pool of molten NaNO3 (with a $\frac{dy}{dT} = -0.056$ dyne/(cm°C) and a transmission range of 0.35–3 μm); 2) a defocused CO2 laser beam (10.6 μm wavelength) to heat up the pool surface (without penetrating into the pool because 10.6 μm is outside the transmission range) to induce $dT/d\ell$ and hence $\frac{dy}{dT}$; 3) a He-Ne laser (0.63-μm wavelength) light sheet, either vertical (Fig. 2A) or horizontal (Fig. 2B), to cut through the pool to illuminate the tracer particles suspended in the pool and reveal the flow pattern; and 4) a molten NaNO3 bath held in a bottom-heated square glass beaker (not shown) to act as a transparent heater around the pool and to eliminate optical distortions caused by the lens effect of the pool (hemispherical). Buoyancy flow is much slower than Marangoni flow and can be neglected (Ref. 20).

Limmaneevichitr and Kou (Refs. 20, 22) conducted flow visualization (Fig. 2A) to demonstrate Marangoni flow and its reversal caused by a surface-active agent in a simulated weld pool as shown in Fig. 3. With $\frac{dy}{dT} < 0$ the surface flow is outward as indicated by the arrows — Fig. 3A. This is because the warmer and hence lower-$\gamma$ surface liquid heated by the CO2 laser (6-mm diameter and 2.5 W) is pulled outward by the cooler and hence higher-$\gamma$ surface liquid near the pool edge. According to the similarity law of hydrodynamics (Ref. 24), this Marangoni flow can be similar to that in steel and Al weld pools in conduction-mode laser spot welding (Ref. 20).

With 3 mole-% of C2H5COOK added to NaNO3 as a surface-active agent, the surface flow is now reversed (Fig. 3B) (Ref. 22). The liquid flows inward along the pool...
surface toward the center, where it turns and goes down toward the pool bottom. The C$_2$H$_5$COOK heated by the CO$_2$ laser (6-mm diameter and 1.6 W) decomposes and lose its ability to reduce the surface tension of NaNO$_3$. Consequently, the surface tension is now higher near the center of the pool surface where the temperature is also higher, that is, $d\gamma/dT > 0$. This verifies the effect of $d\gamma/dT$ on Marangoni flow proposed by Heiple and Roper (Fig. 3C, D) (Ref. 2). Limmaneevichitr and Kou (Ref. 21) also welded cast plates of NaNO$_3$ and observed deeper welds with C$_2$H$_5$COOK.

Kou et al. (Ref. 23) also made, by flow visualization (Fig. 2B), the first demonstration of oscillatory Marangoni flow in a weld-pool-like liquid pool as shown in Fig. 4. The surface flow is outward ($d\gamma/dT < 0$), so the return flow below the surface is inward. With the CO$_2$ laser at 4 W and 1.5 mm diameter, the inward return flow is steady and converges to the center (Fig. 4A). At 6 W and 1.5 mm diameter, however, two inward streams oppose each other. The directions of streams keep alternating with time at a period of about 1.4 s (Fig. 4B). In a separate experiment, a NaNO$_3$ sessile drop on a graphite substrate was heated from above with a CO$_2$ laser beam to induce oscillatory Marangoni flow. The drop shook visibly, indicating surface oscillation accompanying flow oscillation (Ref. 23).

**Effect of Surface-Active Agent on Weld Pool Surface and Ripples**

Kou et al. (Ref. 23) discovered that the surface-active agent can affect pool-surface deformation, pool-surface oscillation and ripple formation — not just weld penetration as in Heiple’s theory (Ref. 2). To keep other forces (such as the Lorentz force, arc force and recoil force) from interfering with Marangoni flow, conduction-mode laser welding was adopted. In
the conduction mode, the laser beam is defocused (e.g., to 6.4 mm in diameter at the workpiece surface) instead of being sharply focused as in the keyhole mode, where a deep narrow vapor hole is created in the pool by the laser beam for deep-penetration welding of a thick workpiece.

304 stainless steel (~Fe-18Cr-8Ni) plates (6.4 mm thick) with low (42 ppm) and high (140 ppm) S contents were spot welded with a YAG laser beam 6.4 mm in diameter at the workpiece surface — similar to the diameter of a gas tungsten arc in order to allow arc effects to be inferred in future studies. A halogen light was directed through a fiber optic-bundle onto the pool surface in order to tell if the pool surface was concave (large light spot), convex (small light spot) or oscillatory (moving light spot).

The significant effect of the surface-active agent S on the welding of 304 stainless steel is summarized in Fig. 5. The laser power was 2800 W and the welding time 5 s. With low S, the light spot is large and its position with respect to the pool boundary varies with time (Fig. 5A), indicating a concave and oscillatory pool surface (shaking of the pool surface clearly visible in recoded movies). The resultant weld shows clear ripples (Fig. 5B) and a shallow penetration (Fig. 5C). With high S, however, the light spot is small and steady, indicating a convex and calm pool surface (Fig. 5D). The resultant weld shows no clear ripples (Fig. 5E) and a deeper penetration (Fig. 5F). This S effect on weld penetration is consistent with the theory of Heiple and Roper (Ref. 2).

The evidence of pool-surface oscillating during solidification is shown in Fig. 6 for the low-S stainless steel. As mentioned previously, the laser power was 2800 W and the welding time 5 s. Upon turning off the laser, oscillation of the low-S pool surface, though weakened, still continued briefly (Fig. 6A–F). The multiple images of the light spot on the pool surface indicate pool-surface oscillation at a frequency above 30 Hz (30 frames/s). The oscillation frequency can be estimated based on the time for the pool diameter to decrease slightly (from the movie) and the number of ripples corresponding to that decrease (from the resultant weld surface). Taking an 8.5-mm-diameter pool of low-S stainless steel as an example, the oscillation frequency seems closer to that of Mode-2 oscillation than Mode-1, based on the equations derived by Xiao and den Ouden (Refs. 25, 26) and the properties of liquid stainless steel — Fig. 6G. To identify the mode of oscillation, further investigation with a high-speed camera may help. In contrast, after the laser was turned off, the high-S pool showed a convex pool with a light spot significantly smaller and without multiple images (Ref. 23).

Kou et al. (Ref. 23) proposed a theory on the effect of $\frac{dy}{dT}$ beyond Heiple’s theory (Ref. 2), as shown in Fig. 7. With $\frac{dy}{dT} < 0$, the fast outward surface flow can make the pool surface concave (Fig. 7A) even with volume expansion caused by melting and thermal expansion. The raised pool surface near the pool edge is unsupported and thus unstable, and it can oscillate along with oscillatory Marangoni flow in the pool. (In fact, a concave pool surface may be even more unstable if the average pool-surface height is above the workpiece surface due to volume expansion.) Surface oscillation near the pool edge can disturb solidification at the pool edge (Fig. 6B–F) and cause clear ripple formation — Fig. 7B. With enough surface-active agent to make $\frac{dy}{dT} > 0$, the pool surface is now convex (mainly caused by volume expansion) and is raised only slightly near the pool edge (Fig. 7C), much more stable and much less likely to cause ripples during solidification (Fig. 7D).

**Solidification in Fusion Zone**

**Weld Pool Quenching**

Kou and Le (Refs. 27–29) demonstrated a simple technique to reveal the microstructural development, microsegregation, and nucleation mechanisms during welding, that is, by quenching the weld pool and its surrounding area with liquid tin or water. The more recently developed in-situ X-ray diffraction (Refs. 30–34) provides an advanced tool for studying phase transformations during welding. The quenching technique, however, is low-cost, does not require synchrotron radiation, and can be used for studying microsegregation and nucleation mechanisms as well as phase transformations.

Quenching was achieved by pouring onto the weld pool and its surrounding area a much cooler liquid and simultaneously extinguishing the arc. For aluminum...
alloys, ice water can be satisfactory. However, for higher-temperature materials such as steels and stainless steels, water tends to be separated from the target area by steam, which prevents effective cooling and hence quenching. Liquid Sn (232°C melting point) at about 250°–300°C, though warmer than ice water, is still far cooler than the target area, and it works much better because of its low vapor pressure. This Sn quenching technique has been adopted frequently by subsequent investigators (Refs. 35–39).

**Quenching to Preserve Solidification Microstructure**

Kou and Le (Ref. 27) Sn-quenched the weld pool during gas tungsten arc welding of stainless steels, as shown in Fig. 8, to help understand the fusion-zone microstructure, which can be changed by post-solidification phase transformations and thus sometimes become difficult to interpret. This Sn quenching technique has been adopted frequently by subsequent investigators (Refs. 35–39).

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before solidification) — Fig. 10A. The quenched microstructure of 2219 Al (~Al-6.3Cu) shows the partially melted zone (S + L) in the form of partially melted grains around the leading portion of the pool boundary — Fig. 10B. Behind the trailing portion of the pool boundary, that is, the solidification front, columnar dendrites and interdendritic liquid (L) coexist as the mushy zone — Fig. 10C. The quenched weld pool has a very fine solidification structure, much finer than the adjacent microstructure. Outside the mushy zone the partially melted zone (S + L) exists along the fusion boundary and eventually solidifies completely as the partially melted zone (S).

The quenched microstructure of 6061 Al (~Al-1Mg-0.6Si) shows TiB₂ particles in the weld pool (Fig. 10D, E). TiB₂ is an effective grain refiner routinely used in casting Al alloys. Based on these results from quenching, the microstructure around the weld-pool boundary of an Al alloy can be established — Fig. 10F. This microstructure, as will be demonstrated subsequently, is the key to understanding nucleation in the fusion zone, and liquation and liquation cracking in the partially melted zone.

**Nucleation Mechanisms Deduced from Quenched Microstructure**

Based on the quenched microstructure around the weld pool (Fig. 10F), three nucleation mechanisms are possible for
equiaxed grains to form in the fusion zone as illustrated by Kou and Le (Ref. 28) in Fig. 11. In the case of an alloy (Fig. 11A), it is possible for strong fluid flow to break dendrite tips off or detach the partially melted grains from the partially melted zone (S+L) to act as nuclei in the weld pool in addition to heterogeneous nuclei. These nuclei can lead to the formation of equiaxed grains that block off columnar grains growing inward epitaxially from the fusion boundary (columnar dendritic structure inside the columnar grains not shown). Fluid flow can be enhanced, for instance, by electromagnetic stirring of the weld pool. For comparison, the microstructure around the weld pool of a high-purity metal (of melting point $T_M$) is also included for comparison (Fig. 11B). Unlike in the case of an alloy, immediately outside the weld pool is 100% solid instead of semisolid, and no equiaxed grains form (assuming negligible thermal undercooling ahead of the S/L interface).

Nucleation mechanisms in welding are rarely identified with proof. Heterogeneous nucleation has been proved in welds of Al alloys by water-quenching (Fig. 10D, E) and ferritic stainless steel by Sn quenching (Ref. 38). Pearce and Kerr (Ref. 40) suggested grain detachment as the nucleation mechanism of equiaxed grains in Al welds made by gas tungsten arc welding with electromagnetic pool stirring. In principle, stirring may also cause dendrite fragmentation but this may be more difficult than grain detachment.

Kou and Le (Ref. 29) demonstrated the effect of welding parameters on the grain structure in 6061 Al welds as shown in Fig. 12. Columnar grains dominate the entire fusion zone at low travel speed and heat input (Fig. 12A), but equiaxed grains can dominate the central portion of the fusion zone as the travel speed and heat input are increased.
creased (Fig. 12B). At higher heat input and travel speed, the growth rate $R$ is higher while the temperature gradient $G$ is lower along the weld centerline, as shown by experimental results (Ref. 1). According to the constitutional supercooling theory (Ref. 41), constitutional supercooling is less with higher $G/R$ (Fig. 12C) and more with lower $G/R$ — Fig. 12D. However, constitutional supercooling is not the nucleation mechanism in view of the TiB$_2$ particle at the center of the dendritic equiaxed grain — Fig. 12E. Heterogeneous nucleation is still the mechanism though it is aided by constitutional supercooling as pointed out by Ganaha et al. (Ref. 42).

**Improving Solidification-Cracking Resistance and Mechanical Properties**

Kou and Le (Refs. 43–45) demonstrated the concept of suppressing solidification cracking with a wavy path that is difficult for crack propagation. This was demonstrated by transverse magnetic arc oscillation during gas tungsten arc welding of Al sheets, as shown in Fig. 13 for 2014 Al alloy (~Al-4.4Cu-0.8Si-0.5Mg). This was likely the first application of magnetic arc oscillation to Al welding. An electromagnetic arc oscillator was mounted around a gas tungsten arc welding torch that traveled in a straight line. Without arc oscillation, normal columnar grains are observed — Fig. 13A. Under identical welding conditions except for arc oscillation in the transverse (lateral) direction at 1 Hz, columnar grains follow the oscillating arc and weld pool. They grow inward from one side of the fusion zone and then from the opposite side (Fig. 13B), and this keeps on alternating along the weld. This is because columnar grains tend to grow essentially normal to the pool boundary (Ref. 1). Solidification cracking (as shown by the fishbone method) is significantly reduced at 1 Hz — Fig. 13C. The weld strength and ductility are also significantly improved, as shown by tensile testing of welds without cracks — Fig. 13D.

The improvement is explained in Fig. 14. Solidification cracking is intergranular (Ref. 1), that is, it propagates along grain boundaries rather than cutting through columnar grains. Thus, the pattern of columnar grains caused by arc oscillation forces the crack path to become wavy as schematically illustrated (Fig. 14B) and actually observed in experiments. A wavy crack path makes it difficult for cracks to propagate, and hence, suppresses solidification cracking.

Arc oscillation also makes the weld pool faster moving and smaller in size. Transverse arc oscillation adds a lateral velocity component ($v$) to the longitudinal velocity component ($u$) of the weld pool (Fig. 14B) and increases its actual moving speed to a higher...
value \( w \) (e.g., from 4.2 to 8.7 mm/s). This results in a higher cooling rate during welding and hence two things. First, the dendrite arm spacing is finer (Ref. 1), which improves mechanical properties (Ref. 18). Second, the partially melted zone (PMZ) and the heat-affected zone (HAZ) as well as the weld pool become smaller. This means less damage caused by grain-boundary segregation in the PMZ and grain growth or over-aging in the HAZ.

**Solidification in Partially Melted Zone**

**Grain-Boundary Segregation**

Huang and Kou (Refs. 46–50) demonstrated that in arc welding high-strength Al alloys tend to develop a wide partially melted zone (PMZ) with thick grain-boundary liquid, and that the liquid solidifies in a planar mode to cause severe grain-boundary segregation to weaken the PMZ severely. Unlike most steels and stainless steels, Al alloys have a very high thermal conductivity \( k \) to reduce temperature gradients during welding, and a wide melting temperature range \( \Delta T \) relative to their low liquidus temperature \( T_L \). For instance, \( \Delta T/T_L \) is \( (927–839 \text{ K})/927 \text{ K} \) or 0.095 for Al-4.5Cu alloy as compared to \( (1797–1768 \text{ K})/1797 \text{ K} \) or 0.016 for 1018 carbon steel and \( (1723–1673 \text{ K})/1723 \text{ K} \) or 0.029 for 304 stainless steel. Thus, Al alloys tend to have a wide PMZ with a thick grain-boundary liquid.

The PMZ in Al alloys is illustrated in Fig. 15 using the gas tungsten arc weld of a simple binary alloy Al-4.5Cu as an example. The PMZ (Fig. 15A, B) is the region immediately outside the weld metal (the fusion zone). Upon reaching the eutectic temperature \( T_E \) (Fig. 15C) during welding, any \( \theta \) (Al\(_2\)Cu) particles still remaining in the Al-rich \( \alpha \)-matrix can react with \( \alpha \) to form a small amount of liquid around the particles, that is, constitutional liquation (Refs. 51, 52). Thus, liquid formation or liquation occurs during welding along grain boundaries and at isolated spots within grains, where \( \theta \) (Al\(_2\)Cu) is located. Upon further heating to above the solidus temperature \( T_S \), however, much more liquid can form (up to 100% liquid at the pool boundary \( T_L \)). With a wide PMZ, the grain-boundary liquid can thus be thick.

Clear microstructural evidence of grain-boundary segregation can be seen in the PMZ (Fig. 15D) — not just the dark-
etching grain boundary shown by numerous previous studies, but also a light-etching \(\alpha\) band next to it. The etching here is effective enough to bring out the \(\alpha\) band with high contrast. Huang and Kou (Ref. 47) suggested that the dark-etching grain boundary is the Cu-rich eutectic while the light-etching band is Cu-depleted \(\alpha\) (Fig. 15D), that is, these are clear microstructural evidence of severe grain-boundary segregation. In fact, close examination shows that the \(\alpha\) band turns darker near the eutectic grain boundary. This suggests that the Cu content in the \(\alpha\) band increases significantly near the eutectic grain boundary.

The thick \(\alpha\) band and the grain-boundary eutectic (Fig. 15D) further suggest the existence of a thick grain-boundary liquid and its solidification by the planar mode toward the grain boundary. Upon cooling, planar solidification of the grain-boundary liquid begins as a solute-depleted \(\alpha\) band free of eutectic and finishes as a solute-rich eutectic grain boundary.

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The PMZ is wide as can be seen from the numerous grains with a light-etching \(\alpha\) band outside the weld metal — Fig. 15E, F. Note that the planar solidification of the thick grain-boundary liquid is directional, that is, inward and upward toward the weld metal (opposite to the direction of heat flow from the arc during welding). On the left side of the PMZ the \(\alpha\) bands face right and up (Fig. 15E), while on the right side, they face left and up — Fig. 15F. Without directional solidification, the eutectic grain boundary might have been in the middle between two neighboring grains, cutting the thickness of the band (and hence the distance over which solute segregation develops) in half.

Huang and Kou (Refs. 48, 50) confirmed grain-boundary segregation in the PMZ of commercial Al alloys by composition measurements. Alloy 2219 Al (~Al-6.3Cu) is shown in Fig. 16 as an example. In the SEM image (Fig. 16A), the thick \(\alpha\) bands appear dark and eutectic light, just the opposite of optical micrographs — Fig. 15D. The composition profile (Fig. 16B) shows a peak Cu content of 30 wt-% Cu (Fig. 15C), suggesting the grain-boundary segregation is caused by solidification-induced microsegregation instead of solid-state diffusion. The rising Cu content in the \(\alpha\) band near the eutectic grain boundary is consistent with the darker-etching of the \(\alpha\) band near the eutectic grain boundary — Fig. 15D.

Huang and Kou (Refs. 46–50) identified planar solidification of thick grain-boundary liquid — not solid-state diffusion — as the cause for the formation of a severely segregated microstructure to weaken the PMZ of Al welds. Note that solute segregation is most severe with the planar solidification mode (Ref. 1) — cellular or dendritic solidification would have left some eutectic throughout the \(\alpha\) band to reduce segregation across the band. Occasionally, at locations of lower temperature gradients, planar solidification did break down into cellular solidification. But this occurred only near the end of the solidification of a very thick grain boundary liquid, and the resultant eutectic grain boundary was cellular in shape (Ref. 49).

The breakdown of planar solidification into cellular has been described quantitatively by two classic theories: the constitutional supercooling theory of Chalmers et al. (Ref. 41), and the interface stability theory of Mullins and Sekerka (Ref. 53). The severely segregated microstructure caused by planar solidification of thick grain-boundary liquid is weak and brittle. Under tensile loading, the soft solute-
depleted α-phase yields easily, but the brittle solute-rich eutectic fractures and initiate numerous cracks to cause premature failure (Fig. 16C). Most cracks are along eutectic grain boundaries, though some are in large eutectic particles within grains, caused by liquation at large θ (Al2Cu) particles present in 2219 Al before welding. This can explain the severe loss of ductility and strength in the PMZ of Al alloys.

Susceptibility to Liquation Cracking

Kou and coworkers (Refs. 54–58) proposed a simple robust criterion for predicting and eliminating the susceptibility of Al welds to liquation cracking. Liquation cracking is cracking in the PMZ along grain boundaries where liquid formation (i.e., liquation) occurs during welding. As mentioned previously, Al alloys often tend to have a wide PMZ with thick grain-boundary liquid during arc welding, and thus intergranular cracking can occur in the PMZ under tension.

The criterion is illustrated in Fig. 17A. As mentioned previously, the density of liquid Al is lower than that of solid Al by about 6% for aluminum (Ref. 18). Thus the weld metal contracts during solidification. The solidifying and contracting weld metal (WM) pulls and induces tension in the solidifying PMZ — Fig. 17A. In view of the decreasing strength of a semisolid Al alloy with decreasing fraction solid fS (Ref. 18, 59), it seems reasonable to assume that a semisolid with a lower fS may have less resistance to cracking under tension, unless the semisolid still has little strength, e.g., fS < −0.3 (Ref. 18). Thus, it is assumed that liquation cracking may occur if, at any point along the fusion boundary between TL and TE, the solidifying weld metal has a higher fS than the solidifying PMZ (Fig. 17B) but not when it has a lower fS — Fig. 17C. For simplicity, the morphological difference between the dendrites on the mushy-zone side of the fusion boundary and the partially melted grains on the PMZ side is neglected, and fS alone is considered. The following criterion is thus proposed for Al arc welds:

\[
\text{PMZ susceptible to liquation cracking if } WM \ f_S > PMZ \ f_S \quad (1)
\]

A few comments are made as follows: First, since a semisolid develops significant strength only after fS reaches about 0.2–0.4 (Ref. 18), the criterion is applicable after fS reaches about 0.3 during solidification. Second, Equation 1 does not imply solidification cracking in the weld metal if WM fS < PMZ fS. The PMZ is narrower than the fusion zone and the grain-boundary liquid in the PMZ is small in volume as compared to the weld pool. Thus, PMZ contraction is not expected to induce much tension in the solidifying weld metal. Third, Gittos et al. (Ref. 60) considered liquation cracking based on the equilibrium solidus temperature, but equilibrium solidification does not occur in welding. Curves of T-fS (temperature vs. fraction solid) of the weld metal and the PMZ can be plotted to check the criterion. The PMZ T-fS curve is the same as the workpiece T-fS curve because of no mixing between the PMZ and the filler metal. The T-fS curves can be easily calculated using commercial computer software and databases, such as those provided by CompuTherm and ThermoCalc. The simple Scheil model for nonequilibrium solidification, which is provided by all software, works well. The Scheil model assumes local equilibrium at the S/L interface and no solid-state diffusion (Refs. 1, 18).

Surprisingly, the simple criterion works very well, its validity verified by numerous experimental data (Refs. 54–58). The criterion can help guide the selection of filler metals to eliminate liquation cracking.

Before calculating the T-fS curve of the weld metal, the weld metal composition needs to be determined. Dilution is the extent the filler metal is diluted by the melted base metal in the weld metal and is often expressed as the wt-% of the base metal inside the weld metal. It can be easily determined from the transverse macrograph of the weld, that is, dilution equals the area of the solid base metal inside the weld divided by the total area of the weld (since mixing in the weld pool is essentially complete) (Ref. 1). The weld metal composition is determined as follows: (wt-% of solute A in weld metal) = (wt-% of solute A in base metal) × (dilution) + (wt-% of solute A in filler metal) × (100% − dilution) (Ref. 1).
The criterion led to a new filler metal that eliminated liquation cracking in 7075 Al (Ref. 57), which is highly susceptible to liquation cracking. The criterion can also shed light on backfilling of liquation cracks. Alloy A357 (~Al-7Si) welded with filler 1100 Al (~pure Al) at 63% dilution can be susceptible to liquation cracking — Fig. 18C. However, the cracks are backfilled. Liquation cracking is most likely to occur above about 570°C, where WM $f_s > > PMZ f_s$ and the PMZ has at least 50% liquid to backfill its cracks. When a crack opens up, a vacuum is created in it to suck in the liquid nearby. The PMZ (and base metal) appears dendritic because Alloy A357 was welded in the as-cast condition.

**Solidification with Dissimilar Filler Metals**

Kou and Yang (Refs. 66-73) established new mechanisms for macrosegregation caused by welding with a dissimilar filler metal, that is, a filler metal different from the workpiece in composition. Dissimilar filler metals are routinely used, to help avoid solidification cracking, liquation cracking, hydrogen cracking, and corrosion (Ref. 1). Even when good mixing of the filler metal exists in the bulk weld pool, it may not exist near the fusion boundary. Macrosegregation can cause hydrogen cracking, corrosion, and stress corrosion cracking (Ref. 1).

Savage and coworkers (Refs. 74, 75) first discovered macrosegregation near the fusion boundary as shown in Fig. 19. The weld interface is the fusion boundary, within which complete melting occurs during welding. A stagnant or laminar-flow layer of liquid base metal can exist at the pool boundary and solidify, without mixing with the bulk weld pool, as a “beach,” which is called the unmixed zone. This remained as the only macrosegregation mechanism proposed for about forty years. The “folds” are “peninsulas” extending from the beach into the bulk weld metal but different from the bulk weld metal in composition. Though not shown, “islands” different from the bulk weld metal in composition have been observed near the fusion boundary by other investigators (Refs. 76, 77). How peninsulas and islands form has not been explained so far.

**New Macrosegregation Concept**

Kou and Yang (Ref. 66) introduced the new macrosegregation concept illustrated in Fig. 20, that is, macrosegregation caused by quick freezing of one liquid metal in another. Liquid metal 2 is below the liquidus temperature of liquid metal 1 — Fig. 20A. Thus, liquid metal 1 can begin to freeze quickly with limited mixing upon entering liquid metal 2 (Fig. 20B) and thus causes macrosegregation — Fig. 20C.

**New Macrosegregation Mechanisms**

Kou and coworkers (Refs. 66–73) proposed two macrosegregation mechanisms for dissimilar-filler-metal welding, based on the liquidus temperature of the bulk weld metal $T_{\text{lw}}$ vs. that of the base metal $T_{\text{lb}}$. Figure 21 shows Mechanism I for the case of $T_{\text{lw}} < T_{\text{lb}}$, that is, for a filler metal that decreases the liquidus temperature of the base metal upon mixing with it. The composition of the bulk weld metal is somewhere between those of the base metal and the filler metal, depending on the extent of dilution of the filler metal by the melted base metal — Fig. 21A. The bulk weld pool, in which the filler metal is mixed uniformly with the melted base metal, begins to solidify at $T_{\text{lw}}$ (if undercooling is negligible) — Fig. 21C. However, a stagnating (or laminar-flow) layer of liquid base metal can exist along the pool boundary and begin to solidify at $T_{\text{lb}}$ into a beach. The beach is filler-metal deficient because it receives no contribution from the filler metal through mixing with the bulk weld pool.

Thus, unlike in welding with a matching filler metal, the pool boundary is no longer isothermal at $T_f$. Instead, it is at $T_{\text{lw}}$ in the bulk weld pool but at $T_{\text{lb}}$ in the liquid base-metal layer. Furthermore, ahead of the solidification front of the bulk weld metal ($T_{\text{lw}}$), there exists a region of liquid weld metal cooler than $T_{\text{lb}}$ (simply because $T_{\text{lw}} < T_{\text{lb}}$). Thus, when the liquid base metal in the stagnant layer is carried by flow into the cooler region, it can freeze quickly with no or partial mixing and results in a peninsula or even island (Fig. 21D). The flow needs not to be exactly in a vertical plane. Any flow that can carry the liquid base metal into the cooler region is enough.

Mechanism I is verified in Fig. 22. In welding 1100 Al (commercially pure Al) with filler metal 4145 Al (~Al-4Cu-10Si) at 50% dilution (Fig. 22A), the bulk weld-metal composition is essentially Al-2Cu-5Si and the condition of $T_{\text{lw}}$ (620°C) < $T_{\text{lb}}$ (660°C) is met (Ref. 66). Composition measurements at the crosses show no Cu and Si. As compared to the bulk weld metal, the peninsula and the islands are deficient in Cu and Si from the filler metal, that is, filler deficient. They originate from the liquid base metal, freezing quickly before mixing with the bulk weld pool (at least at the crosses inside the peninsula and islands). The etching failed to bring out the solidification microstructure inside the peninsula and islands.

Similarly, in welding 1100 Al (commercially pure Al) with filler metal 4047 Al (~Al-12Si) at 61% dilution (Fig. 22B), the bulk weld metal composition is essentially Al-4.7Si and the condition of $T_{\text{lw}}$ (633°C)
<TLB (660°C) is met (Ref. 72). Composition measurements at the circles show no Si. This indicates the beach and the island originate from the liquid base metal, freezing quickly before mixing with the bulk weld pool. It is worth mentioning that the presence of the beach and the island represents macrosegregation, while that of the cellular solidification structure inside them represents microsegregation (of impurities in commercially pure Al).

Figure 23 shows Mechanism II for the case of TLW > TLB, that is, for a filler metal that increases the liquidus temperature of the base metal upon mixing with it — Fig. 23A. The bulk weld pool begins to solidify at TLW (if undercooling is negligible) — Fig. 23C. However, the stagnant layer of liquid base metal near the pool boundary begins to solidify at TLB into a filler-deficient beach. Again, the pool boundary is nonisothermal. Since TLW > TLB, this layer is cooler than the bulk weld pool. Consequently, when pushed by a downward flow (such as that caused by the filler metal droplets impinging the bulk weld pool) into this layer, the liquid weld metal can freeze quickly as intrusions (Fig. 23D). The liquid base metal remaining in the layer can solidify with no or partial mixing as a beach with an upper boundary of irregular shape.

Mechanism II is verified in Fig. 24. In welding pure Cu with filler metal Cu-30Ni at 45% dilution (Fig. 24A), the bulk weld metal composition is essentially Cu-16.5Ni and the condition of TLW (1175°C) > TLB (1085°C) is met (Ref. 66). Three sharp intrusions into the beach are visible, indicating the liquid weld metal intruded into the liquid base metal layer. The intrusions make the upper boundary of the beach irregular in shape. Intrusions have not been reported previously. The solidification structure of the intrusions is much finer than that of the nearby bulk weld metal (Fig. 24B). This verifies quick freezing of the liquid weld metal in the layer of cooler liquid base metal because faster cooling results in a finer solidification structure (Refs. 1, 18). Composition measurements at the circles show no Ni. This indicates the liquid base metal outside the intrusions solidified before mixing with the bulk weld pool.

Summary and Conclusions

Fundamental research was conducted on fluid flow and solidification in welding, and the following things were demonstrated for the first time:

1) Computer models capable of calculating the weld-pool shape. The pool shape no longer had to be assumed as in previous studies. The effect of fluid flow on the pool shape was demonstrated, including flow caused by dy/dt (Heiple’s theory), the Lorentz force and the buoyancy force.

2) Visualization of Marangoni flow, including its reversal and its oscillation. Visualization of Marangoni flow and its reversal by the surface-active agent is the most direct proof of Heiple’s important theory. Visualization of oscillatory Marangoni flow suggests Marangoni flow is likely to oscillate in a real weld pool.

3) A theory on the effect of surface-active agent beyond Heiple’s. Through dy/dt the surface-active agent can affect pool-surface deformation, pool-surface oscillation and ripple formation in addition to weld penetration, which Heiple’s theory addresses. This was demonstrated by conduction-mode laser welding of 304 stainless steel.

4) Quenching of the weld pool to reveal the microstructure development during welding. This simple inexpensive technique helps understand phase transformations, microsegregation, and nucleation mechanisms during welding, as demonstrated by stainless steels quenched with liquid Sn and Al alloys quenched with water.

5) Suppression of solidification cracking with a wavy crack path. Slow lateral oscillation of a moving weld pool that forms columnar grains can cause a wavy path difficult for crack propagation and faster cooling to improve weld mechanical properties, as demonstrated by magnetic arc oscillation in the transverse direction at 1 Hz in gas tungsten arc welding of Al sheets.

6) Weakening of the PMZ by severe grain-boundary segregation. As demonstrated in arc welding of high-strength Al alloys, their high thermal conductivity and large melting range promote a wide PMZ with a thick grain-boundary liquid, which solidifies with the planar mode to cause severe solute segregation and hence weakening.

7) Prediction and elimination of the liquation-cracking susceptibility. An Al arc weld is susceptible to liquation cracking if weld-metal f_lg > PMZ f_lg (after f_lg ∼ 0.3). This simple robust criterion was verified by numerous experimental data and used successfully to guide selection of existing filler metals and development of new ones to eliminate the susceptibility.

8) Fundamental concepts in welding with dissimilar filler metals. It was explained that the weld pool boundary is no longer isothermal as in welding with matching or no filler metals and that macrosegregation can occur by quick freezing of one liquid in another.

9) Macrosegregation mechanisms beyond Savage’s. Based on these concepts, two mechanisms were established for the formation of beaches, peninsulas, and islands different in composition from the bulk weld metal. This is beyond Savage’s mechanism for the formation of beaches (Ref. 74).

Dedication

The author would like to dedicate this work to Robert F. Sekerka, who is University Professor at Carnegie Mellon University, Pittsburgh, Pa., for helping him start his research and teaching in welding. He hired the author in 1979 as an Assistant Professor at Carnegie Mellon to teach and work on welding, encouraged him to publish his class note Welding Metallurgy (now 2nd edition, published by John Wiley), and helped him as a mentor and a world-renowned authority on solidification theories.
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