Introduction

The automotive industry has been facing regulatory requirements for improved fuel economy of their automobiles to reduce fuel consumption and harmful greenhouse gas emissions. This is being accomplished a number of ways including development of more efficient engines, switching to new hybrid and all-electric drive systems, as well as reducing overall vehicle weight, such as light-weighting, through increased use of lighter materials such as polymers, composites, aluminum, and magnesium, and development of advanced manufacturing technologies such as tailor welded blanks. For example, up to 30% savings in weight and scrap have been realized by making tailor welded blanks (TWB) from different automotive steel sheets and stamping them in a single forming operation to make many different automotive body parts (Refs. 1–3). Even greater savings in weight is possible by making TWBs using much lower density AA5754 and AA5182 aluminum alloy sheet (Refs. 2, 4, 5).

However, the ability to make acceptable welds between these aluminum alloy sheets at welding speeds comparable to those used to produce laser welded steel tailor welded blanks and with the ductility and formability required of welds in TWBs has presented considerable challenges due to the unique properties of aluminum alloys relative to steels. Most wrought aluminum alloys have much higher thermal conductivity; low absorptivity to laser beam wavelengths; a tenacious, high melting point oxide; and a pronounced susceptibility to hydrogen porosity, solidification shrinkage porosity, and other solidification defects (Ref. 6).

The weldability of aluminum sheet for TWB applications has been studied using various autogenous fusion welding processes including gas tungsten arc welding (GTAW) (Refs. 7–10), variable-polarity plasma arc welding (VP-PAW) (Refs. 11–13), double-sided arc welding (DSAW) (Refs. 14–16), Nd:YAG laser beam welding (LBW) (Refs. 17–25), nonvacuum electron beam welding (NVEBW) (Refs. 7, 23, 24, 26), and other processes (Refs. 3, 27).

More recently, the weldability of aluminum alloy sheet by the solid-state friction stir welding (FSW) process for production of aluminum TWBs has been explored (Refs. 10, 28, 29). Each of these welding processes have been found to offer specific advantages and disadvantages. Overall, however, successful welding of aluminum TWBs has been found to be significantly more challenging than conventional CO2 laser beam welding of steel TWBs (Refs. 10, 28, 29). Each of these welding processes have been found to offer specific advantages and disadvantages. Overall, however, successful welding of aluminum TWBs has been found to be significantly more challenging than conventional CO2 laser beam welding of steel TWBs (Refs. 3, 20). For example, the lower power density arc welding processes can actively remove the surface oxide by cathodic etching; however, they are typically slower than LBW processes. Also, the wrought microstructure of the sheet is transformed to a solidification microstructure in the weld, which is typically softer than the original wrought alloy sheets. Softening will usually also occur in the heat-affected zone (HAZ) if the sheets are cold worked or heat treat-
ed. Finally, the weld quality can be reduced due to the presence of a number of defects such as hydrogen and solidification shrinkage porosity, as well as solidification, centerline, and liquation cracking (Refs. 7, 14, 16).

While the high energy density LBW and NVEBW processes offer comparable welding speeds as those typically used when making steel TWBs using the CO₂ LBW process, they also are susceptible to fusion weld defects including hydrogen porosity, and solidification, centerline, and liquation cracking. In addition, the high energy density beams are known to cause defects such as alloy element depletion through preferential vaporization of elements such as Mg or Zn, etc., occluded vapor pores due to keyhole instabilities, undercutting, root sag, and very rough, irregular weld bead surface quality (Refs. 17–26). Finally, the relatively new solid-state FSW process has the advantages that it maintains a wrought microstructure so that gas porosity and other solidification defects are not produced; however, it has until recently been too slow for use in making TWBs in the automotive industry (Refs. 5, 10, 29).

In general, most studies of fusion welding of nonheat-treatable AA5182-O and AA5754-O aluminum alloy sheet for TWB applications have reported similar hardness, yield, and ultimate tensile strengths in the welds as the unwelded sheet; however, significantly reduced ductility and formability with failure occurring in the weld were also observed (Refs. 7–9, 14–26). It has frequently been suggested that the reduced ductility and formability can be attributed to the presence of various weld defects such as gas porosity, solidification porosity, occluded gas porosity in laser welds, and geometric defects in the weld bead and surface quality that create localized stress concentrations.

These observations are consistent with earlier studies of the influence of hydrogen porosity on the structural properties of aluminum alloy welds (Refs. 30–32). For example, Ashton et al. (Ref. 32) did not observe any loss in strength in structural aluminum welds when the welds contained <1–2 vol-% hydrogen porosity, but above this threshold, weld strengths were found to decrease in proportion to the decrease in cross-sectional area of weld metal (Ref. 31). However, hydrogen porosity had a much greater effect on weld metal ductility; while only 4 vol-% hydrogen porosity reduced the weld strength by 17%, weld ductility was reduced by 51%. Even when there was <1 vol-% hydrogen porosity and no change in weld strength, there was a significant loss of weld metal ductility (Ref. 32).

Some attempts have been made to model the effects of weld defects on ductility and strain-to-failure in TWB welds. Davies et al. (Refs. 7, 8) developed a model of the effects of the hydrogen gas porosity observed in their GTA and NVEB welded AA5182-O TWBs on the ductility and strain to failure when subjected to uniaxial tension and showed there was a correlation between the level of porosity observed in their welds and the observed decrease in TWB ductility. They also used this model to predict the effects of weld defects in TWBs on forming-limit diagrams (Ref. 9).

In later studies, Bayley and Pilkey (Refs. 33, 34) used a bifurcation criterion in a finite element model of NVEB welded AA5754-O aluminum TWBs to predict the effects of weld bead surface roughness and internal weld porosity.

| Table 1 — Measured Mechanical Properties of the A5182-O Base Metal Sheets |
|------------------|------------------|------------------|
| Yield Strength (MPa) | Ultimate Tensile Strength (MPa) | Ductility (%) |
| 125 ± 2.5 | 288.6 ± 0.8 | 22.5 ± 1.4 |

![Fig. 1 — Schematic of the autogenous DSAW configuration used with a PAW torch above and GTAW torch below the shimmed and clamped 1.0- and 1.5-mm-thick AA5182-O aluminum sheet specimens.](image1)

![Fig. 2 — The range of welding speeds and welding powers used to produce the longitudinal weld metal tensile specimens and hemispherical dome height specimens.](image2)
on the onset of shear localization and failure of the TWB welds when loaded in the longitudinal direction. They found that even small levels of internal porosity had a significant detrimental impact on the onset of shear localization that quickly leads to initiation of through-thickness shear bands, reduced ductility, and premature failure in the weld. They also found that surface defects would reduce weld ductility, but not as much as internal defects.

There have been many experimental studies of the weldability of the nonheat-treatable AA5182-O and AA5754-O aluminum alloy sheets for tailor welded blanks using a range of autogenous fusion welding processes in which the ductility and formability of the welds have been found to be substantially lower than the base metal sheets leading to premature failure of the weld during deformation and forming. The presence of one or more internal or external surface weld defects has frequently been cited as the cause of the observed reduction in weld metal ductility and formability (Refs. 7–9, 14–26, 33, 34). It is noteworthy, however, that no attempts have been made to reduce or eliminate these defects and thereby demonstrate improved ductility and formability. The objective of the present study (Ref. 15), therefore, was to identify and develop techniques that might be used to reduce weld defects and improve ductility and formability in autogenous fusion welded AA5000 series aluminum tailor welded blanks.

**Experimental Methods and Procedures**

The material used in this study was 1.0- and 1.5-mm-thick AA5182-O aluminum alloy sheet. This nonheat-treatable alloy was solution strengthened by adding 4.5 wt-% Mg. Following rolling to these sheet thicknesses, the sheets had been annealed and recrystallized to the O-temper. This sheet alloy is considered to have good strength, ductility, and weldability for use in structural automotive body applications (Refs. 35–37).

To produce welded TWB specimens for longitudinal tensile tests and hemispherical dome-height (HDH) formability tests, the aluminum sheet was sheared into 150 mm × 430 mm specimens. The rolling direction of the sheets were kept perpendicular to the weld joint. This has been shown to facilitate effective cathodic cleaning of the hydrated aluminum oxide from the weld pool surface during welding (Refs. 12–16). The specimens were cleaned with acetone and then a bench grinder with a 200-mm-diameter wire wheel made with a 0.3-mm-diameter stainless steel wire was used to breakup and clean the hydrated oxide from the sheet surfaces in the vicinity of the weld. This was required to improve arc coupling and the weld bead quality and to significantly reduce or eliminate hydrogen porosity defects in the weld metal (Refs. 12–16, 38).

The double-sided arc welding (DSAW) process, as shown schematically in Fig. 1, was used to make high quality, autogenous fusion welds between the cleaned AA5182-O aluminum sheet specimens. The DSAW process was originally developed and patented in 1999 by Zhang and Zhang (Ref. 39) for the welding of relatively thick plate using lower currents. The process uses only one welding power supply, but two torches; frequently, a plasma arc welding (PAW) and gas tungsten arc welding (GTAW) torch each connected directly to one of the two power supply terminals.

As shown in Fig. 1, the torches were positioned on opposite sides of the two clamped sheets. During welding, the clamped sheets were moved between the two stationary torches at the welding speed while the welding current flowed through the welding arc from one torch to the workpiece, through the workpiece, and then through the arc on the opposite side of the sheets to the other torch. In conformance with various welding safety standards, the sheet specimens and clamps, and so forth, were electrically...
grounded and not part of the electric welding circuit.

While Zhang et al. (Refs. 40–45) have examined the use of the DSAW process to make uphill, keyhole-mode welds in 6- to 12-mm-thick plain carbon steel, stainless steel, or aluminum alloy plates, more recently, Weckman et al. (Refs. 14–16) have studied the use of the autogenous, conduction-mode DSA process to make complete-joint-penetration, square welds between much thinner AA5182 aluminum alloy sheet for TWB applications. They found the process is well suited to aluminum TWB applications because visually acceptable DSA welds can be made at speeds comparable to those used for laser welding, but at much lower capital and operating costs (Refs. 14–16).

When using a variable polarity power supply, the two arcs above and below the joint provide cathodic cleaning of the oxide from the top and bottom weld pool surfaces simultaneously and produce better weld bead surface quality than is typically seen in laser or electron beam welds of this alloy (Refs. 12, 13, 17–27). At high DSAW speeds, two small partial-joint-penetration, conduction-mode fusion welds were produced simultaneously on the top and bottom of the sheets; however, at slower welding speeds, the welds become larger and become thermally coupled and merge to form a single symmetric, complete-joint-penetrating double-sided (DSA) weld with arc weld widths about half that normally required for conventional one-sided, complete-joint-penetration GTAW (Refs. 7, 8) and significantly reduced weld distortions.

In the present study, the cleaned 1.0- and 1.5-mm-thick AA5182-O sheets were clamped in the butt joint configuration and shimmed as shown in Fig. 1 so that the top surfaces of the specimens were level with each other. This reduced weld metal drop-through (Refs. 15, 16). The clamped specimens were then welded using the autogenous, conduction-mode DSAW process with a variable polarity power supply and additional trailing shielding gas to further minimize hydrogen porosity. A more detailed description of the experimental apparatus and welding procedures used may be found (Refs. 15, 16).

As shown in Fig. 2, initial conduction-mode DSA welds were made using welding powers of 2.6 and 4.2 kW within a range of welding speeds indicated by the solid lines that had previously been identified to result in visually acceptable, “good” welds in these AA5182-O aluminum sheets (Refs. 15, 16). Following welding, metallographic examination of sectioned, polished, and etched weld specimens was done using an optical microscope and a SEM. Etching was performed using Beck’s reagent for the grain structure and Keller’s reagent for the solidification microstructure (Ref. 46).

Mechanical properties of the base metal sheets including yield strength, ultimate tensile strength, and elongation were measured using standard sheet metal tensile-test specimens (Ref. 47) and a tensile-test machine with a cross-head speed of 10 mm/min. All tensile tests were repeated five times. Longitudinal tensile tests of only the weld metal were also done using quarter-size tensile specimens (Ref. 47). This tensile test specimen geometry allowed direct quantitative comparisons to be made between the properties of the DSA welds made using different weld process conditions and those of the base metal.

Postweld formability tests of the TWB specimens were done using a hemispherical dome-height (HDH) test apparatus (Ref. 2). In this test, the formability of the weld and TWB is tested in a biaxial strain state. All welds and HDH tests were repeated five times. The TWB weld specimens were sheared to 203 × 203 mm (8 × 8 in.) size such that the weld ran down the center of the HDH specimens. A 0.5-mm-thick shim was used over the thin sheet to obtain consistent clamping of the unlubricated TWB sheets during the HDH test. A 440-kN blank clamping force was used. The diameter of the standard HDH punch was 101.6 mm (4 in.) and the punch speed used was 0.25 mm/s (0.01 in./s).
Results and Discussion

Figure 3 shows a typical top weld bead of a DSA weld made using 2.6-kW power and a welding speed of 40 mm/s. The bottom weld beads were similar to those of the top surfaces. The oxide had been cleaned from the weld bead surface. The weld bead quality was good with no surface defects such as rough surface, undercut, or centerline solidification shrinkage pores.

Figure 4 shows a typical transverse section and microstructure of the DSA welds. The welds were complete-joint-penetration, symmetric welds with straight or slight hourglass shaped weld fusion boundaries. In the weld metal, epitaxial growth initially occurred from the base metal grains at the fusion boundary and these grew as columnar grains toward the center of the welds where there is a beneficial transition from columnar grains to equiaxed grains.

Solidification cracking or hydrogen porosity was not observed in any of these welds. In addition, there is only a small amount of sagging of the weld bead due to gravitational forces acting on the molten metal during welding. Surface tension on both the top and bottom weld bead surfaces has kept the weld relatively flat with no sudden changes in contour that might act as a stress concentration and surface defect sight for initiation of strain localization and shear banding during deformations leading to decreased ductility and formability.

Overall, the surface quality and bead geometry of the DSA welds are significantly better than has been reported for welds produced using the keyhole-mode LBW (Refs. 12, 13, 17–25) and NVEBW (Refs. 23, 24, 26, 33, 34) processes. This is expected to facilitate improved ductility and formability of these DSA fusion welded specimens (Refs. 7–9, 33, 34).

Longitudinal Tensile Properties of the Weld Metal

Figure 5 shows the stress-strain curves for the median of the five replications of the tensile tests of both of the AA5182-O base metal sheets as well as the median of five replications of tensile tests of the weld metal from DSA welds made using 4.2-kW welding power as well as 50, 60, and 70 mm/s welding speeds. There was no significant nor measurable difference between the properties of the 1.0-mm and 1.5-mm-thick base metal sheets. The measured mechanical properties of the A5182-O base metal sheets obtained from this stress-strain curve are shown in Table 1. The stress-strain curves for the three welds shown in Fig. 5 are initially similar to the base metal. This is expected, since the base metal is in the O temper; however, in all cases, the weld metal stress-strain curves are significantly foreshortened by fracture of the specimens at strain values less than a third of the base metal fracture strain. Thus, both the ultimate tensile stress and the percent elongation of the weld metal of all three welds were significantly less than the base metal sheets.

Figure 6 is a plot of the longitudinal tensile properties of the welds normalized as a percentage of the base metal properties. The yield strengths of the weld metal were very similar to the stress of the base metal.

Table 2 — Hemispherical Dome Height Test Results for 1.0- to 1.5-mm-Thick AA5182-O Aluminum Alloy Tailor Welded Blanks Welded with a Power of 2.6 kW

<table>
<thead>
<tr>
<th>Specimen/Welding Conditions</th>
<th>Hemispherical Dome Height (mm)</th>
<th>Fraction of 1.0-mm-Thick Base Metal Formability (%)</th>
</tr>
</thead>
<tbody>
<tr>
<td>1.5-mm base metal</td>
<td>28.0 ± 0.29</td>
<td>—</td>
</tr>
<tr>
<td>1.0-mm base metal</td>
<td>27.4 ± 0.33</td>
<td>—</td>
</tr>
<tr>
<td>2.6 kW, 30 mm/s</td>
<td>14.3 ± 0.66</td>
<td>51.9</td>
</tr>
<tr>
<td>2.6 kW, 40 mm/s</td>
<td>14.4 ± 2.41</td>
<td>52.3</td>
</tr>
<tr>
<td>2.6 kW, 50 mm/s</td>
<td>10.6 ± 1.93</td>
<td>38.4</td>
</tr>
</tbody>
</table>
while the ultimate tensile strengths ranged from 65 to 82% of the base metal strength. In all cases, the ductility of the weld metal was less than 60% of the base metal and decreased with increasing welding power and speed to less than 30% of the base metal ductility. These results are consistent with those reported in most studies of autogenous fusion welding of the non-heat-treatable AA5182-O and AA5754-O aluminum alloy sheet for TWB applications, i.e., the hardness, yield, and ultimate tensile strengths were similar to the unwelded sheet; however, significantly reduced ductility with failure occurring in the weld was observed (Refs. 7–9, 14–26).

**Hemispherical Dome Height Formability Tests**

Hemispherical dome-height formability tests were performed on weld specimens made using 2.6-kW power and 30, 40, and 50 mm/s welding speeds because these welds had exhibited the best strength and ductility. The HDH test results are shown in Table 2.

The maximum dome heights of the two base metal sheet thicknesses used were similar (28 and 27.4 mm). Necking of the sheets consistently began along the line where the sheet just left contact with the punch. Fracture initiated in this necked region and quickly propagated circumferentially in both directions from this point. However, the TWB welded specimens had significantly lower hemispherical dome heights ranging from only 52% to 38% of the base metal values. Also, as shown in Fig. 7, all DSA welded TWB specimens fractured along the centerline of the weld with little or no evidence of necking. These results are similar to those reported by Shakeri et al. (Refs. 23, 26) and Buste et al. (Ref. 24) for their NVEB welded AA5754-O aluminum alloy TWBs. Their HDH values were only 45 to 55% of the base metal sheet values when failure occurred in the NVEB weld; however, they were >80% of the base metal values when the TWB failed in the thinner sheet.

**Metallographic and SEM Analysis**

The fracture surfaces and microstructures of the DSA welded TWBs were examined in greater detail to further elucidate the reasons for the observed reduced ductility and formability of the fusion welds in the AA5182 aluminum TWB specimens. Figure 8 shows the fracture surface morphology of the weld made with 2.6-kW power and 30 mm/s welding speed. The entire weld metal fracture surface exhibited a classic ductile fracture morphology. This is consistent with the longitudinal tensile test results, i.e., these welds had the largest weld metal ductility.

The fracture surfaces of TWB specimens made using welding speeds less than 30 mm/s had a mix of ductile fracture and dendritic surface morphologies. Figure 9 shows a representative fracture surface of a weld made using 4.2-kW power and 60 mm/s welding speed. At low magnification (see Fig. 9A), the entire fracture surface appears to have a ductile fracture morphology; however, at higher magnification, a mixture of ductile fracture surfaces (e.g., see Fig. 8) and dendrites (see Fig. 9B) were observed. The presence of regions of dendritic morphology on the fracture surface is direct evidence that solidification shrinkage microporosity was present at the weld centerline where the DSA welded specimens fractured.

To further verify the existence of solidification microporosity in the DSA welds, polished, but unetched, transverse sections were examined using an optical microscope. Solidification shrinkage pores can be identified by their irregular shapes that follow the local solidification morphology, whereas gas pores have smooth spherical shapes. Also, solidification shrinkage microporosity is most often observed in clusters in direct registry with the solidification microstructure, whereas gas pores are more often isolated and randomly spaced in the solidified metal (Ref. 48).

Solidification microporosity was observed in all of the DSA welds to varying degrees. Figure 10A shows an example of a weld made using a welding power of 4.2 kW and 50 mm/s welding speed. This high-power, high-speed weld has a significant amount of microporosity scattered throughout the weld. Welds made using lower welding powers and speeds exhibited much less microporosity. At higher magnification, networks of interdendritic microporosity of the order of 10 μm in size and...
smaller such as shown in Fig. 10B were evident throughout the weld. Hydrogen porosity-induced solidification micropores of the order of 10 \( \mu \text{m} \) diameter with very irregular boundaries were also seen (see Fig. 10C).

The small gas pores nucleated during the last stages of solidification when there was very little liquid metal remaining so that the pore walls took on the irregular geometry of the surrounding dendrites rather than a smooth spherical gas pore interface. Microporosity of the order of 10 \( \mu \text{m} \) in size and smaller would not normally be detected using conventional nondestructive detection techniques, rather their detection would require use of more specialized nano-focus x-ray tubes or destructive examination techniques such as optical metallography and SEM (Ref. 49), and thus frequently goes undetected during conventional nondestructive examinations (NDE).

There is a paucity of information in the welding literature regarding solidification microporosity in fusion welds. Pastor et al. (Ref. 19) observed hydrogen porosity and microporosity in their AA5182 and AA5754 laser welds, but attributed them to hydrogen from their use of wet helium shielding gas. Shakeri et al. (Refs. 23, 26) and Buste et al. (Ref. 24) observed significant microporosity in their AA5754 NVEB welded specimens and suggested that the reduced ductility and formability of their welds was caused by these defects. This was later supported by the analysis of the role of micron-scale defects in reducing ductility and formability of these welds by Bayley and Pilkey (Refs. 33, 34).

In subsequent related work, Worwicks et al. (Ref. 50) modeled damage development and percolation originating from clusters of submicron scale second-phase particles during deformation of the wrought AA5182 and AA5754 base metal sheet. The development of damage during deformation by nucleation, growth, and coalescence of adjacent microvoids was shown to have a very detrimental effect on the material’s ductility and formability. They found that the rate of damage accumulation was dominated by the nucleation stage; however, the accuracy of their predictions were very sensitive to the very subjective assumptions made regarding void nucleation.

The presence of solidification microporosity in the form of clusters or closely spaced networks of interdendritic, irregular-shaped voids or pores such as those shown in Fig. 10 helps to explain why the ductility and formability of the DSA welds were much lower than the wrought base metal. During forming and tensile plastic deformation of the TWB, nucleation of microvoids is not required; rather, the...
preexisting solidification micropores such as those shown in Fig. 10 would have created stress concentrations that localize the strain between adjacent pores and this would quickly lead directly to void growth, void coalescence, and finally premature fracture of the weld. Also, the volume fraction of solidification microporosity was observed to increase with increasing welding speed and this in turn caused the decrease in weld metal ductility with increasing welding speed evident in Fig. 6.

Solidification microporosity is a well-known problem in metal castings (Ref. 51). While it is normally undetectable using conventional NDE techniques, this small defect is known to be quite detrimental to the ductility and fatigue properties of castings (Ref. 52). Solidification microporosity is caused primarily by the inability of liquid metal to flow fast enough down the long constricted interdendritic spaces created during dendritic solidification of long-freezing range alloys.

If the pressure, \( P \), in the liquid during the last stages of solidification drops below the local vapor pressure of the liquid due to viscous drag as the liquid tries to flow down the long, narrow interdendritic spaces, then a microvoid with the added liquid-gas interfacial energy will nucleate and grow in the remaining interdendritic space, i.e., assuming no restriction to pore nucleation, a pore of radius, \( r \), will nucleate when

\[
P \leq P_a - \frac{2 \sigma_{lv}}{r}
\]

where \( P_a \) is the alloy vapor pressure, and \( \sigma_{lv} \) is the liquid-vapor surface tension (Ref. 51). When there is a dissolved gas such as hydrogen in the liquid, the contributions of dissolved gas and pressure drop due to interdendritic flow are additive, and it can be difficult to clearly distinguish between solidification shrinkage-based porosity and gas porosity (for example, see Fig. 10C). In this case, \( P_a \) becomes the larger of the equilibrium partial pressure of the dissolved gas, and the alloy vapor pressure resulting in pore nucleation sooner at higher pressures (Ref. 51). To avoid dissolved hydrogen from contributing to the nucleation of microporosity, the dissolved hydrogen concentration in the melt must be less than the equilibrium solubility of hydrogen in solid aluminium, i.e., < 0.04 ppm (Refs. 6, 51).

The susceptibility of alloys to microporosity is known to increase in proportion to the freezing range of the alloy because the longer freezing range alloys typically solidify with a columnar-dendritic solidification morphology with relatively large distances between the dendrite tips close to the liquidus temperature, \( T_L \), and the root or base of the dendrites where solidification is complete at the nonequilibrium solidus temperature, \( T_S \). This creates a two-phase mushy zone with long, narrow interdendritic channels between the dendrites that restrict fluid flow to the dendrite roots. This choking off of the supply of molten metal to the dendrite roots ultimately causes nucleation and growth of interdendritic and intergranular voids at the dendrite roots during the final stages of solidification.

Under equilibrium solidification conditions, the freezing range is given by the difference between the equilibri-
Reducing Solidification Microporosity in Autogenous Fusion Welded AA5182-O Aluminum

There are two possible mechanisms for formation of solidification microporosity in welds: 1) solidification microporosity that can occur anywhere in a weld (for example, see Fig. 10A) during solidification of longer freezing range alloys while solidifying at high rates with a cellular dendritic or dendritic microstructure and low thermal gradients, and 2) solidification porosity and cracking of longer freezing range alloys along the centerline of welds made at high welding speeds with pronounced long, tear-drop shaped weld pools. This frequently leads to premature failure along the weld centerline such as shown in Fig. 7.

Reducing Solidification Microporosity in Fusion Welds

It is well known that solidification microporosity in castings can be avoided by increasing the thermal gradient between the liquidus and solidus temperature isotherms during solidification (Ref. 51). This shortens the mushy zone, thereby reducing the distance that the interdendritic liquid must flow between the dendrites and facilitating adequate supply of molten metal to the final metal to solidify at the dendrite roots. It can also cause a transition from a dendritic solidification morphology to more favorable morphologies such as cellular dendritic or cellular. In casting, this is accomplished using a combination of risers, casting design, and chills (Ref. 51).

However, these techniques are not appropriate in welding. Instead, initial insight into the weld process parameters that might be used to increase the thermal gradient and shorten the mushy zone during solidification of a fusion weld may be obtained by examining the analytical equation for the cooling rate down the centerline of a thin sheet heated by a line heat source developed by Adams (Ref. 54), i.e.,

\[
G_{CL} = \frac{2\pi k \rho C_p}{\eta v_{ws}} \left( \frac{t}{H_{net}} \right)^2 (T_c - T_0)^3
\]

where \(G_{CL}\) is the cooling rate (K/m at \(T_c\)); \(k\) is thermal conductivity (W/mK); \(\rho\) is density (kg/m³); \(C_p\) is average specific heat (J/kgK); \(t\) is the sheet thickness (m); \(T_c\) is the temperature where the gradient exists (K) (for example, the liquidus temperature \(T_L\)); \(T_p\) is the preheat temperature (K); \(E\) is the welding power (W); \(I\) is the welding speed (m/s); \(\eta\) is the arc efficiency; \(V\) is the welding voltage (V); \(P\) is the welding power squared, i.e., \(P = E \times I\) and welding speed, \(v_{ws}\), gives

\[
G_{CL} = 2\pi k \rho C_p v_{ws} \left( \frac{t}{\eta E I/v_{ws}} \right)^2 \left( T_c - T_0 \right)^3
\]

Thus, from Equation 3, \(G_{CL}\) is predicted to increase in proportion to the welding speed, i.e., \(G_{CL} \propto v_{ws}\) whereas \(G_{CL}\) is inversely dependent on the welding power squared, i.e., \(G_{CL} \propto 1/P^2\). This suggests that the greatest increase in \(G_{CL}\) would be realized by decreasing the welding power, \(P\).

It should be noted that there is no apparent limit to the welding speed or power that can be used in these equations because a weld pool is always predicted to exist by this model due to the mathematical singularity that exists.
ists at the line heat source. In reality, most heat sources are Gaussian distributed over the weld surface and this and other factors result in practical limits to the values of power and welding speed that can be used while still producing complete joint penetration welds.

Improved estimates of the thermal gradients at the tail of the welds made in the present study may be obtained using the analytical heat transfer model of the DSAW process by Kwon and Weckman (Ref. 55). This model includes use of Gaussian distributed arc heat sources on both the top and bottom of the sheet rather than a line heat source as used in Adam’s solution (Ref. 54). This model and the properties and parameters used in the model are described in greater detail in Refs. 15, 16, and 55.

Figure 11A shows the predicted weld pool and isotherms on the top surface of the DSA weld made at 50 mm/s using 4.2 kW. The location of the plasma arc on the top surface is indicated where the radius of the circle is the distribution coefficient of the Gaussian distributed arc. Also shown are the equilibrium solidus isotherm, 859 K (586ºC), and the nonequilibrium solidus isotherm predicted using the Scheil equation, 718 K (445ºC). The distance between the weld pool tail and these isotherms is an indication of the size of the mushy zone in the weld pool and the thermal gradient during solidification. In this case, G-cl at the liquidus temperature is predicted to be 20,700 K/m.

From Equation 3, G-cl will increase with increased welding speed. This may be seen in the predicted isothermal plot of the weld made using 4.2 kW, but at a higher welding speed of 80 mm/s — Fig. 11B. The weld width and weld pool size are smaller than the weld shown in Fig. 11A, because H-net is smaller. Note that the weld pool length/width ratio or aspect ratio remains about the same, but the weld width and weld pool size are smaller because H-net has decreased.

Also, the distance between the tail of the weld and the solidus temperatures has decreased noticeably, thereby increasing the thermal gradient. In this case, the thermal gradient was predicted to increase from 20,700 to 26,300 K/m with the increased welding speed. This trend is consistent with that predicted by Equation 3.

From Equation 3, G-cl will also increase in inverse proportion to the square of the welding power, i.e., G-cl ∝ 1/P². The effects of this may be seen by comparing the isothermal plots shown in Fig. 11A for the weld made using 4.2 kW at 50 mm/s and Fig. 11C for the weld made at the same speed using only 2.6 kW power. In this case, the decrease in power has resulted in a weld that is smaller and the distance between the tail of the weld pool and the solidus isotherms is much smaller. This decreased distance between the liquidus and solidus isotherms is responsible for the large increase in the predicted thermal gradient, which was predicted to increase significantly from 20,700 to 36,900 K/m. The large increase in thermal gradient with decreased welding power is consistent with the observation that the severity of microporosity was much lower in the welds made using the lower welding powers.

Predictions of the centerline thermal gradient at the tail of the welds,
GCr, made using 2.6- and 4.2-kW power vs. welding speed are shown in Fig. 12. The 4.2-kW welds had thermal gradients ranging from 20,700 to 25,000 K/m. The gradients increased initially with welding speed, but appeared to asymptotically approach a constant maximum value at the higher welding speeds. The thermal gradients of the lower 2.6-kW power welds were much higher, ranging from 31,800 K/m to an asymptotic maximum value of 36,900 K/m. Again, there appears to be a maximum thermal gradient possible with increasing welding speed.

In the DSAW process, the maximum welding speed will also be limited by either the onset of inconsistent arc coupling, or decrease in the weld size and width to the point that thermal coupling between the top and bottom weld pools is lost and separate smaller, partial-joint-penetration welds will be made on the top and bottom of the sheets leaving an incomplete fusion defect in the middle of the sheets (Refs. 14, 55).

Considering the trends for significant increase of the thermal gradient by decreasing welding power, a number of new low-power welding process conditions were identified by projecting the range of good welding conditions shown in Fig. 2. Note that use of decreased weld power also required a reduced welding speed in order to provide sufficient energy input per unit distance, \( H_{net} \), to heat and melt the metal and form a weld. In this case, the welding speeds were reduced such that \( H_{net} \), and therefore the weld width, were held about the same as the weld made using 2.6 kW at 30 mm/s (i.e., \( H_{net} \approx 29 \times 10^{-3} \) J/m and width \( \approx 5 \) mm). While using all procedures necessary to minimize hydrogen gas porosity, three acceptable lower power welding conditions were identified experimentally that produced good DSA welds using lower powers and welding speeds ranging from 2.2 kW at 25 mm/s to 1.3 kW at 15 mm/s. Welds could not be made using welding powers less than 1.3 kW. The predicted isotherms and weld pool shape of the weld made at the lowest welding power of 1.3 kW are shown in Fig. 11D. While the weld width was slightly larger than the weld produced using a higher power and welding speed shown above in Fig. 11C, the aspect ratio of the weld pool is smaller and the distance between the weld pool tail and the solidus isotherms is the smallest of all welds shown. The predicted thermal gradient in this lowest power weld has increased even more significantly from 20,700 to 45,500 K/m. This is the highest thermal gradient of all welds thus far.

The predicted thermal gradients of these three new low-power welds are shown in Fig. 12. All three of the new welds have much higher thermal gradients with the highest gradient of all welds being 45,500 K/m made using the lowest power and speed at 1.3 kW and 15 mm/s. Welds produced using these conditions can be expected to have the lowest propensity for hydrogen porosity and solidification microporosity and, therefore, the best ductility and formability of all welds thus far.

### Reducing Centerline Solidification Porosity and Cracking in Fusion Welds

There is a natural desire in production environments to weld as fast as possible and, as indicated in Fig. 2, such increased welding speeds require use of increased welding power to maintain adequate energy input/unit distance to obtain complete penetration welds with the desired width, \( W \). However, as the welding power and speed is increased, the length of the weld, \( L \), increases and long tear-drop weld pool shapes are produced such as the weld shown in Fig. 13.

Weld pool geometries with high length-to-width or aspect ratios are known to result in macrosegregation that occurs during normal solidification of the weld. This creates high solute concentrations at the centerline with solidus temperature depression and subsequently solidification porosity and centerline cracking along the
center of the weld (Refs. 6, 56). While centerline solidification cracking was not observed in any of the DSA welds in this study, the presence of solidification porosity near the center of the weld and a contiguous grain boundary with high solute concentrations down the weld centerline would lower ductility and formability of the weld and cause centerline cracking similar to that in the HDH specimen shown in Fig. 7 (Ref. 6).

The propensity for solidification pores and cracking along the center of welds can normally be reduced by decreasing the welding speed so that the weld pool aspect ratio is decreased to produce a more rounded weld pool shape (Ref. 6, 56); however, as shown in Fig. 2, this would also require reduced welding power so that excessive weld width and melt-through defects are not produced.

As indicated in Fig. 13, the length and width of a DSA weld pool can be measured approximately by examining the top weld bead surface at the end of the weld, i.e., the final crater. The weld width, \( W \), and leading edge of the weld pool are easily identified and measured. The shape of the fusion boundary and ultimately the point of final solidification at the weld pool tail can be estimated by following the small ripples on the weld bead surface formed by normal fluctuations of the weld pool surface during solidification. From this, the length, \( L \), and ultimately the weld pool length-to-width or aspect ratio can be estimated. This can be a very subjective measurement.

An alternative and somewhat less subjective parameter that can also be used to characterize the weld pool aspect ratio is the nondimensional weld pool Peclet number, \( Pe \), defined as

\[
Pe = \frac{v_{ws}L_C}{\alpha}
\]  

(4)

where \( \alpha = k/\rho C_p \) is the thermal diffusivity of the metal and \( L_C \) is a characteristic length dimension that, in this case, we chose to be the easily measured weld width. The Peclet number is derived by nondimensionalizing the governing energy equation in the solid, and it represents the ratio of energy transport in the plate by mass advection over energy transport by thermal conduction (Refs. 57, 58).

A characteristic of the Pe number is that welds with the same Pe number will be geometrically similar, e.g., have the same weld pool aspect ratio, even though they have different values for \( v_{ws}, L_C \), and \( \alpha \). Thus, there is a direct, linear relationship between the weld pool aspect ratio and the Peclet number. In this case, however, the Peclet number can be easily determined by measuring the weld width, \( W \) (see Fig. 13), and knowing the welding speed and thermal diffusivity of the alloy.

Figure 14 is a plot of the measured and predicted weld pool aspect ratios vs. weld pool Peclet numbers for all of the DSA welds. In both cases, the weld pool aspect ratio increases linearly with the weld pool Peclet number; however, the predicted aspect ratios are always less than the measured values. While the analytical solution is deliberately tuned to predict the same weld width as the measured width, the predicted weld length will always be less than the actual weld pool length because the analytical model does not include the effects of latent heat of fusion.

The absorption and release of latent heat of fusion during melting and solidification of the weld metal will increase the length of the actual weld pool, thereby creating a larger measured weld pool aspect ratio. Irrespective, all other trends are consistent and linear. The measured aspect ratios ranged from two for the lowest power 1.3-kW weld to about six for the 4.2-kW welds, whereas the predicted aspect ratios varied over a smaller range from 1.2 to 2.2. In both cases, the weld pool aspect ratio and Pe decreased linearly as the welding power and speed were decreased. This general relationship is also evident in the isothermal plots shown in Fig. 11.

In both cases, the weld with the lowest propensity for solidification shrinkage pores and centerline cracking defects, for example, the weld produced using the lowest power and welding speed (1.3-kW and 15 mm/s), was also the weld with the lowest weld pool aspect ratio and Pe value. Note also that the aspect ratio and Pe did not change significantly with welding speed at a given welding power, e.g., 4.2-kW.

The relationships between weld aspect ratio, \( Pe \), and the predicted centerline thermal gradient, \( G_{CL} \), for all welds are shown in Fig. 15. In this plot, the thermal gradient is inversely proportional to the weld pool aspect ratio and the Pe number where the highest \( G_{CL} \) and, therefore, the conditions with the lowest propensity for solidification shrinkage microporosity occurs in welds with the lowest weld pool aspect ratio and lowest Pe. These conditions also minimize the probability of creating centerline solidification pores and cracking defects. In all cases, the welding power used had the greatest effect. At any given power, the variation of the welding speed within the limits of producing acceptable welds had much less effect. Thus, the weld produced using the highest power (4.2 kW) and speed (80 mm/s) is most likely to be subject to...
microporosity and centerline defects that would severely limit weld ductility and formability, while the weld produced using the lowest power (1.3 kW) and welding speed (15 mm/s) is least likely to be susceptible to these defects and exhibit the best ductility and formability.

**Properties of the Low-Power Welds**

The weld bead surfaces and weld bead geometry and grain structures observed in transverse sections of all of the low-power welds were very similar to those shown in Figs. 3 and 4, respectively; however, there was no evidence of hydrogen gas porosity or solidification microporosity or cracking at this or higher magnifications in the DSA welds made using the lower welding powers of 2.2, 1.8, and 1.3 kW. This is consistent with the discussions in the previous section. In addition, the yield strengths of the low-power welds and base metal were the same. This suggests that the yield strength of the A5182-O alloy weld is not influenced significantly by the presence of solidification microporosity in the weld fusion zone and is consistent with many previous studies of the effects of porosity on weld properties (Refs. 30–32).

The measured ultimate tensile strength, ductility, and hemispherical dome-height measurements of all DSA welds normalized with respect to the base metal properties vs. welding power are shown in Fig. 16. In all cases, the first three welds made using the lowest powers and speeds had significantly improved properties relative to the initial welds produced using the higher 2.6- and 4.2-kW powers and higher welding speeds.

The ultimate tensile strength (UTS) and ductility of the weld metal improved with decreasing weld power and speed. The UTS values have increased to about 90% of the base metal UTS values. The specimen made with 1.8-kW power and 20 mm/s welding speed had a weld pool Peclet number of 2.3 and the best weld metal ductility. The percent elongation of this weld was 19.2%, which is > 47% better than all of the welds made using 2.6- and 4.2-kW powers and is very close to the base metal ductility of 22.5%.

While the measured UTS and ductility of the lower power welds have significantly improved, they are still slightly below the base metal properties. This may be due to the differences between the weld metal and base metal microstructures. As shown in Fig. 4, the weld metal solidification microstructure has a larger average grain size that would result in less grain boundary strengthening than the base metal, which has a smaller grain size (Ref. 59).

The smaller number of grains across the subsized longitudinal tensile specimens can also be expected to make the weld metal more susceptible to grain boundary embrittlement. In addition, the second phase Fe- and Mn-based intermetallic particles were larger in the weld metal. This would be expected to promote strain localization and void nucleation in the metal during deformation that would decrease the weld metal ductility relative to the base metal (Ref. 50).

As shown in Table 3 and Fig. 16, there was a very significant increase in HDH formability of the low-power weld specimens. The TWB made using 1.8 kW welding power and 25 mm/s welding speed had the best HDH formability with a HDH value at fracture of 24.3 mm or 88.8% of the base metal formability. Also, the HDH values of the other two low-power welds was much better than all welds made using 2.6- and 4.2-kW welding powers. All of the low-power welds had formability values > 80% of the base metal values. This represents an increase of > 64% in formability relative to the welds made using the higher 2.6-kW and 4.2-kW welding powers.

There was a change in the failure mode of the three low-power HDH specimens compared to the specimens welded using the higher powers. Photographs of typical failure modes are shown in Fig. 17. None of the low-power welds failed in the weld. There was increased base metal deformation followed by necking in a circumferential direction in the base metal sheet at the first point of contact with the punch similar to that observed in the base metal specimens (see Fig. 17A and B).

Figure 17A shows the HDH specimen that was welded using 2.2 kW welding power and 25 mm/s welding speed. As is indicated, fracture of the specimen initiated at the intersection between the weld and the circumferential line or ring of first contact between the sheet and the HDH punch. The fracture then propagated circumferentially toward the thinner base metal sheet. The specimens welded using 1.8-kW welding power failed in the thin sheet circumferentially along the first point of contact between the punch and the sheet (see Fig. 17B).

Figure 17C shows the typical failure mode observed in the welding specimens made using 1.3-kW welding power. These HDH specimens failed parallel to the weld in the thinner base metal sheet. This mode of failure has been shown to be related to the restraint to strain in the direction parallel to the weld provided from the thicker sheet that shifts the strain state in the thinner sheet from a biaxial strain state toward a plane strain condition that has lower formability.

It is clearly evident from Figs. 16 and 17 that eliminating surface defects such as undercutting and geometric stress concentrations as well as internal weld defects such as hydrogen porosity and solidification microporosity in the AA5182-O DSA welds has significantly improved the ductility and formability of the fusion welded TWBs. This is supported by others who have used the GTAW process to weld aluminum alloy TWBs (Refs. 7–10) and by models of the influence on weld defects on ductility and formability of welds (Refs. 7–9, 33, 34, 50); however, these researchers have not offered an explanation for the presence of these internal weld defects nor provided guidance as to how to minimize or eliminate these weld defects.

The results of the present study (Ref. 15) suggest that DSAW and other fusion welding processes should be capable of producing high-quality, defect-free welds with good mechanical properties and formability provided very stringent contaminant and oxide precleaning and shielding gas procedures are used to eliminate hydrogen porosity in the weld metal and provided the welds are made using the welding powers and welding speeds low enough to avoid formation of solidification micropores, and solidification pores and centerline cracking defects in the weld metal. Finally, the welding process must be capable of producing very good weld bead surface quality with low surface
roughness on both sides of the weld and with little or no geometric stress concentrations such as undercuts.

In this study, the autogenous, conduction-mode DSA fusion welding process has been shown to be capable of satisfying all of these combined criteria and producing TWBs with superior ductility and formability in A5182-O aluminum sheets. It is expected that these same principles and procedures would apply when using any other fusion welding process.

Conclusions

While the weld metal yield strength of autogenous, conduction-mode DSA fusion welds made between 1.0- and 1.5-mm-thick AA5182-O aluminum alloy sheets was found to be similar to the base metal sheets, the ultimate tensile strength, percent elongation, and hemispherical dome height (HDH) formability of the DSA tailor-welded blank (TWB) specimens were low. These unacceptable properties were attributed to the presence of weld metal defects including hydrogen porosity, solidification microporosity, and solidification porosity clustered near the weld centerline and centerline cracking that collectively acted as strain localizers that obliterated the normal void nucleation stage during plastic deformation and ductile fracture and quickly led to void growth, void coalescence, and final premature fracture of the welds during plastic deformation and forming.

Hydrogen porosity was prevented by using established techniques for fusion welding of aluminum alloys. However, a new technique was developed for reduction and elimination of solidification microporosity in the autogenous DSA fusion welds in the AA5182-O alloy sheets that involved welding at very low welding powers and speeds to create high thermal gradients at the weld interface during solidification. Centerline solidification porosity and cracking was also prevented using these same low welding powers and speeds by changing the weld pool shape from a long teardrop shape to a smaller, more circular shape with low weld pool length/width aspect ratios and nondimensional weld pool Peclet numbers.

Using these combined techniques and procedures, acceptable DSA fusion welds made at low powers ranging from 1.3 to 2.2 kW and low speeds ranging from 15 to 25 mm/s were found to have no hydrogen porosity, solidification microporosity, or solidification porosity and cracking at the weld centerline. The properties of these high-quality, defect-free welds were found to be significantly improved, with weld metal UTS increased from < 80% to > 90% and ductility increased from < 60% to > 80% of the base metal sheets properties. Also, the maximum hemispherical dome height increased significantly from < 52% to > 80% of the base metal dome height. In all cases, the tensile and HDH test specimens failed in the thin sheet, not in the weld. Thus, the ductility and formability of these fusion welded specimens do not appear to be limited by the DSA fusion weld.

The results of this study suggest that the autogenous, conduction-mode DSA fusion welding process can be used to make tailor-welded blanks in AA5182-O aluminum alloy sheets with good weld metal ductility and formability provided that all internal weld metal defects and all weld bead surface defects are avoided. This requires careful cleaning and removal of hydrated oxides prior to and during welding to prevent hydrogen porosity as well as welding at low welding powers and speeds, so that the formation of solidification microporosity and centerline solidification porosity defects and centerline cracking in the weld are prevented. It is expected that these same principles and procedures could be used beneficially for all fusion welding processes that might be used to make aluminum tailor welded blanks with improved ductility and formability.

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different welding techniques.


