



American Welding Society®

SUPPLEMENT TO THE *WELDING JOURNAL*, APRIL 2022
Sponsored by the American Welding Society

WELDING
RESEARCH

The Effects of Postweld Processing on Friction Stir Welded, Additive Manufactured AlSi10Mg

An examination of postweld heat treatments and hot isostatic pressing on the overall joint quality and mechanical performance

BY M. EFF, H. HACK, D. SHIPLEY, AND S. SHIRA

Abstract

Given the limited build volumes for most current additive manufacturing (AM) machines, a method for taking advantage of the unique capabilities offered by AM while combining it with traditional manufacturing methods is needed. Welding of AM-produced components is a solution to this challenge. The objective of this study was to determine the feasibility of using friction stir welding (FSW) to join AlSi10Mg melted with laser powder bed fusion (PBF-L) and examine the effects of postweld heat treatment (HT) and hot isostatic pressing (HIP) on overall joint quality and mechanical performance. Samples were examined using optical microscopy, scanning electron microscopy, hardness testing, and tensile testing. Examination of the samples that underwent a postweld annealing HT revealed cracking along the stir zone and the thermomechanically affected zone (TMAZ) boundary. Examination of the crack revealed evidence of liquation near single-phase silicon precipitates in the TMAZ despite the annealing temperature being 27°C (81°F) below the solidus temperature of the material according to the material specification. Using a calculated pseudobinary phase diagram of AlSi10Mg, the annealing HT was determined to be in the partial liquation regime for AlSi10Mg. The voids and crack formation mechanisms were determined to be caused by constitutional liquation coupled with the unique TMAZ microstructure and stress state. The as-welded and HIP coupons were void and defect free, and FSW was determined to be a feasible

method of joining PBF-L aluminum alloys with minimal knockdown in tensile strength.

Keywords

- Friction Stir Welding
- AlSi10Mg
- Additive Manufacturing
- Additive Manufacturing Welding
- Constitutional Liquation

Introduction

Additive manufacturing (AM) of components offers advantages in design flexibility, rapid prototyping, and small-volume production of components. AM features a range of build complexities and build times with the two often having a negative correlation. Detailed and higher-quality builds take hours or even days depending upon build size (Ref. 1). As the volume of the AM build increases, the time and cost and chances for a quality issue are all negatively impacted (Ref. 1). A method for taking advantage of the unique capabilities offered by AM while combining it with traditional manufacturing methods, such as extrusions, castings, machined components, pre-forms, and forgings, is needed to optimize production speed, improve flexibility of the overall method, improve quality, and ensure a cost-effective solution (Refs. 2, 3). Welding of AM-produced components is one such solution that enables traditional manufacturing techniques to be mated with AM. Welding also allows for the production of larger AM parts than can fit within the build volume of current AM machines.

Welding of laser powder bed fusion (PBF-L) parts and other AM-produced parts is challenging. As the parts are created by PBF-L, only small volumes of material are melted, and they are often highly compositionally segregated as a result of solidification. Additionally, the repeated heating

<https://doi.org/10.29391/2022.101.009>

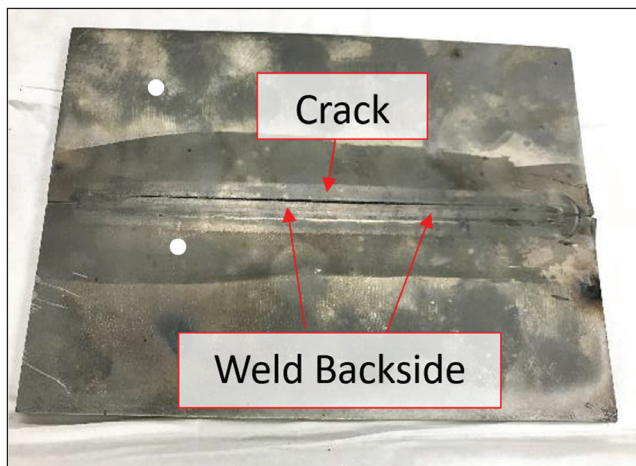


Fig. 1 — Crack formed during quenching (anneal and quench postprocessing).

and cooling of the part can lead to a unique residual stress profile within the component. Materials that are more resistant to solidification cracking, such as steel and titanium, have been successfully printed using AM and welding (Refs. 4–6). Cracking of AM components when welded has been observed in 304 stainless steel despite the material being readily joined via fusion-based processes. The severity of cracking increases with melt pool size during laser welding (Ref. 7). Solid-state welding has been demonstrated to join PBF-L aluminum (Al) parts and results in a defect-free joint (Refs. 8, 9). Solid-state welding technologies, including friction stir welding (FSW), alleviate the potential for liquation cracking because no melting typically occurs (Ref. 10). FSW can join complex geometries and results in reduced residual stresses when compared to fusion welding (Refs. 10–13).

This work focused on the joining of additive manufactured AlSi10Mg using FSW. Alloy AlSi10Mg is a common choice for PBF-L of Al because it is readily printed. As printed, it possesses an ultimate strength of 440 MPa and a yield strength (YS) of 240MPa, which are changed to 320 and 250 MPa, respectively, after a solution anneal at 530°C (986°F) followed by an aging step at 165°C (329°F) (Ref. 14). Ductility is improved from 4 to 8% by the heat treatment (HT), which also results in a reduction in residual stress (Ref. 14). HT at high temperatures is recommended because it increases plasticity during failure by values up to 2.8 times and greatly improves fatigue performance (Refs. 15, 16). The microstructure of AlSi10Mg as printed consists of eutectic silicon (Si) along the dendrite boundaries in a matrix of supersaturated α -aluminum, but it is typically nonhomogenous due to multiple heating and cooling cycles as well as solidification behavior

(Refs. 17, 18). Micropores and other large discontinuities are often observed in the microstructure (Refs. 16, 19, 20). During HT, the eutectic Si diffuses into discontinuous precipitates of spheroidized primary Si, which has been observed in both stress relief cycles (250–350°C/482–662°F) and solution anneal treatments (530°C/986°F) (Refs. 17, 21). The spheroidized particles grow in size with higher elevated temperatures and during aging cycles (Ref. 15).

In this work, plates printed with PBF-L were joined via FSW in the butt joint configuration. The effects of postweld processing on mechanical performance and weld microstructure were evaluated. Aluminum alloys have limited AM applications because most commercial alloys are prone to solidification defects, have poor as-built quality, and require high laser power to break down the Al oxide layer on the powder (Refs. 11, 22). Alloy Al10SiMg is one of the few Al AM materials because it is a eutectic alloy with a small solidification temperature range, which helps suppress cracking during the PBF-L process (Refs. 23–25). PBF-L AlSi10Mg is strong as built but possesses numerous pores and a reduced fatigue life without postprocessing (Refs. 26–28). FSW has not only shown the ability to join Al alloys that are difficult to fusion weld but also the ability to process material to heal pores, increasing density, refining the microstructure, and improving mechanical performance (Refs. 10, 29–31).

The objective of this study was to determine the feasibility of FSW to join PBF-L AlSi10Mg and examine the effects of postweld HT and hot isostatic pressing (HIP) on overall joint quality and mechanical performance. Welds were examined using optical microscopy, scanning electron microscopy (SEM), and tensile testing.

Experimental Procedure

Material

The material of interest for this study was AlSi10Mg. The nominal composition of the powder used for the PBF-L process is displayed in Table 1. The material has a reported solidus temperature of 557°C (1035°F). The powder particle size was nominally 20 μ m in diameter.

PBF-L Build Approach

The panels for subsequent welding and postprocessing trials were built in an EOS M 290 PBF-L machine with dimensions of 57.2 $\times$ 177.8 $\times$ 7.6 mm (2.25 $\times$ 7 $\times$ 0.3 in.). The laser power remained constant throughout the building process at 700 W with a spot size of 0.1 mm (0.004 in.). The laser tra-

Table 1 — AlSi10Mg Material Composition (wt-%)

Al	Si	Fe	Cu	Mn	Mg	Ni	Zn	Pb	Sn	Ti
Bal.	9.0–11.0%	≤0.55%	≤0.05%	≤0.45%	0.2–0.45%	≤0.05%	≤0.10%	≤0.05%	≤0.05%	≤0.15

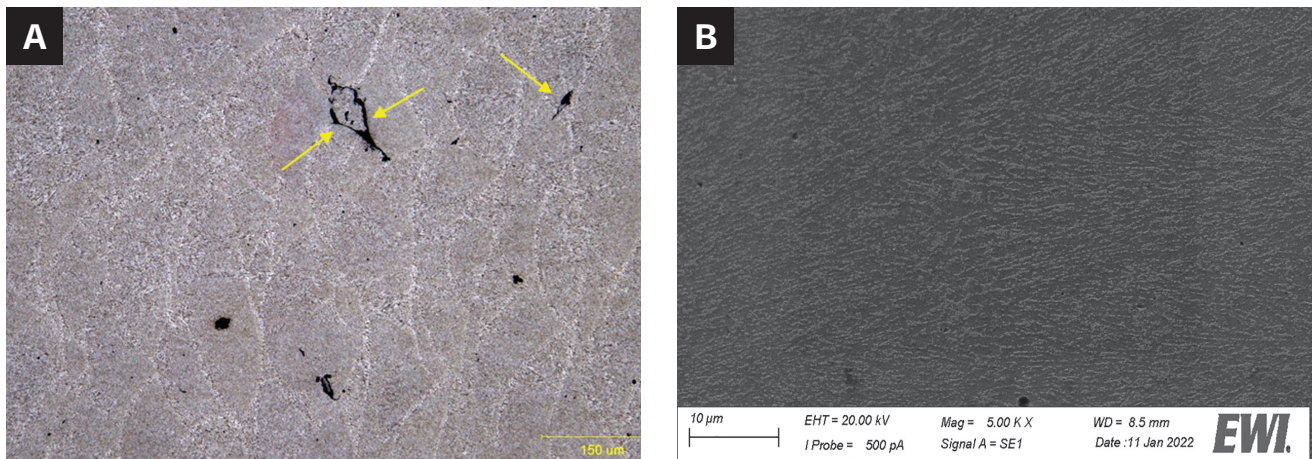


Fig. 2 — A — As-built microstructure (optical). Porosity and unbonded regions are highlighted by yellow arrows. Build direction from left to right; B — as-built microstructure (SEM). Note the nonequiaxed precipitation morphology.

versed at a constant speed of 1.3 m/s (4.27 ft/s) with a scan spacing distance of 0.19 mm (0.007 in.) to build individual layers 30 µm (0.0012 in.) thick up to a final part height of 57.2 mm (2.25 in.). Prior to removal from the build plate, the parts were stress relieved at 288°C (550°F) for 45 min and then air cooled.

FSW Approach

The AlSi10Mg plates were machined to a final dimension of 176.5 × 55.9 × 6.4 mm (6.95 × 2.20 × 0.25 in.). The plates were joined in a butt joint configuration along their long axes. Immediately prior to welding, the joint was cleaned using an Al oxide abrasive pad and wiped with alcohol to remove debris.

All FSW occurred on an AccuStir™ gantry-style machine. The friction stir tool used was made of H13 steel and possessed a truncated, threaded, three flat pin; radiused convex scrolled shoulder; and nominal working depth of 6.1 mm (0.24 in.). Fixturing included toe clamps, an end stop, and side pushers to ensure compression along the abutting faces.

After initial process development, a spindle speed of 600 rpm and a travel speed of 203.4 mm/min (8 in./min) at zero degrees of tilt was selected for welding based on iterative trials and evaluations involving visual inspection and metallography. All welds underwent radiography, and no indications were detected.

Postweld Processing

After FSW, the welds were processed in one of four ways, as summarized in Table 2. The as-welded condition underwent no additional processing. The HTs were performed in air. The first HT was an anneal and quench treatment. The next HT was an anneal and age cycle to avoid the stress of the quench. The last treatment was a HIP cycle. All annealing steps were conducted at 530°C (986°F) for 6 h. The quenched sample was immediately removed from the furnace and placed into a deionized water bath for 10 min. Natural aging occurred in the atmosphere at room temperature. In the aged sample, artificial aging occurred in an atmospheric furnace at 160°C (320°F) for 12 h, and it was air cooled upon removal. The HIP

Table 2 — Postweld Processing Schedules

Postweld Processing	Annealing	Liquid Quench	Natural Age	Artificial Aging
As welded	N/A	N/A	N/A	N/A
Anneal and quench	530°C, 6 h	Water bath	24 h	N/A
Anneal and age	530°C, 6 h	N/A	24 h	160°C, 12 h
HIP		510°C for 3 h at 101.7 MPa		

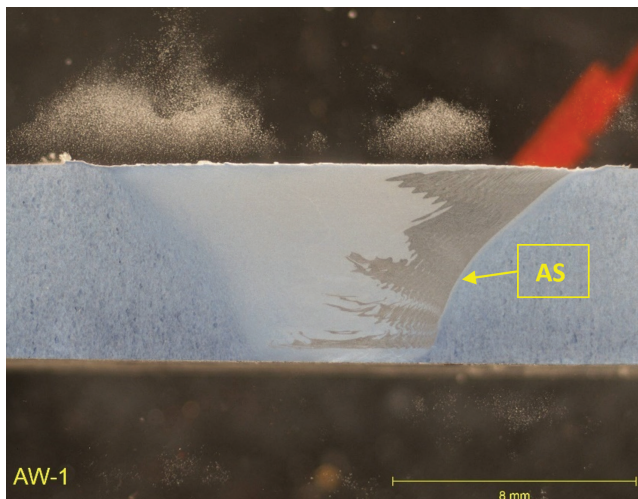


Fig. 3 — As-welded macrograph. The letters AS denote the advancing side.

cycle was done in an Ar environment at 510°C (950°F) for 3 h at 101.7 MPa (14.75 ksi) of pressure.

Microscopy and Mechanical Testing

Postprocessing welds and as-built material were cross sectioned, mounted, and polished using standard metallographic techniques. Samples were etched using Keller's reagent applied via a cotton swab. The cross sections were then subject to optical microscopy, hardness testing, and SEM. Secondary election imaging and energy dispersive x-ray spectroscopy (EDS) were used to evaluate the local composition of the microstructure at select points. Vickers hardness testing was performed near the top and bottom of the weld along a line at 2.98 N (300 g). Three tensile tests were removed from each postprocessed configuration and the as-built material. Dog-bone tensile testing occurred according to ASTM E8, *Standard Test Methods for Tension Testing of Metallic Materials*. The reduced section was nominally 31.75 mm (1.25 in.) in length with a cross section of 6.35 × 6.35 mm (0.25 × 0.25 in.). Ultimate tensile strength (UTS), 0.2% offset YS, and elongation were recorded.

Results

Visual examination of the samples after postprocessing did not reveal any notable changes except for the annealed and quenched sample. This weld displayed a full-thickness crack along the advancing side (AS) as soon as the plate was quenched, as displayed in Fig. 1. The crack was not observed when removed from the furnace, but it immediately formed as the weld was immersed into the water quench. Since the weld was cracked, artificial aging was not performed because the plate had already failed. A cross section was removed near the start of the weld, since the crack did not fully extend into the beginning of the weld, and at the weld center, where the crack was fully formed. No catastrophic failure occurred in any other postweld processing condition, but cracking and pores were observed metallographically in the annealed and aged welds as well.

Microstructural Analysis

The as-built microstructure is displayed in Fig. 2A and B. The PBF-L process resulted in a highly densified structure with small intermittent porosity and unbonded areas noted by the yellow arrows in Fig. 2A. Figure 2B shows a discontinuous network of Si precipitates in the Al matrix. Weld cross sections were examined on a macroscopic and microscopic scale. The as-welded macrostructure is displayed in Fig. 3, which shows an indication-free microstructure within the weld. Based upon optical measurements at the center of the plate thickness, the stir zone (SZ) was 8.4 mm wide, the AS thermomechanically affected zone (TMAZ) was approximately 350 μm, and the retreating side TMAZ was approximately 1 mm. Macroscopic examinations showed consolidation in all postprocessing conditions except for the samples that underwent the annealing HT at 530°C (986°F). The samples that were annealed contained pores and cracking near the edge of the SZ within the AS TMAZ when examined at higher magnification. Metallographic images of the AS SZ/TMAZ for each postprocessing condition are provided in Fig. 4. The as-welded (Fig. 4B) and HIP (Fig. 4F) processed samples showed no flaws along the SZ/TMAZ interface. Figure 4C and D are different locations of the annealed and quenched sample, which both show cracking and defects in the TMAZ.

Table 3 — EDS Point Compositional Results from Various SEM Images (at.-%)

Element	EDS 1	EDS 2	EDS 3	EDS 4
Al	43.6%	20.5%	17.6%	15.7%
Si	56.1%	79.5%	80.7%	83.2%
Mg	0.3%	0%	1.7%	1.1%

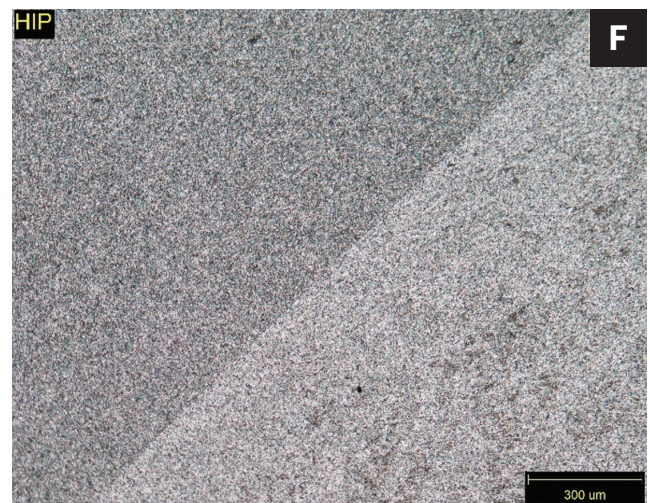
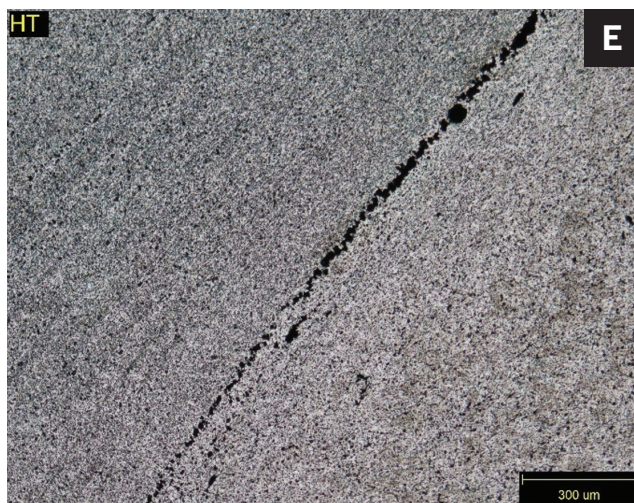
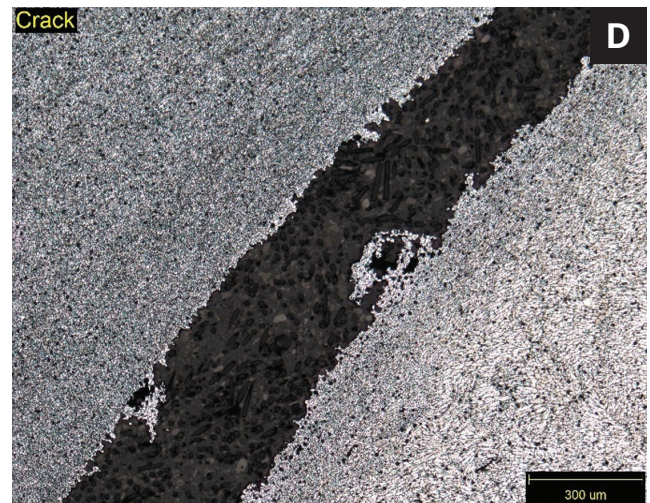
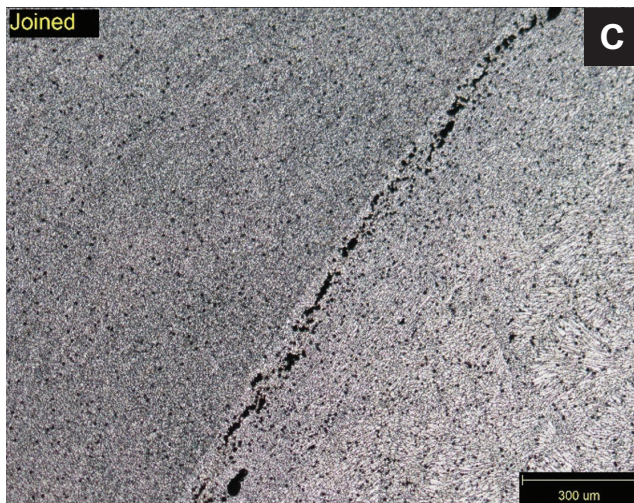
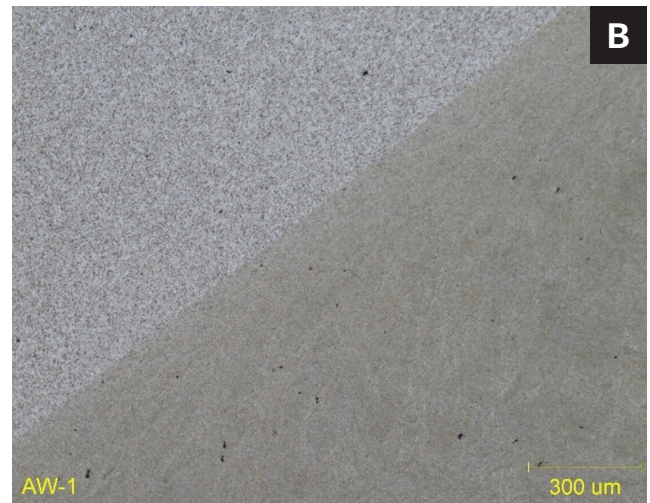
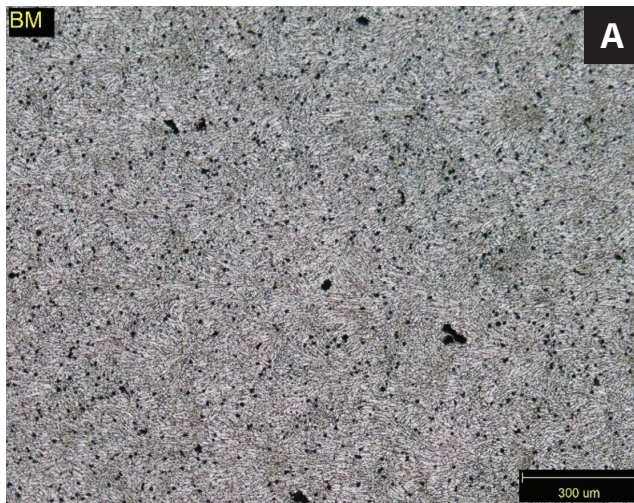


Fig. 4 — SZ/TMAZ interface for all postweld processing conditions and the PBF-L-deposited microstructure. A — As built; B — as welded; C — annealed and quenched near the start of the weld; D — annealed and quenched full crack; E — annealed and aged; F — HIP.

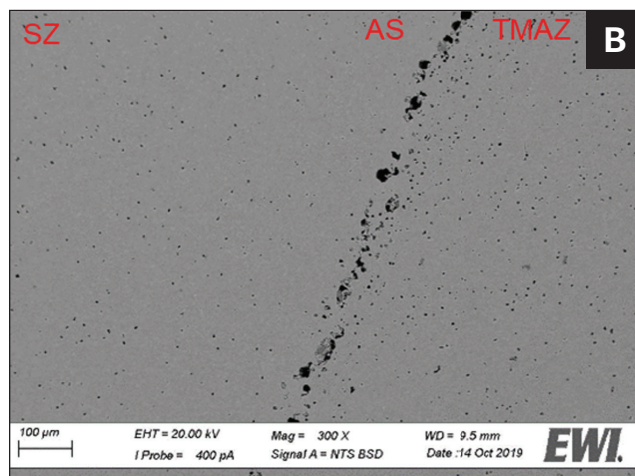
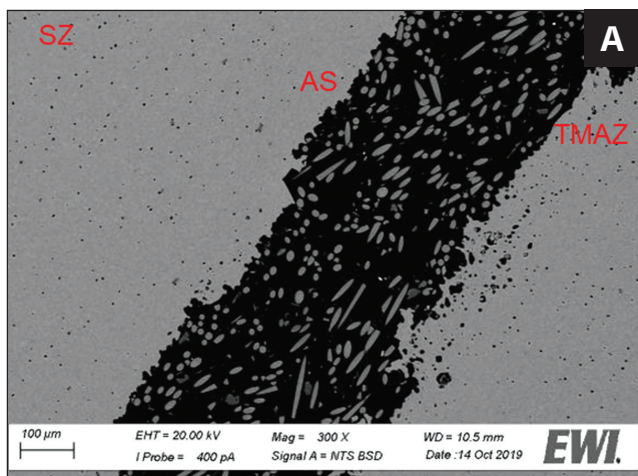


Fig. 5 — High-magnification of the crack (A) and near-crack (B) morphology in the annealed and quenched condition.

The pores observed in Fig. 4C were clearly in the TMAZ and outside of the SZ by approximately 10 μm (0.0004 in.) to 30 μm (0.0012 in.), depending on location. A nearly continuous line of pores was observed along the TMAZ in the annealed and aged (Fig. 4E) condition.

SEM was used to evaluate the pores and cracking in the TMAZ in more detail. The focus of the SEM analysis was the annealed and quenched and the annealed and aged samples due to the cracking and pore formation. High-magnification images of cracked and near-crack regions along the AS SZ/TMAZ are displayed in Fig. 5. The crack appeared to be a tensile failure that propagated from pore to pore. As the pore volume reached a critical point, the remaining bonded material could not support the stress. Cracking was observed as the part was removed from the oven, and the thermal stress of rapid cooling was enough to cause the crack to propagate. A high-magnification view of the pores observed is displayed in Fig. 6. The pore walls were rich in what is believed to be single-phase Si precipitates, an equilibrium constituent in this alloy, based upon the EDS compositional analysis summarized in Table 3. The EDS points of the precipitates were high in Si and very low in magnesium (Mg). The excited volume of the EDS spot likely exceeded the Si precipitate size of 2–3 μm , and, therefore, the EDS composition included parts of the surrounding Al matrix, effectively lowering the measured amount of Si (Ref. 32). The Si precipitate size and distribution are representative of both the as-built and SZ microstructures observed in the samples. Further examination of the pores in the annealed and aged samples is displayed in Figs. 7 and 8. Figure 7 displays the interior of a single pore in the TMAZ. This interior is rich in Si according to the EDS. The interior pore had soft, smooth features, suggesting the pore may have formed as a result of liquation. The pore also appeared to be lined with Si precipitates.

Individual precipitates were identified near pores and are displayed in Fig. 8. The precipitates had a smooth, glossy surface, which is characteristic of liquation. EDS analysis revealed that this layer was primarily composed of Si. Localized liquation was also apparent near the Si precipitates.

SEM of the SZ and base material after the anneal and age HT were also conducted. The resultant micrograph of the SZ is displayed in Fig. 9, and the PBF-L region is shown in Fig. 10. The SZ displayed small pores randomly distributed throughout the weld. However, the pores were adjacent to, or were lined with, Si precipitates. The PBF-L region displayed thin, elongated voids that appeared on grain boundaries, which is characteristic of the AlSi10Mg parts built with PBF-L (Ref. 24).

Mechanical and Hardness Testing

Tensile test results are summarized in Table 4. The as-welded coupon displayed a minor loss in UTS, and the postprocessing greatly decreased strength in all cases. Minimal scatter in the recorded strength data was observed throughout the tensile testing. The failure location did not change in any of the three repetitions of the tensile test for any of the postweld processing conditions. The maximum scatter (as measured by the standard deviation [σ]) occurred in the YS of the annealed and aged condition at 11.89 MPa (0.29 ksi). The maximum scatter resulted in a relative standard deviation (RSD), which is defined as the standard deviation divided by the mean, of 26.27%. The standard deviation values for the annealed and aged condition are relatively low, but due to the low average strength values, the RSD is high.

The remainder of the postprocessing methods remained under 2% RSD for all strength values. Failures for the as-welded, annealed and quenched, and annealed and aged conditions all occurred directly at the transition between the SZ's AS and the TMAZ. The HIP postprocessed sample failed in the TMAZ near the AS but away from the interface of the two regions.

Hardness testing was performed on two HT conditions, as-welded as well as annealed and aged, to determine the effects of welding and the anneal cycle on local strength. The results, which are plotted in Fig. 11, showed about a 20% drop in hardness from the heat-affected zone (HAZ) into the SZ for the as-welded sample. The annealed and aged sample showed little variation in hardness from the HAZ into the

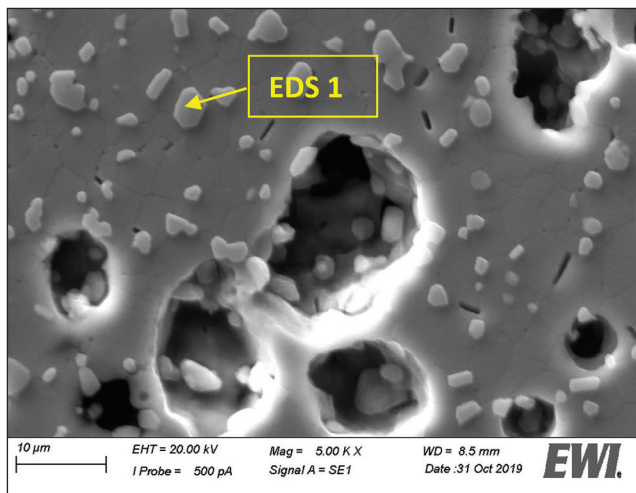


Fig. 6 — Annealed and quenched TMAZ pores with EDS precipitate analysis. Particles rich in Si based on EDS analysis.

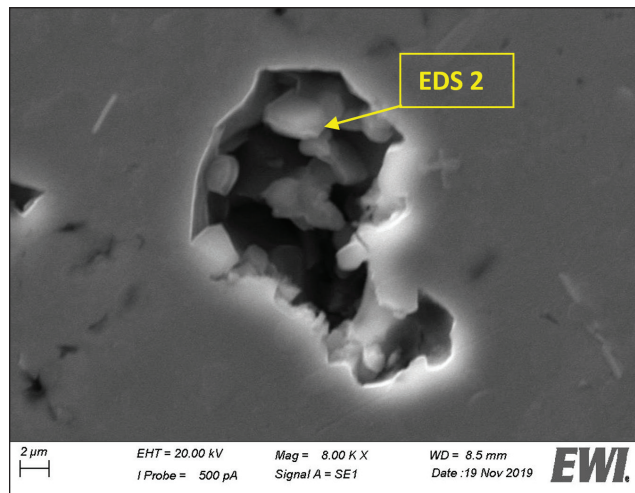


Fig. 7 — Pore on the annealed and aged sample with EDS analysis. Image taken from the AS TMAZ. Interior rich in Si based on EDS with smooth features suggesting liquation.

SZ as expected after an annealing HT. Hardness variations through thickness were not noted on either sample as the top and bottom had very similar hardness profiles. The annealing and subsequent aging also resulted in a notable hardness drop when compared to the as-welded condition in the SZ and HAZ. Additional points were taken near the top of the as-welded sample to show transition in the critical AS SZ/ TMAZ where the crack was found.

Discussion

Cracking Mechanism

The pores and cracking were observed on both samples that were subjected to the postweld anneal at 530°C (986°F). The anneal and quench treatment resulted in a larger crack, but the complete failure was likely a result of the thermal

Table 4 — Average Tensile Test Results with Relative Standard Deviation

Condition	UTS	0.2% Yield Strength	Elongation (%)	Failure Location
As built	294.3 MPa (42.7 ksi) RSD: 1.53%	242.7 MPa (35.2 ksi) RSD: 1.66%	13.40% RSD: 4.00%	Reduced section
As welded	279 MPa (40.5 ksi) RSD: 0.59%	162.3 MPa (23.6 ksi) RSD: 0.58%	8.93% RSD: 2.11%	AS/TMAZ interface
Annealed and quenched	0 MPa (0 ksi) RSD: 0%	0 MPa (0 ksi) RSD: 0%	0.00% RSD: 0%	AS/TMAZ interface
Annealed and aged	59.0 MPa (8.5 ksi) RSD: 10.81%	50.7 MPa (7.3 ksi) RSD: 23.48%	1.53% RSD: 26.27%	AS/TMAZ interface
HIP	151.7 MPa (22.0 ksi) RSD: 0.31%	98.3 MPa (14.2 ksi) RSD: 1.92%	25.56% RSD: 3.94%	TMAZ on AS

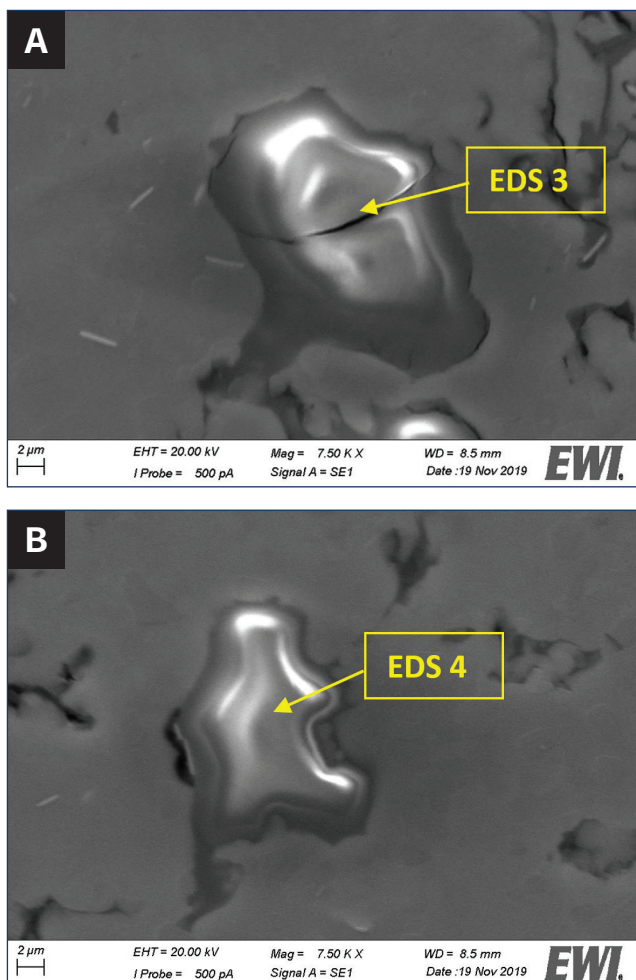


Fig. 8 — Evidence of liquation near precipitates.
*Annealed and aged AS TMAZ. A — Precipitate one;
 B — precipitate two.*

stress during quenching, which overloaded the material between the voids formed during annealing.

The voids and cracking observed appeared to be the result of liquation near single-phase Si precipitates. The glossy surfaces observed around Si precipitates in Fig. 8 were the result of a thin melt layer solidifying upon cooling. However, according to the material supplier, the AlSi10Mg possesses a solidification temperature range of 557°C (1035°F) to 596°C (1105°F), which is well above the annealing temperature used in this study. As the evidence of liquation does not align with the reported material solidification range, a pseudobinary phase diagram was calculated — Fig. 12. Computer coupling of phase diagrams and thermochemistry (CALPHAD) was used to generate a pseudobinary phase diagram for the AlSi10Mg system. Thermo-Calc software was used, and the material was input as a ternary system of Al, Si, and 0.45 wt-% Mg. Al and Si were variable and served as the x-axis of the pseudobinary phase diagram. The pseudobinary phase diagram stated that AlSi10Mg has a solidification range of 521°C (970°F) to 596°C (1105°F). The calculated phase diagram stated that melting can begin to occur at a temperature as low as 521°C (970°F), which is below the annealing temperature of 530°C (986°F).

The lower solidification temperature explains why partial melting occurred at the annealing temperature of 530°C (986°F) as this temperature places the alloy into a region of four-phase equilibrium between single-phase Al, Mg₂Si, liquid, and single-phase Si. According to the phase diagram, the partial melting could occur anywhere in the build.

Partial liquation of the matrix around the Si precipitates at the annealing temperature was a result of constitutional liquation, which is defined as nonequilibrium eutectic melting in a heterogeneous structure (Ref. 33). In this case, the diffusion of Si from the Si precipitate into the Al matrix created a local composition primarily of Al containing roughly 0.4% of Mg with evaluated levels of Si. Once the Al matrix containing Mg reached a Si content greater than 5%, local melting could occur at the annealing temperature. Since the single-phase Si precipitates are the most plentiful source of Si, the melting occurred at the boundary of the Si precipitate and the Al matrix because this location resulted in the shortest diffusion distance to reach the critical local composition. As single-phase Si precipitates were observed throughout the build, it begs the question as to why liquation and cracking were only observed in the AS TMAZ. The answer lies in the thermomechanical history resulting in a unique microstructure and stress state at this location.

First, pores were observed throughout the SZ as well as in the TMAZ for samples annealed at 530°C (986°F). Figure 9 displays pores in the SZ that were smaller in size and appeared almost at random locations throughout the SZ. As temperature is key in pore formation, it is likely that the FSW process was exceeding the eutectic temperature, albeit at much shorter times than the HT. Partial melting has been observed in FSW of other alloys with precipitates that cannot be solutionized during welding (Refs. 10, 34–36). Thus, the peak FSW temperature was likely more than 530°C (986°F), resulting in partial liquation in the SZ and near the SZ. The partially liquated areas were supersaturated in Si and had higher amounts of Mg as well. The partially liquated region had a lower melting temperature upon reheating as a result. The liquation-prone areas likely originated in hot portions of the weld and were distributed by the stirring action of the tool. Additionally, SZs typically possess compressive residual stresses (Refs. 10, 37). Thus, if liquation does occur during HT, the area lacks the tensile stresses required to form voids or cracks.

In comparison to the SZ, the pores in the TMAZ occurred over a very defined region within 30 μm of the SZ. As the SZ reached temperatures of more than 530°C (986°F), a narrow region of the TMAZ also reached these temperatures due to conduction and the AS being the hottest region in a friction stir weld (Ref. 31). These high-temperature regions within the TMAZ did not experience dynamic recrystallization or movement via stirring, which resulted in a narrow region with a unique thermomechanical history as well as precipitation morphology and distribution. In this narrow region, partial liquation can occur during FSW and HT. These liquation sites become enriched in solute and become preferential liquation sites upon reheating. The TMAZ also typically possesses tensile residual stresses due to the lack of full recrystallization and the lower temperatures experienced compared to the SZ. Unlike in the SZ, the liquation-rich regions were aligned in

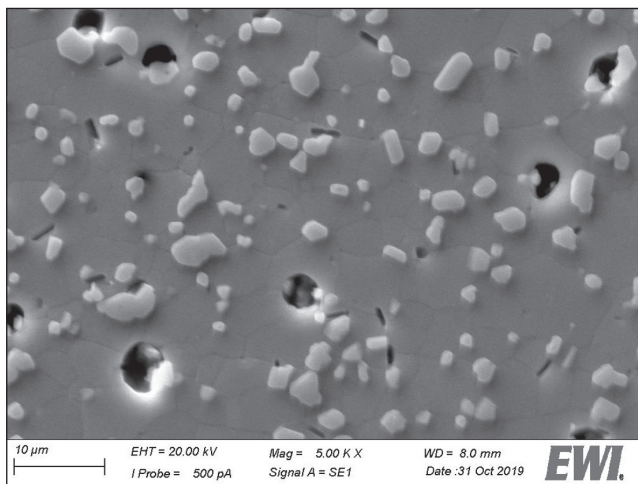


Fig. 9 — SEM micrograph of the SZ after annealing and aging.

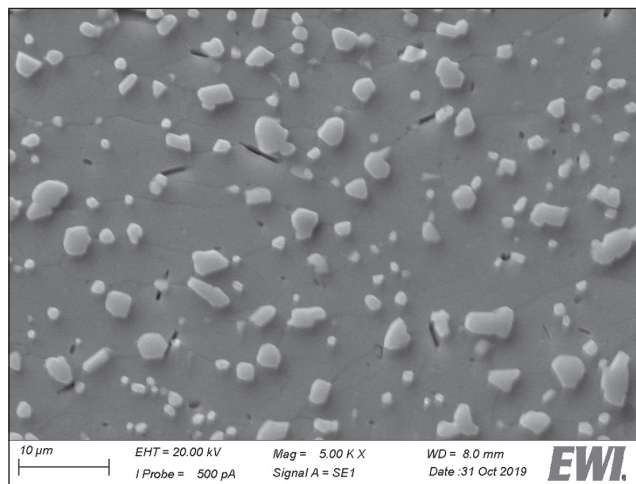


Fig. 10 — SEM micrograph of the PBF-L region after annealing and aging.

a narrow band and were not redistributed by the friction stir tool. A larger volume of pores was observed near the surface of the weld, where the weld temperature and residual stress caused by the thermal gradient was higher, rather than at the root of the weld, where temperature and, therefore, residual stress was lower.

Next, the stress state in the crack region was unique. The AS/TMAZ border had a sharp change in mechanical properties, as shown by the hardness traverse for the as-welded sample in Fig. 11. The SZ was softer and, therefore, exhibited a lower yield strength than the base material and HAZ. The transition from soft to hard occurred rapidly on the AS in the TMAZ. As a result, a metallurgical notch exists on the advancing side TMAZ. As the part was heated and attempted to relieve stress during the HT, the metallurgical notch acted

as a stress riser between the compliant SZ and the stiff PBF-L-built matrix, creating a high-stress region that already contained tensile residual stresses coupled with the highly aligned susceptible microstructure described above, leading to the creation of voids and cracking.

To summarize the cracking mechanism, cracking occurred at the SZ/TMAZ boundary because the alloy was heat treated in a partial melting regime at 530°C (986°F). Liquation occurred in the Al matrix near Si precipitates due to constitutional liquation and the readily available supply of Si provided by the Si precipitates. The TMAZ near the AS SZ had a unique thermomechanical history over a thin region, which contributed to the local microstructure's liquation susceptibility and its tensile residual stresses. The TMAZ was located at a microstructural notch between the soft SZ and hard PBF-L

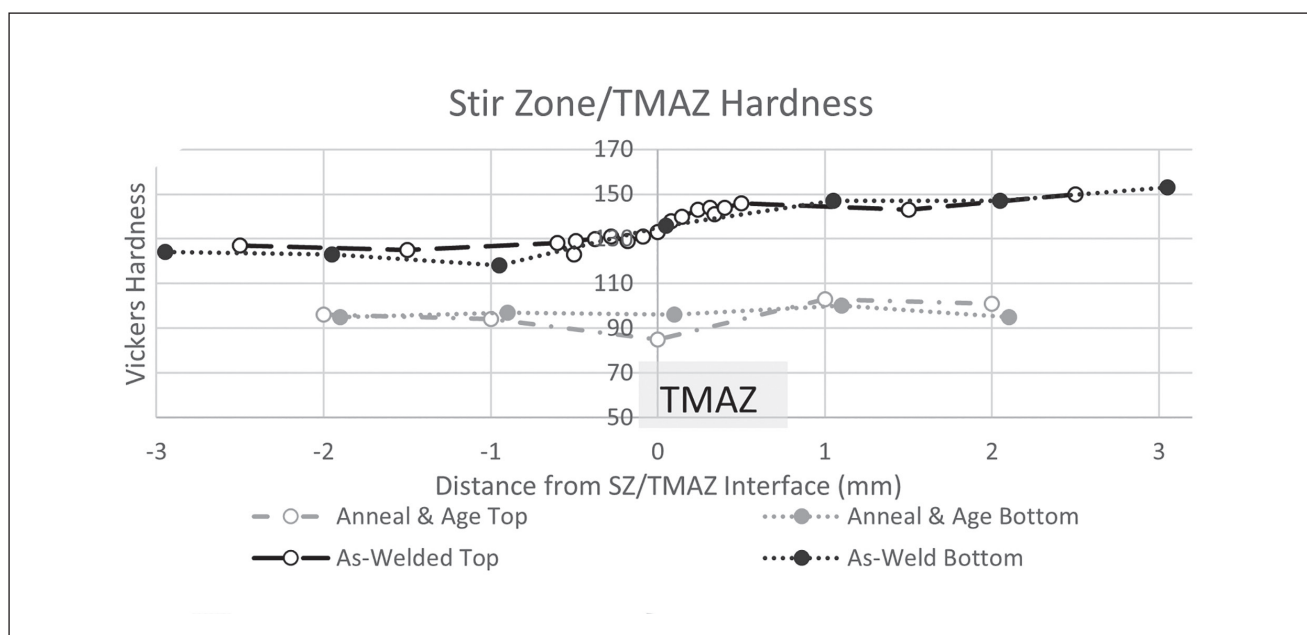


Fig. 11 — Hardness profiles from the SZ into the HAZ.

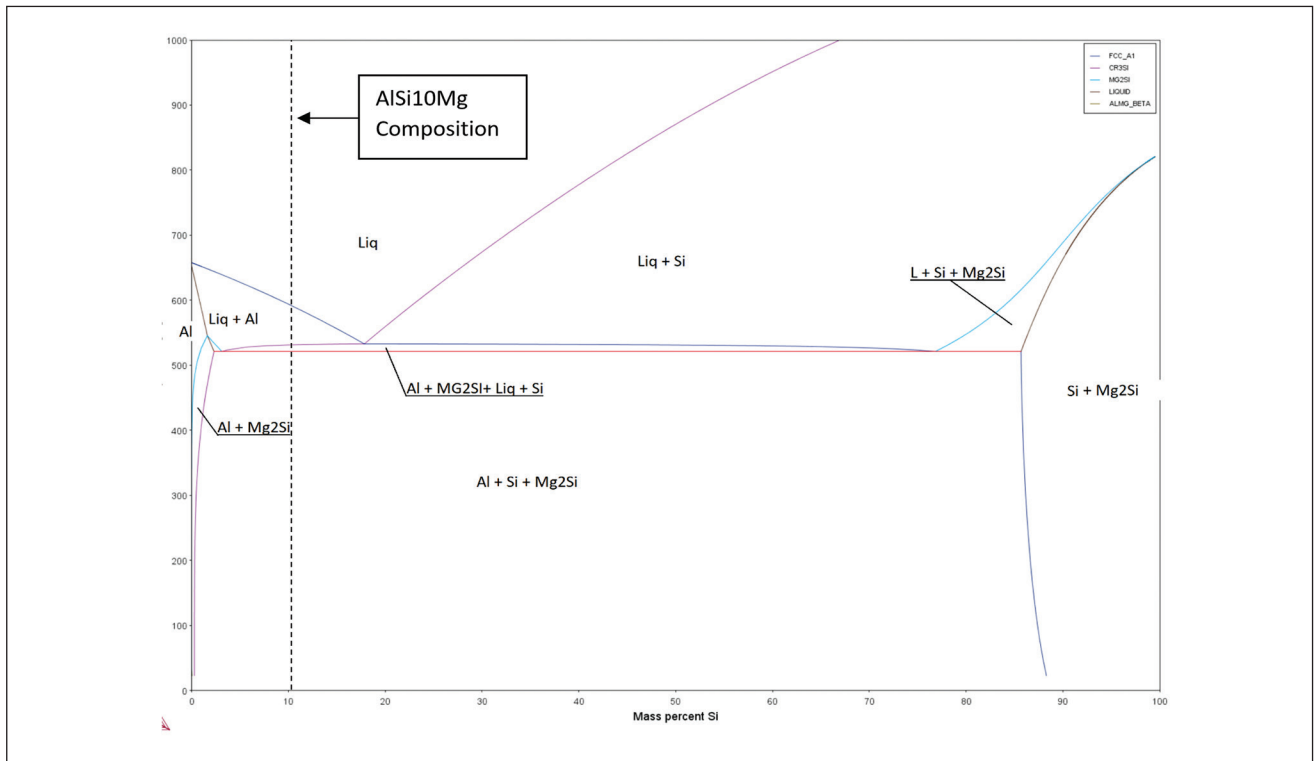


Fig. 12 — Calculated pseudobinary diagram for the AlSi10Mg system.

material. Thus, the TMAZ had a susceptible microstructure at a high-stress region of the weld. All of the above factors combined led to voids and cracks forming along the TMAZ.

The as-welded and HIP samples did not show any signs of cracking or pore formation in the TMAZ. The HIP was performed at a slightly lower temperature of 510°C (950°F) and was below the solidus temperature of 521°C (970°F) according to the calculated phase diagram. As the temperature was below the solidus, no liquation near precipitates occurred and the sample emerged from postprocessing without cracks or voids. The as-welded coupon also was void and crack free and performed best in tensile testing of the postprocessing conditions.

Effects of Postprocessing on Strength

The as-welded coupon displayed a minor loss in UTS as compared to the base metal, and postprocessing greatly decreased weld strength in all cases. The strength knockdowns can be explained by Si morphology, distribution, and size as AlSi10Mg is a precipitation-strengthened alloy. In all cases, the failures occurred near the AS TMAZ.

The as-welded microstructure contained a fine distribution of spheroidized Si precipitates and a refined equiaxed grain size. This microstructure was weaker than the as-deposited dendritic microstructure with eutectic Si (Ref. 21). Because the SZ is the weakest part of the weld, as verified by hardness testing, the failure will occur within this region. The failure occurred next to the TMAZ as the strength rapidly changed in this region, creating a metallurgical notch. The metallurgical notch acted as the initiation point for failure.

The generation of porosity and cracks along the SZ/TMAZ contributed to the reduction in strength in the annealed and quenched as well as the annealed and aged conditions. Based upon the hardness of the annealed and aged condition, even without cracks, an approximate 20% reduction in strength compared to the as-welded condition would be expected. The drop in hardness can be explained by additional coarsening on the Si precipitates during the HT as single-phase Si was stable during the annealing HT.

The HIP process greatly reduced the strength by 50% compared to the PBF-L as-built material. The HIP cycle resulted in large, near-spheroidized Si precipitate size and thus the knockdown in strength. Subsequent aging cycles may allow for minor strength gains after HIP. As it stands, the HIP greatly reduced the tensile performance of the as-welded AlSi10Mg. Ideally, AlSi10Mg should undergo a solution anneal, quench, and subsequent aging to provide the best performance while reducing residual stress in the build. However, Si precipitates cannot be dissolved in the solid state, reducing the effectiveness of the annealing step. A shorter anneal cycle at a slightly lower temper may be required to limit coarsening of Si while still providing the benefits of stress relief and solutionizing of Mg₂Si precipitates.

Conclusions

AlSi10Mg sheets were built using PBF-L and subsequently welded together using FSW. The as-welded coupons were subjected to varying HTs and postweld HIP of the weldment. The various welded and postprocessed samples were analyzed, cracking was noticed in all samples subject to an

annealing cycle at 530°C (986°F), and no cracking was noted in the HIP sample or the as-welded coupon. Strength knockdowns were noted but could be improved upon with additional development of HT procedures. The following conclusions can be drawn from this work:

1) FSW is feasible on PBF-L Al components. The as-welded sample exhibited 80% of the UTS of the as-PBF-L-printed coupon.

2) Cracking occurred at the SZ/TMAZ boundary during the annealing HT because the alloy is in a partial melting regime at 530°C (986°F) as Mg is a large eutectic point suppressant when added to the Al-Si system. Liquation occurs in the Al matrix near Si precipitates due to constitutional liquation and the readily available supply of Si present in the single-phase, single Si precipitates. The TMAZ near the AS SZ has a unique thermomechanical history, which may contribute to the local microstructure's liquation susceptibility. All of the above factors lead to voids and cracks forming along the TMAZ.

3) HIP of the as-welded component resulted in a defect-free weldment and normalized the properties between the SZ and as-PBF-L-printed microstructures. The HIP process is not optimized for strength and resulted in a 50% knockdown in UTS compared to the as-PBF-L-printed condition.

4) Based upon the CALPHAD modeling, HT development below the critical temperature of 521°C (970°F) is a viable solution to prevent cracking during postweld HT of welded PBF-L-printed AlSi10Mg.

References

1. Thomas, D., and Gilbert, S. 2015. Costs and cost effectiveness of additive manufacturing: A literature review and discussion. National Institute of Standards Technology. DOI: 10.6028/NIST.SP.1176
2. Frazier, W. E. 2014. Metal additive manufacturing: A review. *Journal of Materials Engineering and Performance* 23(6): 1917–1928. DOI: 10.1007/s11665-014-0958-z
3. Karunakaran, K. P., Suryakumar, S., Pushpa, V., and Akula, S. 2010. Low cost integration of additive and subtractive processes for hybrid layered manufacturing. *Robotics and Computer-Integrated Manufacturing* 26(5): 490–499. DOI: 10.1016/j.rcim.2010.03.008
4. Fieger, T., Sattler, M., and Witt, G. 2018. Developing laser beam welding parameters for the assembly of steel SLM parts for the automotive industry. *Rapid Prototyping Journal* 24(8): 1288–1295. DOI: 10.1108/RPJ-12-2016-0204
5. Huber, F., Mann, V., Kaufmann, F., and Schmidt, M. 2019. Laser beam welding of Ti-6Al-4V hybrid-parts from additively manufactured elements and sheet metal. *Proceedings of the Lasers in Manufacturing Conference* 2019. Hannover, Germany: German Scientific Laser Society (WLT).
6. Wits, W. W., and Becker, J. M. J. 2015. Laser beam welding of titanium additive manufactured parts. *Procedia CIRP*. 28: 70–75. DOI: 10.1016/j.procir.2015.04.013
7. Matilainen, V.-P., Pekkarinen, J., and Salminen, A. 2016. Weldability of additive manufactured stainless steel. *Physics Procedia* 83: 808–817. DOI: 10.1016/j.phpro.2016.08.083
8. El Hassanin, A., Velotti, C., Scherillo, F., Astarita, A., Squillace, A., and Carrino, L. 2017. Study of the solid state joining of additive manufactured components. *2017 IEEE 3rd International Forum on Research and Technologies for Society and Industry (RTSI)*, pp. 1–4, New York, N.Y.: IEEE. DOI: 10.1109/RTSI.2017.8065967
9. Prashanth, K. G., Damodaram, R., Scudino, S., Wang, Z., Prasad Rao, K., and Eckert, J. 2014. Friction welding of Al-12Si parts produced by selective laser melting. *Materials and Design* 57: 632–637. DOI: 10.1016/j.matdes.2014.01.026
10. Mishra, R. S., De, S. P., and Kumar, N. 2014. *Friction Stir Welding and Processing*. London, UK: Springer.
11. Ericsson, M., and Sandström, R. 2003. Influence of welding speed on the fatigue of friction stir welds, and comparison with MIG and TIG. *International Journal of Fatigue* 25(12): 1379–1387. DOI: 10.1016/S0142-1123(03)00059-8
12. Fahimpour, V., Sadrnezhad, S. K., and Karimzadeh, F. 2012. Corrosion behavior of aluminum 6061 alloy joined by friction stir welding and gas tungsten arc welding methods. *Materials & Design* 39: 329–333. DOI: 10.1016/j.matdes.2012.02.043
13. Chowdhury, S. M., Chen, D., Bhole, S. D., Cao, X., Powidajko, E., Weckman, D. C., and Zhou, Y. 2010. Tensile properties and strain-hardening behavior of double-sided arc welded and friction stir welded AZ31b magnesium alloy. *Materials Science and Engineering A* 527(12): 2951–2961. DOI: 10.1016/j.msea.2010.01.031
14. EOS. EOS aluminium AlSi10Mg material data sheet. Retrieved from eos.info/03_system-related-assets/material-related-contents/metal-materials-and-examples/metal-material-datasheet/aluminium/material_datasheet_eos_aluminium-alsi10mg_en_web.pdf.
15. Jiang, X., Xiong, W., Wang, L., Guo, M., and Ding, Z. Heat treatment effects on microstructure-residual stress for selective laser melting AlSi10Mg. *Materials Science and Technology* 36(2): 168–180. DOI: 10.1080/02670836.2019.1685770
16. Bagherifard, S., Beretta, N., Monti, S., Riccio, M., Bandini, M., and Guagliano, M. 2018. On the fatigue strength enhancement of additive manufactured AlSi10Mg parts by mechanical and thermal post-processing. *Materials & Design* 145: 28–41. DOI: 10.1016/j.matdes.2018.02.055
17. Alghamdi, F., and Haghshenas, M. 2019. Microstructural and small-scale characterization of additive manufactured AlSi10Mg alloy. *Sn Applied Sciences* 1(3): 255. DOI: 10.1007/s42452-019-0270-5
18. Zhang, C., Zhu, H., Liao, H., Cheng, Y., Hu, Z., and Zeng, X. 2018. Effect of heat treatments on fatigue property of selective laser melting AlSi10Mg. *International Journal of Fatigue* 116: 513–522. DOI: 10.1016/j.ijfatigue.2018.07.016
19. Wu, J. J., Wang, L. Z., and An, X. G. 2017. Numerical analysis of residual stress evolution of AlSi10Mg manufactured by selective laser melting. *Optik* 137: 65–78. DOI: 10.1016/j.ijleo.2017.02.060
20. Ngnekou, J. N. D., Nadot, Y., Henaff, G., Nicolai, J., Kan, W. H., Cairney, J. M., and Ridosz, L. 2019. Fatigue properties of AlSi10Mg produced by additive layer manufacturing. *International Journal of Fatigue* 119: 160–172. DOI: 10.1016/j.ijfatigue.2018.09.029
21. Yu, X., and Wang, L. 2018. T6 heat-treated AlSi10Mg alloys additive-manufactured by selective laser melting. *Procedia Manufacturing* 15: 1701–1707. DOI: 10.1016/j.promfg.2018.07.265
22. Louvis, E., Fox, P., and Sutcliffe, C. J. 2011. Selective laser melting of aluminium components. *Journal of Materials Processing Technology* 211(2): 275–284. DOI: 10.1016/j.jmatprotec.2010.09.019
23. Buchbinder, D., Schleifenbaum, H., Heidrich, S., Meiners, W., and Bultmann, J. 2011. High power selective laser melting (HP SLM) of aluminum parts. *Physics Procedia* 12A: 271–278. DOI: 10.1016/j.phpro.2011.03.035
24. Yang, P., Deibler, L. A., Bradley, D. R., Stefan, D. K., and Carroll, J. D. 2018. Microstructure evolution and thermal properties of an additively manufactured, solution treatable AlSi10Mg part. *Journal of Materials Research* 33(23): 4040–4052. DOI: 10.1557/jmr.2018.405
25. Wu, Y., Liao, H., Liu, Y., and Zhou, K. 2014. Dynamic precipitation of Mg₂Si induced by temperature and strain during hot extrusion and its impact on microstructure and mechanical properties of near

eutectic Al–Si–Mg–V alloy. *Materials Science and Engineering A* 614: 162–170. DOI: 10.1016/j.msea.2014.07.023

26. Schneller, W., Leitner, M., Springer, S., Grün, F., and Taschauer, M. 2019. Effect of HIP treatment on microstructure and fatigue strength of selectively laser melted AlSi10Mg. *Journal of Manufacturing and Materials Processing* 3(1): 16. DOI: 10.3390/jmmp3010016

27. Vrancken, B., Kempen, K., Thijs, L., Kruth, J.-P., and Van Humbeeck, J. 2014. Adapted heat treatment of selective laser melted materials. *Proceedings of the Euro PM 2014 Congress*, pp. 1–6. Shrewsbury, England: European Powder Metallurgy Association (EPMA).

28. Rosenthal, I., Tiferet, E., Ganor, M., and Stern, A. 2015. Post-processing of AM-SLM AlSi10Mg specimens: Mechanical properties and fracture behaviour. *The Annals of “Dunarea de Jos” University of Galati. Fascicle XII, Welding Equipment and Technology* 26: 33–38.

29. Brown, R., Tang, W., and Reynolds, A. P. 2009. Multi-pass friction stir welding in alloy 7050-T7451: Effects on weld response variables and on weld properties. *Materials Science and Engineering A* 513–14(1): 115–121. DOI: 10.1016/j.msea.2009.01.041

30. Dolatkhan, A., Golbabaie, P., Givi, M. K. B., and Molaiekiya, F. 2012. Investigating effects of process parameters on microstructural and mechanical properties of Al5052/SiC metal matrix composite fabricated via friction stir processing. *Materials & Design* 37: 458–464. DOI: 10.1016/j.matdes.2011.09.035

31. Ren, J. G., Wang, L., Xu, D. K., Xie, L. Y., and Zhang, Z. C. 2017. Analysis and modeling of friction stir processing-based crack repair-

ing in 2024 aluminum alloy. *Acta Metallurgica Sinica (English Letters)* 30(3): 228–237. DOI: 10.1007/s40195-016-0489-8

32. Williams, D. B., and Carter, B. C. 2009. *Transmission Electron Microscopy*, second ed. New York, N.Y.: Springer Science.

33. Pepe, J. J., and Savage, W. F. 1967. Effects of constitutional liquation in 18-Ni maraging steel weldments. *Welding Journal* 46(9): 411-s to 422-s.

34. Gerlich, A., Yamamoto, M., and North, T. H. 2007. Local melting and cracking in Al 7075-T6 and Al 2024-T3 friction stir spot welds. *Science and Technology of Welding and Joining* 12(6): 472–480. DOI: 10.1179/174329307X213873

35. Gerlich, A. P., and Shibayanagi, T. 2011. Liquid film formation and cracking during friction stir welding. *Science and Technology of Welding and Joining* 16(4): 295–299. DOI: 10.1179/1362171811Y.0000000005

36. Cao, G., and Kou, S. 2005. Friction stir welding of 2219 aluminum: Behavior of θ (Al₂Cu) particles. *Welding Journal* 84(1): 1-s to 8-s.

37. Chao, Y. J., Qi, X., and Tang, W. 2003. Heat transfer in friction stir welding — Experimental and numerical studies. *Journal of Manufacturing Science and Engineering* 125(1): 138–145. DOI: 10.1115/1.1537741

MICHAEL EFF (meff@ewi.org) and **DREW SHIPLEY** are with EWI, Columbus, Ohio; **HARVEY HACK** is with Northrop Grumman Corp., Falls Church, Va.; and **SETH SHIRA** is with Path Robotics, Columbus, Ohio.

Guidelines for Submitting a Research Paper

Peer review of research papers is now managed through an online system using Editorial Manager software. Papers can be submitted into the system directly from the *Welding Journal* page on the American Welding Society (AWS) website (aws.org) by clicking on “submit papers.” You can also access the new site directly at editorialmanager.com/wj. Follow the instructions to register or log in. This online system streamlines the review process and makes it easier to submit papers and track their progress. By publishing in the *Welding Journal*, more than 60,000 members will receive the results of your research.

Additionally, your full paper is posted on the AWS website for FREE access around the globe. There are no page charges, and articles are published in full color. By far, the most people, at the least cost, will recognize your research when you publish in the world-respected *Welding Journal*. The authors' names, along with the companies they work for and their positions, should be listed.

If you'd like to discuss a particular idea or email a submission for evaluation, please contact Managing Editor Kristin Campbell at kcampbell@aws.org.

Correction

The Research Supplement titled The Toughness of High-Strength Steel Weld Metals starting on page 67-s of the February 2022 issue was missing Tables 4–6.

The online version at <https://s3.amazonaws.com/WJ-www.aws.org/supplement/2022.101.006.pdf> has been corrected. The *Welding Journal* regrets this error.