Influence of texture and grain refinement on the mechanical behavior of AA2219 fabricated by high shear solid state material deposition


ABSTRACT

Issues with rapid grain growth, hot cracking and poor ductility have hindered the additive manufacturing and repair of aluminum alloys. Therefore, this is the first investigation to spatially correlate the processing-structure-property relations of a precipitation hardened aluminum alloy 2219 (AA2219) material with respect to deposition orientations and build layers. The AA2219 material was processed by a high deposition rate (1000 cm³/h) solid-state additive deposition process known as Additive Friction Stir Deposition or MELD. An equiaxed grain morphology was observed in the three orientations, where Electron Backscatter Diffraction (EBSD) identified a layer-dependent texture with a strong torsional fiber A texture in the top of the build transitioning to weaker textures in the middle and bottom layers. Interestingly, the tensile behavior reflected the texture layer-dependence with tensile strength increasing from the bottom to the top of the deposition. However, there were no statistically significant differences in hardness measured from the top to the bottom of the deposition. Furthermore, no orientation dependence on mechanical properties was observed for compression and tension specimens tested at quasi-static (0.001/s) and high (1500/s) strain rate. Transmission Electron Microscopy (TEM) determined a lack of θ′ precipitates in the as-deposited cross-section, therefore resulting in no precipitation strengthening.

1. Introduction

The solid state material deposition process known as MELD or Additive Friction Stir Deposition (AFS-Deposition) is a transformative process for solid-state additive manufacturing (AM), repair, joining, or adding secondary features, which is being reported for the first time on deposited aluminum alloys, specifically Aluminum Alloy (AA) 2219. This solid state deposition process shares some similarities to other high shear and elevated temperature solid-state processes such as shear extrusion or friction stir welding and processing (FSW/P), but is significantly different in the aspect that feedstock material flows through a rotating hollow tool with the feedstock material being deposited onto a substrate while remaining completely in the solid state. As described in previous research by Rivera et al. [1] on MELD processed IN625, the solid state material deposition can either feed from solid or metal powder material, pushing it through a non-consumable rotating cylindrical tool. This will generate heat and plastically deform the feedstock material under controlled pressure from the tool while layers are built upon a substrate. Once a layer is added, the tool height increases to begin the deposition for the next layer, creating a strong metallurgical bond between layers. Some advantages of this process are grain refinement, homogenization, and reduction of porosity. During the process, temperatures are similar to those in the stir zone (SZ) of FSW, which are estimated to be between 0.6 and 0.9 Tm, where Tm is the melting point of the material [2]. Being a highly scalable process, with deposition rates for aluminum alloys over 1000 cm³/h, this process allows for repairing, coating, and/or building fully-dense materials.
AA2219 electron beam freeform (EBF) AM has been reported in the open literature [10–12], in which they observed that, by varying the process parameters, the microstructure varied from fine equiaxed grains to dendritic. In addition, the researchers reported tensile mechanical properties of the as-deposited AA2219 using EBF had a yield strength (YS), ultimate tensile strength (UTS), and elongation of 100 MPa, 275 MPa, and 18%, respectively [10–12].

FSW of AA2219, which is based on a similar solid-state concept, provides a starting point for understanding the microstructure evolution that occurs during solid-state processing. FSW has been a useful option for welding similar and dissimilar metals that were originally presented in 1991 by TWI [13], and with prior FSW research on AA2219 examining the effectiveness, optimal parameters, and mechanical properties [3,8,14,15]. FSW is popular in multiple industries, such as aerospace and automotive, due to higher effective weld properties when compared to other welding techniques due to better retention of base material mechanical properties, less distortion and less weld defects [16,17]. Additive FSW has even been investigated by a number of researchers where plates of stacked material are joined together by a pin-tool penetrating into the successively stacked plates [18–20]. However, MELD depositions significantly differ from additive FSW where now material is extruded through a hollow rotating tool. The microstructure of FSW components has also been a point of interest among researchers with specific emphasis on the stir zone (SZ), also known as the nugget [3,21]. This is the area where the grains recrystallize, and grain refinement is observed [3,8,21]. The mechanical properties in this region are desirable, and for that reason a processing technique called FSP was also developed [22–25]. The FSP technique has also been a topic of interest experiencing significant research [22,24–26]. In both FSW and FSP as well as MELD, the microstructure may experience dynamic recrystallization (DRX) [23–25]. There are two types of DRX in the literature, continuous DRX and discontinuous DRX. The difference between them is that, in continuous DRX, new grains are formed by a gradual increase in the misorientation between the subgrains [24,25]. Under discontinuous DRX, grains with high-angle grain boundaries form through dynamic nucleation and growth from a previously deformed microstructure. Previous research has demonstrated that it is common for aluminum alloys to undergo dynamic recovery (DRV) during hot deformation [24,25]. Additionally, research by Su et al. [21–26] on AA7075 has shown that the FSW and FSP will experience different mechanisms such as discontinuous DRX, DRV, and continuous DRX at different stages of the microstructural evolution [21,24,25].

Due to a variety of end uses for AA2219, the mechanical properties of multiple tempers were investigated in previous FSW research [3,8,14]. Tensile and microhardness testing on FSW butt welds of thick AA2219-

### Table 1

<table>
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<tr>
<th>Elements</th>
<th>Al</th>
<th>Cu</th>
<th>Mn</th>
<th>Ti</th>
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**Fig. 1.** (A) Schematic of the solid state material deposition (MELD) process with solid feedstock material extruded through a hollow tool. (B) Representative figure of the as-deposited AA2219 identifying the longitudinal, transverse, and build directions.

**Fig. 2.** Locations of the Vickers microhardness testing along the as-deposited AA2219.

**Fig. 3.** Schematic of the microindentation grid used to analyze the transverse cross-section of the as-deposited AA2219 in the three different longitudinal locations.

Unique features of the AA2219 is that the material is a precipitation strengthened alloy with a variety of beneficial properties such as high strength-to-weight ratio, suitable weldability, resistance to stress corrosion cracking, and superior properties at cryogenic temperatures [3,4]. The mechanical properties of AA2219 spread through a wide temperature range from −250 to +250 °C [5], that in addition to the aforementioned benefits makes this alloy attractive for aerospace applications. AA2219 has been successfully joined by different conventional joining processes, such as gas metal arc welding [6], gas tungsten arc welding [6], plasma arc welding [7], and FSW [3,4,6,8,9].

Little AM research on AA2219 has been reported to date. However,
O and –T87 plates have been investigated previously [3,6] with microhardness testing showing a maximum value of 95 HV [3]. Additional research on FSW butt welds of AA2219-T6 thick plates with standard tensile specimens reporting tensile YS, UTS and elongation values of 345 MPa, 410 MPa, and 15%, respectively [8]. For comparison purposes, and due to range of tempers studied in literature, AA2219-O will be used as a baseline. Additionally, since MELD is a solid state deposition process with similarities to FSW and FSP, the prior FSW research discussed subsequently aides in elucidating mechanisms occurring during this solid state deposition of AA2219. Specifically at the microstructural level, prior FSW AA2219 research has shown a refined equiaxed microstructure in the weld zone is attainable, which improves strength and hardness [3,8,9,27]. Additionally, the aforementioned investigations observed coarsening of the \( \text{Al}_2\text{Cu} \), or \( \theta \), particles, where the \( \theta \) particles are present in the base material but smaller in size than in the weld zone. Li and Shen [27] attributed growth of the \( \theta \) particles from a combined action of three different formation mechanisms, i.e., Aggregation Mechanism and Diffusion Mechanism I and II. Furthermore, research by Cao and Kou [14] concluded that there was no evidence of liquation in FSW AA2219, since the welds only contain \( \theta \) particles and no eutectic particles. Subsequently, there is a wide range of reported sizes of the \( \theta \) particles dependent on the FSW processing parameters, with sizes reaching up to 150 \( \mu \text{m} \) [4,8,14,27]. Previous research by Kang et al. [28] performed Transmission Electron Microscopy (TEM) in FSW of AA2219-T8, where they found \( \theta' \) precipitates in the base material and only \( \theta \) phase precipitates in the SZ.

This is the first manuscript to report the microstructure and resultant mechanical properties of AA2219 produced by the MELD solid state deposition process. Electron Backscatter Diffraction (EBSD) and TEM are used to quantitatively characterize the microstructure of as-deposited AA2219. Mechanical properties in tension and compression in quasi-static and high rates are reported.

2. Materials and methods

Aeroprobe Corporation, who created and patented this technology, provided MELD fabricated samples by pushing a solid AA2219-T851 bar through a hollow rotating tool to deposit the material onto an AA2219-T851 plate substrate. During the MELD process, the solid feedstock material was added, and heat generated by friction between the feedstock material and the tool shoulder under hydrostatic pressure and the substrate plastically deformed both feedstock and substrate as they were stirred together to metallurgically bond the material to the substrate and the successive layers of AA219 as depicted in Fig. 1. The average chemical composition of the rod-stock material is shown in Table 1. In this study, 100 mm long MELD AA2219 depositions consisted of 6 layers that were approximately 1 mm in thickness per layer.
From this point on, the build direction will be referred to as BD, the transverse direction as TD, and the longitudinal direction as LD.

A wire EDM milled flat dogbone sub-compact tensile specimens from the AA2219 depositions that has previously been used for both high strain rate (HR) and quasi-static (QS) tensile experiments on cast, wrought, and AM alloys [30–32]. The flat dogbones had a gauge length of 4.5 mm, width of 2.0 mm and thickness of 1.5 mm [1]. The tensile specimens were machined from the bottom, middle, and top of the build in the longitudinal direction (LD) and transverse direction (TD). An EDM also machined the compression specimens with a diameter of 6 mm and a height of 6 mm. The compression specimens were machined from the build direction (BD), TD and LD orientations to examine directional dependence of mechanical properties in the specimens (see Fig. 9). As a first investigation, cross-sections were taken from the beginning, middle, and end of the as-deposited material (see Fig. 10) to quantify spatial dependence of the initial microstructure and Vickers microhardness (see Fig. 2) in the deposition. For microstructural characterization, the three cross-sections were stepwise polished using an aqueous lubricant down to a 1200 grit SiC paper, then polished using 1- and 0.3-µm diamond suspension and a final polish of 0.05 µm with a colloidal silica suspension.

Vickers microindentation (Fig. 3) analyzed the hardness through the cross-section of the as-deposited material in the 3 different locations. A load of 0.1 kgf held for 10 s was used for the hardness tests, as described in ASTM E384-16 [33]. An indentation grid (Fig. 11) consisting of 10 rows with spacing between the rows of 1 mm and 9 columns spaced at 2.5 mm, which complied with ASTM E384-16 [33] probed the as-deposited cross-section. Minimum and maximums values represent the error bars, and are taken into consideration when determining statistical differences in the data [34,35].

A Tescan Lyra FIB-FESEM equipped with an EDAX Hikari Super EBSD camera and an Octane Elite Silicon Drift Detector was used for EBSD, electron dispersive X-ray spectroscopy (EDX), and fractography. EBSD scans were run at the same magnification along the build direction over an area of approximately 50 × 50 µm, with a step size of 0.1 µm. The scans were performed at 20 kV and a beam current of 5.5 nA. Line scans were performed from top to bottom of the as-deposited cross-section to spatially quantify grain size. In addition, images
were taken in the longitudinal, build direction, and long transverse directions to elucidate the 3D microstructure of the deposition. EBSD data were post-processed using the Neighbor Pattern Averaging and Reindexing (NPAR) option in the EDAX TEAM software, followed by the grain dilation algorithm with setting of a minimum grain size of 2 pixels, and a grain must contain multiple rows of pixels. For grain reference orientation deviation (GROD) maps, any pixel with an pattern indexing confidence index of < 0.2 was removed from the map. A transmission electron microscope (TEM) FEI TECNAI F-20 was used in scanning mode (STEM) at 200 kV to study the size, geometry, and type of the precipitates. In particular, STEM experiments used a high-angle-annular dark field (HAADF) detector, imaging in STEM mode, a spot size of 6, camera length of 100 mm, C2 aperture 2, and a dwell time of 16 μs. The specimens for the TEM were prepared using the lift-out method with a Focused Ion Beam (FIB) FEI Quanta 200 3D Dual Beam microscope. TEM specimens were lifted out of the feedstock material for comparison purposes and from three locations (top, middle and bottom) in the as-deposited cross-section.

An electromechanical Instron 5185 load-frame with a 50 kN load cell performed the ambient temperature QS tensile and compression experiments at a strain rate of 0.001/s. The HR tensile and compression experiments were conducted on a Kolsky tension/compression bar. The 12.7 mm diameter 350 Maraging steel striker, incident and transmitted Kolsky bar system performed both the HR tension and compression experiments. The dynamic tensile and compression experiments were performed at a strain rate of 1500/s. Strain was measured using a video camera, where the video strain data was processed using a Matlab routine that employs a normalized cross-correlation technique (normxcorr2) [36,37]. The same tensile specimen was used for testing of both QS and HR for comparison purposes of material behavior especially elongation to failure.

3. Results and discussion

EBSD analysis of the feedstock material and the as-deposited microstructures provides a comparison of texture and grain size that result from the solid state material deposition process. The EDAX software calculated the grain diameter by measuring the area of the equiaxed grains, and reported an average grain size of approximately 30 µm for the feedstock material (Fig. 4A). In the deposited material, the measured average grain size of approximately 2.5 µm shows significant grain refinement (Fig. 4B) as compared with the feedstock material.

To further elucidate the microstructural changes occurring during the solid state deposition process, Fig. 5 shows a 3D EBSD representation of the feedstock material and the as-deposited material. In Fig. 5A, EBSD orientation maps show that the feedstock material started as a rolled plate before being machined into a rod for the solid state material deposition process. In addition, EBSD scans in Fig. 5B were performed across the as-build cross-section of specimen from the middle of the build to show a 3D reconstruction of the microstructure. Additionally, EBSD was performed from the top to bottom of the as-built cross-section, and the grain size was uniform with no variations between layers. The EBSD data suggests that DRX is taking place in the stirred region of the material during the MELD process. The significant reduction in grain size observed in the EBSD data (Figs. 4 and 5) is the first point indicating recrystallization. In addition, many of these grains have low levels of intragranular misorientation, as evidenced by the GROD maps (Fig. 6B, D and F). Recrystallized grains should display lower levels of intergranular misorientation due to their lower dislocation densities. While the temperature during MELD was not directly measured, it is generally thought that friction stir processes increase the temperature of the material in the stir zone to temperatures between 60% and 90% of the melting point [2], high enough to produce recrystallization during the intense shear deformation of the solid state deposition process. The grain boundary maps in (Fig. 6A, C, and E) provide further information about the possible nature of the recrystallization process. The presence of a significant fraction of low angle boundaries (red lines in Fig. 6A, C, and E) in an otherwise recrystallized microstructure suggests that new grain boundaries are being formed during the
Dynamic deformation process and that the migration of these boundaries will produce DRX. It should also be noted that these low angle grain boundaries are directly associated with the regions of high intragranular misorientation in the GROD maps (Fig. 6B, D, and F). Doherty et al. [38], describes DRX as a process in which new grain structures are produced in a deformed material via the creation and migration of grain boundaries driven by the stored energy of deformation. Dynamic recrystallization happens during the deformation process, and as such, one would expect to see some low angle grain boundaries in some grains during the DRX process. This description describes the data presented here in this paper. Humphreys distinguishes between a discontinuous recrystallization process in which the grain nucleation and growth occur heterogeneously throughout the material and a continuous process in which recrystallized grain nucleation and growth happen uniformly and gradually; and as such, there is no identifiable nucleation and growth stages [39]. Baker et al., uses the term “continuous dynamic recrystallization” as defined by Humphreys, to describe the stir zone microstructures produced by FSW in ODS steels [40]. Similar, DRX microstructures have been observed for FSW of AA7075 aluminum alloys as well [21].

The hardness results from the microindentation experiments are depicted in Fig. 7 as a bar plot from top to bottom of the as-deposited cross section. Error bars showing the minimum and maximum hardness values for each respective probed layer have also been included in the plot. The average hardness values plotted in Fig. 7 reflect a trend of
slightly higher hardness values in the top of the material with a gradual decrease towards the bottom of the deposition. However, when also examining the standard deviation in the data, there is no clear statistical difference in hardness from the top to the bottom of the as-deposited cross-section. Additionally, in Fig. 7, EBSD IPF maps of the grain morphology and size are shown next to the hardness bars to indicate that the top portion of the deposit (from 0 to 2 mm) has a grain size of 2.6 µm, the middle region (2–4 mm) has a grain size of 2.5 µm, and the bottom region (4–6 mm) has a grain size of 2.5 µm. Since no difference in the grain size is observed from the top to the bottom of the as-deposited cross-section, one may then deduce that there is no measurable Hall-Petch effect in the material from the top to the bottom. It should also be noted that the hardness of the feedstock material is significantly higher than any of the as-deposited material measured since the feedstock is heat-treated.

EBSD was also used to assess the microtextures generated in the as-deposited microstructures. The pole figures (PF) in Fig. 8 show the variation in texture across the as-deposited cross section. As seen in Fig. 8A, the material has a strong texture (MRD = 8.3) in the top of the build that weakens toward the middle (Fig. 8B) and the bottom (Fig. 8C) of the deposition (MRD = 3.9). Previous research by Montheillet et al. [41] showed the four typical torsional textures that may be found in FCC aluminum, and Fig. 9 shows the texture fiber type-A and type-C in the {111} PF from which the fiber texture A can be correlated to the {111} PFs in Fig. 8A. In Fig. 8A, the {111} fiber component and a split of the {111}<110> fiber component are evident. In the middle of the deposition (Fig. 8B) and in the bottom of the deposition (B-C), there is a combination of fiber A and C texture. For comparison, a weak Goss type rolling texture in the feedstock material was seen. Fig. 10 shows the {111} PFs from the top, middle, and bottom at a different angle to elucidate the torsional texture (torsional axis is out of the plane of the page). Research by Fonda et al. [42], reported similar torsional textures for FSW of AA2195.

Crystallographic texture in a polycrystalline material refers to the preferential orientation of the grains in the material. Strong texture in a material may cause anisotropic behavior during plastic deformation, and, when the mechanical behavior of a material changes due to texture, it is known as texture strengthening [43–45]. As seen in Figs. 8 and 10, the stronger texture in the top of the as-deposited material may result in variations in mechanical behavior when compared to the middle and the bottom layers.

TEM analysis further identifies how the feedstock material changes from the bottom, middle, and top regions of the deposition, which correlates to the EBSD IPF maps shown in Fig. 7. STEM-HAADF images from the feedstock material (Fig. 11A) show many, small (≪ 100 nm) θ′ precipitates, as expected from AA2219-T8 and T851. These precipitates are responsible for the high yield strength levels measured in the feedstock material. For the AFS-deposited material, Fig. 11B–D show STEM-HAADF micrographs for the bottom, middle, and top regions, respectively, that contain larger θ (Al2Cu) precipitates. All the STEM micrographs from the bottom, middle, and top depict similar size and type (θ phase) precipitates. No θ′ precipitates were observed indicating that there should not be any precipitation strengthening after the solid state deposition. It should be noted that it is the spacing between precipitates that is responsible for increased yield strength for large precipitates. The average distance between precipitates in the solid state deposition material is much larger than in the feedstock material, and as such, there is very little precipitation strengthening. As mentioned before, previous research by Kang et al. [28] in FSW of AA2219 found the same θ′ precipitates in the base material and θ precipitates in the SZ.

These TEM results suggest that the temperatures during MELD processing are high enough to first dissolve the θ′ precipitates and then the cooling rate is slow enough that larger, θ phase particles form. According to the work of Lorimer [46], the solvus temperature of θ′ precipitates in the Al-Cu system is around 753 K (480 °C). Friction stir welding and friction stir processing can generate temperatures in the stir zone ranging from 60% to 90% of the melting point [2]. These temperatures could exceed the solvus, thus dissolving the θ′ phase. In fact, the data presented here suggests that the stir zone temperatures reach at least 80% of the melting point of the material, the approximate solvus temperature for the Al-Cu system. As the material does not cool quickly during further solid-state material deposition processing, the θ phase will nucleate and grow. Future work will monitor the temperatures during the solid state material deposition process to provide further insight about this precipitate evolution.

Next, the average QS and HR tensile behaviors for LD oriented specimens from the top, middle, and bottom layers with the associated uncertainty bands corresponding to the maximum and minimum experimental values are plotted in Fig. 12A and B. For both QS and HR (Fig. 12A and B), the top layer exhibits higher YS and tensile strength (TS) than the middle and bottom layers (see Table 2 for results).

Additionally, in Fig. 12A a clear delineation is observed with maximum YS and TS observed in the top layer specimens that decreases for
Fig. 11. TEM micrographs (A) feedstock material showing θ’ precipitates, (B) bottom, (C) middle, and (D) top, TEM micrographs showing θ precipitates and grain structure from the build cross-section Zone 2 for the as-deposited AA2219 (see Fig. 7 for reference of locations). (Note: magnification in (A) is much higher than B–D).

Fig. 12. As-deposited tensile results comparing deposition layer dependency in (A) quasi-static and (B) high rate. For both strain rates, the plots identify that the top layer is strongest followed by the middle and bottom layer.
Multiple factors can contribute to the YS of a material as shown in Eq. (1). The quasi-static long transverse, short transverse, and longitudinal directions exhibited a YS of 145, 140, and 150 MPa, respectively. The high rate long transverse, short transverse, and longitudinal achieved a YS of 295, 238, and 225 MPa, respectively. All compression tests were stopped at a strain of 0.25.

Furthermore, compression QS and HR experiments examined the stress-state and strain rate dependence of the as-deposited material. Fig. 13 plots the average (3 specimens) QS and HR compressive behavior up to a strain of 0.25 for the TD, DB, and LD directions with uncertainty bands corresponding to minimum and maximum experimental values. The QS compressive data shows a near isotropic response of the as-deposited AA2219, with an average YS of approximately 145 MPa for the TD, 140 MPa for the BD, and 150 MPa for the LD directions. For the HR compressive material behavior, the TD has a higher yield than the BD and LD directions, with an average YS of approximately 295 MPa for the TD, 238 MPa for the BD, and 225 MPa for the LD directions. In QS, the material experiences more strain hardening and higher TS than at HR. However, as expected the HR achieves higher YS than QS, but the HR does not experience as much strain hardening and eventually plateaus. The QS and HR compression experiments were stopped at a maximum strain of 0.25 mm/mm since the material did not fracture.

To compare stress-state asymmetry, the average QS and HR tensile behavior for the TD and LD directions with the associated uncertainty bands corresponding to the maximum and minimum experimental values are plotted in Fig. 14. The QS tensile TD specimens achieved an average YS of 148 MPa, an TS of 363 MPa, and εf of 0.24 mm/mm. The QS tensile LD specimens achieved an average YS of 143 MPa, an TS of 355 MPa, and εf of 0.26 mm/mm. The QS tensile specimens in both directions exhibited a ductile failure (Fig. 15) with strain hardening and softening after yielding. Similar to the compressive behavior, the material shows more strain hardening in the QS case, when compared to the HR, which after yielding exhibits a stress-strain curve plateau. The HR tensile TD and LD specimens achieved average YS of 235 and 240 MPa, TS of 240 and 240 MPa, and εf of 0.37 and 0.37, respectively. The HR fractography (Fig. 16) revealed a ductile behavior with a cup-cone fracture surface. For both QS and HR cases, isotropic behavior in the two orientations are observed for the tensile mechanical behavior.

SEM fractographic analysis of both QS and HR (Figs. 15 and 16, respectively), show ductile failure behavior for the as-deposited AA2219 material. SEM QS fractography images with increasing magnifications are shown in Fig. 15A–D. Similarly, the HR SEM fractography images are shown in increasing magnifications in Fig. 16A–D. Both QS and HR fracture surfaces exhibited microvoids and dimples.

Based on the data in this paper, the YS of as-deposited AA2219 appears dominated by the grain size, i.e. Hall-Petch type strengthening. Multiple factors can contribute to the YS of a material as shown in Eq. (1).

\[ \sigma_Y = \sigma_0 + \sigma_{SSS} + \sigma_{WN} + \sigma_{IP} + \sigma_T + \sigma_{TXT} \]  

where, \( \sigma_Y \) is the material YS, \( \sigma_0 \) is the Peierls-Nabarro stress \([47]\), \( \sigma_{SSS} \) is the solid solution strengthening \([2,8,9,48]\), \( \sigma_{WN} \) is the work hardening based upon dislocation density \([38,49]\), \( \sigma_{IP} \) is the Hall-Petch strengthening \([2-4,50]\), \( \sigma_T \) is the precipitation strengthening \([2,8,51,52]\), and \( \sigma_{TXT} \) is the texture strengthening \([43-45]\).

Since these samples were fabricated by a solid-state process, a
change in chemical composition from the feedstock material is not expected, and thus, the $\sigma_W$ and $\sigma_{SS}$ contributions are unlikely to be different between the feedstock and the as-deposited material. Any work hardening present in the feedstock material is removed by the MELD process, thus negating the contribution of $\sigma_{WH}$. From Fig. 5, it is clear that the grain size in the as-deposited material is significantly refined from the feedstock and as such, the $\sigma_{HP}$ contribution should be large in the as-deposited material. The fine precipitates present in the feedstock material are all removed by the solid-state deposition process, and so, $\sigma_P$ will be quite small in the as-deposited material. Lastly, there is a difference in the texture between the feedstock and the as-deposited material, so there could be a texture strengthening component, $\sigma_{TXT}$, but its magnitude is unclear.

Using similar reasoning, our current explanation for the small decrease in YS from the top layer of the deposition to the bottom is that the YS is controlled by texture strengthening. The grain size, shown in Fig. 7, is the same for all of the layers (top, middle, and bottom). None of these three layers possess the fine precipitates necessary for precipitation strengthening. The composition and dislocation density are not expected to be different between the three layers. In contrast, the top layer was the most textured of the three layers. This texture reduced significantly in the middle and bottom layers; and, in addition, the nature of the texture changed from an A-fiber only to a mixture of A- and C-fibers. Further work would be necessary to estimate the direct impact of these textures on YS.

The averaged as-deposited AA2219 tensile properties are compared to AA2219 tensile properties from various manufacturing or joining techniques in Table 3. The MELD process (highlighted in gray in Table 3) resulted in higher YS, UTS (converted from TS to engineering UTS for comparison purposes) and strain to failure than AA2219-O and as-deposited EBF AA2219. Additionally, MELD AA2219 resulted in increases in the same three properties compared to FSW 2219-O. The MELD AA2219 mechanical properties are below those of AA2219-T851, but as shown previously in the TEM data (Fig. 11), the nano-precipitates that provide the precipitation strengthening of AA2219 were eliminated during the MELD. Previous research on EBF by Taminger and Hafley [10] showed that after depositing the AA2219 and performing the T62 heat treatment; the material properties of wrought AA2219-T62 were achieved, suggesting that MELD AA2219 would be heat treatable as with the EBF AA2219. The strain to failure was not directly compared to published literature data since this study used a subscale specimen for both QS and HR experiments instead of a larger ASTM E8 tensile coupon, but a column of strain to failure is provided for a qualitative reference in Table 3.

4. Summary and conclusions

This is the first manuscript to detail the microstructure and mechanical properties of solid state material deposition AA2219 fabricated from AA2219-T851 feedstock material.

- EBSD results revealed a strong grain refinement from the high shear
and elevated temperature process. A uniform grain size was found from the top to the bottom of the as-deposited cross-section.

- The top layer had the highest YS and TS, followed by the middle and bottom, respectively. This small difference can be attributed to texture strengthening.
- EBSD PFs showed that the material has a strong torsional fiber A texture in the top of the build, and a mixture of A and C texture in the middle and bottom sections.
- TEM showed that there are no θ' precipitates in the as-deposited cross-section, therefore no precipitation strengthening should be expected. The mechanical properties of the as-deposited AA2219 material are higher than AA2219-O, but the values are below those of AA2219-T851 since there are no θ' precipitates providing strength to the material.
- As mentioned above, previous research by Taminger and Hafl ey [10] showed that after applying the T62 temper to the as-built EBF material, the materials mechanical properties matched those of a AA2219-T62. This is extremely important, since being able to regain those microstructural and mechanical properties that the T62 and T851 tempers provide is crucial for the applications of this alloy.
- The strain rate dependence was examined, and at HR, the material experienced higher yield strength and lower tensile strength. At HR, the material strain hardening was less than when tested at QS where the HR material stress-strain curve plateaus after yielding [53].
- Texture strengthening appears to be the mechanism responsible for the changes in behavior between the top, middle, and bottom layers of the deposition.
- The tensile specimens in HR specimens also exhibited more strain to failure than in the tensile specimens in QS.

### References

3. W. Xu, J. Liu, G. Luan, C. Dong, Temperature evolution, microstructure and


