Rationalization of anisotropic mechanical properties of Al-6061 fabricated using ultrasonic
additive manufacturing¹

Niyanth Sridharan^{1,2*}, Maxim Gussev² Rachel Seibert², Chad Parish², Mark
Norfolk³, Kurt Terrani², S.S.Babu^{1,2}

¹ University of Tennessee, Knoxville, TN 37916

² Oak Ridge National Laboratory, Oak Ridge, TN 37831

³ Fabrisonic LLC, Columbus, OH 43221

*Corresponding author.

All correspondences regarding this manuscript may please be directed to

niyanth.sridharan@gmail.com

Dept. of Mechanical, Aerospace & Biomedical Engineering,

1512 Middle Drive, 414 Dougherty, Knoxville, TN 37996-2210,

Phone: 614-961-7436

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ABSTRACT:	3
1. INTRODUCTION:	4
2. EXPERIMENTAL WORK:	6
2.1 FABRICATION OF SAMPLES:.....	6
2.2 MECHANICAL TESTING:.....	7
2.3 MICROSTRUCTURE CHARACTERIZATION	8
3. RESULTS AND DISCUSSIONS:	9
3.1 MECHANICAL TESTING AND FRACTOGRAPHY:	9
3.2 CHARACTERIZATION OF THE AS RECEIVED SAMPLES:	10
3.2.1 <i>Optical and Scanning Electron Microscopy:</i>	10
3.2.2 <i>Orientation Imaging Microscopy and selection of regions for TEM</i>	11
3.2.3 <i>Transmission Electron Microscopy</i>	13
3.3 CHARACTERIZATION OF THE SAMPLES AFTER TESTING	14
3.3 IN-SITU TENSILE TESTING	15
3.3.1 <i>Feasibility of strain localization</i>	15
3.3.1 <i>Microstructure at regions of strain localization</i>	15
3.4 RATIONALIZATION OF STRAIN LOCALIZATION AND FRACTURE BEHAVIOR OF UAM	16
SUMMARY AND CONCLUSIONS:	17
ACKNOWLEDGEMENT:	18
REFERENCES:	18
LIST OF FIGURES :	23

Abstract:

Ultrasonic additive manufacturing (UAM) is a solid-state process, which uses ultrasonic vibrations at 20 kHz along with mechanized tape layering and intermittent milling operation, to build fully functional three-dimensional parts. In the literature, UAM builds made with low power (1.5 kW) exhibited poor tensile properties in Z-direction, i.e., normal to the interfaces. This reduction in properties is often attributed to the lack of bonding at faying interfaces. The generality of this conclusion is evaluated further in 6061 aluminum alloy builds made with very high power UAM (9 kW). Tensile deformation behavior along X and Z directions were evaluated with small-scale in-situ mechanical testing equipped with high-resolution digital image correlation, as well as, multi-scale characterization of builds. Interestingly, even with complete metallurgical bonding across the interfaces without any discernable voids, poor Z-direction properties were observed. This reduction is correlated to coalescence of pre-existing shear bands at interfaces into micro voids, leading to strain localization and spontaneous failure on tensile loading.

Keywords

Ultrasonic additive manufacturing, Electron Back Scatter Diffraction, Digital Image correlation, tensile tests

1. Introduction:

Additive manufacturing is an emerging technology with many industrial applications that is not achievable by traditional manufacturing methodologies. Ultrasonic additive manufacturing (UAM) is a solid state joining process that uses high-frequency (20 kHz) vibrations to fabricate parts in a layered manner. Compared to traditional industrial technologies, UAM allows for complex shaped parts with internal channels or cavities and with insertions of other materials with special functional properties. Usually UAM combines mechanized tape layering and concomitant machining to fabricate fully functional parts [1]. Due to the nature of the UAM process, the builds have relatively low strength level when loaded perpendicular to the welding layers (often referred as Z-direction). This low strength has been attributed to the presence of voids and the lack of metallurgical bonding between the layers.

Much of the published work attempts to address the above challenge by increasing the UAM energy input. The ultrasonic energy is supplied through a sonotrode vibrating at a frequency of 20 kHz. While the frequency of vibration is usually fixed, by adjusting the normal force, vibration amplitude and the sonotrode travel speed controls the energy input [2]. The linear weld density (LWD), i.e., the ratio of the well-bonded area to the total length of the interface inspected, is most often used to quantify the effectiveness of the processing parameters. Usually a build with a higher linear weld density was hypothesized to have better mechanical properties [3]. There have been several studies to investigate the effect of process parameters (normal force, vibration amplitude, etc.) on the mechanical properties of the UAM builds [4-6]. A wide range of testing methodologies including peel tests [4, 5], tensile and shear tests [6] have been used to evaluate the mechanical strength [7]. Often, 85% reduction in strength along the Z-direction has been observed in comparison to that of bulk monolithic material [2]. The low

strength in this direction has been correlated to the linear weld density, i.e., builds with a higher LWD failed at higher loads while those with a lower LWD failed at lower loads. Despite the fact that the samples failed in a brittle fashion macroscopically, fractography analysis revealed localized regions with ductile fracture with dimpled features. These were interpreted as regions where metal to metal bonding had occurred. Based these results, an emerging hypothesis among researchers is to correlate the strength along the Z-direction to both LWD and the efficiency of metal-to-metal bonding [8].

The term metallurgical bonding implies metal-to-metal bonding at an atomic level. In the ultrasonic additive manufacturing process this is achieved by the ultrasonic vibration, which has been hypothesized to remove and disintegrate the oxide layers and bring the metal atoms in intimate contact [9]. Attempts have been made to study the phenomena of oxide dispersal occurring during the UAM process, however the phenomena has not been well understood [1, 10]. For example, even when the processing conditions lead to adequate metal-to-metal contact, the presence of oxide particles at the interface will change strength of solid-state welds [11]. It was also hypothesized that these regions are the first to break while tensile testing [11]. Adding to the complexity, a strong correlation between the mechanical properties and non-oxide defects have also been postulated in other solid-state processes, i.e., friction stir welding [12, 13]. The nature of bonding in UAM builds have been characterized before using multi scale characterization including TEM [1, 10] and dual beam FIB and electron backscatter diffraction [9, 14]. They all seem to indicate that the bond formation was a result of dynamic recrystallization at the bond interface [1, 9, 14, 15]. This conclusion is based on analyzing the grain structure and crystallographic micro textures at the interface and comparing it with the bulk [9, 14].

Furthermore, dynamic recrystallization was hypothesized to occur driven by adiabatic heating promoted by the severe high-strain rate plastic deformation. Interestingly, effects of adiabatic heating on microstructure and mechanical properties have been studied during high strain rate deformation of metals before. Adiabatic heating occurs because the heat dissipation is not as rapid as the strain energy deposition at high rates, often leading to shear bands [16]. Adiabatic shear bands, once initiated, often lead to localization of strain inside the material resulting in a catastrophic failure. Although, there are evidences for adiabatic deformation heating occurring under conditions set up by the UAM process [1, 2, 9, 14], it is not clear whether these shear bands influence the mechanical properties UAM builds and forms the motivation for the current research.

The above literature review poses the following fundamental question: is the reduction in Z-strength related to presence of (i) voids, and/or (ii) oxide particles or oxide films, and/or (iii) presence of adiabatic shear bands. The present attempts to rationalize the above using a Al-6061 builds made by UAM with detailed in-situ small-scale tensile tester equipped with digital image correlation and multi-scale microstructure characterization of the interface regions before and after mechanical testing [17].

2. Experimental work:

2.1 Fabrication of samples:

Commercially available 150- μm thick tapes of Al-6061 H-18 alloy were used in this study. The composition of this feedstock material is shown in Table 1. The UAM-builds were fabricated using the 9kW UAM machine at the Fabrisonic LLC. Vibration amplitude of 35 μm , normal force of 5000N, and travel speed of 200 inches/minute (84.7 mm/s) were used for fabrication of builds. During deposition, the build was maintained at 75°C.

Table 1. Alloy compositions (mass %) determined by direct current plasma emission spectroscopy per ASTM E 1097-12.

Al	Mg	Si	Fe	Cu	Cr	Ti	Mn	Zn	Ni
Balance	0.86	0.66	0.35	0.23	0.13	0.084	0.082	0.077	0.015

2.2 Mechanical Testing:

Tensile dog-bone specimens (see Fig. 1) were manufactured using electro-discharge machining (EDM) and special care was taken to avoid any occasional loading or bending of the specimens. Specimens were machined in three directions: X – along the direction of the travel of the sonotrode, Y – along the sonotrode vibration direction, and Z – along the build direction where the load will be applied normal to the interfaces as shown in Figure 1(a). At least 5 specimens were tested per each direction. Tensile tests were performed on a MTS Insight 2-52 one-column tensile screw machine. All specimens were shoulder loaded using special grips and tested at room temperature with a strain rate of 10^{-3} s^{-1} .

After initial scoping studies, interrupted tensile tests were also performed on the specimens in the Z direction where the specimens were loaded to a pre-determined load value, which is less than the load to failure. This was done to investigate the effects of voids on the deformation localization and fracture processes. Non-contact strain measurements were performed using Digital Image Correlation (DIC) to reveal the strain localization regions. The details of DIC method are presented elsewhere [18]. Prior to the testing, the specimens were painted with white and black paint forming random speckle pattern. An Allied Vision GX3300 digital camera was employed to get images during the tensile tests; the resolution was $\sim 12 \text{ } \mu\text{m}$ per pixel. After the

test, strain fields were calculated using VIC-2D commercial software and a custom program utilizing common DIC algorithms.

2.3 Microstructure characterization

The specimens were cut from the build using a low speed diamond saw with lubricant to protect the dislocation substructures formed during processing. The specimens were then cold mounted and polished using SiC paper followed by diamond paste. To remove the deformation layer caused by the diamond paste polishing, the samples were further polished with colloidal silica using a Beuhler Vibromet. The samples were examined using optical microscopy to look at the void fraction in each sample. Following optical microscopy **O**rientation **I**maging **M**icroscopy (OIM) using **E**lectron **B**ack **S**cattered **D**iffraction (EBSD) was performed on a JEOL 6500 FEG Scanning electron microscope operating at 20kV using a probe current of 4nA. After OIM, thin foils were extracted using a FEI Helios **D**ual **B**eam **F**ocused **I**on **B**eam (DB-FIB). **T**ransmission **E**lectron **M**icroscopy (TEM) performed using a JEOL 2100F 200kV TEM. The acceleration voltage used was 200kV with an emission current of 111 μ A.

After tensile testing, the fracture surface was examined using a Hitachi S3400 SEM with an accelerating voltage of 20kV. Following fractography, regions showing ductile failure were identified and sectioned using a low speed diamond saw to preserve the dislocation substructures, as mentioned earlier. They were later cold mounted in epoxy and polished using similar techniques described earlier and analyzed using optical microscopy and EBSD. EBSD was performed in these samples to compare the extent of deformation in the different regions of the build and to understand the reason for failure at the interface where the failure occurred.

3. Results and Discussions:

3.1 Mechanical testing and fractography:

Figure 2 shows typical engineering stress-strain curves for the specimens prepared using UAM. The builds were fabricated using Al-6061 tapes in the H-18 condition. Average strength and ductility values are given in Table 2. The data for a commercial wrought alloy of the same composition are given to provide a baseline for comparison. The wrought alloys were in the T-6 temper. As follows from the results, UAM-produced specimens had lower strength and smaller ductility, compared to the wrought alloy. Z-specimens demonstrated strong decrease in the strength level and macroscopically brittle fracture.

Table 2. Mechanical properties of the UAM-specimens and wrought commercial alloy (T-6)

Type	Yield stress, MPa	Ultimate stress, MPa	Uniform elongation, %	Total elongation, %
Wrought alloy	294	315	7.2	15.4
X	217	225	1.1	5.1
Y	221	224	0.5	6.0
Z*	46	—	—	—

* Fracture stress is given for the Z-direction specimens.

The fractographs shown in figure 3(c)-(d) displayed equiaxed dimples away from the interfaces showing that the sample failed by a ductile mode. However, near the interfaces the dimples were absent. This may be due to the fact that delamination at the interface had immediately occurred on loading. This phenomenon has been observed and documented in Al alloys fabricated using ARB [19]. De bonding in the interfaces while loading parallel to the interfaces could be attributed to interfacial shears in-between layers on loading. If delamination had occurred as proposed after loading, there would not be any signs of localized plastic deformation

in these locations of the build. In the case of UAM this lack of deformation and de bonding of layers would have contributed to the lower strength of the build in the X direction.

As expected while testing along the ‘Z’ direction the samples fail by de-bonding of the tapes without any elongation. The fractography of the tested samples is shown in the figure 3(a-b). The samples inspected showed the presence of local regions with dimpled features. The presence of these features demonstrated that, despite the absence of macroscopic ductility, some regions in the test specimen failed by micro void coalescence. These results can be interpreted as the evidence for metal-to-metal (metallurgical) bond between the two tapes [8]. However, this interpretation has to be confirmed with detailed characterization of samples in the as-received and post-testing conditions.

3.2 Characterization of the as received samples:

3.2.1 Optical and Scanning Electron Microscopy:

The as received samples were examined using optical microscopy and inspected for voids. Figure 4(a) shows the regions where microstructure analysis was performed schematically. As shown in figure 4(b) and (c) the as received sample shows close to a 100% LWD at the bottom region. It is pertinent to mention that the current UAM process parameters were adopted from the literature where the authors had optimized the process parameters using design of experiments [20]. Nevertheless, some voids were observed at the top regions of the builds as shown in figure (4c). This is due the fact that as the build height increases, only a fraction of the deposited energy is used for creating the bond between the layers. This phenomenon is related to compliance of the build and to compensate for this decrease in energy dedicated for bonding, the input energy needs to be progressively increased [20], which was not done in the current research.

Several defect features, observable by optical and scanning electron microscopy, are reported after UAM processing. Ram et.al [3] have classified them as line, point and parabola shaped defects. The line defects were hypothesized to occur as a result of insufficient energy input to deform the interfaces and are not observed in the present study due to the optimized parameters used. However, the fact that such defects were observed at the top regions of the build does indicate that the energy level does drop down as hypothesized previously. The parabola voids are related to the roughness induced on the top surface of the foil by the sonotrode and not the input energy. Extensive electron back-scattered diffraction performed on the defects indicates significant deformation at the regions below the parabola shaped defects. However no deformation could be detected on the foil on the top region of the defect [14]. Thus the formation of these defects was hypothesized to occur as a result of the inability of the Al to flow into cavities created due to the action of the sonotrode. These were also reported to be potential sites for crack nucleation during loading [1-5], but no in situ study was performed to validate this hypothesis. Apart from those there are several regions where micro cracks and voids had developed at the interface of the builds as shown in figure 4(d). These micro porosities were observed previously during ultrasonic spot welding of Al 6061 and were hypothesized to occur as a result of dispersion and trapping of interfacial voids. However no mechanism for the hypothesis was put forward [21].

3.2.2 Orientation Imaging Microscopy and selection of regions for TEM

The primary aim of the multiscale characterization performed is to understand if metallurgical bonding occurred at the interface. For metallurgical bonding to occur two conditions need to be met viz. (i) Complete oxide layer removal (ii) Plastic deformation to effect collapse of asperities to bring metallic surfaces to intimate contact [11, 22]. To understand if plastic deformation

occurred at the interface, the grain structure at the interface was analyzed using EBSD. The data is presented shown in figure 5 (a). The interface exhibited a refined equiaxed microstructure, which is different from the pancake structure of the bulk tapes. This microstructure is typical of that at the interface in materials after ultrasonic additive manufacturing [1, 9, 10, 15]. The observed microstructure is reported to form as a result of dynamic recrystallization driven by adiabatic heating [1, 2, 9, 14]. However there is no consensus if the recrystallization at the interface is a continuous or discontinuous dynamic recrystallization [16, 23]. Only by accurately measuring the temperature rise at the interface at high temporal resolutions (20KHz) can one address the nature of the recrystallization phenomena. This refinement in grain size has also been observed during other solid state welding processes such as Accumulative Roll Bonding (ARB) where the grain size reduction is proportional to strain [19, 24]. The grain refinement observed using EBSD shows the presence of plastic deformation at the interface.

The above orientation imaging was used as a reference to preparation of transmission electron microscopy samples by focused ion beam imaging. Considering the nature of the strong texture that arises at the interface as a consequence of the UAM process, to confirm the exact location of the interfaces on the extracted TEM samples a series of Selected Area (SA) diffractions were performed. A total of 5 SA diffraction patterns covering several regions of the samples were performed. The results show that nearly all the grains that were sampled were oriented with their $\langle 110 \rangle$ zone axis parallel to the electron beam. This is consistent with the results from previous work done using EBSD where the $\langle 110 \rangle$ is parallel to the direction of vibration of the sonotrode [14]. Figure 5(b-c) shows typical diffraction patterns seen at the interface without any tilting. This result confirmed that the TEM samples extracted by FIB are exactly from the interface region.

3.2.3 Transmission Electron Microscopy

The STEM results shown in figure 6(a) shows that there is no distinguishable line that clearly denotes the interface between the two bonded tapes. Since the interface is not clearly visible it is evident that Al grains may have migrated across the interface as hypothesized previously [1]. To understand the distribution of oxygen, energy dispersive spectrum maps were generated (see Fig. 7). Previous work on friction stir welding of 5XXX Al alloys showed the presence of Mg rich oxides at the interface [25]. Characterization using FIB and TEM has shown that the MgO could form during friction stir welding [12]. Such oxide formation occurring at the interfaces during the UAM process could significantly affect the mechanical properties of the build. However, the EDS maps shown in figure 7 (a-d) did not show any signs of oxygen or Mg enrichment at the interface as previously observed and reported in Al-6061 [14]. This, along with the fact that the interface line could not be identified using TEM, shows the complete oxide removal and atomistic contact of the Al atoms at the interface. The presence of plastic deformation along with the removal of the oxide layer confirmed metallurgical bond at these interfaces.

This leads to a fundamental question, why do we see poor Z-properties (see Fig. 2), even with the presence of metallurgical bonding. Furthermore, the grain refinement seen in Fig. 5(a) should have lead to an increase in strength. Extensive grain refinement has shown to lead to significant strengthening in Al alloys fabricated using ARB [19, 24], with a caveat that most of the mechanical tests in ARB samples were performed along the X and Y orientation. Though studies made using nano indentation showed an increased rise in hardness at the interfaces fabricates using UAM [26] unlike ARB the strengths do not show a corresponding increase. The reason for such a behavior is not well understood.

3.3 Characterization of the samples after testing

To rationalize the drop in mechanical strength in Z-direction, further characterization was performed on the fractured samples (tested in both X and Z orientation) using EBSD. The interface close to the failure was targeted, because we seek to understand deformation behavior during uniaxial loading. The investigated regions are shown in figure 2(b). Both the samples tested along ‘X’ and ‘Z’ directions show an equiaxed microstructure with a refined grain size at the interface as shown in figure 8(a-c). No major difference at the interfaces could be seen and the microstructure was uniform throughout the interface region. The extent of the local deformation was estimated by calculating the Kernel Average Misorientation (KAM). KAM is a measure of the misorientation between the selected kernel and the adjacent kernels. An increase in the average KAM value corresponds to a higher degree of misorientation between adjacent kernels and can be interpreted as a sign of enhanced dislocation activity. A lower KAM corresponds to a lower degree of plastic deformation and vice versa. A region failed in a ductile fashion should show significant plastic deformation and thus possess a higher average KAM [27]. Surprisingly the KAM distribution obtained from the interfaces of the X and Z directions and the as fabricated samples did not show any difference (see Fig. 8). This shows that locally at the interface both the samples tested along the ‘Y’ and ‘Z’ do not exhibit any deformation since the KAM distribution is the same. In corollary, similar phenomenon was observed in accumulative roll bonding [19], which leads to catastrophic de-bonding of layers along the bond-line during tensile testing.

There are two ways to interpret the above results with reference to UAM. First, all the samples tested by us (see Fig. 2) may not have metallurgical bonding and the failure occurs catastrophically across the mechanically locked interfaces. However, the results from

characterization (see Figs. 5 and 6) confirmed complete metallurgical bonding occurring. Second, the failure may occur due to pre-existing strains at the interface regions that are barely below a critical level to induce catastrophic failure. On tensile loading, with small amount of strains, the sample exceeds the criticality by strain localization and led to catastrophic failure. Therefore, to delineate the exact mechanism interrupted tensile tests with in-situ spatial strain characterization are required.

3.3 In-situ Tensile Testing

3.3.1 Feasibility of detecting strain localization

Tensile specimens were pre-loaded to a particular stress value (40 MPa, i.e. below the fracture load; see point M in Figure 9) along the Z direction and unloaded (Point U in Figure 9). Strain analysis was performed using digital image correlation (DIC) during the loading and unloading steps. The strain field analysis revealed an average strain of $\sim 0.06\%$ in the entire gauge length. The strain was confirmed to be elastic since the strain recovered after unloading (U). The fluctuations in strains ($\sim 0.02\%$) is a result of noise or for the given test configuration. However a region with significantly higher strains ($\sim 0.12\%$) appeared in the gauge (marked with a black arrow) and did not recover after the unloading. This shows that though macroscopically The appearance of plastic strain indicates that damage accumulation processes (e.g., crack nucleation and opening) may occur below the **actual** fracture stress in the specimens loaded in Z-direction.

3.3.1 Microstructure at regions of strain localization

To understand the strain localization, microstructure characterization was performed at the peak strain areas and the results are presented in figure 10. Post deformation, optical microscopy showed several cracks nucleating at various regions across the length of the build as shown in figure 10(a). The results are surprising because of the following two observations: (i) presence of voids did not guarantee nucleation of cracks; and (ii) voids were not present at the locations

where strain had localized resulting in crack growth. EBSD showed that there are no signs of plastic deformation as the KAM of the areas where the cracked and un-cracked regions are similar. The average KAM distributions are shown in the figure 10 (d). This clearly demonstrates that there is no microstructural difference between the cracked and un-cracked regions at the interface.

3.4 Rationalization of Strain Localization and Fracture Behavior of UAM

In order to understand the reason for the initiation of cracks in the samples it is necessary to understand the changes that the material undergoes during UAM. The UAM process as stated previously is a high strain rate deformation process that employs high frequency sliding between the tapes to initiate a solid-state bond [2]. The high strain rate deformation leads to adiabatic heating with temperature exceeding $0.5T_m$. An increase in temperature prevents the material from strain hardening by preventing dislocation interaction, while plastic deformation continues to localize at regions where strain hardening is minimum or absent [28],[29]. With continued deformation this results in the formation of micro voids, which on further straining can coalesce and form cracks. The presence of micro voids observed could be a result of plastic instability set up due to the formation of adiabatic shear bands [16]. The exact mechanism as to when the voids and cracks nucleate is a subject of much debate and is an active area of research [16] [23, 28] . Characterization of the as received microstructure shows the presence of micro cracks and micro voids at the interface. The threshold value for the onset of plastic instability varies with strain rate and for Al-Li alloys it was determined to occur at a strain of 0.10-0.17 after $80\mu s$ when tested at a strain rate of $2.3 \times 10^3 \text{ s}^{-1}$ [16, 23]. The reported strain rate, strain and the time seem to agree with the UAM regimes. Subsequently when the build is loaded in tension the voids grow until they reach the edge of the region where instability has occurred. The voids are unable to

grow beyond the interface region since the adjacent regions can strain harden. Thus the voids become elliptical as they elongate finally nucleating as cracks by coalescing as shown in the schematic, Fig. 11. This results in the familiar dimpled fracture surface. A schematic of this process is illustrated in the figure 11 (a-e). It is noteworthy that the above schematic mechanism is based on available data from this work. For confirmation of the mechanism, there is a need for in-situ tensile testing coupled with diffraction contrast tomography and phase contrast tomography [30]. We also propose that to offset this deleterious effect of proposed pre-existing shear banding, heat treatment of the builds can anneal the strain localization [31] and thereby improve the mechanical properties when loaded normal to the interface.

Summary and Conclusions:

Al-6061 builds were fabricated and the mechanical properties were evaluated. The mechanical properties along the X and Z directions were evaluated. The properties showed a marginal decrease while loading parallel to the interfaces (X direction) while a brittle failure was observed while loading perpendicular to the interfaces (Z direction). EDS did not pick up any oxygen peaks and there was no line that could clearly delineate the interface. Both of these observations indicate metallurgical bonding between the tapes and hence the decrease in properties was not due to the lack of bonding. The builds failed without any detectable deformation at the interface. The brittle failure was attributed to the onset of strain localization during the UAM process. High strain rate deformation results in strain localization, leading to the formation of micro-voids at the interface. Upon loading normal to the interface these voids coalesce to nucleate cracks that propagate through the interface resulting in a mode-I type fracture without any macroscopic ductility.

Acknowledgement:

The aid and technical insight of Kevin Field is gratefully acknowledged. This work was sponsored by Laboratory Directed R&D funds at Oak Ridge National Laboratory. Scanning electron microscopy was performed at Center for Nano phase Materials (CNMS), Oak Ridge National Laboratory, sponsored by the Department of Energy, Office of Basic Energy Sciences. This research was performed, in part, using instrumentation provided by the Department of Energy, Office of Nuclear Energy, Fuel Cycle R&D Program and the Nuclear Scientific User Facilities. The authors also gratefully acknowledge the U.S. Department of Energy, Office of Energy Efficiency and Renewable Energy, Advanced Manufacturing Office.

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List of figures :

Figure 1: Schematic showing the (a) Sample extraction for mechanical testing. Since X and Y direction are the same they are used interchangeably throughout this manuscript (b) showing the dimensions of the tensile test samples used

Figure 2: Typical engineering tensile curves for the UAM-produced specimens in comparison with bulk commercial alloy. Note the drastic decrease of strength along the Z direction. Areas where post deformation characterization was performed are also marked in the figure

Figure 3: Shows the fractographs of the samples (a) Tested along the ‘Z’ direction (b) Shows the ductile regions in figure 3(a) at a higher magnification. The presence of dimples at the interface indicative of a ductile failure (c) Samples tested along the ‘X’ direction (d) Samples tested along the ‘X’ direction ((c) shown at a higher magnification) showing de bonding of the tapes at the interface

Figure 4: Shows the (a) Schematic showing the regions where microstructure analysis was performed (marked using a blue square) (b) Shows the microstructure at the bottom region where no voids were visible (c) Shows the optical micrograph at the sample at the top region showing the presence of ‘parabola shaped defects’ (d) Shows a high magnification SEM micrograph indicating the presence of micro voids at the interface. Note that the voids are only ~ 1.5-2.0 microns in length

Figure 5: (a) EBSD showing the microstructure of various interfaces. Note the equiaxed grains at the interface marked with the black arrows. The direction of accumulation of tapes is marked as “build direction” (b-c) Shows the SA diffraction patterns from the interfaces showing that for all the regions the sample is oriented with the $\langle 110 \rangle$ zone axis parallel to the electron beam confirming the observed texture for the Al fabricated using UAM

Figure 6: Shows the STEM micrograph in the (a) STEM mode showing that it is not possible to delineate the interface (b) HAADF shows the differential contrast arising from the presence of Ga atoms from the sample preparation stage. The structure of the grain boundaries are similar to those reported during severe plastic deformation of Al alloys

Figure 7: Shows the approximate location of the interface (a) Shows that it is not possible to delineate the interface clearly. The differential contrast in the HAADF image is due to the presence of Ga ions (b) and (c) Shows the EDS maps of Oxygen and Magnesium respectively. The map shows that oxygen and Mg maps show only noise. However the Si clusters formed at the interface. It is not clear if this is a result of precipitation or if the Si was present in the form of clusters in the start

Figure 8: Shows EBSD of the interface in (a) As received interface (b) Tested in the ‘Z’ direction (c) Tested along the ‘Y’ direction. For both b and c the interface region is marked with black arrows inside the. The arrow indicates the direction of loading (d) KAM distributions of the (i) as received (ii) Loaded along the Z axis (iii) Loaded along the Y axis. Note that there is no major change in the KAM distribution indicating that there is no deformation at the interface during tensile loading

Figure 9: Shows the strain distribution in the sample during interrupted tensile testing. The regions marked M are in the fully loaded condition and U is in the fully unloaded condition. The load vs. displacement data also shows difference in slope while loading indicative of the absence of elastic behavior in the samples loaded along the ‘Z’ direction. The DIC data shows that a maximum peak plastic strain of 0.12 exists locally though the overall strain to which the sample was loaded is 0.06. After unloading it is also seen that the plastic strain does not relax completely and there is significant detectable peak present. The region marked with the white arrows in the

DIC image indicated the areas where maximum strain is present and the region where the microstructure analysis was performed.

Figure 10: Shows the post deformation microstructure analysis at the region where maximum strain concentration had accumulated. The optical micrographs (a-c) show extensive cracking at the interfaces. Regions where cracks originate don't correspond to the voids. (d) EBSD shows that there is no change in KAM distribution when the cracked regions are compared to the un-cracked regions.

Figure 11: Shows the schematic of fracture mechanism in UAM builds (a) Shows the interface (marked with the blue dashed box) with micro voids indicative of the onset of plastic instability (marked with the filled circles) (b) Sample loaded parallel to the direction of build accumulation (c) On loading the micro voids start to grow in parallel to the interface since adjacent material can strain harden and voids cannot grow into the bulk (d) Voids finally coalesce to form cracks (e) Finally fracture occurs without any significant macroscopic ductility.

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Defect free
build fabricated

Low ductility suggests poor
bonding between layers

However characterization shows presence of a bond

Anisotropy in properties arise due to the
coalescence of shear bands present in the builds

Mechanical testing using digital image correlation used to
identify and characterize regions with strain localization

Graphical Abstract