QUANTITATIVE EVALUATION OF CRYSTALLOGRAPHIC TEXTURE IN ALUMINUM ALLOY BUILDS FABRICATED BY VERY HIGH POWER ULTRASONIC ADDITIVE MANUFACTURING

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Abstract

Very high power ultrasonic additive manufacturing (VHPUAM) has shown good bond quality over traditional ultrasonic consolidation processes. However, the stability of microstructure in bulk and interface regions is unknown. Our earlier research showed a large difference in grain growth kinetics between bulk and interface regions. Therefore, we have performed in-situ studies of crystallographic texture evolution using a neutron beam line, before, during, and after heat treatment at 343°C for 2 hours. Shear texture in the as-received condition was found to be stronger with higher vibration amplitudes. We also observed rapid reduction of rolling textures in the initial material and presence of shear textures even after heat treatment.

Introduction

Ultrasonic additive manufacturing (UAM) or ultrasonic consolidation (UC) is a new solid state joining process used to make a solid part (also referred as build) by repeatedly adding thin metal foils (0.1-0.2 mm thickness) on top of each other [1]. The first layer of metal foil is added and joined on top of a solid block or substrate. This process is similar to ultrasonic seam welding. Then the foil is pushed down with a certain level of normal force by a roller-shaped sonotrode. The sonotrode has rough surface features to provide substantial friction and prevent the slippage between foil and sonotrode surfaces during the application of lateral ultrasonic vibration. During each joining cycle, the sonotrode rolls over the top surface of the added foil while pressing and vibrating at the ultrasonic frequency. As a result, the oxide layers (which are always present on aluminum surfaces) are broken and the nascent metal-to-metal surface contacts are formed creating a solid-state bond between foil and the substrate [1,2]. After joining, a new layer of metal foil is added on top of the previous layer and the process is repeated until the part reaches its final dimension. The UAM process also has the ability to machine the part while joining and stacking each layer to create channels or contour surface as well as the ability to add different material or to embed wires, sensors, or fibers. This ability allows one to create a composite or advanced material for sensing and structural applications. With UAM, the manufacturing cost that is associated with other conventional manufacturing processes can be potentially reduced [1,3].

Previous researchers have investigated the correlations between process parameter ranges and the bond quality of the UAM parts [4-9]. By varying different processing parameters

(normal force, vibration amplitude, and weld speed), and determining the bond quality in terms of linear weld density (the ratio of bonded area over the entire interface) or bond strength, they have come up with ranges of optimal process windows for different materials such as Aluminum 3003 and 6061 alloys [4-9]. Their results showed an increase in linear weld density and bond strength with higher normal force, higher vibration amplitude, and lower welding speed [7,8]. However, the welding speed cannot be too low as it can lower the rate of productivity as well as cause localized melting or sticking of the foil material on the sonotrode surface. Such effects can result in down time of the equipment because the sonotrode needs to be cleaned, and in some cases resurfaced [1]. Difficulties were also observed when attempting to bond harder metals or alloys together in addition to softer face centered cubic material such as aluminum or copper [11]. As a result, the upgraded version of UAM machine called VHPUAM (Very High Power Ultrasonic Additive Manufacturing) was created to increase the amount of ultrasonic power capable of producing larger normal force and larger vibration amplitude. VHPUAM can be used to bond thicker foil as well as harder or advanced materials together in which UAM could not achieve good bonding [11-12].

While the focus of previous UAM researches was to maximize the linear weld density and achieving flat and smooth bonded interface, there is insufficient evidence to claim that the bond strength significantly increase as a result of higher linear weld density [8,9]. Therefore, understanding the bonding mechanism at the microstructure level at the bonded interface locations is also important subject of research. However, we also do not know the effect of parameters on the microstructure and properties across the interfaces and in the bulk from top to bottom of the UAM or VHPUAM parts. The first subgrain refinement at the interface of Al 3003-O UAM part was reported by Johnson [13] because of severe plastic deformation. This refinement is attributed to the interaction between the surface texture of the sonotrode and the foil surface. His results also show little changes in the bulk microstructure of the foil after UAM process indicating little interaction of UAM process on the bulk properties of bonded foil layer [13]. Similar results were reported in Aluminum 6061 O and Al 3003 H18 where some local interface locations show very small and fine recrystallized grains as compared to original large equiaxed grains in Al 6061 O and thin and elongated grains in Al 3003 H18 [14-21]. The interface microstructure of VHPUAM material was first reported by Sriraman et al. [11] in Copper C11000 build. In addition to finer grains at the interface region showing some shear and flowing phenomena as a result of ultrasonic vibration during VHPUAM process, the bulk microstructure was found to be softer than the original copper foil. Soft bulk microstructure in VHPUAM of Aluminum 3003 H18 builds, made with larger vibration amplitude, was also found by Sojiphan et al [18]. This softening in the bulk of VHPUAM foil has never been reported in UAM material. Instead, Schick et al [9,17] reported harder microstructure in Al 3003 H18 UAM build as compared to original Al 3003 H18 foil.

The interface and bulk microstructure at different locations from the top to bottom of the VHPUAM build was first reported by Fuji et al [19]. The build was made with 26 μm vibration amplitude, 5.6 kN normal force, and 35.6 mm/s weld speed and composed of eight layers of Al 3003 H18 foil [19]. Their results show very similar fine equiaxed grains with shear texture at all selected interface locations from the top, middle, and bottom regions of the build [19]. A similar microstructure was also found in two-layer VHPUAM build made with the same processing parameters [19]. They also reported little changes in the bulk microstructure of VHPUAM build

as demonstrated by the presence of elongated grains with rolling texture components similar to that of original Al 3003 H18 foil [19].

Post processing heat treatment at 343°C for 2 hours was first performed by Schick with the goal to minimize the void fraction presented along the interface of Al 3003 H18 UAM builds [17]. Although the result did show slight improvement on the linear weld density, the microstructure, after heat treatment, showed a certain significant phenomena of stable very fine grains at the interface as opposed to the very large grains present in the bulk [17]. It was not clear why the grains at the interface did not grow and remained small (less than a few microns in diameter). In contrast, the grains in the bulk of the foil grew to 20 µm or larger in the bulk. It was speculated that dispersed oxides and intermetallic particles may be pinning the grain boundary motion close to the interface regions but this hypotheses has not yet been tested [17]. A similar study on VHPUAM of Al 3003-H18 was also performed and showed similar results of very fine grains at the interface and large grains in the bulk [18].

The above results clearly show that there is a difference of recrystallization and grain growth phenomena in the bulk and at the interface regions. While the details of these measurements have been confirmed [22] with detailed electron backscattered diffraction (EBSD) analyses, the current paper focused on confirming these phenomena using neutron diffraction technique. The goal of this research is to identify whether crystallographic textures (i.e. the shear texture in the interface region and the retained rolling texture in the bulk region) observed in EBSD results are the true representation of the bulk of VHPUAM [18,19,22]. To investigate the kinetics of the overall microstructure and texture evolution, the in-situ texture experiment with the sample at T=343°C was also performed with a time resolution of 5 minutes during the 2 hours heat treatment.

Experimental

Commercial 3003 aluminum alloy (Al-1 Mn-0.7 Fe-0.12 Cu, wt.%) tapes of 0.15mm thick and 25.4mm wide in H18 temper condition were selected for the current work. Build consisting of 80 layers were made using the Very High Power Ultrasonic Additive Manufacturing (VHPUAM) system in EWI [12]. Each build was made on top of an Al6061 baseplate, and made of up to 80 layers height and 20 cm in length. Three different sets of processing parameters including vibration amplitude and normal force are listed in Table 1 with constant levels of 35.6 mm/s speed and 20 kHz frequency. The vibration amplitude is listed in terms of the percentage of the maximum vibration amplitude produced by the new VHPUAM system. It is important to note that only 66 layers instead of 80 layers can be made in Build A where the vibration amplitude is lower. Above 66 layers, the foil could not be bonded to the previous layer and can be hand-peeled off easily from the build.

Table 1: VHPUAM processing parameters used to fabricate three VHPUAM builds

Sample	Vibration amplitude (% max amp)	Normal Force (kN)
Build A	60%	5.34
Build B	75%	5.34
Build C	75%	4.00

Optical images of as-processed VHPUAM builds were obtained using an Olympus-GX51 Optical Microscope. Hardness measurements were done on each mounted sample using Leco AMH-43 microhardness machine. Vicker's indents with 10 g load were made around the middle location of each foil layer of VHPUAM builds to create hardness maps.

Two rectangular samples per one build with cross section of 10.25mmx10.25mm and 9.6mm thick (from the top surface of the build) were machined using wire EDM process. These three samples were referred as Build A, Build B, and Build C. One sample from each build was then machined into thinner rectangular pieces with the same cross section and 4mm thickness from the bottom surface of the machined sample. The first sample of each build was referred as the "whole build" whereas the smaller second sample was referred as "bottom layers". The original aluminum 3003-H18 tapes were also EDM machined into 10.25mm wide and 50mm long pieces and stacked together using bolts and nuts through the machined holes at 10mm away from each side of the length as reference samples.

In-situ neutron diffraction was performed on both original tapes and VHPUAM samples using the high-pressure preferred orientation (HIPPO) at the Los Alamos National Laboratory [23,24]. The bulk measurement of texture or crystal orientations of overall grains in VHPUAM samples can be obtained as opposed to local textures previously obtained from electron backscatter diffraction (EBSD). Thus, the texture data obtained from neutron diffraction includes both texture from the interfaces and the bulk regions in VHPUAM builds. The samples were oriented in the instrument such that the 10mm diameter neutron beam impinges the sample on the RD-TD planes and penetrates through the sample thickness (ND). The sample was clamped into a vanadium strip sample holder and inserted vertically downward into the high temperature vacuum furnace, which can be installed into the large sample chamber. Diffraction data were acquired from 45 detector panels arranged on five detector rings of nominal angles $2\theta = 40^o$, 60°, 90°, 120°, and 150°. Three different rotations of 45°, -22.5^o, and -45° were used to increase the pole figure coverage. A Rietveld refinement using the MAUD program [25] was performed on the diffraction patterns using E-WIMV method with a 10° resolution to simultaneously fit all diffraction data and obtain the pole figures and orientation distribution function (ODF) for each sample [26,27]. The *mtex* program [28,29] was used to correct the alignment and plot the pole figures based on the MAUD data of each sample analyzed in this study.

Results and Discussion

Al3003-H18 VHPUAM Builds and Microhardness Measurement

The cross sections of the three Al3003-H18 VHPUAM builds are shown in Figure 1. The optical image results show that larger void- size distribution or non-bonded regions are present in the layers near the top of the build especially in Build A made using lower vibration amplitude. The void fractions also decrease significantly in Build A in the bottom layers as shown in Figure 1 indicating better bonding occurs in the bottom layers as compared to the top layers in the same build. This is probably due to the accumulative thermal-mechanical cycles during the VHPUAM process. In other words, more passes have been run over the bottom layers while adding and bonding top foil layers. Similar results of high bonded ratio are also found in Build B and Build C in which higher vibration amplitude (75%) is used except the size and the fractions of voids

are much lower. Better bond is also found when larger normal force is used in Build B (5.34kN) as compared to Build C (4kN). The hardness maps on the three VHPUAM builds are presented in Figure 1 below with the color scale ranging from 40VHN to 80VHN. The original Al3003-H18 hardness is around 70VHN [18].
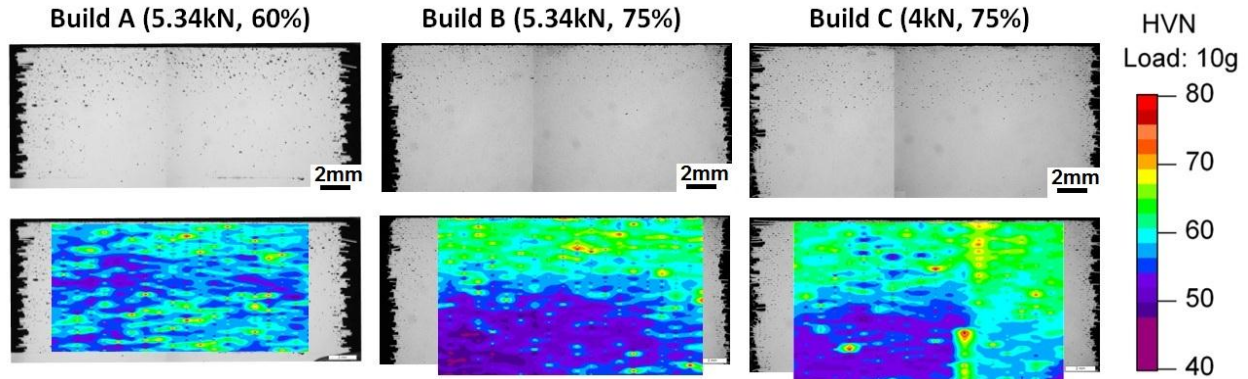

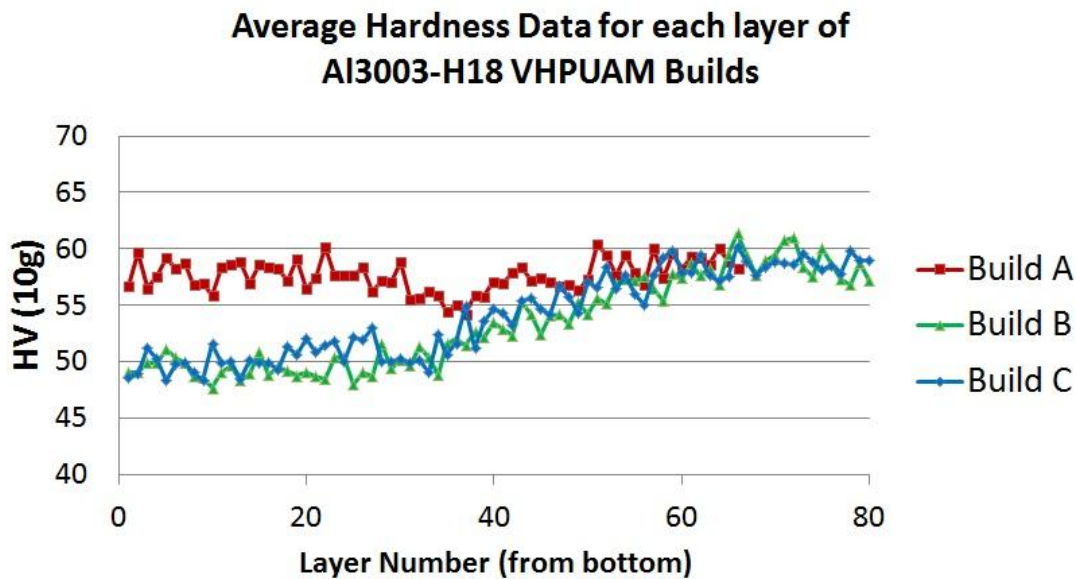**Figure 1: Optical Micrographs and hardness maps of three VHPUAM builds.**

Figure 2: Average hardness distribution of each layer in VHPUAM builds

The average values of hardness for each layer starting from bottom to top are also plotted in Figure 2. It can be seen that the hardness map and the average hardness values in Build A is in the range of 55-60VHN and are consistent from the bottom of the build to the top of the build. In Build B and Build C where higher vibration amplitude is used (75%), the softening in the bulk of VHPUAM builds are much more significant in the bottom layers of the build, i.e. the first 30-40 layers where the average hardness is around 50VHN as compared to 55VHN in the top layers of both builds. It is noted that there is not much difference between the hardness values as well as the void distributions in the higher vibration amplitude builds where the normal forces are different. Therefore, we can suggest that larger vibration amplitude has greater effect on softening of microstructure in VHPUAM builds and improve the area of bonded area. As the build height increases, it might be necessary to adjust the process parameters such as increase

normal force or reduce speed in order to improve bond quality in the top of the build as well as maintain the bulk properties such as hardness to be uniform across the entire VHPUAM build.

<u>Crystallographic Textures</u>

The crystallographic texture results are displayed as 111 pole figures (with the ND direction in the center). By comparing the 111 pole figure of the Al3003-H18 foil as a reference, we can study and evaluate the changes in crystallographic orientations due to in-process deformation and heating as well as post-processing heat treatment. The typical 111 pole figures of rolled fcc metals are shown in Figure 3. The color markers on the schematic diagram indicate the major crystallographic texture components of rolled fcc metals including copper, brass, Goss, and S3 components in the 111 pole figures. The {111}<110> shear texture components which was found in the interface region of VHPUAM build was also located in the middle of 111 pole figure in the same schematic diagram [19]. The 111 pole figures of original Al3003-H18 foils obtained before and after heating at 343°C for 2 hours are shown in Figure 3. The crystallographic texture result of original Al3003-H18 foils shows the strong crystallographic texture components typical for fcc rolling textures (copper, brass, S3) as expected. During in-situ heating, the crystallographic textures rapidly change and show stronger recrystallized cube texture. This inferred by the reduction in the intensity of retained rolling texture and slight increase in Goss texture component. As expected, in absence of any shear deformation, the pole figures of the foil material do not show shear texture components.
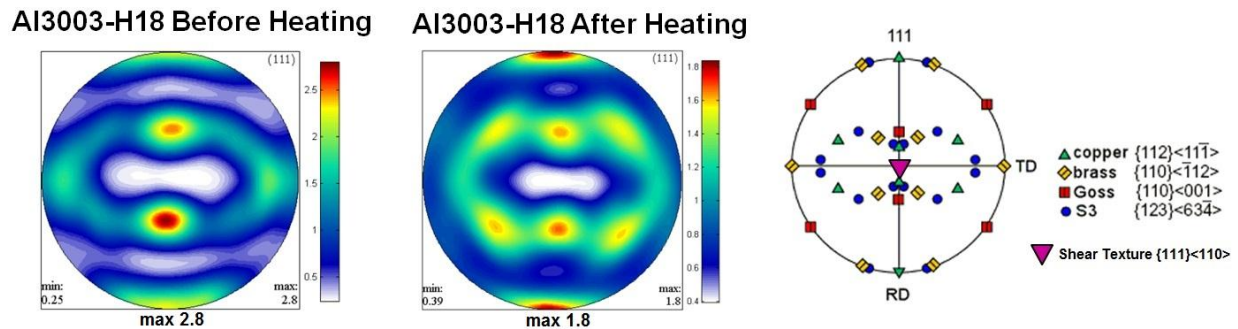

Figure 3: 111 pole figures obtained from neutron diffraction of Al3003-H18 foil before and after heat treatment at 343°C for 2 hours.

The crystallographic textures of VHPUAM builds before and after heat treatment are displayed in Figure 4. The results show the crystallographic textures of the whole build and bottom layers[1]. In all cases, the textures exhibit components of the rolling process found also in the unprocessed original foils as well as the {111}<110> shear texture components. The shear deformation reorients the {111} plane normal parallel to the RD-TD planes and the <110> is parallel to the RD vector [19, 22]. From our previous result [18, 19, 22], it can be concluded that these shear texture components belong to the material within the interface region mainly below the bonded interfaces within the bulk VHPUAM builds. Below the interface region, i.e. within the bulk region of each layer, little changes in crystallographic textures take place, retaining the rolling texture for most part. However, the quantitative evaluation of thermal effect and/or

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[1] The data from bottom layers of Build B have not been measured yet and thus not shown in this work.

thermomechanical effects on the crystallographic textures components requires further research [22].
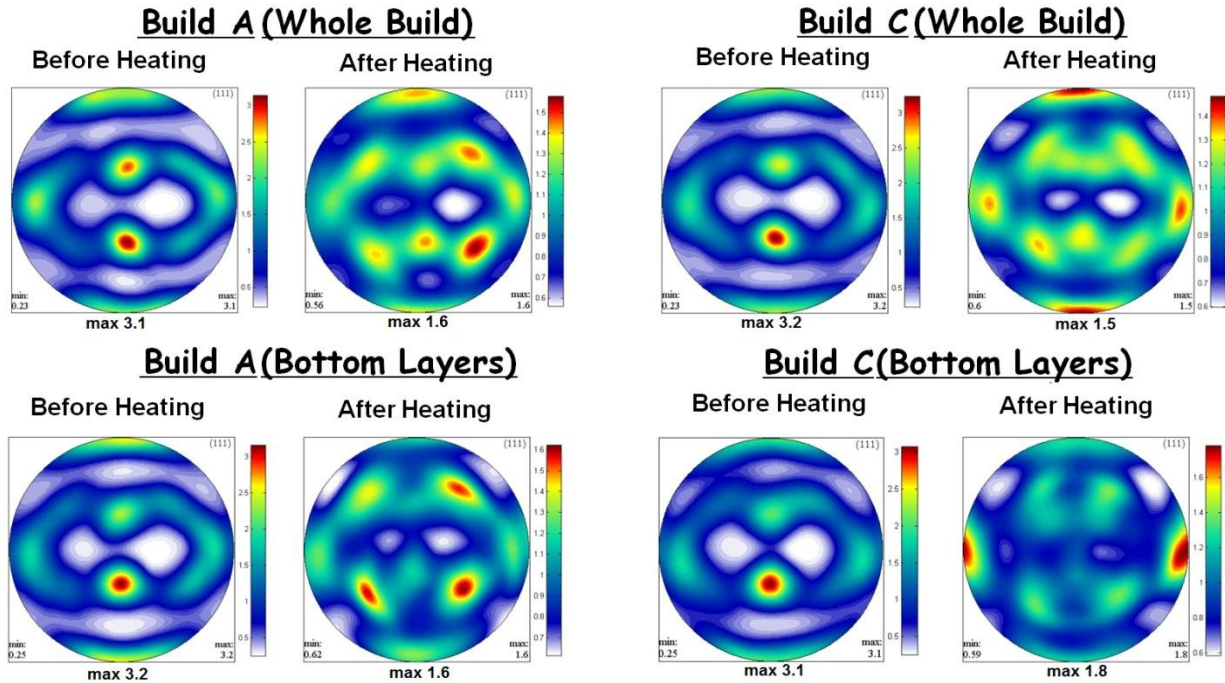

Figure 4: 111 pole figures obtained from neutron diffraction of Al3003-H18 VHPUAM builds (whole build and bottom layers) before and after heat treatment at 343°C for 2 hours.

While comparing the results of the bottom layers to the whole build, we observe that the {111}<110> shear texture components are stronger in the bottom layers sample than in the whole build samples. These results are also consistent in both Build A and Build C where higher intensity of textures are present and shifted towards this shear texture components although the intensity is much less than the overall rolling texture components. These results support the microhardness results. In other words, the bottom layers could undergo excessive plastic deformation during the layer-by-layer building operation. The deformation conditions may include shear within the interface regions resulting in localized adiabatic heating [20]. This localized heating may trigger softening in VHPUAM builds, especially when higher vibration amplitude (75%) is used. In addition, the crystallographic textures after heat treatment also confirmed the retention of {111}<110> shear texture components, indicating that grains affected by shear deformation within the interface regions are relatively stable and resist recrystallization and grain growth. The more in-depth quantitative analysis and discussion of possible mechanisms of this result will be the subject of our future research [22].

Summary and Conclusions

We present preliminary bulk crystallographic textures evaluation obtained from neutron diffraction as a function of processing parameters, location in the sample, and heat treatment. The microhardness measurement was also used to assess the change and gradient in microhardness of Al3003-H18 layers as compared to original foil. The major findings are summarized as follows:

- VHPUAM induces large thermo-mechanical cycles on the Al3003-H18 layers resulting in softening of microstructure in all three builds. It is interesting to note that at lower vibration amplitude (60%), we cannot fabricate the build more than 66 layers and the microhardness is relatively uniform ~55-60VHN in the bulk of each layer. At larger vibration amplitude (75%), the bottom 30-40 layers are much softer ~50VHN and the hardness increase to the same 55-60VHN in the top 20-30 layers of the build. Within these 20-30 layers of the builds, the bond quality is lower and voids are scattered along the interfaces.
- The crystallographic textures of the entire VHPUAM builds contain contributions from rolling texture and the {111}<110> shear texture components. These shear texture are attributed to the grains within the interface regions below the bonded interface in which shear deformation occurs during VHPUAM process.
- The shear texture component is larger in the bottom layers of the build and appears to increase when larger vibration amplitude is used. This suggests that repeated thermomechanical cycles by VHPUAM processes result in more shear deformation in the interface region and softening in microstructure in the bulk of each layer.
- The shear texture component remains relatively strong even after heat treatment in all the samples similar to before heat treatment whereas the rolling texture components undergo larger changes during heating. This confirms our EBSD results that the interfacial grains composed of the shear texture components are relatively stable as compared to the grains in the bulk regions. During heat treatment, relatively less recrystallization and grain growth take place within the interface region whereas larger degree of recovery, recrystallization, and grain growth was observed in the bulk regions of Al3003-H18 UAM and VHPUAM builds.

Acknowledgements

References

[1] White DR. Adv Mater Process 2003;161:64.
[2] Yang Y, Ram GDJ, Stucker BE. J Mater Process Technol 2009;209:4915.
[3] Li D, Soar RC. J Eng Mater Techol 2009;131: 021016-1.
[4] Kong CY, Soar RC, Dickens PM. J Mater Process Techol 2004;146:181.
[5] Kong CY, Soar SR, Dickens PM. Composite Struct 2004;66:421
[6] Ram GDJ, Yang Y, Stucker BE. J Manufac Sys 2006;25:221.
[7] Kulakov M, Rack HJ. J Eng Mater Technol 2009;131:021006-1.
[8] Hopkins CD, Walcott PJ, Dapino MJ, Troug AG, Babu SS, Fernandez SA. J Eng Mater Technol 2012;134:011004-1.
[9] Schick DE, Hahnlen RM, Dehoff R, Collins P, Babu SS, Dapino MJ, et al. Weld J 2010;89:105s.
[10] Zhang C, Li L. Metall Mater Trans B 2009;40B:196.

[11] Sriraman MR, Babu SS, Short M. Scripta Mater 2010;62:560.

[12] Graff KF, Short M, Norfolk M. Proc. SFF Symp 2010;82.

[13] Johnson K. Ph.D. Dissertation. Loughborough University; 2008.

[14] Li D, Soar RC. Mater Sci Eng A 2008;498:421.

[15] Mariani E, Ghassemieh E. Acta Mater 2010;58:2503.

[16] Dehoff RR, Babu SS. Acta Mater 2010;58:4305.

[17] Schick DE M.S. Thesis. The Ohio State University; 2009.

[18] Sojiphan K, Sriraman MR, Babu SS. Proc SFF Symp 2010;362.

[19] Fujii HT, Sriraman MR, Babu SS. Metall Mater Trans 2011;42A:4045.

[20] Sriraman MR, Gonser M, Fujii H, Babu SS, Bloss M. J Mater Process Techol 2011;211:1650.

[21] Sriraman MR, Fujii H, Gonser M, Babu SS, Short M. Proc. SFF Symp 2010;372.

[22] Sojiphan K, Babu SS. The Ohio State University. Unpublished research. 2012.

[23] Wenk HR, Lutterotti L, Vogel SC. Nucl. Instr. Methods Phys. Res., Sect. A 2003;515:575.

[24] Vogel SC, Hartig C, Lutterotti L, Von Dreele RB, Wenk HR, Williams DJ. Powder Diffr 2004;19:65.

[25] Lutterotti L, Matthies S, Wenk HR, Schultz AS, Richardson JW. J Appl Phys 1997;81:594.

[26] Matthies S, Pehl J, Wenk HR, Lutterotti L, Vogel SC. J Appl Crystallogr 2005;38:462.

[27] Wenk HR, Lutterotti L, Vogel SC. Powder Diffr 2010;25:283.

[28] Hielscher R, Schaenben H. J Appl Crystallogr 2008;41:1024.

[29] Bachmann F, Hielscher R, Schaeben H. Solid State Phenomena 2010;160:63.