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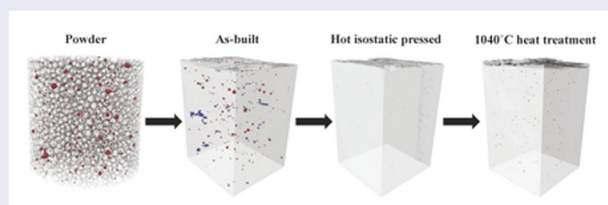
Analyzing the effects of powder and post-processing on porosity and properties of electron beam melted Ti-6Al-4V

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ABSTRACT

Metal additive manufacturing techniques such as the powder-bed systems are developing as a novel method for producing complex components. This study uses synchrotron-based X-ray microtomography to investigate porosity in electron beam melted Ti-6Al-4V in the as-built and post-processed state for two different powders. The presence of gas porosity in the starting powder was shown to correlate to porosity in the as-built components. This porosity was observed to shrink after a hot isostatic press treatment, but grow following a subsequent heat treatment. Crystal plasticity simulations were used to observe the effects of various observed pore sizes on mechanical behavior under loading.



IMPACT STATEMENT

This work combined μ SXCT and crystal plasticity simulations to characterize porosity through each processing stage of EBM Ti-6Al-4V. Gas porosity was detected in the as-HIPed state, and growth was observed after subsequent heat treatment.

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KEYWORDS

Additive manufacturing; electron beam melting; defects; X-ray tomography; crystal plasticity

1. Introduction

Additive Manufacturing (AM) is a rapidly developing field of technologies with the goal of manufacturing functional, near-net-shape metallic parts from a three-dimensional computer model by the layer-wise addition of material [1]. These technologies have generated considerable interest for their potential benefits of enabling advanced part geometries, and cost savings through reduced material waste and time to market. Powder-bed processes are one such class of technology, which operate by sequentially spreading powder layers of controlled thickness across a build area, where the appropriate cross-section is selectively melted via an electron beam (EBM) or laser beam (SLM). These processes are popular due to their high resolution and dimensional

tolerance, and the availability of automated commercial machines capable of producing functional parts with reduced operator involvement.

However, the replacement of a structural component using AM is significantly more complex than simply having the 3-D geometry of the component; variability in performance as a function of materials and AM processes also poses a major challenge. Even if the required form and fit tolerances are met, minor variations in the AM process can produce undesirable mechanical properties that are unsuitable for safety critical components. For EBM Ti-6Al-4V, defects deriving from the starting materials, process control, and insufficient post-processing can significantly impact structural integrity and durability and may result in a component

that appears satisfactory, but in fact does not meet the rigorous requirement for safety-critical applications [2–6]. More troublesome is quantifying the uncertainty associated with these properties and accurately calculating the resulting risks. A better understanding of the effects of powder properties, processing parameters, and post-processing on defect population and properties will help define post-processing and inspection requirements that will help in managing these risks. For EBM Ti-6Al-4V, efforts by the authors and others have investigated the effects of processing parameters on defect properties [3,4,7]. Building upon this body of work, this study uses synchrotron-based X-ray microtomography to investigate the comparatively unexplored area of powder and post-processing on defect properties and population. Additionally, based on the results of the defect characterization, crystal plasticity simulations were employed to determine the role of the defects on the mechanical behavior upon loading. Past studies, including [8,9], have analyzed the role of pores within a crystal plasticity framework, but these studies do not directly address the defects produced by AM within the associated microstructures. Here, fast Fourier transform (FFT)-based crystal plasticity modeling was used to calculate full-field solutions for stress and strain and to quantify the perturbing effect of voids.

Previous work, including extensive research in the related fields of welding and powder metallurgy, has identified a number of possible defect formation mechanisms that can contribute to a final defect population in AM components. These can be loosely categorized into processing defects and raw material defects. The cause of processing defects are generally simplified into two extremes; an extremely low energy input results in insufficient melting or ‘lack of fusion’ porosity, whereas an extremely high energy input causes ‘keyholing’ [2,10,11]. Lack of fusion, as the name suggests, is the result of incomplete welding of layers or adjacent melt pools, resulting in an irregular morphology [3,12]. Keyholing is a welding term used to describe the deep, narrow vapor depression that forms under high-energy density melting conditions due to the vaporization of metal under the heat source [13]. This is usually accompanied by keyhole porosity, believed to form when fluctuations in the vapor cavity lead to its collapse and the subsequent entrainment of vapor in the melt pool in the form of a bubble [14]. King et al. observed the transition from conduction to keyhole mode in laser powder-bed AM, with the associated porosity, for single track experiments in 304 stainless steel [11]. However, there is limited work published on keyhole pore formation in the EBM process [2].

A frequently cited source of raw materials-related defects in AM metals is ‘trapped gas’ porosity, commonly

believed to be a result of inert gas (argon) trapped in the starting powder during its manufacturing process, which is transferred to the as-built part. Studies with direct metal laser sintering systems such as the Laser Engineered Net Shaping process have shown a correlation between powder and part porosity, but there is little information on the effects of varying powders with EBM systems [15]. In addition to trapped inert gas, the presence of absorbed soluble gases, particularly hydrogen, cannot be ignored as these have been linked to porosity in electron beam welds of Ti-6Al-4V [16–18]. The relative lack of research in the effects of powder is likely influenced by the fact that the only commercially available EBM system comes with process parameter sets designed to be used with machine-manufacturer-supplied powder. However, the interest in using alternate sources of powder is rapidly increasing, thus encouraging the identification of the characteristics of powders that meet the requirements of the EBM process.

The quantitative characterization of these different types of defects found in AM components requires significant information on spatial distribution and morphology that is difficult to achieve through traditional means of cross-section analysis or bulk density (Archimedeon density, ultrasonic inspection) [2,19]. X-ray micro-tomography (μ XCT) has been used effectively in characterizing defects in welds [14,20], and it has increasingly become the tool of choice for investigating defects in AM materials [3,21–24]. Recent work has utilized synchrotron-based micro-tomography (μ SXCT), as it offers another level of detail, as the high-energy X-rays provide large penetration, a high signal to noise ratio, as well as the possibility of sub-micron voxel resolution [4,11,12,25,26]. Previous work by some of the authors has shown that this technique is capable of resolving defects in EBM Ti-6Al-4V below the resolution limit of lab-scale μ XCT ($\sim 5\text{--}10\text{ }\mu\text{m}$) [4]. Lastly, the non-destructive nature of these techniques also allows re-scanning after or during additional processing of the same sample, as was done in this study.

The goal of this study was to utilize these unique capabilities of μ SXCT to further investigate the influence of powder properties and post-processing on defects in EBM Ti-6Al-4V. While eliminating or reducing the powder porosity through better raw material selection has obvious benefits, the presence of processing flaws such as lack of fusion porosity demand that all parts destined for service undergo hot isostatic pressing (HIP). Previous work with μ XCT by Tammis-Williams et al. has shown that both gas porosity and lack of fusion porosity are eliminated within the resolution limit of lab-scale μ XCT ($\sim 5\text{ }\mu\text{m}$) [5,27]. However, the microstructural coarsening of the α -phase resulting from the high-temperature

HIP process lowers the yield strength. Therefore, if the goal is to retain the fine microstructure characteristic of as-built AM parts, then one must also perform a high-temperature β -solution heat treatment followed by a quench [6]. Tammás-Williams et al. showed that exposure to the high temperatures required for β -solution treatment resulted in regrowth of the pores, a problem known to the powder metallurgy community as thermal-induced porosity, but one that has seen limited discussion within the AM community [27]. However, to date no μ XCT observation of pores in the as-HIPed condition has been reported within the resolution limits of lab-scale instruments. Therefore, this study aims to utilize the higher resolution of μ SXCT to determine if pores are observable in the HIPed condition, and measure their re-grow in a subsequent β -solution heat treatment.

2. Experimental procedure

To examine the contribution of powder porosity to the final defect population at different stages of post-processing in EBM Ti-6Al-4V, μ SXCT was performed on the powder, an as-built sample, and a HIPed sample for two different powders. The HIP treatment was performed at 900°C for 2 h at 103 MPa in argon, in accordance with ASTM F 2924. To investigate potential regrowth, the HIPed sample underwent a β -solution heat treatment of 1050°C for 10 min after initial imaging, and was rescanned at the same locations. Test blocks were fabricated using plasma-atomized Arcam Ti6Al4V

ELI powder from AP&C and plasma rotating electrode process (PREP) powder from TIMET. Compositions provided in the powder quality certifications are given in Table 1. PREP powder is generally accepted as having lower internal porosity compared to other powder-manufacturing methods, as it does not employ an atomization gas, but it is fabricated in an inert gas environment [28]. Often different processing parameters are required when changing powders to accommodate necessary changes in layer thickness, but this has been shown to affect both the gas and lack of fusion porosity present in the sample [3,4]. To limit this effect, the powders used in this study had approximately the same size range, albeit with different size distributions. Particle size distributions measured via SEM images taken on a Joel 5900 and Genesis Particle analysis software are given in Figure 1. Additionally, the same processing parameters and version of machine were used for both; in this case, the recommended parameters for Ti-6Al-4V in an ARCAM A2X (see Table 2) were used. Synchrotron imaging samples were sectioned via EDM from the bulk specimens with 1 mm $\times$ 1 mm $\times$ 30 mm dimensions, with the long axis parallel to the build direction. Samples of the powders were sealed at tap density inside 1.5 mm Kapton tubing for μ SXCT.

Synchrotron X-ray microtomography was performed at the 2-BM beamline at the Advanced Photon Source at Argonne National Lab operating in white beam mode. A total of 1500 projections were taken over 180° with a 100 ms exposure time resulting in approximately a

Table 1. Compositions of virgin Arcam Ti6Al4V ELI and TIMET PREP powder.

Powder	Element (Wt.%)														
	Ti	Al	V	Fe	Y	C	O	N	H	Cu	Mn	Mo	Sn	Zr	Other
AP&C	Bal.	6.41	3.89	0.18	< 0.001	0.01	0.08	0.02	0.001	–	–	–	–	–	< 0.4
TIMET	Bal.	6.09	3.98	0.17	–	0.02	0.18	< 0.01	0.002	< 0.01	< 0.01	0.01	< 0.01	< 0.01	< 0.05

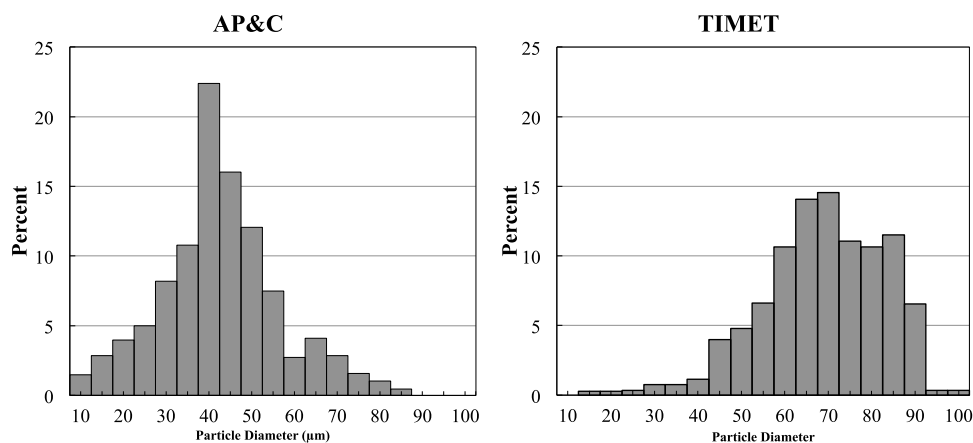


Figure 1. Number-weighted size distributions of powders used in this study measured via SEM image analysis and Genesis Particles software.

Table 2. Arcam A2X processing parameters for Ti-6Al-4V. Ranges cover automated process adjustments that occur during the build.

PreHeat			Focus offset	125 ± 75 mA
			Heating focus offset	250 ± 150 mA
PreHeat	PreHeat 1		Offset to part	5 mm
			Max beam current	30 ± 10 mA
			Beam speed	11,000 ± 3000 mm/s
			Max No. repetition	39 ± 4
			Ave current	14.8 ± 4 mA
	PreHeat 2		Max beam current	38 mA max
			Beam speed	11,000 ± 3000 mm/s
			Max No. repetition	21 ± 4
			Ave current	16.8 ± 4 mA
			Max heat time	28 ± 7 sec
Melt	Contours = 3	Outer contour	# Spots	50
			Spot time	0.8 ms
			Multispot overlap	0.5 mm
			Current	5 mA
		Inner contour	Focus offset	Nominal post calibration ± 10 mA
			Speed function	6
			Current	12 mA
			Focus offset	0
	Hatch		Speed function	30
			Current	17 mA
			Focus offset	Nominal post calibration ± 10 mA
			Speed function	36
			Line order	1
			Line offset	0.2 mm
			Max heat time	25 s
Heating				

4 min scan time. A 0.65 μm voxel size (edge length) was obtained. The radiographs were reconstructed and filtered using TomoPy 0.0.3 [29]. Avizo version 9 was used for segmentation and analysis. A minimum of eight face-connected voxels were used as the minimum feature size, establishing a minimum feature size of approximately 1.5 μm .

3. Results and discussion

The combined μSXCT results of the samples at different stages of post-processing are shown in Figure 2. Relevant statistics are summarized in Table 2. The plasma-atomized powder (Figure 2(a)) has a significantly larger population and maximum size porosity than the PREP powder (Figure 2(d)), and is comparable to previously reported μXCT results from plasma-atomized powder [3]. In the PREP powder, numerous pores were observed that had an irregular morphology inconsistent with the highly spherical gas porosity frequently observed in the as-built condition. While more investigation is required to determine the nature of this porosity, the irregular shape suggests it may be shrinkage porosity, and therefore would not contribute to gas porosity in the as-built parts. For these reasons, only porosity within the powder with sufficiently spherical morphology to be realistically considered gas porosity (anisotropy < 0.5) was compared to the spherical porosity observed in the fabricated samples. However, highly spherical gas pores were still detected in the PREP powder, in contrast to the widely

held belief that PREP powder is free of gas porosity. It should be noted that, in contrast to the trapped argon, the gas pores in the PREP would likely contain helium, which is the gas used in most PREP powder-manufacturing processes.

μSXCT results of the as-built parts are shown in Figure 2(b) and 2(f). Multiple μXCT scans were taken throughout the height of the samples, with no significant variation in porosity observed along the build direction, so the scans from the top ~ 1.5 mm were used to compare the samples. As with the powder porosity, pores were segregated by morphology, with low sphericity pores attributed to lack of fusion defects. Results show a similar trend in porosity to that of the powder, with the AP&C part having considerably more defects with morphologies consistent with that of gas porosity. Comparison of the gas pores in the AP&C powder with the as-built part (Figure 3) shows a similar distribution but lower number density and volume fraction, consistent with previous results and the current assumption that some pores are eliminated during processing, either through combination in the melt pool and/or escape [3]. While the limited number of pores detected in the PREP powder and parts makes it difficult to conclude that transfer is occurring from the powder to the parts, the size is comparable, and the presence of spherical pores in the as-built PREP samples suggests that some level of gas porosity does exist in the powder, albeit at a significantly reduced size and number density compared to the plasma-atomized powder.

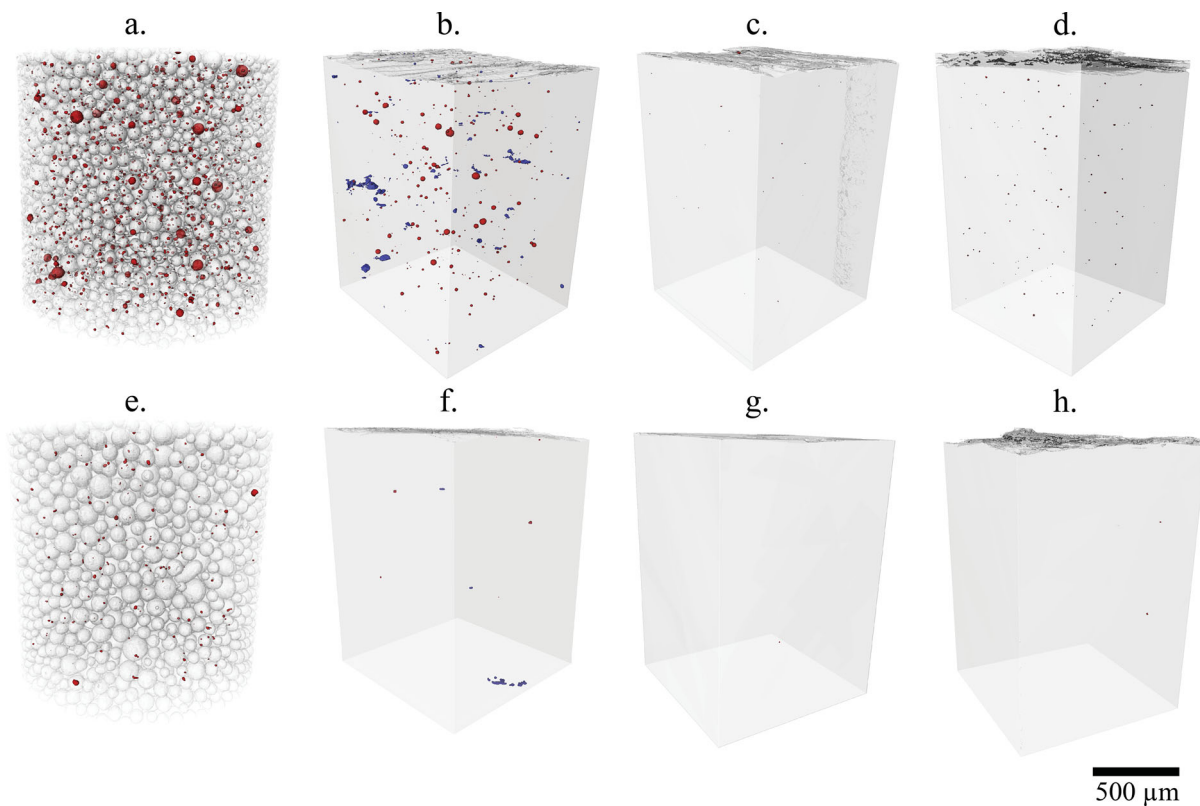


Figure 2. μ SXCT results showing porosity in powder and samples built from AP&C (a–d), and TIMET (e–h). From left to right, the sequence of samples is powder (a and e), as-built (b and f), HIPed (c and g), and HIPed + β solution heat treatment (d and h).

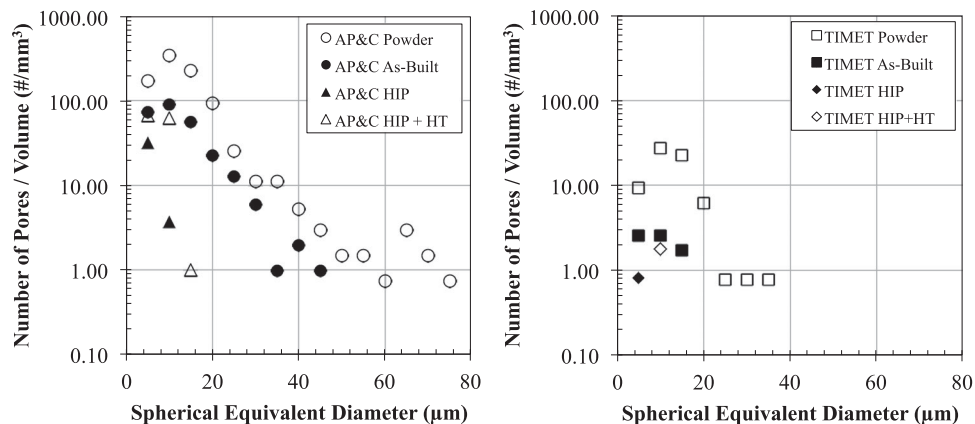


Figure 3. Size distributions of gas pores detected by μ SXCT, expressed as number densities.

While this does not conclusively prove trapped inert gas in the powder is the source of the defects in these as-built parts, there is sufficient evidence to largely disqualify the other two likely defect formation mechanisms, namely keyholing and absorbed gas porosity. The dramatic difference in porosity between the two as-built specimens despite their same processing parameters suggests that keyholing is not a factor, as a similar effect would be expected in both. Gong et al. speculated that

the control software present in Arcam systems, which acts to stabilize the melt pool geometry, makes it difficult to achieve the conditions required for keyholing [2]. Similarly, the gas analysis of the powder (Table 1) shows that with the exception of nitrogen, the TIMET powder had higher concentrations of absorbed gas, including hydrogen [16–18], making it unlikely that absorbed gases were the cause of the increased gas porosity seen in the AP&C samples. While these should not be discounted as

possible defect formation mechanisms in the EBM process, in this case the inert gas porosity appears to be the dominant factor.

Another notable result is that the builds made with AP&C powder display a significantly higher number density (Table 3) of lack of fusion pores. This is a counterintuitive result because the expectation is that, with a similar size range for the two powders combined with the identical processing parameters and machine, a similar degree of melting should occur. This then suggests that other powder properties such as variations in the size distribution, associated packing density and spreadability may also significantly affect the frequency of lack of fusion defects.

Results from the HIPed and post-HIP β -solution treatment samples are shown in Figure 2(c,d,g,h). Previous work using lab-scale μ XCT with feature resolution of $\sim 5\ \mu\text{m}$ did not detect any porosity in the HIPed condition, although cross-section analysis suggested the presence of incompletely closed pores [5,30]. Given the assumption that the gas porosity originates from insoluble gas transferred from the powder, it is reasonable to expect the HIP process to shrink but not entirely eliminate pores. To the authors' knowledge, the higher resolution of the synchrotron μ SXCT, with feature resolution of approximately $1.5\ \mu\text{m}$, has for the first time detected pores in the HIPed condition using μ XCT, all of which are at or below the resolution limit of the previously used instruments ($\sim 5.5\ \mu\text{m}$). While this was to be expected with the plasma-atomized powder samples, pores were also detected in the PREP samples, which provides evidence for the presence of inert gas porosity originating from the PREP powder. In both cases, however, the lack of fusion porosity, which is assumed to be truly void because of the vacuum environment in the machine, was eliminated with the HIP process within the resolution of this experiment.

This continued existence of this porosity after HIPing raises a concern for post-HIP heat treatments or high-temperature operating conditions because of the potential for pore regrowth. This process has been utilized

to good effect in the manufacture of titanium foams, and Tammas et al. showed evidence from μ XCT of pore regrowth with EBM Ti-6Al-4V after a β -solution treatment; however, they were not able to observe pre-existing pores in the HIPed condition [31,32]. Figure 4 shows that pre-existing pores detected in the HIPed samples coarsen after heat treatment. In both cases, the number density of pores in the post-HIP heat-treated samples displayed a similar, albeit smaller number of gas pores compared to the as-built sample. This is to be expected, as pores near the resolution limit in the as-built state would likely not regrow to within the resolution limit after the HIP/heat treat cycle, resulting in fewer pores detected in the final state. The fact that these pores are not eliminated during the HIP process, and their subsequent regrowth, provides further evidence that the gas pores observed in the as-built state are a result of inert gas transferred from the powder. These results suggest that powder with low porosity such as PREP would be the ideal choice for parts that must be subjected to a high temperature post-HIP heat treatment, or are exposed to high operating temperatures, as there would be less risk of re-growth of pores after HIPing.

To investigate the deleterious role of porosity on the mechanical properties of the material, crystal plasticity simulations were performed on microstructures with the maximum pore size obtained from samples characterized in the as-built and HIPed conditions. Samples of the Ti-6Al-4V material were mechanically sectioned, polished to a mirror finish, and characterized using electron backscatter diffraction (EBSD). The microstructure for the as-built and HIPed Ti-6Al-4V samples were statistically similar with an average grain size of the β phase of $290\ \text{nm}$ and β area fraction of 13%. To isolate the role of porosity on the mechanical behavior of the local microstructure, the same base microstructure was used in the analysis of the as-built and HIPed conditions. To enable a large field of view, a step size was set to $0.43\ \mu\text{m}$ in the EBSD scan, thus only the α phase was considered in this analysis. The data were cleaned by Dream.3D by removing any voxel with a confidence

Table 3. Porosity statistics from μ XCT analysis of samples in Figure 2.

	Lack of fusion				Gas porosity			
	Volume fraction ($\times 10^4$)	# Pores per mm^3	Ave. Eq. diam.	Max Eq. diameter	Volume fraction ($\times 10^4$)	# Pores per mm^3	Ave. Eq. diam.	Max Eq. diameter
AP&C powder	n/a	n/a	n/a	n/a	23.5	892	11	72
AP&C as-built	2.97	187	7.81	49	3.78	263	10	42
AP&C HIP	n/a	n/a	n/a	n/a	7.65×10^{-3}	35	3	6
AP&C HIP + HT	n/a	n/a	n/a	n/a	1.36×10^{-1}	131	5	10
TIMET powder	n/a	n/a	n/a	n/a	7.15×10^{-1}	68	10	33
TIMET as-built	1.70×10^{-1}	12	9	28	2.27×10^{-2}	7	7	13
TIMET HIP	n/a	n/a	n/a	n/a	5.09×10^{-4}	1	5	5
TIMET HIP + HT	n/a	n/a	n/a	n/a	3.89×10^{-3}	2	7	8

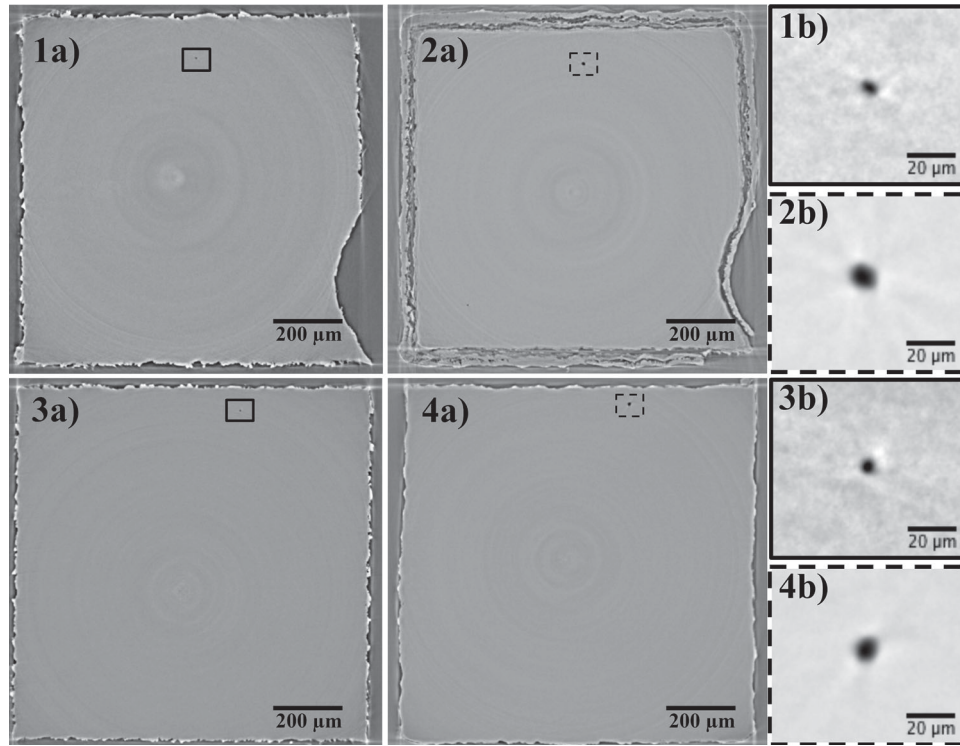


Figure 4. Reconstructed slices of the (1a) APC HIP, (2a) APC HIP + heat treatment, (3a) TIMET HIP and (4a) TIMET HIP + heat treatment, showing the same pore before and after solution heat treatment, with corresponding magnified images of the pores (1b–4b). Pore diameter grows approximately 200% from the HIPed condition after β -solution heat treatment.

index below 0.02 and/or an image quality below 50, thus essentially removing any points corresponding to the β phase. The material exhibited a basketweave or Widmanstätten microstructure with lathes of the α phase present. A threshold of 5° was used to consider neighboring voxels as part of the same grain and the α lathes were fit as ellipses via the software MTEX. The average major axes of the ellipses were measured at $6.135 \mu\text{m}$ and the minor axes at $2.466 \mu\text{m}$. Simulations were constructed using an elasto-viscoplastic formulation based on fast Fourier transforms (EVP-FFT) by Lebensohn et al. [33], based on the original formulation by Moulinec and Suquet [34]. It would be difficult to represent the small features of the β phase in the simulations; thus the simulations were homogenized to only represent the α phase. The crystal plasticity parameters were calibrated based on the macroscopic uni-axial stress-strain curves for dogbone samples from the as-built and HIPed conditions. Based on the Burgers Orientation Relationship [35], the slip systems of the β phase can be transformed into the hexagonal α phase and represented as part of the basal slip systems, thereby accounting for deformation in both the α and β phases. The anisotropic elastic constants for the α phase were obtained from [36]. A power-law-based flow rule is used [33], such that $\dot{\gamma}_0 = 0.001$ and

$n = 15$ [37]. A Voce hardening law is used to describe the hardening:

$$\tau(\Gamma) = \tau_0 + (\tau_1 + \theta_1 \Gamma)[1 - e^{-\Gamma\theta_0/\tau_1}],$$

where $\tau_0 = 75.2 \text{ MPa}$ and $\theta_0 = 3000$ are the initial yield stress and hardening rate, respectively, and $\tau_1 = 14.1 \text{ MPa}$ and $\theta_1 = 16$ are the parameters that describe the asymptotic behavior of the material (as calibrated based on the current experimental macroscopic stress-strain data, with constants shown pertaining to the prismatic $\langle a \rangle$ slip systems in the α phase). The ratios of the hardening parameters for the asymmetric family of slip systems were obtained from [37], such that: $\tau_a^{\text{prismatic}} = 1.14 * \tau_a^{\text{basal}}$ and $\tau_a^{\text{prismatic}} = 1.59 * \tau_{c+a}^{\text{pyramidal}}$. The similar EBSD scan representing a typical Ti-6Al-4V microstructure produced via EBM was used as input for the EVP-FFT simulations, in which additional material is added in the x -direction to provide continuity to transmit load, while in the y -direction a buffer zone with zero strength was used to break periodicity (but also ensure the strain fields produced by the imposed porosity did not interact across the periodic boundary conditions). For more details about the set-up, please refer to [38]. From the pore dataset (Figure 2(b,c),

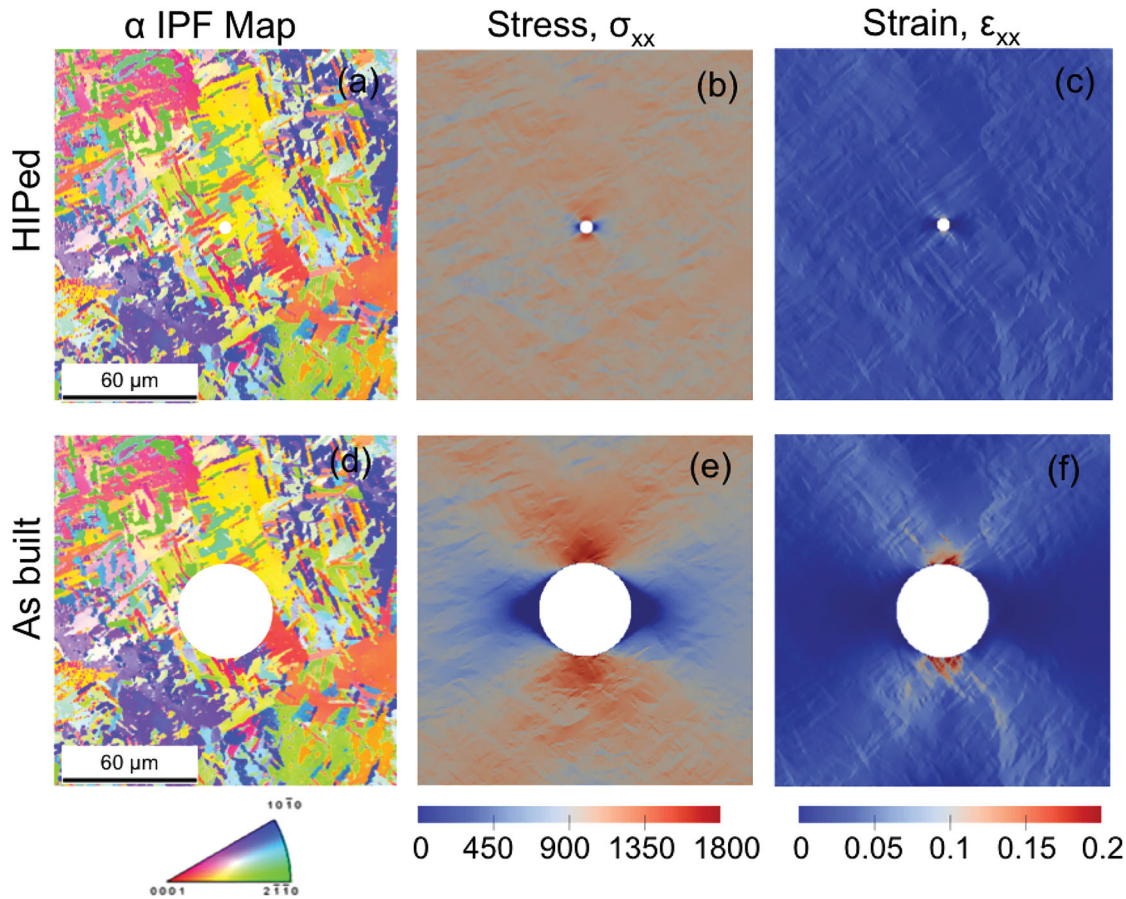


Figure 5. Crystal plasticity results for the HIPed and as-built condition, where the maximum void size is imposed on the microstructure from EBSD (original IPF maps for the α phase is shown). The simulations are loaded in the horizontal direction to 2.5% strain. The axial component of the stress and strain is displayed about the loading direction.

and Table 3), the maximum size for both HIPed (6 μm) and as-built (42 μm) materials was inserted into the EBSD scans as an added gas phase (i.e. void space) as an individual pore at the center of the scanned area.

Both the HIPed and non-HIPed simulations were uni-axially loaded to 2.5% strain, to enter the macroscopic plastic regime, in the $\times$ (horizontal) direction (Figure 5). As expected, the larger void leads to a greater extent of plasticity around the void. The results of the stress and strain contour plots are similar to the continuum theory. As discussed in Bataille et al. [9], a ratio for pore diameter to average grain size larger than unity resembles the homogeneous solution with local perturbations due to the local microstructure. In this case, for the HIPed and as-built conditions, the ratio of the maximum pore diameter to average grain size is approximately 1 and 7, respectively. For the HIPed case, the strain around the pore is similar in value to the deformation of the microstructural features away from the pore, whereas the stress concentration of the pore and the resulting deformation in the as-built case is

significantly greater than that caused by the microstructural features. This demonstrates that the largest pore size for the as-built case significantly increases the micromechanical fields near the pore and subsequently has a deleterious affect on the mechanical behavior of the material, thus indicating the benefits of the HIPing process, including its effectiveness in reducing the pore size and the resulting micromechanical fields. The crystal plasticity simulations demonstrate the relatively large stress concentrations around the maximum size pores measured by the μXCT technique. It should be noted that these pore sizes are below the resolution limit of many traditional CT methods, thus emphasizing the importance of the μXCT characterization within this study.

4. Conclusions

This study utilized non-destructive synchrotron-based X-ray microtomography to investigate the effects of

powder and post-processing on porosity in Arcam EBM Ti-6Al-4V. The main results of this investigation were as follows:

- (a) Lower spherical porosity in the powder resulted in a similarly reduced porosity in the as-built and post-HIP heat-treated conditions.
- (b) The size distribution of gas pores found in the parts in the as-built condition was similar to that of pores found in the powder used to make the parts, but at a reduced volume fraction.
- (c) The hot isostatic press treatment decreases the pore size and number density drastically such that the largest observed pore was approximately 5 μm , but did not entirely eliminate the gas porosity. Lack of fusion porosity was apparently eliminated within the μXCT resolution limit after HIP treatment.
- (d) A β -solution heat treatment of the HIPed material resulted in regrowth of the gas pores by a factor of about 200% from the HIPed state. Lack of fusion porosity was not observed to regrow.
- (e) Based on crystal plasticity simulations, the largest pore size in the sample characterized in the as-built condition exhibited a significant increase in the stress concentration near the pore, whereas the largest pore in the HIPed condition resulted in spatial variations in deformation similar to those caused by the microstructure itself.

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