Deformation Behavior of Additively Manufactured GP1 Stainless Steel

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Abstract.

In-situ neutron diffraction measurements were performed during heat-treating and uniaxial loading of additively manufactured (AM) GP1 material. Although the measured chemical composition of the GP1 powder falls within the composition specifications of 17-4 PH steel, a fully martensitic alloy in the wrought condition, the crystal structure of the as-built GP1 material is fully austenitic. Chemical analysis of the as-built material shows high oxygen and nitrogen content, which then significantly decreased after heat-treating in a vacuum furnace at 650 °C for one hour. Significant [austenite-to-martensite](#) phase transformation is observed during compressive and tensile loading of the as-built and heat-treated material with accompanied strengthening as martensite volume fraction increases. During loading, the initial average phase stress state in the martensite is hydrostatic compression independent of the loading direction. Preferred orientation transformation in austenite and applied load accommodation by variant selection in martensite are observed via measurements of the texture development.

Introduction.

Additive manufacturing (AM) is a rapidly developing processing pathway that produces components by selectively melting and solidifying feedstock to build a desired geometry, rather than subtractive machining from cast or wrought stock material [1,2]. A major difference between the two routes is the microstructure of the material in the final component [3]. Casting and thermo-mechanical processing to produce wrought stock material has been optimized over centuries and a wide variety of thermal and deformation processing options are commonly used to obtain a desired component microstructure, which defines performance. In contrast, there are more limited opportunities to optimize microstructures of AM materials after manufacture, primarily because components are built to near-net shape to minimize the amount of material and post-build machining needed. As a result, most bulk thermo-mechanical treatments, (e.g., rolling, forging, etc.) cannot be readily applied. Options for modifying AM component microstructures after manufacture are generally limited to surface deformation treatments (e.g., shot or laser peening) and bulk or surface thermal/chemical treatments [4]. AM processing parameters have perhaps the strongest influence on component microstructure, and significant research is being performed to establish both empirical connections between the input parameters and the achieved microstructure and properties, and physics-based models to guide AM process optimization [5].

The difference in microstructures between AM and wrought components can be significant, similar to the changes observed within welded microstructures. In some cases, welding can lead to introduction of additional phases, such as the appearance of ferrite in austenitic stainless steels [6], and welding processes often result in substantial residual stresses in and around the welded region [7] due to large thermal gradients and rapid cooling. These features are often present in AM materials as well [8,9]. The

GP1 powder meets the chemical specifications of martensitic precipitation hardenable 17-4 PH stainless steel, but is specifically made for AM production. Wrought 17-4 PH is fully martensitic with a martensite finish temperature of 105 °C [10]. However, when used in AM production, 17-4 PH material has shown to contain a significant fraction of retained austenite [11]. Obviously, this leads to significantly different mechanical properties as the retained austenite transforms to martensite during deformation via transformation induced plasticity (TRIP) [12], and thus can exhibit lower yield stress but higher ductility and toughness relative to wrought 17-4 PH [13].

In-situ neutron diffraction measurements have proven to be an invaluable tool to monitor the internal strain and microstructure development during deformation, providing quantitative grain level information that can be used to develop and validate constitutive models [14,15]. Diffraction inherently provides phase specific information, which is especially useful when phase transformations take place during the deformation [16,17]. The non-destructive nature of neutron diffraction, due to the bulk penetration of neutrons into most materials, allows for monitoring of the evolution of the microstructure through multiple steps, providing a unique window into the processing-structure-property-performance (PSPP) relationship of additively manufactured materials.

In the present work, we have applied in-situ neutron diffraction measurements to monitor the microstructure development of additively manufactured GP1 material starting from the powder (EOS), following the build process, during the vendor recommended “stress-relief” heat treatment and, finally, during deformation, touching on the processing, structure, and property relationship of the PSPP. The employed in-situ neutron diffraction measurements allow us to follow the phase development and phase stresses as well as texture in the material throughout [18].

Experimental.

Sample Preparation. The AM material was produced from EOS GP1 powder feed material [19] using a powder-bed, laser sintering technique, specifically employing an EOS M280 direct metal laser sintering (DMLS) machine operating in a nitrogen atmosphere. The chemical composition of the GP1 is within specification of the composition of the martensitic precipitation hardening 17-4 PH stainless steel [20,21]. Several specimens with conventional Charpy test specimen geometry were built from the GP1 powder on a single build plate as described in [8]. Compression and tension samples were electro-discharge machined (EDM) from two of the Charpy specimens to determine the macroscopic strength properties of the material. The compression specimens were cylinders 6.3mm in diameter by 15mm long following the ASTM E9 standard [22]. The 2.4:1 l:d aspect ratio of the compression samples is required in order to minimize shading of the incident and diffracted neutron beams during the in-situ measurements. MoS₂ grease was used as lubrication but a small amount of barreling was observed at the ends of the samples, whereas no significant buckling was observed. The neutron gauge volume is centered on the length of the sample minimizing the effect of the barreling on the diffraction results. The tension samples were threaded-end specimens following ASTM E8 [23] with a 3 mm diameter in the gauge section. The axes of the deformation samples were in the plane of the build plate and parallel to the long axis of the specimen from which they were removed. One of the compression cylinders was used for in-situ stress relief heat treatment.

Microscopy. The microstructure of the additively manufactured material was examined in five conditions (designations in parenthesis): as-built (A), as-built + compression (AC), as-built + tension (AT), as-built + heat-treated (AH), and as-built + heat-treated + compression (AHC). Compression indicates ~10% strain at room temperature, tension indicates ~12% strain at room temperature, and heat-treated indicates a

vacuum heat treatment for 1 hour at 650°C, based on the vendor recommended stress relief treatment [19] (Note, vendor does not specify vacuum or ambient.). The microstructures were evaluated using standard metallographic sampling and preparation techniques. The metallographic samples were mounted in epoxy with the polish planes radial to the cylinder axis, but with an unknown relation to the build direction as that was unfortunately not tracked during the machining process. Metallographic preparation was performed by grinding with successively finer SiC grinding papers using rotary grinding wheels from 60 grit to 800 grit papers. Ground samples were then polished through 6, 1, and 0.25µm diamond polishing paste on metallographic preparation cloths. Polished samples were etched using a 10% Oxalic acid solution.

Table 1: Chemical specifications of 17-4 PH [20,21], GP1 powder [19], and measured chemical compositions of as-built GP1 AM material and heat treated GP1 AM built material.

Material	Cr	Ni	Cu	Mn	Si	Nb + Ta*	Mo	P	S	O	N	C	Fe
17-4 PH, Spec.	15-17.5	3-5	3-5	<1	<1	0.15-0.45	-	<0.04	<0.03	-	-	<0.07	Bal.
GP1, Powder spec.	15-17.5	3-5	3-5	<1	<1	0.15-0.45	<0.5	-	-	-	-	<0.07	Bal.
As-built AM GP1, Measured, Sample A	15.56	4.61	4.19	0.61	0.80	0.28 (No Ta detected)	0.12	0.011	0.004	0.13	0.13	0.045	73.4
Heat-treated** AM GP1, Measured, Sample AH	15.62	4.62	4.18	0.61	0.82	0.28 (No Ta detected)	0.12	0.011	0.003	0.022	0.020	0.046	73.6

A '-' signifies that a range is not given in the specs for this element.

* For GP1 only Nb content is prescribed, but for 17-4 PH the sum of Nb and Ta content is prescribed.

** 1 hour at 650°C in vacuum.

Neutron Diffraction. The in-situ neutron diffraction measurements were performed using the Spectrometer for Materials Research at Temperature and Stress (SMARTS) instrument at Los Alamos Neutron Science Center (LANSCE) [24]. A detailed description of the instrument has been given elsewhere [24], and here only the salient details of the experimental setup will be described. The in-situ stress relief measurements were carried out in a custom built high temperature and high vacuum furnace [24] that employs tungsten mesh heating elements symmetrically placed above and below the sample position as the heating source. The sample was positioned on a ceramic platform well insulated from the chassis of the furnace as described in [25]. The furnace temperature was controlled by a sheathed thermocouple in the furnace hot zone and the sample temperature monitored by a thermocouple in mechanical contact with the end of the cylindrical sample. The sample was heated from room temperature to 650°C at an average rate of roughly 12°C/minute. Vacuum was maintained at 10⁻⁶ Torr for the duration of the thermal cycle.

For the in-situ loading measurements, the horizontal Instron load frame was oriented with the loading axis at 45 degrees to the pulsed (20Hz) "white" incident neutron beam. The incident beam was defined by motorized BN slits such that only the gauge section of the samples was illuminated, 10 mm wide and 4 mm tall during tension, 4 mm wide and 4 mm tall during compression. Bragg's law is satisfied by effectively scanning the wavelength (energy) of the neutrons through each pulse using the time-of-flight (TOF) technique. Diffracted neutrons were collected in two fixed detector banks placed at ±90° relative to the incident neutron beam, resulting in two diffraction patterns with scattering vectors along the loading and transverse axes [24]. Lattice strains were determined along these two directions; parallel and transverse to the applied load. The load on the sample was recorded by at 250kN maximum load

cell and strain in the gage section determined by an Instron 2620-603 extensometer with a gauge length of 10 mm. The deformation test was completed in displacement control with the test interrupted at defined intervals of crosshead displacement for nominally 10 minutes to allow the collection of satisfactory diffraction patterns.

Neutron diffraction texture measurements were performed using the HIPPO instrument at LANSCE [26]. HIPPO employs multiple detector banks at various scattering angles and a series of sample rotations to obtain excellent orientation space coverage, and taking advantage of the TOF technique it provides full Orientation Distribution Function (ODF) measurements in minutes for most engineering materials [27].

Data Analysis. Lattice parameters were determined using Rietveld refinement [28] of the full diffraction patterns using the GSAS software [29] and the SMARTSware batch refinement routines [30]. The lattice strains are calculated from the change in lattice parameter as

$$\varepsilon = \frac{a - a_0}{a_0}. \quad (1)$$

The austenite phase is fit using the FCC (face centered cubic) crystal structure and the martensite is fit using the BCC (body centered cubic) crystal structure as the small tetragonality cannot be resolved with the inherent resolution of the SMARTS instrument 90° banks. The lattice strains determined from the Rietveld refinement of the entire pattern have been shown to accurately represent the macroscopic elastic strain (type I strain) in the sample [31]. The macroscopic stress in the constituent phases in the sample can then be calculated using the bulk isotropic Young's modulus (E) and Poisson's ratio (ν) according to Hooke's law as

$$\sigma_i = \frac{E}{(1 + \nu)(1 - 2\nu)} [(1 - \nu)\varepsilon_i + \nu(\varepsilon_j + \varepsilon_k)] \quad (2)$$

For uniaxial loading and weakly textured samples it is appropriate to assume that the two transverse strains are identical, and thus equation (2) can be reduced to the following expressions for the longitudinal and transverse stresses:

$$\sigma_L = \frac{E}{(1 + \nu)(1 - 2\nu)} [(1 - \nu)\varepsilon_L + 2\nu\varepsilon_T] \text{ and } \sigma_T = \frac{E}{(1 + \nu)(1 - 2\nu)} [\varepsilon_T + \nu\varepsilon_L] \quad (3)$$

The phase specific Young's modulus and Poisson's ratio used in the calculations are given in Table 2. The martensite properties were taken from wrought 17-4 PH as it is fully martensitic, and the austenite properties were taken for 304 stainless steel as its main alloying components (Cr and Ni) are relatively close to the composition of GP1 and 17-4 PH.

Table 2: Elastic constants used for austenite and martensite in the stress calculations.

Material	Young's modulus (E , GPa)	Poisson's ratio (ν)
Austenite (SS304) [32]	193	0.29
Martensite (17-4 PH) [33]	197	0.272

For a general stress state, it is necessary to measure at least 6 elastic strain components in order to fully determine the stress tensor, but in the case of uniaxial loading we have employed the assumption of

The reference lattice spacing of the austenite phase was taken to be the observed value under nominal holding stress ($\sim 10\text{MPa}$). As the martensite phase is not present initially in the material, a measure of the stress free lattice parameter (a_0) for the phase strain calculations is not immediately apparent. In the present work, we have chosen to determine the reference lattice parameter of the martensite as that which makes the phase fraction weighted summed phase stresses equal to the measured macroscopic stress at the highest applied load (where the phase fraction of the martensite is the highest). Typically this condition is used as a validity check of the measurements, but it cannot be applied in this case as only austenite is present from the start.

Results.

a) b) c) d)

Optical metallography of samples in the as-built, heat-treated, compressed and heat-treated + compressed conditions are shown in Figure 1. Very little difference is observed in the overall microstructure, with evidence of the additive building process observed in all photomicrographs.

profile. A set point for the control thermocouple of 660 °C was used, and a thermocouple in physical contact with the sample read 624 °C during the 1 hour hold time. While the average heating rate of the sample was ~10 °C/min, below 300 °C, the control thermo-couple oscillates about the set ramp temperature because of the poor thermal coupling of the radiative heating elements at low temperature. As evidenced by the evolving lattice parameter, the sample temperature lags behind the control thermocouple, again until roughly 300 °C, then follows smoothly until the target temperature of 650 °C.

Figure 4: Lattice strain variation and temperature profile during the in-situ stress relief procedure (Sample AH).

During the initial heating from room temperature to a control temperature of 660 °C, the austenite lattice parameter increased by $\sim 12070 \pm 50 \mu\epsilon$ (microstrain). From the known CTE of 304L steel [39], this corresponds to a sample temperature of 665 ± 3 °C. Clearly, while the temperature of the furnace was held constant at 660 °C, the lattice parameter of the austenite continuously decreased by roughly 500 $\mu\epsilon$ over the duration of the hold at temperature (more than 10 times the uncertainty). Moreover, upon returning to room temperature a permanent change of roughly -500 $\mu\epsilon$ is observed in the lattice parameter consistent with the evolution while at temperature. No significant change in diffracted peak width nor peak intensity was observed during the hold at temperature. Finally, no evidence of the martensitic phase was observed during or following the heat-treatment.

Figures 5a-c show the observed macroscopic stress/strain curves collected during in-situ uniaxial compression and tension measurements performed on the AM GP1 samples. Specifically, Figures 5a and b show the compressive flow curves of the as-built and heat-treated samples, respectively, while Figure 5c shows the tensile flow curve of the as-built sample. The dips in the flow curves indicate relaxation taking place during the nominally 10 minute long hold time for the diffraction measurements, but as the time average value is close to the bottom of the relaxation parts, it is evident that most of the relaxation takes place very early on in the hold time and that the conditions are nearly constant over most of the hold time.

Figure 5: Measured macroscopic stress-strain curves: a) Sample AC, b) Sample AHC, and c) Sample AT. The blue line is the extensometer data, red dots are the time average value of true stress and strain during the diffraction measurements. Green dashed line with square symbols are the martensite phase weight fraction from the Rietveld refinement of the diffraction data.

The flow curves are sigmoidal in nature with an initial yield stress of the AM GP1 material of nearly 600 MPa, followed by a plateau in the flow stress (zero hardening) to 6% strain in the as-built material, 4% strain in the heat-treated material. Beyond this point, significant hardening is observed until the completion of the tests at 10-12% strain. Though not shown, each of the samples unload in a linear elastic manner.

Figures 5a-c also show the observed martensitic volume fractions during the deformation tests. During compression of both the as-built and heat-treated samples, the deformation induced phase transformation initiates at 2-2.5% strain and the martensitic fraction increases monotonically to 0.4 and

0.7, respectively at 10% strain. In contrast, when loaded in tension the phase transformation starts much earlier, at roughly 0.5% strain, and reaches a martensitic fraction of 0.9 at 12% strain.

Figures 6a-c show the applied stress vs. lattice strain curves during compression of the a) as-built and b) heat-treated material as well as during c) tension of the as-built material. In each case, the austenite loads linearly with a slope of 190 ± 5 GPa until yield, at which point the martensite phase appears. In all cases to varying degrees, the martensite forms with a significantly reduced lattice strain relative to that in the austenite. Indeed, in the case of compression of the as-built material, the martensite forms reverse loaded, with positive (negative) strain in the longitudinal (transverse) direction despite the -550 MPa applied longitudinal stress. The other two cases are less extreme, but in each case the incipient martensite forms in a relaxed strain state relative to the applied tractions.

a) b) c)

Figure 6: Measured applied stress versus lattice strain plots for; a) Sample AC, b) Sample AHC and c) Sample AT.

Significant relaxation of the elastic strain in the austenite phase is observed with the inception of the deformation induced phase transformation during the stress plateau observed in the macroscopic flow curves, see Figure 5. Concomitantly, the martensite plays the role of a reinforcing phase accumulating large increments of elastic strain, which is supporting load shed from the relaxing austenite phase. It is important to recall that the austenitic and martensitic volume fractions are decreasing and increasing, respectively, in particular during this portion of the deformation test. Eventually, as the macroscopic hardening again increases (see figure 4), the elastic strain begins to accumulate once more in the austenite (the phase hardens), but the martensite continues to accumulate strain at a faster rate throughout the tests

Figure 7 highlight the texture evolution along the loading direction observed for the austenite and martensite during deformation tests. The first column shows an initial inverse pole figure of the austenite, the second and third columns show final inverse pole figure of the austenite and martensite, respectively.

a) b) c)
d) e) f)
g) h) i)

Figure 7: Inverse pole figures for the AC, AHC and AT samples; Austenite before deformation, and Austenite and Martensite post deformation.

From the IPF's, it is clear that the texture development in the austenite phase for all three samples, regardless of the sense of deformation, results in a stronger 111 orientation along the straining direction. The strength of the 111 orientation scales with the degree of transformation, i.e. the strongest and weakest 111 texture component is observed following tension and compression of the as-built materials, respectively. The martensite, on the other hand, shows one distinct deformation texture

following compression, i.e. a strong increase of the 100 orientation, and another in tension; nearly random with the highest point close to the 110 orientation.

Discussion.

The properties of AM GP-1 steel do not match that of 17-4 PH despite being within the chemical specifications. Moreover, the mechanical properties of the AM GP1 evolve following the vendor recommended stress-relief heat-treatment, resulting in significantly different mechanical properties than the as-built material, but still not matching those of wrought 17-4 PH. The in-situ neutron diffraction results, combined with metallurgical studies provide a window into the microstructural evolutions that result in the observed mechanical properties.

It has been suggested in the literature that parts produced using 17-4 PH powder via AM experience large solidification undercooling due to the rapid quench rates inherent to the AM process [40]. This would lead to a significant fraction of metastable austenite in the microstructure [41]. However, this seems unlikely in light of the fact that the austenite is stable through the long heat treatment and slow cooling carried out in this work. Nitrogen has been shown to be a strong austenite stabilizer in stainless steels [42] which, given the high nitrogen content of the as-built material, can explain the fully austenitic structure for a composition that otherwise would be martensitic. While the nitrogen content of the powder is not published, given that it is 98% austenite, it is likely that significant nitrogen contamination is present in the GP1 powder as it is processed under a nitrogen atmosphere [19].

The optical metallography shows that the layered microstructure created by AM persists through the proscribed heat-treatment. No decrease in dislocation density or evolution of texture is observed during the heat treatment, indicating that neither recovery nor recrystallization has occurred [43,44]. However, the permanent change in lattice parameter observed during the in-situ heat treatment is a strong indication of a chemical change during the thermal cycle. The chemical characterization pre- and post-heat treatment confirms that both the oxygen and the nitrogen content have decreased during the stress-relief treatment. It is likely that mostly interstitial oxygen and nitrogen has been removed from the material as only a few of the nitrides, e.g. Fe_2N , can be dissolved at the relatively low temperature of the heat-treatment [45]. It is important to note that the heat treatment was done under high vacuum ($\sim 10^{-6}$ Torr), while the vendor suggestion does not cite atmosphere. While no phase transformation from [austenite-to-martensite](#) was observed during the heat treatment, it is clear that the reduction of dissolved nitrogen in the AM material reduces the stabilization of the austenitic phase. During subsequent compressive deformation, the stress-induced phase transformation occurs at an increased rate in the heat-treated material relative to the as-built.

The neutron diffraction also elucidates the details of the stress induced phase transformation. The transformation initiates at lower strains when loaded in tension than when loaded in compression; roughly 0.4% strain compared to 2.5% strain. This is consistent with the predictions of the finite element based elasto-plastic phase field model with grain boundary effect developed by Malik et al. [46], where the phase transformation was predicted to start earlier in tension than in compression. It was also predicted that deformation induced transformation would asymptotically approach limits of about 45 vol.% in compression and about 90 vol.% in tension, again consistent with the levels of transformation observed in the as-built material in this work.

The transformation from [austenite-to-martensite](#) is associated with a Bain strain that includes a uniform dilatation of 4.7% due to the lower density of the martensite [47], as well as a shear strain of 0.26 [48].

These size and shape changes, respectively, will result in significant grain scale stresses on the incipient martensite, as well as reaction stresses on the constraining austenite. Figures 8a-c show the austenite and martensite phase stresses, calculated from the lattice strains, as a function of true macroscopic strain; the evolving martensite fraction and phase-weighted average stresses are also shown. Recall that the phase-weighted average longitudinal stress determined from the diffraction data is forced to be identical to the measured applied stress at the highest load as explained earlier, in order to determine the stress free lattice parameter of the martensite.

Figures 8a-c show the calculated phase and average stress vs. the true strain during compression of the a) as-built and b) heat-treated material as well as during c) tension of the as-built material. Throughout all three tests, the phase-weighted average transverse stress is close to zero and the longitudinal stresses are roughly equal to the macroscopic stress, giving an indication of the achieved accuracy of the phase stress determination and the validity of the assumptions. Moreover, early in the test when the material is single phase austenite, the austenite stress calculated from the lattice strains matches the macroscopic stress well. Furthermore, in each test, the initial stress state of the incipient martensite in the diffraction pattern (about 1% weight fraction detection limit) is approximately hydrostatic compression, even when the sample is loaded in tension. At higher applied strains the stress state in the martensite of the as-built compression and tension samples, Figures 8a) and c), approaches the expected uniaxial stress state, with a near zero transverse stress, but the martensite in the heat-treated and compressed sample, Figure 8b), shows significant tensile transverse phase stress at higher loads. In the as-built samples, the stress state in the austenite remains close to uniaxial with a slight hydrostatic component (the transverse phase stress has the same sign as the longitudinal phase stress). In the heat-treated sample the austenite phase stress has a significant compressive hydrostatic component above 5% strain, and the martensite shows a matching tensile hydrostatic stress component.

The deformation texture of the austenite is also controlled by the transformation. Tensile and compressive deformation of FCC metals results in distinct texture components. For example, tensile plastic deformation results in the slip direction, in this case the 111 direction, aligning with the straining direction while compressive deformation aligns the slip plane normal, 220, with the straining direction

[54]. Contrarily, in both cases of compression and tension of the AM GP1, a moderate 111 texture component develops in the austenite parallel to the straining direction, indicating that the texture development is significantly influenced by preferential transformation. Grains with their 200 plane normal parallel to the straining direction transform first, with 111 oriented grains showing more resistance to transformation and remaining untransformed to a large degree at the end of the tests, regardless of the loading sense.

The initial texture of the as-built GP1 material is rather atypical for AM materials with cubic crystal structure as it shows a strong 220 intensity along the build direction, whereas reported textures for other cubic materials typically shows a moderate 200 intensity along the build direction for the majority phase. The texture of relatively short (low build height) AM materials is typically governed by the almost one-dimensional heat flow into the base plate [37,38]. As discussed in [8], the present samples were built on a relatively open substructure that significantly decreased the direct heat flow into the base plate. This change in heat flow during the build may be the cause of the atypical texture observed. The initial texture can certainly play a role in the deformation behavior, but as the loading direction is not along the build direction with the strong 220 intensity, the effect will be a second order effect. It is also notable that there is no obvious orientation relationship between the austenite and the martensite as has been observed between austenite and ferrite in AM 304L stainless steel [55].

Conclusions.

In-situ neutron diffraction measurements were performed during heat-treating and deformation of material additively manufactured from GP1 powder [19]. The detailed phase specific information obtained from the diffraction data has enabled us to draw the following conclusions regarding the properties of the material:

1. The crystal structure of the GP1 powder is 98% austenitic even though the measured chemical composition of the AM GP1 falls within the specifications for the 17-4 PH steel, which in the wrought condition has a fully martensitic structure. Note that the Nitrogen content is not explicitly given in the 17-4 PH specification.
2. The as-built texture shows a strong 220 pole density along the build direction, and there is no obvious direct orientation relationship between the austenite and the martensite.
3. The vendor recommended stress relief heat-treatment of the AM material (630 °C/1hr) permanently changes the lattice parameter, indicating that chemical changes take place. The measured chemical composition prior to and following heat-treatment shows that the nitrogen and oxygen content is reduced by almost an order of magnitude after heat-treating, most likely caused by removal of interstitials within the austenite as the temperature is not high enough to dissolve nitrides and oxides.
4. No decrease of dislocation density or evolution of texture is observed during the heat-treatment indicating that neither recovery nor recrystallization has occurred.
5. No phase transformation is observed during the heat treatment. This indicates that the austenite is not a metastable structure caused by the rapid solidification, but stabilized by the high nitrogen content. Post-heat treatment nitrogen levels are still relatively high (200 ppm by weight), which is apparently sufficient to stabilize the austenite.
6. Significant phase transformation is observed during compressive and tensile loading of the as-built and heat treated material (compression only for the latter).

7. The degree of transformation is lower in compression than in tension, and the transformation process starts at earlier strains in tension. Both these findings validate model predictions presented in the literature [46].
8. The rate and final value of transformation is much higher in the heat treated material than in the as-built material suggesting that removing interstitial nitrogen and oxygen from the austenite facilitates the transformation process, that is destabilizes the austenite. The heat treatment was completed in a high vacuum furnace, and it is not known if the result would be different if the heat treatment were completed in air or a cover gas.
9. Texture development in the austenite and martensite phases suggest that there is preferred transformation depending on the orientation of the austenite grain relative to the loading axis as well as variant selection in the martensite due to the applied load.
10. The initial average phase stress state in the martensite is close to being hydrostatic compression rather than close to the overall stress state of the polycrystal as a whole (compressive or tensile). As phase transformation continues with increasing enforced deformation, the phase stress state becomes closer to the macroscopic uniaxial stress state.
11. In the heat-treated sample, there is a clear hydrostatic component to the phase stresses at high applied loads; compressive in the martensite and tensile in the austenite.

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a) Sample A

b) Sample AC

c) Sample AH

d) Sample AHC

Figure 2

Figure 3

GP1 in-situ Heat Treating

Figure 4

Figure 5

a)

b)

c)

Figure 6

Figure 7

a)

b)

c)

Figure 8