Strain rate sensitivity and mechanical anisotropy of selective laser melted 17-4 PH stainless steel

Tyler LEBRUN*, Kenichi TANIGAKI*, Keitaro HORIKAWA*, and Hidetoshi KOBAYASHI*

*Graduate School of Engineering Science, Osaka University
1-3 Machikaneyama-cho, Toyonaka-shi, Osaka 560-8531, Japan
E-mail: tyler.lebrun@impact.me.es.osaka-u.ac.jp

Received 5 June 2014

Abstract
Quasi-static and dynamic tensile testing was performed upon machined tensile specimens fabricated from bulk primitives produced by the consolidation of water atomized, 17-4 precipitation hardened stainless steel powder by layer-based, selective laser melting. Such mechanical evaluation was performed by a screw-driven uniaxial tension testing machine and a split-Hopkinson tensile bar apparatus. Strain rates evaluated include $10^{-3}$, $10^{-1}$ and $10^3$ $s^{-1}$. Prior to tensile testing, specimens underwent additional thermal processing in accordance with industry standards. Evaluations of the solution heat treatment and peak-age conditions were made alongside similarly prepared, but traditionally processed specimens meeting the same material standard for chemistry (drawn rod). Tensile strength across all strain rates is higher for these selective laser melted (SLM) specimens as a result of microstructure refinement through rapid melt, solidification and cooling during processing. Ultimate tensile and yield strengths increase with increasing strain rate and show no preferential direction relative to the building direction. Elongation anisotropy is observed as a consequence of directional porosity stemming from pores limited to within individual layers. Specimens loaded normal to the SLM building plane commonly rupture at small elongation because of this mechanical fibering from the selective laser melting process.

Keywords: Additive manufacturing, Selective laser melting, Split-Hopkinson bar, 17-4 PH, Stainless steel

1. Introduction

As a result of continued development by industry and academia, the various iterations of additive manufacturing (AM) in both consumer and industrial sectors have seen steady improvement, forwarding the progress of this burgeoning technology. From this ongoing work, AM implementation has expanded in limited cases beyond “rapid prototyping” and “rapid tooling” to now incorporate “rapid manufacturing”. Niche applications of AM have proven successful at economically producing low-quantity and customized, near-net shaped hardware (Allen, 2006). Additionally, AM is turned to in cases where geometries are otherwise impossible or cost-prohibitive to fabricate by traditional means. Recently, attention has been drawn to the ability of AM to produce functionally gradient or otherwise engineered microstructures within the material being processed. Much of the attention towards these tools is merited, but the state of this suite of materials processing methods is very much a work in progress.

Of the variety of AM technologies commercially available, this study is focused on the form known as selective laser melting (SLM). Interest into this kind of AM across a variety of industries is buoyed because of its ability to produce near net shapes and designs of high complexity with sufficient precision (Relvas, et al., 2012). Principally, this process consolidates a powdered working material into bulk form by way of layer-based, full-melt and solidification. Such a processing method requires the deconstruction of a three-dimensional computer generated model into two-dimensional slices of a predefined thickness (typically accomplished by an AM machine-paired, assistive software). Sequential reconstruction of the slices by the layer-based process recreates the full geometry. Surface curvature accuracy in the building direction without additional post-processing is as accurate as a piecewise
interpretation limited in resolution only by the layer thickness of each slice. Fabricating geometries by this additive method permits the inclusion of features that are otherwise impossible to machine in a subtractive way (drilling, milling, and etc.). Such examples include integral cooling channels and high complexity, non-extrudable thermal heat sinks (Simchi, Petzoldt, and Pohl, 2003). Additionally, as the fundamental understanding of these processes improve, engineered microstructural performance and properties become more controllable. Aspects such as phase presence, functionally gradient properties, and microstructural texture and direction – all tailor-made features potentially includable within a SLM near-net design – are possible.

This investigation is focused on the SLM processing of 17-4 PH stainless steel (AISI 630, JIS G4303 SUS630). Some interest into 17-4 PH by other researchers has brought to light a variety of resultant microstructural properties uniquely characteristic of this method of processing. The morphological dependence of resulting microstructure on processing atmosphere (Murr, et al., 2012), variations in phase presence of α and γ-iron (Starr, et al., 2012), and porosity prevalence across processing parameters (Gu, et al., 2008) are just some examples of these findings. 17-4 PH and other precipitation hardenable stainless steels are attractive for their flexibility at fine-tuning additional mechanical strength through the careful heat treatment and resultant growth of precipitates.

The aim of this study is to investigate the mechanical properties of bulk material produced by SLM via commercially available hardware. To make this evaluation, it was necessary to examine the resultant material irrespective of external, shape-driven defects (e.g. surface texture/finish, surface cracks, and external build faults resulting from any geometries fabricated). To achieve this, tensile specimens were machined from bulk material primitives. The effect of heat treatment on mechanical performance relative to a control material was also investigated so as to ascertain the effectiveness/sensitivity of additional thermal processing per industry standards. The resultant microstructure from the SLM process and its evolution following subsequent heat treatments are detailed below. Additionally, the characteristics of internal defects, their shape, size, and morphology are assessed and contextualized with respect to results from tensile testing.

2. Experimental procedure
2.1 Working material characterization

Pre-alloyed 17-4 PH stainless steel powder was supplied by Kobe Steel (Kobe City, Hyogo Prefecture, Japan). The powder used in this study was produced by “V-jet” water atomization from a molten stream. As-atomized powder was further processed by a mechanical sieve to reduce the upper-limit particle size. Particle chemistry was provided by the supplier and its comparison against the ASTM A564/A564M standard is shown in Table 1. Particle size distribution measured by the supplier using a laser-based particle size analyzer, Microtrac-HRA 9320-X100, is detailed in Table 2. Figure 1 shows an example image of the received powder as observed by scanning electron microscopy (SEM) against a carbon tape background. Particle morphology was of two distinctly different shapes: irregular and jagged particles commonly seen in water atomized stainless steels and the less common smooth, spheroids. Alloy composition of both particle shapes was measured to be the same by energy dispersive x-ray spectroscopy.

As a basis of comparison throughout the evaluation of the SLM material, additional specimens of 17-4 PH hot-drawn rod were fabricated. Hereafter these specimens are referred to as the “control material”. This material meets the same standard for chemistry and minimum properties as outlined by the same ASTM standard mentioned above. For completeness, its chemistry is provided in Table 1.

<table>
<thead>
<tr>
<th>Working Material</th>
<th>C</th>
<th>Mn</th>
<th>P</th>
<th>S</th>
<th>Si</th>
<th>Cr</th>
<th>Ni</th>
<th>Nb</th>
<th>Cu</th>
<th>Fe</th>
</tr>
</thead>
<tbody>
<tr>
<td>Control (Drawn Rod)</td>
<td>0.054</td>
<td>0.18</td>
<td>0.011</td>
<td>0.005</td>
<td>0.41</td>
<td>16.51</td>
<td>4.16</td>
<td>0.42</td>
<td>3.97</td>
<td>bal.</td>
</tr>
<tr>
<td>ASTM A564/A564M Limits</td>
<td>0.07</td>
<td>1.00</td>
<td>0.040</td>
<td>0.030</td>
<td>1.00</td>
<td>15.00-17.00</td>
<td>3.00-5.00</td>
<td>NS</td>
<td>3.00-5.00</td>
<td>bal.</td>
</tr>
</tbody>
</table>

Limits are in percent maximum unless shown as a range
NS – Not Specified
Table 2  17-4 PH powder size distribution

<table>
<thead>
<tr>
<th>Powder Diameter Range (µm)</th>
<th>&lt;2.8</th>
<th>2.8-3.9</th>
<th>3.9-5.5</th>
<th>5.5-7.8</th>
<th>7.8-11</th>
<th>11-16</th>
<th>16-21</th>
<th>21-31</th>
<th>31-44</th>
<th>&gt;44</th>
</tr>
</thead>
<tbody>
<tr>
<td>Percent Weight (%)</td>
<td>0.0</td>
<td>0.0</td>
<td>0.0</td>
<td>0.5</td>
<td>2.8</td>
<td>15.2</td>
<td>36.9</td>
<td>31.9</td>
<td>9.5</td>
<td>3.2</td>
</tr>
</tbody>
</table>

Fig. 1 Water atomized 17-4 PH stainless steel powder imaged atop a carbon tape background. Note the two morphology types, spherical and irregular.

2.2 Selective laser melting

Bulk material primitives were fabricated by SLM using a M2 LaserCusing® machine (Concept Laser GmbH, Germany) operated by JMP (Takatsuki City, Osaka Prefecture, Japan). The entire lot of bulk material was produced in the same operation under the same parameters. Laser power and scanning velocity of the fiber laser were set at 190 W and 700 mm/s respectively (1075 nm wavelength). Active circulation of nitrogen gas from an external tank source ensured that residual oxygen content to be less than 0.7% throughout the build. Individual layers were set at 30 µm and controlled by the incremental growth of the build chamber. Powder distribution was performed by a wiper mechanism equipped with a steel blade.

The entire volume of bulk material was fabricated as a series of rectangular tetrahedrons and cylinders – each primitive shape corresponding to a particular tensile specimen loading direction. Figure 2 details their shape and size with respect to the tensile specimens machined from within their volume. Rectangular shaped specimens of 40 x 15 x 35 mm were built such that two tensile specimens with loading axes oriented parallel to the theoretical building plane (X-Y, horizontal reference plane) could be machined following SLM. Cylindrical specimens of dimensions 13 dia. x 40 mm were oriented with their axis of symmetry normal to the building plane. These yielded a single tensile specimen with a loading direction oriented normal to the SLM process. These two primitive shapes were designed with sufficient excess stock such that surface defects would be entirely machined away and have no influence on the tensile response during testing.

Fig. 2 Primitive blocks from which tensile specimens were fabricated. Building direction indicated by the red arrows. Thin line outlines indicate figurative tensile specimen orientation and location. All dimensions are in mm.
Other studies have focused on the externalities stemming from raster patterns used to consolidate material in each discreet layer, attempting to quantify or minimize the effects of warpage, porosity, or residual thermal stresses (Kruth, et al., 2004). While the pattern utilized in this study was fixed and not varied as a topic of study, its detailing here will help explain the presence and morphology of embedded flaws. Of the available patterns the M2 is capable, Fig. 3 outlines the “checkerboard” pattern used to produce the bulk material in this study. Each layer’s cross-section was divided into 5 x 5 mm squares of alternating scanning directions such that no adjacent square had the same scanning orientation. Additionally, each individual scanning track overlapped the previous, adjacent path by 30% of the laser diameter (a laser diameter spot of 150 µm used). To minimize the formation of seams between the edges of each checkerboard square, slight overlaps of 15% of the laser diameter were made at the start and finish of each laser scan path. This scanning pattern was repeated from one layer to the next, but displaced by 1 mm in both the principle axes of the machine build plane (+X and +Y relative to the operator’s oriented view). This was done in an effort to minimize overlapping features that may yield defects capable of transcending beyond one layer and persist through the building direction.

Fig. 3 Scanning pattern used on an arbitrary cross-section of a cubic primitive.

In practice, the SLM process is a series of discreet steps within the working chamber of the machine. Figure 4 details a schematic outline of the internal workings of the SLM machine, highlighting the sequential procedure for each layer. A steel plate affixed to the actuated floor of the build chamber acts as the base upon which powder is initially spread and laser scanning performed. The floor of the build chamber is lowered a prescribed amount equivalent to the layer thickness. The floor of the working powder reservoir, adjacent to the build chamber, is raised a slightly greater amount than the build chamber is lowered. This volume of powder now available to be collected by the distribution mechanism exceeds the volume necessary for the individual layer. Powder is then collected by the wiping mechanism (in this case a metal or rubber blade, or fiber brushes) and distributed into the available volume of the build chamber (250 mm x 250 mm x 30 µm). Excess powder is swept beyond this space and collected in a catchment. The scanning pattern for the current layer’s cross-section is then rastered by the laser system. The process is repeated, typically over the course of several hours or days until the entire volume of the submitted design is complete. Finished parts or volumes of material must then be physically separated from the build plate. Unused powder may be collected and re-used.
Mechanical testing of material fabricated by SLM was performed across a matrix of combinations of testing speeds, heat treatment conditions, and material orientations. Because of the different testing apparatuses necessary to test at various speeds and the unknown size effect(s) associated with the building process, tensile specimen geometry was opted to be of the same design across all tests. The most limiting machine used in this study, the split-Hopkinson tensile bar used in tensile testing at dynamic speeds, ultimately constrained the tensile specimen to that shown in Fig. 5. As previously mentioned, specimens of this configuration were machined from bulk material following SLM fabrication. No heat treatments were performed on the bulk material until following initial machining. All tensile specimens, regardless of orientation relative to the building direction, were heat treated to Condition A status per ASTM A564/A564M (1040 ± 15 °C for 30 min.). Half of these specimens were then aged to the H900 state per the same material standard (480 ± 5 °C for 1 hour). In deviation from the standard, all heat treatments were performed in vacuum within a rotary pump evacuated, quartz chambered electric furnace. Immediately following the specified heat treatment, specimens were exposed to air and quenched in water.

**2.4 Tensile testing**

The aforementioned tensile specimens were tested at three different strain rate orders: $10^{-3}$, $10^{-1}$, and $10^{3}$ s$^{-1}$. Testing conducted at quasi-static strain rates, $10^{-3}$ and $10^{-1}$ s$^{-1}$, was done by a standard uniaxial, screw-driven tensile testing machine (Shimadzu Autograph, DSS-10T-S). A custom testing jig was configured to accommodate the specimen grips. Because of the short gage length of the tensile specimens, elongation was measured indirectly by displacement of the cross-head via a laser displacement sensor and calibrated against the elastic deformation of the testing apparatus. Crosshead displacement rate (speed) was set at 0.5 and 50 mm/min to achieve the necessary strain rates of $1.04 \times 10^{-3}$ and $1.04 \times 10^{-1}$ s$^{-1}$, respectively.

Testing conducted at strain rates on the order of $10^{3}$ s$^{-1}$ required a split-Hopkinson tensile bar (SHTB) (Kolsky bar). Figure 6 illustrates a schematic outline of the SHTB and its configuration used in the dynamic tensile testing of the material in this study. The design of a SHTB for tensile testing of metallic materials requires two, axially supported cylindrical metal bars attached to the material specimen at its ends. A third component, known as the striker tube, is a...
hollow cylinder situated within a concentric vessel surrounding the rest of the test equipment at one end. This vessel remains open towards the test apparatus’s end, attached to a pressurized gas source and valve, and pointed away from the specimen. At strategic and carefully measured locations, strain gages in two wheatstone bridge configurations are affixed to both bars as they are attached to the tensile specimen (2 mm gage length, foil-type gages are used).

Fig. 6 Schematic diagram of the split Hopkinson tensile bar used in this study.

Initiation of a dynamic tensile test requires the release of the pressurized air behind baffles attached to the striker tube. The striker tube then travels at high speed across the short distance between the open end of its containment vessel, along the axis of the test setup, and makes contact with a striker plate attached at the end of one of the two collinear bars. The bar affixed with the striker plate is known as the “incident bar”. The impact of the striker tube onto the striker plate initiates a tensile pulse that then travels down the length of the incident bar. Amplitude of this pulse is proportional to the striker tube velocity at the time of impact. The tensile pulse continues and passes the initial set of foil gages, 1350 mm distant from the bar-specimen interface (Fig. 6, from bottom-right to upper-left). Upon reaching the bar-specimen interface, some of the initial stress pulse is transmitted into and beyond the specimen. This fractional wave is carried further into the second bar known as the “transmission bar”, while the remainder is reflected and travels back down the incident bar. The transmitted stress wave is recorded by the second set of foil gages located 350 mm further down the axis from the specimen. What fraction of the initial tensile pulse is transmitted and not reflected is principally dependent on the response of the material being tested.

Figure 7 shows an example of the time-varying stress state of the incident and transmission bars. The stress state is calculated using the strain measurements made by the foil gage arrays affixed to the incident and transmission bars, the elastic modulus for these bars, and Hooke’s law. Figure 7 also highlights the distance of the gages as the transmitted stress wave is a time-delayed signal along the two collinear bars. A rectangular incident pulse and smooth transmitted pulse are observed. Since the ratio between the cross-sectional areas of SHTB and the specimen ($A_s/A$) is large (roughly 20 times), Eq. 1 may be adopted as the symbolic expression of the nominal stress within the specimen ($\sigma_n$). In this expression $\sigma_I$ is the transmitted stress. The nominal strain rate, $\dot{\varepsilon}_n$, and nominal strain, $\varepsilon_n$, were calculated by Eqs. 2 and 3 respectively. In these expressions, $t$ is time, $l$ is the gauge length of the specimen, $C$ and $\rho$ are the speed of the elastic wave and density of the SHTB bars, respectively. The values of these constants used in this study are $E = 1.96 \times 10^5$ GPa, $\rho = 8.03 \times 10^3$ kg/m$^3$, and $C = \sqrt{E/\rho}$, $C = 4.94$ km/s.

\[ \sigma_n = \frac{A_s}{A} \sigma_I \]  \hspace{1cm} (1)

\[ \dot{\varepsilon}_n = \frac{2}{\rho Cl} (\sigma_I - \sigma_t) \]  \hspace{1cm} (2)

\[ \varepsilon_n = \frac{2}{\rho Cl} \int (\sigma_I - \sigma_t) dt \]  \hspace{1cm} (3)
The load on the tensile specimen is relatively constant and can be seen by the initial stress pulse shown in part of Fig. 7. As the load bearing capacity of the specimen decreases throughout the test, the true stress and the measured strain rate increases. Figure 8 shows a comparison between a typically calculated stress-strain curve for a SHTB test with the strain rate as tabulated. Because strain rate is non-constant, yet within the same order of magnitude, the reported values alongside mechanical properties when tested by SHTB are listed as an average calculated from test start to finish.

Equations (1) – (3) are only valid under the assumption that the stress states at both ends of the specimen are equal. Figure 9 shows the comparison between the transmitted wave as calculated from measurements by the strain gages located on the transmission bar and the time-shifted difference between the initial incident wave and the reflected wave as measured by the strain gages on the incident bar. Since both curves agree well, the eqs.(1) – (3) can be used to obtain stress-strain curves for each individual test.

3 Results and discussion
3.1 Mechanical testing

Figure 10 shows calculated stress-strain curves for SLM material specimens tested at the quasi-static rate of $10^{-3}$ and the dynamic rate of $10^3$ s$^{-1}$. This figure also outlines tests done on specimens after solution heat treatment (Condition A) and the peak-age (H900). For clarity, each plot contains a single curve from individual tests of their respective orientation and speed. Individual curves are typical of the family of tests at each combination of parameters.
From these plots, several observations about the SLM material can be made. Principally among them, tensile loading direction has little influence on the measured yield and ultimate strength of this set of SLM material. Additionally, controlled aging has the desired consequence of enhancing the strength of the material. However, these two plots begin to show the difference in the orientation dependence to plastic flow post-yield, until failure. Further discussion of which is made in the section detailing porosity.

Figures 11 and 12 illustrate the differences between the parallel oriented SLM specimens and the control material for both Condition A and H900 states, at $10^{-3}$ and $10^3$ s$^{-1}$ respectively. The significant distinctions between the two material types, regardless of thermal processing, can be found in the differences of material strength and elongation to failure. SLM material specimens demonstrate appreciably higher yield and ultimate strengths above the conventionally fabricated control material. Explanation for this behavior is outlined in the later section detailing microstructure. In contrast to the enhanced strength, SLM material demonstrated an appreciable reduction of elongation until failure. However, due in large part to the embedded flaws generated at the time of forming, the premature fracture of test specimens can commonly be attributed to observed porosity and not the inherent limitation of the material itself. Tests of the SLM material frequently failed to exceed 15% elongation in the Condition A state and 10% in the H900 condition.

For additional clarity and summary, Fig. 13 and Fig. 14 plot ultimate tensile and yield strengths (UTS and YS) as a function of strain rate. YS is defined as the proof stress value where a 0.2% linear offset of the linear elastic regime intercepts the stress-strain curve. Individual tests are depicted as data points, while averages of each data set, both of
strain rate and UTS, are plotted as piecewise linear curves through the tabulated mean. These two plots also include test results from the control material for comparison.

From these figures, several observations can be made. SLM fabricated material exhibits higher yield and ultimate tensile strengths compared to specimens made of the control material. This strength increase is independent of orientation tested, as both perpendicular and parallel tensile specimens have negligible differences between them throughout most of the testing performed. Additionally, the absence of tensile strength anisotropy is evident between specimens of the same heat treatment condition. Increases in strength are also independent of thermal processing history, persisting through solution heat treatment and subsequent aging, i.e. both Condition A and H900 states demonstrate higher yield and ultimate tensile strengths compared to the control of the same thermal history. While both heat treatment conditions did demonstrate an increase in relative strength, this difference is not nominally proportional. Specimens tested following only a Condition A heat treatment did show a greater difference in UTS to the control (12.4% at 10⁻³, 8.5% at 10⁻¹, and 8.6% at 10⁻³ s⁻¹) than those aged to the H900 condition (6.1% at 10⁻¹, 5.8% at 10⁻¹, and 5.8% at 10⁻³ s⁻¹). Trend lines from the averaged values of UTS do not suggest convergent or divergent behavior as a function of strain rate up to the strain rate limit of 10³ s⁻¹ tested.

Strain rate dependence may be discussed in a variety of ways. For this study, Equation 4 outlines the strain-rate sensitivity parameter, \( m \) (Picu, et al., 2004):

\[
m = \log\frac{\sigma_1 / \sigma_2}{\varepsilon_1 / \varepsilon_2}
\]

The parameter \( m \) may be used to describe the sensitivity of continued plastic flow during uniform plastic deformation following yield and before ultimate tensile stress and non-uniform deformation (e.g. necking). In many materials where the plastic strain difference between yield and ultimate stress is large, on the order of 10% or more, \( m \) may be used to describe the change in behavior for a single tensile test across increasing values of strain (in cases of constant strain rate, plastic flow stress may be used for \( \sigma \)). However, tensile tests performed at both quasi-static and dynamic strain rates for the material in this study exhibit a narrow range of plastic strain between these two limits, less than 5%. Similarly, the differences between the YS and UTS are also proportionally small, approximately 100 MPa. Therefore, \( m \) is better suited as a parameter to compare between the various thermal processing conditions and SLM orientations across the strain rates evaluated. For this purpose, the measured ultimate tensile strength for an individual test is used for \( \sigma \) when evaluating \( m \). Subscripts 1 and 2 refer to a set of tests for comparison where \( \sigma_1 \) and \( \sigma_2 \) are UTS at the corresponding tests’ strain rates, \( \varepsilon_1 \) and \( \varepsilon_2 \). The mean calculated value for the sensitivity parameter is reported for comparisons between strain rates of the same material kind and orientation. Table 3 shows the results of these calculations across the six different specimen kinds and heat treatment conditions for the changes in strain rate tested at 10⁻¹ to 10³ s⁻¹.

---

Fig. 13 Ultimate tensile strength for all tests plotted against strain rate.

Fig. 14 Yield strength for all tests plotted against strain rate.


© 2014 The Japan Society of Mechanical Engineers

[DOI: 10.1299/mej.2014smm0049]
Table 3  Strain rate sensitivity, \( m \)

<table>
<thead>
<tr>
<th></th>
<th>Strain rate sensitivity parameter, ( m )</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td>Perpendicular</td>
</tr>
<tr>
<td>Condition A</td>
<td>0.0084</td>
</tr>
<tr>
<td>H900 temper</td>
<td>0.0077</td>
</tr>
</tbody>
</table>

Several observations from Table 3 can be made about the SLM material and its general behavior. Firstly, layer orientation had little influence on the sensitivity to changes in strain rate from quasi-static to dynamic rates tested. Additionally, the effect of additional thermal processing was seen as negligible. In contrast, the control material demonstrated marginally greater resistance to plastic flow in the Condition A state, however this difference was not significant. In general, the increases seen in yield and ultimate tensile strengths for the SLM material across all strain rates as a consequence of microstructural refinement did not come with a significant change in sensitivity to strain rate.

### 3.2 Porosity

A common residual of the SLM process is the presence of porosity. The material fabricated under the set of processing parameters outlined in this study behaved in a similar manner. While the material was measured to be of high density (>99.5%), the presence of porosity was found to strongly influence the elongation response of the machined tensile specimens (as shown in Fig. 10). Based on observations made of post-test fracture surfaces and sectioned bulk material, several characteristics of these pores were consistently found. Figure 15 shows measurements made of porosity height observed from bulk material sectioned in the vertical reference plane. Pores did not traverse beyond an individual layer and were primarily contained therein. Of the material that was sectioned in the vertical reference plane (X-Z), individual pore heights were found to be approximately no larger than the predetermined layer height of 30 \( \mu \)m (see Fig. 3). However, similar measurements made of pores in the horizontal reference plane (X-Y) and direct observations of fracture surfaces of perpendicular specimens identified the expanse of pores in-plane to an individual layer were not necessarily limited by any processing parameter or mechanism.

Figure 16 (a) and (b) show fracture surface features indicative of porosity and its orientation relative to the building direction. Figure 16 (a) shows a large pore of the kind commonly encountered in the fracture surface of perpendicular specimens. Alternatively, Fig. 16 (b) shows porosity oriented out of plane to the fracture surface as seen in parallel specimens. Note the difference in observable depth of the porosity between Fig. 16 (a) and (b). Parallel specimen fracture surfaces show occasionally deep porosity in the direction normal to the fracture surface – an example of large and elongated pores as formed within an individual layer. Perpendicular specimens, however, have porosity limited in depth to the approximately layer thickness. Regardless of orientation, the sizes of pores were found to occasionally exceed 100 \( \mu \)m in size of their greatest feature. Figures 16 (c) and (d) show macroscopic images of the rupture surfaces for Fig. 16 (a) and (b) respectively. Areas of higher magnification are identified by the square annotation.

![Fig. 15](image)

**Fig. 15**  Porosity observed in the vertical reference plane (X-Z) with the vertical height limited by the layer thickness (30 \( \mu \)m). Building direction is indicated by the arrow. Solution heat treated sample shown.

Under the settings used in this study, pores were formed during SLM as a result of the incomplete wetting of molten material during this process. Close examination of Fig. 16 (a) shows the surface texturing of the internal pore. The striated patterning is visible and consistent with observations of the uppermost surface of the bulk material, denoting molten flow of liquid metal. Of the porosity found through optical microscopy, small-sized voids typically took the form of spherical or elongated ellipsoidal volumes. Larger, connected porosity chains and networks were commonly in straight lines, occasionally parallel or at right angles to each other in-plane. These coincided well with the checkerboard scanning pattern of parallel scanning paths contained within squares at right angles to their neighbors. Further processing by heat treatment was found to have no observable effect on either of these characteristics.

Mechanical loading normal to the building plane is likely to have the adverse effect of stress concentration at large pores’ peripheries, where the greatest feature of size is oriented normal to the applied load. These flaws are loaded such that they would be undergoing Mode 1 crack separation. This occurs at all pores on every layer, as they are all oriented normal to the load under these conditions regardless of in-plane shape and alignment.
3.3 Microstructure

Optical micrographs across all states of thermal processing, imaged in the horizontal reference plane, are shown in Fig. 17 (a) through (c). Bulk 17-4 PH processed by SLM under this study’s machine settings and without additional thermal processing is dominated by a lath martensitic microstructure with grain sizes sub-10 µm. Artifacts of material processing are clearly visible. Microstructural features such as increased grain boundary and dislocation density gradients are apparent. Individual scanning paths are visible in the as-fabricated condition as dark and light bands at ±45° when viewed in the horizontal reference plane (as shown in Fig. 17 (a)). Following additional thermal processing, no noticeable grain orientation or bias is discernible as a consequence of the SLM process (refer to Fig. 17 (b)). As expected, the micrographs show mild coarsening of the microstructure following this thermal processing. Grain size in the H900 aged condition is approximately 10 µm. Based on these observations, the relative increase in strength over the control material and its traditional forming methods is a result of this highly refined microstructure. As supported by tensile testing, solution heat treatment and subsequent aging of the SLM material has the intended consequence of strengthening. This is achieved through the nucleation of Cu-rich precipitates. However, the relative increase in average UTS from Condition A to H900 was between 2-8% less for SLM specimens than the same intended increases observed in the control material (across all strain rates tested, 24.0-28.9% strength increase for the control material, versus 20.2-25.4% for the SLM material).

![Fig. 17 (a)](image-a) As-fabricated microstructure in the horizontal reference plane. Scanning patterns and laser terminus spots are visible.

![Fig. 17 (b)](image-b) Condition A microstructure, horizontal reference plane.

![Fig. 17 (c)](image-c) H900 temper, horizontal reference plane.

4 Conclusions

After the completion of this experimental investigation of selective laser melted 17-4 PH powder, the following conclusions can be made:

1. SLM processing of powdered 17-4 PH has the potential to generate higher tensile strength over traditional forming methods. The source of this strength is largely a consequence of the highly refined microstructure produced during local melting and rapid solidification during SLM. Understandably, the microstructure and resulting mechanical properties are directly linked to the processing parameters deployed.

2. Orientation of the layers as built during SLM did not significantly affect the measured strength of the specimens evaluated via tensile testing.

3. Tensile strength of both the control and the SLM material increase with increasing strain rate across the range investigated, from $10^{-3}$ to $10^{3}$ s$^{-1}$. Increasing tensile strength with strain rate is consistent regardless of thermal
processing condition.

4. Following solution heat treatment, aging of SLM processed 17-4 PH generated the intended strength enhancements. Relative strength increases were less than those of the control. However, absolute increases in strength were similar.

5. Flaws resulting from SLM were principally of the form of porosity. These voids demonstrated orientation dependence relative to the building plane and were limited in the incremental, layer-based building direction inherent to SLM. Similarly, the elongation anisotropy observed between tensile specimens loaded in-plane to the additive layers and those loaded normal to said plane, highlight the sensitivity of the material to stress concentration at the site of these voids. Because of this, specimens loaded normal to the building plane were found to fail earlier with greater frequency than those loaded in-plane. Even under the best of circumstances, elongation of SLM tensile specimens was less than those made of the control material.

References


