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The Influence of Post-Build Microstructure on the Electrochemical Behavior of Additively Manufactured 17-4 PH Stainless Steel

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Abstract

The additive manufacturing (AM) build process produces a segregated microstructure with significant variations in composition and phases that are uncommon in traditional wrought materials. As such, the relationship between the post-build microstructure and the corrosion resistance is not well understood. Stainless steel alloy 17-4PH is an industrially-relevant alloy for applications requiring high-strength and good corrosion resistance. A series of potentiodynamic scans conducted in a deaerated 0.5 mol/L NaCl solution evaluated the influence of these microstructural differences on the pitting behavior of SS17-4. The pitting potentials were found to be higher in the samples of additively-processed material than in samples of the alloy in wrought form. This indicates that the additively-processed material is more resistant to localized corrosion and pitting in this environment than the wrought alloy. The results also suggest that after homogenization, the additively-produced SS17-4 could be more resistant to pitting than wrought SS17-4 in an actual service environment.

Keywords

Additive Manufacturing; Corrosion; Electrochemistry; Microstructure; Pitting; Stainless Steel

1.0 Introduction

Stainless steels are iron based alloys typically containing between 0.12 and 0.30 mass fraction Cr that are resistant to corrosion in common aqueous environments due to the formation of a Cr-rich passivating oxide film [1]. This class of alloys exhibit low general corrosion rates in aerated aqueous environments, but can become susceptible to forms of localized attack, such as pitting, crevice corrosion, and stress corrosion cracking (SCC) in environments that contain chlorides, or that allow for the local concentration of chlorides [2]. Depending on the alloying elements, the concentrations, and the heat treatments, these alloys can be either ferritic (α), austenitic (γ), duplex ($\alpha+\gamma$), martensitic (M), or precipitation hardened (PH) or a combination of these forms. Since the morphology, composition, and integrity of the passivating film can be expected to vary with the

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underlying microstructure, the resistance of these alloys to localized corrosion can also be expected to vary with composition and processing variables [2].

Stainless steel alloy 17-4 PH (UNS S17400, AISI 630) is a popular alloy for applications requiring higher strength and corrosion resistance. By mass fraction, this alloy contains 0.15 to 0.175 Cr, 0.03 to 0.05 Ni, 0.03 to 0.05 Cu, and 0.0015 to 0.0045 (Nb+Ta) and is a M-PH class stainless steel alloy that can be heat treated to strength levels above 1310 MPa (190 ksi) [3–4]. The corrosion resistance of this alloy is among the best of the hardenable stainless steel alloys and can be considered similar to that of the austenitic alloy 304 (UNS S30400) in dilute aqueous solutions [3–4]. The combination of strength, corrosion resistance and weldability makes this alloy a popular candidate when considering alloys for the manufacture of additive processed parts. However, the microstructure and phase constituents of a component produced via an additive route may be completely different from one obtained through traditional fabrication procedures. Conventionally cast alloys have comparatively slower cooling rates, and wrought materials are thermo-mechanically processed after casting to break up the solidification structure and create a more homogeneous material. Therefore, the high cooling rates (up to 10^7 °C/s) and multiple re-melting events that occur during the AM build process are likely to produce microstructures that are dramatically different than those observed in conventional materials. Specifically, prominent micro-segregation and/or unexpected phases are present, particularly in components where no post-build thermal processing is performed [5–7].

The ability of a passivating film to resist breakdown can vary with composition and microstructure. Consequently, the extent to which the unique processing conditions and thermal history of additive processed SS17-4 can alter the corrosion behavior for this alloy is unclear. As a result, an investigation into the localized corrosion resistance of SS17-4, which compared samples in the wrought, powder bed fused (melted), and post-build heat treated condition, was undertaken. The focus was to test the hypothesis that the additive processing conditions alter the magnitude of the electrochemical conditions required to nucleate localized attack in a chloride containing solution as measured by the pitting potential.

2.0 Experimental

2.1. Materials

Samples of SS17-4 steel were selected for analysis from wrought bar stock and from a bar of additively manufactured (AM) material. The wrought SS17-4 was received in the solutionized, martensitic condition (i.e., condition ‘A’). The AM bar was built from virgin nitrogen-atomized powder with a sieve analysis showing that 95.8 % of the particles had sizes less than 53 μm . The build was done in an EOS M270^a, direct metal laser sintering (DMLS) powder-bed system using a Yb-fiber laser, a laser power of 195 W, a laser scan speed of 800 mm/s, a laser energy density of 97.5 J/mm², and a nitrogen build atmosphere. These conditions produced a solid material with an approximate layer thickness of 20 μm .

^aCertain commercial entities, equipment, or materials may be identified in this document to describe an experimental procedure or concept adequately. Such identification is not intended to imply recommendation or endorsement by the National Institute of Standards and Technology, nor is it intended to imply that the entities, materials, or equipment are necessarily the best available for the purpose.

The chemical compositions of the wrought SS17-4 and the post-build AM-17-4 are shown in Table I. Note that chemical analyses were performed on both the nitrogen-atomized AM17-4 powder and on the post-build solid material. The reported difference in the nitrogen content was within the measurement error for the analysis.

X-ray diffraction (XRD) measurements performed on material from this build by Cheruvathur, Lass and Campbell [8] revealed that the as-built microstructure is two-phase with an approximate relative volume fraction of 0.5 fcc (γ austenite) and 0.5 bcc (α ferrite). The high fraction of retained austenite in this condition is likely due to nitrogen absorption in the powder during atomization. Nitrogen is known to stabilize austenite and lower the martensite transformation start (M_s) temperature in steel [9]. The energy dispersive x-ray spectroscopy (EDS) measurements shown in Fig. 1 also revealed that the as-built microstructure was heavily segregated [8]. The Cr segregation resulting from the AM build process leaves the inter-dendritic regions significantly more Cr-depleted than what is observed in conventionally cast materials. This will contribute to the substantially lower martensite start temperature observed in AM-produced material as compared to conventionally-cast material. As such, a heat treatment protocol consisting of homogenizing at 1150 °C for 120 min prior to solutionizing at 1050 °C for 30 min was applied to the as-built bar. This protocol was shown to eliminate the segregation and produce a uniform microstructure with a volume fraction of austenite of less than 0.1[8].

2.2. Potentiodynamic Polarization Current Measurements

Electrodes of the three conditions (i.e., solutionized wrought, AM as-built, and AM+HT), were constructed by mounting samples in epoxy metallurgical mounts in the following orientations: a) wrought with the plane parallel to the rolled surface (RD plane) exposed, b) AM as-built in the XY plane (hereafter referred to by the ISO/ASTM 52900:2015 designation- the XY plane [10]), c) AM as-built with the plane normal to the build direction exposed (hereafter referred to by the ISO/ASTM 52900:2015 designation- the XZ plane), and d) AM built after a post-build homogenization heat treatment in the XY plane. The AM as-built material was examined in the two orientations to allow comparison of the micro-segregation in the melt-pools, which could produce significant variations in the electrochemical behavior. (Note that there was no need for similar orientations in the AM +HT as the material was homogenized after the heat treatment.) Electrical contact was made through the back of the samples with copper wires that were isolated from the environment by polytetrafluoroethylene (PTFE) sleeves. Sample preparation consisted of polishing the surfaces to a 1 μ m diamond surface finish using standard metallurgical practice prior to the start of each experiment [11].

Potentiodynamic polarization experiments were used to compare the electrochemical behavior of the four SS17-4 conditions. All polarization experiments were conducted in a glass electrochemical cell containing approximately 125 mL of 0.5 mol/L deaerated NaCl solution. After inserting the electrode into the cell, the cell was sealed and the solution was purged of oxygen by bubbling argon gas through it for at least 20 minutes after which the argon flow was diverted to continuously purge the space above the solution. A commercial computer controlled high-gain closed-loop amplifier (potentiostat) was used in conjunction

with a saturated calomel reference electrode (SCE), and a platinum counter electrode to measure and provide the current required for each potential as the potential was scanned. While the potential ranges of these scans were varied to enable better resolution of differences in the alloys, a final set of four scans on each orientation and condition were conducted between $-1.000 \text{ V}_{\text{SCE}}$ and $0.00 \text{ V}_{\text{SCE}}$ to enable a definitive comparison. The scan rate was 1 mV/s for all experiments. When the experiments were concluded, the electrode was removed from the electrochemical cell, the surface was rinsed, dried with alcohol, and preserved for microscopic examination.

3.0 Results & Discussion

The microstructures of the SS17-4 steels in the conditions and orientations evaluated in this study are shown in Fig. 2. Fig. 2a is the solutionized wrought SS17-4 in the RD plane. This microstructure consists of delta-ferrite (δ) stringers and small niobium carbide (NbC) precipitates (these are typically less than $1 \mu\text{m}$ and are not observed in the image) in a martensite matrix. This microstructure is consistent with the literature for solutionized SS17-4 [12]. The AM as-built material is shown in the XY plane in Fig. 2b, and in the XZ plane in Fig. 2c. The melt pool shapes and laser tracks are clearly visible in Fig. 1b. Given the aforementioned energy density of the laser during the build, the maximum width of a melted region in a single track appears to be approximately $100 \mu\text{m}$. Fig. 2c shows the overlapping hemispherical melt pools created by the laser in the perpendicular orientation to the build direction. The heights of the individual melt pools range between $20 \mu\text{m}$ to $50 \mu\text{m}$. Assuming the melt pools are hemispherical suggests that between 50 % and 100 % of the previous layer is re-melted upon the addition of a subsequent layer. The scanning electron microscope (SEM) examinations of the microstructure within individual melt pools (Fig. 1) revealed a dendritic-cellular solidification structure with the dendrites predominantly oriented along the build direction. These examinations also revealed fine Nb-rich carbides in this material [8]. Fig. 2d shows the microstructure of the AM-heat treated steel. A martensitic microstructure is dominant, and XRD results indicate the amount of retained austenite is less than 10 % [8]. The EDS analysis in Fig. 1 shows that the micro-segregation has been eliminated and that coarsened MC-type carbides were present along prior dendrite boundaries. Though the volume fraction of retained austenite is still slightly greater than the (2 to 5) % in the conventional wrought material (Fig. 2a), the two microstructures are quite similar overall. No delta-ferrite was observed in any of the AM 17-4 conditions. Since it forms during the later stages of the solidification process, the initial size of the delta-ferrite pools is on the order of the dendrite spacing. The high solidification rates achieved in the AM process do not allow stringers or large pools to form. Thus, the delta-ferrite in the AM material is confined to submicron regions of the dendrite cores [13].

Fig. 3 shows the isopleth sections of the SS17-4 phase diagram constructed from the powder composition shown in Table I. The isopleths were created using ThermoCalc 2015b thermodynamic software [14] and the TCFE8 database [15]. Fig. 3a is the phase diagram for the composition minus the 0.0015 mass fraction of nitrogen that was determined by a chemical analysis of the post-build material. Fig. 3b is the same composition but the nitrogen was included. The dashed line in each figure indicates the carbon content in the powder. As shown in the figure, in addition to substituting for C in the $M(\text{C}, \text{N})$ (Nb- and Cr-rich)

carbide, nitrogen also promotes the formation Cr-rich $M_2(C, N)$ and the ternary nitride Z-phase ($CrNbN$). Carbo-nitrides are known to lower the fracture strength of some steel alloys [9]. However, similar nitrogen contents in austenitic stainless steels have been shown to have a beneficial influence on the mechanical properties and on the corrosion resistance [16–20].

The results of the potentiodynamic polarization experiments are presented in Fig. 4. Each curve shown in the figure is the arithmetic mean of the five scans conducted over an identical potential range in the deaerated 0.5 mol/L NaCl solution. While variations were observed in both the electrode potential and the current density in the individual scans, the set of scans was quite similar for each condition. The dashed line in the figure is the approximate potential that corresponds to the onset of pitting.

As previously noted, this investigation used measurements of the pitting potential to quantify whether the additive processing conditions could alter the magnitude of the oxidizing conditions required to nucleate localized attack in a chloride-containing solution for the different processing conditions, orientations, and heat treatments. The potentiodynamic scans shown in Fig. 4 indicate that the behaviors of the four electrodes do exhibit similar behavior in the passive range, but differ in the potential that resulted in pit initiation. An examination of the measurement error in the scan data was performed to determine whether these differences are statistically significant. The results of the examination are presented for both the wrought and heat treated AM conditions in Fig. 5. Differences in the pitting behaviors between the two orientations of the as-built SS17-4 were also observed, but they were not as meaningful as the differences between the wrought and the AM+HT conditions. As was the case in Fig. 4, the solid lines in this figure are the means of the individual scans, and the dashed lines represent the associated standard measurement uncertainty ($\pm 1\sigma$) [21]. The inherent statistical noise was reduced by applying a standard non-parametric local regression (LOESS) algorithm with tri-cube weighting [22–23]. The difference in pitting behavior between the wrought and AM+HT conditions is statistically different at about the 5σ level. Therefore, the pitting behavior of the wrought samples was found to be statistically significant from that of the additive processed samples. Overall, the onset of pitting occurred at a higher potential in the AM+HT materials than in the wrought sample; a quantitative analysis of this behavior is presented below.

The surfaces of the electrodes were examined in an optical microscope after the completion of each scan. Representative surfaces for each condition are presented as Fig. 7. The surfaces are presented in the same order as those shown in Fig. 2: a) wrought in the RD plane, b) AM as-built in the XY plane, c) AM as-built in the XZ plane, and d) AM + heat treatment in the XY plane. In the figure, the wrought SS17-4 exhibits significant pitting at what are most likely MnS inclusions as well as some additional pitting at carbide interfaces [24–26]. Both orientations of the AM as-built material exhibit pitting at the carbide interfaces similar to that exhibited in the wrought condition. Some additional fine-scaled pitting is visible within the highlighted melt pools in Fig. 7b and Fig. 7c. Since these pits appear to roughly track the general shape of the melt pools, they were attributed to compositional heterogeneities near the melt pool interfaces in this condition. It is also possible that they are a smaller MCN or M_2CN carbo-nitrides. While the heat-treated AM surface is similar to that of the wrought,

the overall density of observable pits was considerably lower than what was observed for the wrought material.

As shown in the inset in Fig. 6, the potentiodynamic polarization curves in Fig. 5 can be represented by two straight lines on a semi-log plot. Thus, the Tafel kinetics[27] of the pit initiation reactions can be represented using an Arrhenius plot with an activation energy that changes at this intersection indicating a fundamental change in the nature of the reaction that produces the measured current. Here, the slope changes when the passive films break down locally and the anodic dissolution of metal at the tip of propagating pits becomes the dominant source of the measured current. The potential where this change occurs, in an increasing potential scan, is known as the pitting potential (E_{pit}). This potential is directly related to the Gibbs energy driving corrosion reactions (ΔG) by the relationship

$$\Delta G_i = -n_i F E \quad (1)$$

where E is the potential measured with respect to a reference electrode that has a potential fixed by the thermodynamics of a single reaction [28, 29], F is Faraday's constant, and n is the charge exchanged in the reaction designated by the subscript i . Since all electrode reactions are influenced by potential per this relationship and the reactions responsible for pitting are unknown, the measured pitting potential values are not converted to free energies. However, these values are directly related to these driving forces, and any statistically significant difference between the values determined for E_{pit} with the conditions examined must be due differences in composition, segregation, or microstructure.

A value for E_{pit} , and the associated uncertainty, was estimated for each experiment by fitting the two linear portions observed in the semi-log plots to exponential functions with the form, $y = ae^{bx}$. The potential where the two fits intersected was defined as E_{pit} . A Monte Carlo uncertainty propagation algorithm was used to propagate the uncertainties in the fitted exponential functions to obtain the uncertainties in E_{pit} , with the assumption of Gaussian-distributed uncertainties. 10,000 calculated variants were used to sample the uncertainties in E_{pit} , which was adequate to assure convergence. Fig. 7 shows the E_{pit} values determined by this process for the 16 potentiodynamic polarization curves along with their $\pm 1 \sigma$ uncertainties. Each sample condition is grouped along the vertical direction for clarity and plotted using a different color. Table II lists the mean E_{pit} , the standard deviation of the E_{pit} values, the voltage range for E_{pit} , and the arithmetic mean of the 1σ uncertainties, $\overline{\sigma_{\text{und}}}$, obtained using the Monte Carlo analysis for each sample condition. It is worth noting that $\overline{\sigma_{\text{und}}}$ is approximately independent of the sample condition.

The wrought condition exhibits the most negative pitting potential and the E_{pit} values are tightly clustered with a mean potential of (-0.1351 ± 0.0048) V. The standard deviation of the measured values, 0.0048 V, is somewhat larger than $\overline{\sigma_{\text{und}}} = 0.0030$ V, and sample variability is likely a contributing factor. The as-built XY plane and XZ plane conditions are also reasonably well clustered, with standard deviations of 0.0069 V and 0.0063 V, respectively. Not surprisingly, these conditions exhibit slightly larger variability than the wrought condition. The mean pitting potential for the as-built XY plane samples is similar to

the wrought condition, but the as-built XZ plane values are significantly displaced to a less negative mean of (-0.1045 ± 0.0063) V, a 5σ difference.

The behavior of the HT condition is the most unusual. The mean E_{pit} of (-0.093 ± 0.037) V shows the largest deviation from the wrought condition, but the individual values cover a remarkable range of 0.083 volts, compared with 0.0105 V to 0.0158 V for the. Here, current density oscillations were observed indicating the nucleation and repassivation of meta-stable pits in a potential range that is close to that which produced stable pitting in the wrought condition samples [30]. The decrease in the current density to the original passive level above this region suggests that either the passive film has become more stable or that the meta-stable pits are repassivating before they grow large enough to generate large current transients. In either case, stable pit nucleation and growth does not occur until higher electrode potentials are reached. The range of E_{pit} values shown in Fig 6 reflects the stochastic behavior of this process, where the left-most scan exhibited no repassivation (making it comparable to the wrought behavior) and the remaining three scans showed varying numbers of passivation events at different potentials.

The initial expectations at the start of this evaluation were: a) the behaviors of the wrought SS17-4 and the AM-HT would be similar in the NaCl solution because both alloys have similar chemical compositions, both martensitic, and both have a volume fraction of retained austenite on the order of five percent, and b) the dendritic microstructure in the as-built condition would produce the lowest E_{pit} values with the largest scatter. However, the pitting data shown in Fig. 6 clearly demonstrate that the behaviors are considerably different from that expected; and as such, generate several key questions. The most significant question is, what produced the differences in the pitting behavior observed between the four conditions of the SS17-4 shown in Fig. 6? The results imply that the differences can be attributed to the combined influences of the microstructures and the variability in the local composition.

The microstructures shown in Fig. 2 are noticeably different, and this could produce some of the observed changes in the electrochemical behavior. The simplest explanation for the difference in pitting behavior between the wrought SS17-4 and the AM-HT could be the absence of large inclusions and delta-ferrite stringers in the AM-HT condition. It is well known that a “cleaner” microstructure promotes a more homogeneous surface chemistry, which would improve the resistance to general corrosion in the NaCl solution. However, localized attack, such as pitting, tends to initiate at sites where the integrity of the passive film is reduced [27],[24, 31]. A more detailed examination of Fig. 2a and Fig. 2d revealed that while the prior austenite grain sizes are comparable, the relative sizes of the martensite laths in the AM-HT grains are considerably smaller than those in the wrought alloy. Since the passive film in a stainless steel is essentially Cr-Ni oxide [1], a material with a heterogeneously distributed primary constituent such as chromium is likely to have a non-uniform passive film, which would then be susceptible to localized attack (pitting). Such heterogeneity is exhibited in the Cr distributions in Fig. 1. Interfaces, such as grain boundaries, lath boundaries, and more importantly, those between second-phase constituents (e.g., carbides), generally produce discontinuities in the local composition and often serve as preferred sites for pit initiation [31] [25–26]. Typically, niobium is added to martensitic stainless steels to minimize the formation of chromium carbides, which would lead to an

overall enhancement in the stability of the passive film. The EDS analyses also revealed that the NbC precipitates in the AM+HT were approximately an order of magnitude finer than those in the wrought, and that they were more homogeneously distributed. While this could produce an improvement in the pitting resistance for the AM+HT alloy (Fig. 6) it is unlikely that the improvement was solely due to the cleaner microstructure.

Fundamentally, the additive manufacturing process is a repetitive welding process that melts, solidifies, and re-heats the layers of the structure multiple times during any given build. This produced the segregated microstructures shown in Fig. 2b and Fig. 2c. The EDS analyses of the as-built material revealed that the Cr tended to segregate to the dendrite cores. This could account for the larger scatter observed in the E_{pit} values between the as-built surfaces, because the electrodes were ground and polished to remove all the pits formed during the previous potentiodynamic scan. Considering that each additive layer is approximately 20 μm thick, grinding is likely to change the local composition on the surface exposed to the test solution in each experiment.

Local variations in the Cr content account for the differences observed in electrochemical behavior in the as-built condition. However, it does not explain the differences between the wrought SS17-4 and the AM-heat treated SS17-4 alloys because the EDS data also show that the heat treatment eliminated the dendritic microstructure and that the chromium was homogeneously distributed after the heat treatment. The differences in pitting resistance between the wrought and the heat-treated conditions are more substantial so they are likely due to another influence.

Recall that the nitrogen content in the AM17-4 was 0.0015 mass fraction (Table I). Nitrogen can have a beneficial influence on the corrosion resistance in austenitic stainless steels, and similar enhancements have also been reported in martensitic stainless steels [32]. The exact role of the nitrogen in these enhancements is debatable and a considerable amount of research has been performed to identify the active mechanism [33]. In austenitic stainless steels, the literature indicates that nitrogen in solid solution enhances passivity by incorporating into the passive film and/or segregating on the metal/film interface [34], and that nitrogen can form N-H bonds during an active dissolution of metal, which inhibits local acidification and effectively suppresses pit growth [35]. In martensitic stainless steels, the improved pitting resistance is more likely due to the suppression of CrC formation [36], and this appears consistent with what was observed in the AM+HT alloy. The 0.0015 mass fraction of nitrogen in the AM+HT alloy is considerably higher than the solubility of nitrogen in ferrite (approximately 0.001 by mass fraction [37]). The additional 0.0005 mass fraction of nitrogen present is likely to remain homogeneously distributed in the retained austenite (where the solubility of nitrogen is approximately ten times higher), or it could promote the formation NbCN precipitates. Both possibilities could produce a more stable passive film. Based on these findings, the nitrogen in the AM+HT does appear to have a beneficial influence on the pitting resistance in this alloy. However, additional studies including argon-atomized SS17-4 are necessary to confirm this. These experiments are in progress.

While the E_{pit} data in Table II and in Fig. 7 suggest that the enhanced pitting resistance may have been suppressed, it must be noted that the lowest value of the pitting potential observed in the AM+HT is equal to that exhibited by the wrought SS17-4. In contrast, the highest value in the pitting potential range is approximately 10 mV higher than that of the wrought. This suggests that the AM+HT could be more resistant to pitting than the wrought alloy in an actual service environment. Additional evaluations designed to identify the source of the scatter in the E_{pit} data for the AM+HT condition are in development.

4.0 Conclusions

This evaluation was designed to determine whether the repetitive build process in an additive manufacturing process would be detrimental to the localized corrosion resistance of SS17-4 in a chloride-containing solution. It featured a series of potentiodynamic scans conducted over a range of potentials to assess the electrochemical behavior in deaerated 0.5 mol/L NaCl. Measurements were performed on electrodes made from: solutionized wrought material in the RD plane, as-built AM material in the XY plane, as-built AM in the XZ plane, and heat treated AM in the XY plane. After heat treatment, the microstructure of the AM+HT condition was completely homogenized (i.e., no segregation remaining from the additive build process was observed). The compositions of the wrought and the AM alloys were similar except for 0.0015 mass fraction nitrogen in the AM alloy that was retained from the atomization.

The results of the potentiodynamic experiments established that the behaviors of the three conditions of the SS17-4 are different, and that the pitting potentials were generally higher for the AM+HT than those observed for the wrought alloy. While the scatter in the pitting potentials was considerably greater in the AM+HT, the increase was also determined to be statistically significant. The lowest value in the range of pitting potentials observed in the AM+HT is equal to that exhibited by the wrought SS17-4, whereas the highest value in the pitting potential range is approximately 10 mV higher than that of the wrought. The difference was attributed to a) a finer martensite lath structure with more homogeneously distributed NbC precipitates than those in the wrought alloy, and b) a more stable passive film resulting from the absorbed nitrogen in the AM+HT alloy.

This evaluation indicates that at a minimum, the performance of the additively produced SS17-4 was no worse than the wrought material in the chloride environment. The results also suggest that after a homogenizing heat treatment, additively produced SS17-4 could be more resistant to pitting than wrought SS17-4 in an actual service environment. Additional evaluations are in development to determine whether the improvement in the pitting resistance in the heat-treated material was due to the presence of nitrogen or to some other factor in the build process.

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Fig. 1.

Optical micrographs, SE images and EDS maps for Fe, C, Cr, Cu, and Nb showing the composition and micro-segregation of the AM17-4 stainless steel. The images were obtained from as-built AM17-4 (parallel to the X-Z plane), homogenized AM17-4 (parallel to the X-Y plane), and wrought SS17-4 (parallel to the RD plane) respectively.

Fig. 2.
The microstructures of the four electrode conditions: a) wrought SS17-4 parallel to the RD plane, b) as-built AM17-4 parallel to the XY plane, c), as-built AM17-4 parallel to the XZ plane, and d) heat treated AM17-4 parallel to the XY plane (Swede's etch).

Fig. 3. Isopleth sections of the SS17-4 phase diagram constructed from the powder composition in Table I. Fig. 3a contains no nitrogen. Fig. 3b contains 0.0015 mass fraction nitrogen as reported by chemical analysis of the virgin powder.

Fig. 4.

Comparison of the electrochemical behavior of the wrought SS17-4, the as-built AM17-4, and the heat treated AM17-4. The lines are the mean of the potentiodynamic scans performed over the same potential range. Each scan was conducted in deaerated 0.5 mol/L NaCl with a sweep rate of 1.0 mV/s.

Fig. 5. Comparison of the pitting behaviors of the RD plane of wrought SS17-4 and the XY plane of heat treated AM17-4 in deaerated 0.5 mol/L NaCl. The dashed lines represent a measurement uncertainty of ± 1 standard deviation (σ).

Fig. 6.

Pitting produced during the potentiodynamic scans in deaerated 0.5 m/l NaCl. a) wrought 17-4 in the RD plane, b) as-built AM17-4 in the XY plane, c), as-built AM17-4 in the XZ plane, and d) heat treated AM17-4 in the XY plane. The dashed lines in Fig. 6b and Fig. 6c indicate melt pool boundaries.

Fig. 7. Comparison of the pitting behaviors of wrought SS17-4 in the RD plane, as-built AM17-4 in both the XY and XZ planes, and heat treated AM17-4 in the XY plane in deaerated 0.5 mol/L NaCl. The inset is a graph showing the pit initiation-repassivation behavior observed in the AM+HT alloy. The error bars represent a measurement uncertainty of ± 1 standard deviation (σ).

Table I

Composition of the SS17-4 Alloys

Element	Wrought (Solutionized)	Additive (Post-build)
Cr	0.1587	0.1572
Ni	0.0424	0.0481
Cu	0.0387	0.0374
Nb	0.0025	0.0027
Mn	0.0055	0.0067
Mo	0.0012	0.0011
Si	0.0005	0.0080
C	0.0002	0.0003
S	0.0004	0.0001
P	0.0003	0.0002
N	–	0.0015
Fe	Bal.	Bal.

(All values are mass fraction. Measurement uncertainty: ± 0.0001)

Table II

Statistical Analysis of Pitting Potential Data

	Wrought	As-built: XY plane	As-Built: XZ plane	AM heat treated
Mean E_{pit} (V)	-0.1351	-0.1301	-0.1045	-0.0931
Stand. Dev. (V)	0.0048	0.0069	0.0063	0.037
Range (V)	0.0105	0.0158	0.0145	0.0833
$\overline{\sigma}_{ind}$ (V)	0.0030	0.0035	0.0031	0.0036