

Relationship between manufacturing defects and fatigue properties of additive manufactured austenitic stainless steel

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ABSTRACT

Tensile properties, fatigue crack initiation, fatigue crack growth rate, and fatigue life are evaluated in 304L austenitic stainless steel fabricated by directed energy deposition (DED). Large lack of fusion (LoF) defects (often > 1 mm in length) significantly reduce ultimate tensile strength and ductility, as well as accelerate fatigue crack initiation and reduce fatigue life. In comparison, small spherical defects (< 100 μm in diameter) have less effect on tensile and fatigue properties. Fatigue crack growth rate is less severely affected by defects than other properties, showing only local acceleration in the proximity of LoF defects. Therefore, shorter fatigue life is attributed to the role of LoF defects on facilitating fatigue crack initiation and to a lesser extent fatigue crack propagation. Additionally, the fatigue life can be normalized for defects by considering their effect on ultimate tensile strength, suggesting that in the limit of low defect population, the fatigue strength of additively manufactured stainless steel is similar to conventional wrought materials.

1. Background

Directed energy deposition (DED) additive manufacturing (AM) promises to produce geometrically complex engineering components in a single manufacturing step. Difficulty in managing manufacturing defects that are characteristic of the DED process, however, is driving implementation of additional post-processing treatments that are expensive and time-consuming [1]. A recent review of the literature on the fatigue behavior of AM metals found that manufacturing defects and microstructural heterogeneity in as-deposited AM materials lead to degraded fatigue properties compared to wrought materials [2]. Manufacturing defects particularly impact the fatigue performance of DED materials. In DED austenitic stainless steels, the presence of manufacturing defects is reported to increase scatter in fatigue life data [2–4] and decrease overall fatigue life compared to wrought austenitic stainless steels [2]. In contrast, enhanced fatigue life has also been reported for wire + arc additive manufactured (WAAM) 304L compared to wrought and laser-DED 304L materials, despite the presence of processing-induced porosity in the WAAM 304L [5]. Other studies report fatigue crack growth rates (FCGR) that are consistent with wrought austenitic stainless steels (i.e., described well by the Paris relationship) [6]. These conflicting findings in the literature demonstrate the need to develop a more comprehensive understanding of the fatigue behavior of

AM materials and to investigate the influence of processing defects on the measured fatigue behavior.

The goal of this work is to elucidate the influence of manufacturing defects on the fatigue behavior of DED 304L materials. This work correlates fatigue performance with defect character, in particular the mechanical influence of lack of fusion (LoF) defects on fatigue life. Fatigue crack initiation, fatigue crack propagation, and fatigue life are measured and correlated with the influence of defects on ultimate tensile strength. The results highlight the importance of manufacturing defects on fatigue performance of DED austenitic stainless steels, particularly the role of LoF defects on fatigue crack initiation.

2. Materials and methods

Materials were manufactured from gas atomized austenitic stainless steel powder feedstock conforming to the requirements of a 304L grade alloy [7]. The 304L feedstock powder had a nominal composition (wt %) of 19% Cr, 10% Ni, 1.5% Mn, 0.6% Si, and 0.01% C. Rectilinear blocks (Fig. 1) measuring approximately 60 mm $\times$ 40 mm $\times$ 10 mm were additively manufactured via DED in a Laser Engineered Net Shaping (LENS®) 750 workstation utilizing time-invariant input processing parameters, as listed in Table 1. Two fin-shaped and two boss-shaped builds were fabricated, where the fins stand up on the baseplate

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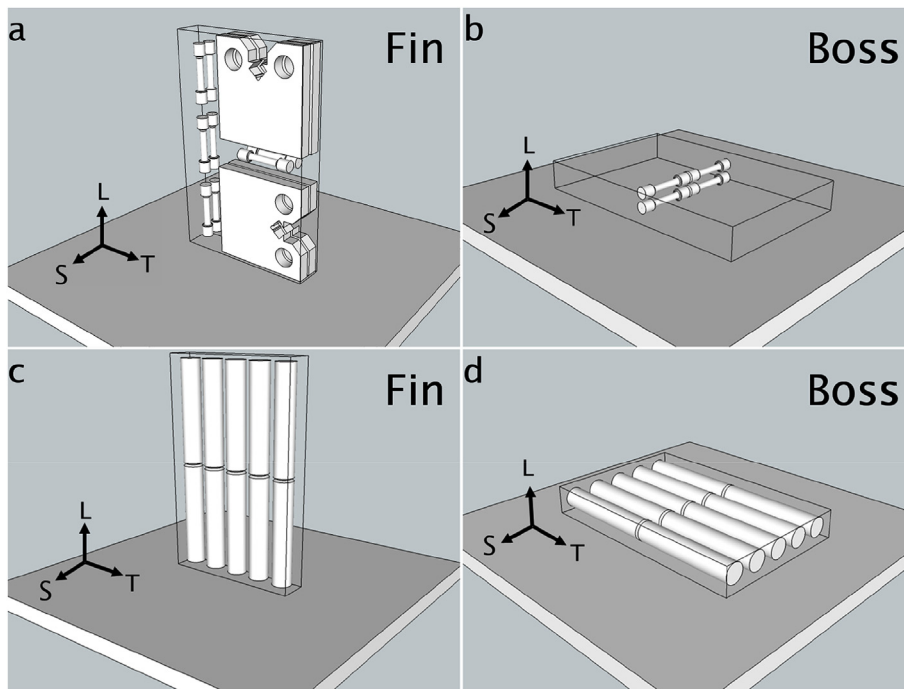


Fig. 1. Schematic depictions of the four DED 304L build geometries from which miniature tensile, compact tension, and cylindrical notched tensile (CNT) specimens were extracted. Axes in each schematic indicated the longitudinal (L, build direction), short transverse (S), and long transverse (T) directions for the builds. Paired specimens define locations in each build. (a) A fin build with L (top, middle, bottom) and T (middle) tensile specimens, as well as TL (top) and LT (bottom) compact tension specimens. (b) A boss build with S (top and bottom) tensile specimens. (c) A fin build with CNT-L specimens. (d) A boss build with CNT-T specimens.

Table 1
Deposition parameters for the DED 304L builds.

Deposition parameter	Value	Unit
Laser	Yb:Fiber	–
Laser power	400	W
Scan speed	11	mm/s
Hatch increment	0.46	mm
Layer increment	0.34	mm
Hatch rotation	90	degrees
Powder feed rate	34	g/min
Oxygen concentration	< 15	ppm

while the bosses lie down on the baseplate, as shown in Fig. 1. The fin and boss terminology are used to distinguish the geometry of the builds with respect to the baseplate from the ambiguity of directions such as horizontal and vertical. Each build was deposited on a 5 mm thick wrought 304L austenitic stainless steel plate. In all cases, the L-direction is defined as the build direction, while the in-plate directions are referred to as T and S for transverse and short transverse in relation to the dimensions of the build.

Mechanical test specimens extracted from the DED 304L builds are depicted schematically in Fig. 1. Miniature cylindrical tensile specimens were extracted with their loading axis parallel to the L direction from near the top, middle, and bottom of one of the fin builds (Fig. 1a). Tensile specimens were also extracted from the middle of builds with their loading axis in the T and S directions, respectively, from a fin and boss (Fig. 1b). Side-by-side specimens at the same distance from the baseplate are assumed to be equivalent, thus as shown in Fig. 1, a pair of specimens defines each location. The density of the specimens was measured by the Archimedes method with reference to a value of $7.90 \text{ g}\cdot\text{cm}^{-3}$ reported in the literature for wrought 304L [8]. The miniature tensile specimens had a gauge diameter of 1.8 mm and a nominal gauge length of 8.4 mm. Tensile tests were performed on a screw-driven universal testing machine (Satec, USA) under ambient conditions at a constant displacement rate (0.0044 mm/s) corresponding to a strain rate of approximately 10^{-3} s^{-1} . Strain was measured via a laser extensometer with a gauge length of approximately 6 mm.

Tension-tension fatigue testing was performed on a servo-hydraulic universal testing machine (MTS, USA) in ambient conditions to

evaluate fatigue life. A notched geometry was employed to investigate crack initiation in the presence of a stress concentration and triaxial stress state [9]. The direct current potential difference (DCPD) method is shown in the literature to be an effective technique to quantify crack initiation [10,11]. Circumferentially notched tensile (CNT) specimens were extracted from one fin (Fig. 1c) and one boss (Fig. 1d). Longitudinal CNT specimens (CNT-L) were extracted from the fin, and transverse CNT specimens (CNT-T) were extracted from the boss. The specimens had a reduced section diameter (D) of 5.7 mm, with a minimum diameter (d) of 4.0 mm at the notch root. The notch root radius was $\sim 0.13 \text{ mm}$ with a notch angle of 60° . This V-shaped notch groove corresponds to a stress concentration factor (K_t) of ~ 3.9 [12]. Load controlled tests were conducted to failure at a load-ratio of 0.1 (ratio of minimum to maximum load, R) with a loading frequency (f) of 1 Hz. The maximum stress for the notched specimens (S_{Nmax}) was calculated from the initial cross-sectional area at the notch root and the maximum applied load, which represents the net-section stress at maximum applied load. In addition to fatigue cycling to failure, crack initiation was evaluated by instrumenting the test specimens for the DCPD method. The same spot welded configuration for the current and voltage sensing leads was utilized as in previous investigations, where high sensitivity to fatigue crack initiation was demonstrated for similar specimens [10,13]. Crack initiation was defined by the transition from the initial value to a rapidly increasing rate of change in the potential difference with increasing number of load cycles for each specimen.

Four compact tension (CT) specimens were extracted from one fin build (Fig. 1a) to measure fatigue crack growth rates. Compact tension specimens were extracted with different orientations near the top of the build (TL) and the bottom of the build (LT), as shown in Fig. 1a. Each CT specimen had a thickness (B) of 3.0 mm and width (W) of 22.2 mm, consistent with geometric guidelines given in ASTM E647 [14]. Notches were machined to give $a = a/W = 0.20$, where a is the length of the crack (or notch) as measured orthogonally from the load line in the plane of the specimen. Specimens were precracked in accordance with the ASTM E647 standard [14] by shedding load (constant load ratio and frequency, $R = 0.1$ and $f = 10 \text{ Hz}$, respectively) to a final stress intensity factor range (ΔK) of $\sim 8 \text{ MPa}\cdot\text{m}^{1/2}$ at a of ~ 0.28 . For the CT geometry, ΔK is calculated from the relationship given in ASTM E647 [14]:

$$\Delta K = \frac{\Delta P}{B\sqrt{W}} \frac{(2 + \alpha)}{(1 - \alpha)^{\frac{3}{2}}} (0.866 + 4.64\alpha - 13.32\alpha^2 + 14.72\alpha^3 - 5.6\alpha^4) \quad (1)$$

Pre-crack lengths were marked on the specimens by heat tinting at 300 °C for 20 min in air. FCGR tests were conducted at constant load amplitude and frequency ($R = 0.1$ and $f = 1$ Hz, respectively). Tests were terminated at a fractional crack length α of ~ 0.6 , corresponding to $\Delta K \sim 30 \text{ MPa}\cdot\text{m}^{1/2}$. Crack length was determined during testing using the compliance method [14] by attaching a clip gauge to knife edges at the load line of the CT specimens. For compliance measurements, the modulus was assumed to be the modulus of fully dense austenitic stainless steel (204 GPa). In porous materials with high relative density, the modulus is proportional to the relative density [3], but this correction was not taken into account in the determination of crack length since the measured crack length agreed with crack length determination from compliance within experimental uncertainty. After breaking open the specimens, the initial crack length (after precracking) and final crack length were measured, using high-resolution optical images of the fracture surface and digital analysis software. All testing (including precracking) was controlled with Fracture Technologies Associates (FTA) software (LTI, USA). Crack length data for determination of the fatigue crack growth rate were corrected post-test using the post-mortem measurements of the initial and final crack lengths from the fracture surface.

Specimen cross-sections and fracture surfaces were evaluated using a Keyence VHX-5000 digital confocal microscope. Cross-sections were prepared by grinding/polishing with incrementally finer abrasives for metallographic examination. The cross-sections were then vibratory polished using a Pace Giga 0900 (Pace® Technologies, USA) in 0.02 μm colloidal silica.

3. Results

3.1. Density and characterization of manufacturing defects

Archimedes measurements revealed a greater population of manufacturing defects in the tensile specimens extracted from near the baseplate for both the boss and fin geometries. All of the material in the boss was within 10 mm of the baseplate and, in general, the boss displayed the lowest density, between 97.1 and 98.0%. In the fin, the density varied along the build direction, ranging from 98.2% to 99.6%, and was greatest near the middle of the build height. Due to the variation of density along the height, the measured density of the L tensile specimens represents an average over the length (~ 16 mm) of each specimen. No effort was made to assess lateral variations in the density, and the boss specimens are assumed to be nominally uniform in density.

Density measurements of CNT specimens were consistent with the measured density of the tensile specimens. For the purposes of this study, the CNT-T specimens are described as low-density material with density values ranging from 96.8% to 97.4%. In contrast, the CNT-L specimens are categorized as high-density material with density values ranging from 99.4% to 99.5%. Although the CNT density values represent an average over a greater specimen length (~ 56 mm), the notched regions of the CNT-T and CNT-L specimens are from equivalent locations in the builds as the low-density (S orientation) and high-density (L orientation) tensile specimens, respectively. Density along the crack plane in CT specimens could not be directly measured; however, density of the TL-oriented compact tension specimens is inferred to be higher than that of the LT-oriented compact tension specimens based on the density of the tensile specimens. This correlation will be confirmed from observation of the fracture surfaces of the compact tension specimens, as discussed in Section 4.3.

The bulk defects in the high-density material (e.g., near the notched region of a CNT-L specimen) are shown in Fig. 2a to be mostly spherical in appearance and fine ($< 100 \mu\text{m}$ in diameter). In contrast, low-

density materials (Fig. 2b) were characterized by irregularly shaped, large defects that could exceed 1 mm length in these materials. Partially melted powder particles are often welded to the surface of these large defects. While the large defects generally incorporate pore volume, the large aspect ratio is characteristic of LoF defects, which also describes predominantly planar defects that do not display significant volume. Therefore, for the purpose of this study, we distinguish between defect populations as either (1) small, round pores, which are characteristic of the higher-density specimens (generally greater than about 98.5% dense) or (2) LoF defects, which are characteristic of lower-density specimens (typically less than 98.5% dense).

3.2. Tensile behavior

Tensile properties exhibited a strong dependence on specimen density. Tensile stress-strain curves of tensile-type specimens extracted from the builds are shown in Fig. 3. The ultimate tensile strength (S_u) is shown to vary with density for all the extraction locations (Fig. 4). The S_u ranges from as low as 450 MPa to almost 700 MPa, while total elongation can be less than 20% or more than 70%.

3.3. Fatigue life behavior

The tension-tension notched fatigue life data (Fig. 5) demonstrate a significant difference between the fatigue behavior of low-density (CNT-T) and high-density (CNT-L) materials. For the same nominal maximum stresses, lower-density CNT-T specimens exhibited shorter fatigue life compared to the higher density CNT-L specimens. The general shape of the fatigue life curves is similar to notched fatigue data for conventional wrought annealed (AN) 316L [15], strain hardened (SH) 304L [15], and SH 316L [16] materials tested under nominally the same conditions (i.e., similar ambient conditions, notch geometry, load ratio, and frequency). In general, fatigue life curves are characterized by a power-law relationship and, in the case of notched specimens, a modified Basquin relationship is used where the stress is represented by the net-section stress at maximum load (S_{Nmax}) and N is the number of cycles to failure [17]:

$$S_{Nmax} = sN^b \quad (2)$$

where s and b are fitting parameters that can be thought of as the intercept and slope, respectively, in a log-log plot. As shown in Table 2, all the materials exhibited essentially the same slope (b), but different intercept values (s).

3.4. Fatigue crack initiation behavior

In addition to the fatigue life to failure, the DCPD instrumentation enables the identification of fatigue crack initiation. The fraction of the total cycles to failure that were required to initiate a crack is shown in Fig. 6 as a function of the maximum net-section stress for both low-density (CNT-T) and high-density (CNT-L) specimens. In both cases, the number of cycles required to initiate a crack becomes a larger fraction of the total life for lower S_{Nmax} . For the same S_{Nmax} , the low-density material requires a smaller fraction of its total fatigue life (and fewer cycles) to initiate a propagating fatigue crack than the high-density material.

3.5. Fatigue crack propagation behavior

Fatigue crack growth rate (da/dN) is shown in Fig. 7 as a function of the fatigue driving force (ΔK , stress intensity factor range). Fatigue crack growth tests were performed following standard practice with specimen pairs in both the TL (high-density) and LT (low-density) orientations. The high-density specimen pair (TL orientation) exhibited similar fatigue crack growth rates as one of the lower-density (LT orientation) compact tension specimens, while the other low-density

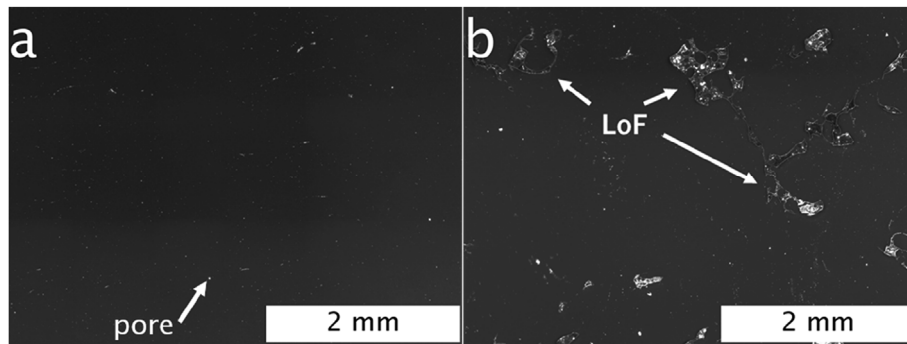


Fig. 2. Micrographs of polished cross-sections near the notch of the (a) high-density (CNT-L) and (b) low-density (CNT-T) specimens. Arrows indicate pores and lack of fusion (LoF) defects, respectively.

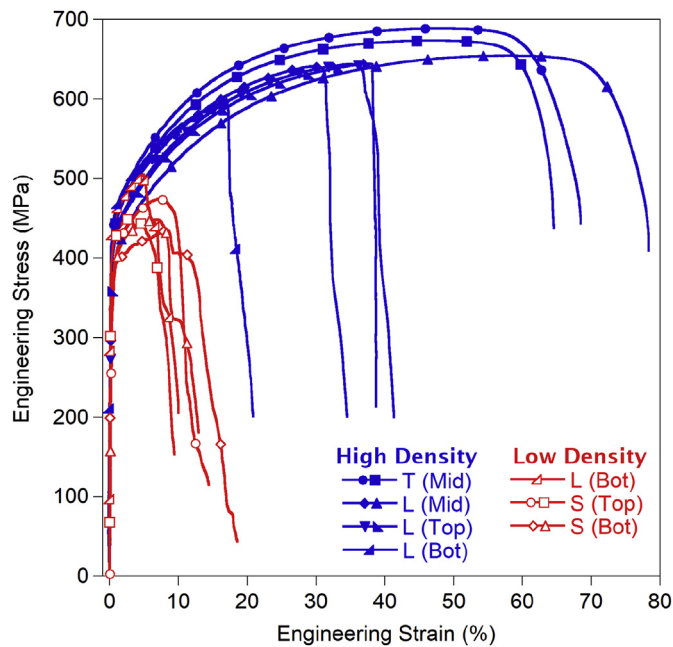


Fig. 3. Tensile curves from the T, L, and S tensile specimens. Specimen distance from the baseplate is indicated in parentheses for top (top), middle (mid), and bottom (bot). Specimens with higher ($> 98.5\%$) and lower ($< 98.5\%$) density are differentiated, respectively, by blue closed symbols and red open symbols. (For interpretation of the references to colour in this figure legend, the reader is referred to the Web version of this article.)

specimen showed a somewhat higher crack growth rate. The slope of the $da/dN-\Delta K$ curves for the high-density TL specimens is consistent with the data for wrought 316L [18]. Additionally, the slope of the FCGR curves is consistent with recommendations for steels from the International Institute of Welding (IIW) [19], although the magnitude is somewhat lower. While the slope is higher in the specimen (LT orientation) with higher FCGR, the fatigue crack growth remains lower than the IIW recommendation for $\Delta K < \sim 20 \text{ MPa}\cdot\text{m}^{1/2}$.

4. Discussion

4.1. Tension

The values of S_u and El_T exhibited in Fig. 3 clearly depend upon the density of the materials in this study: ultimate tensile strength and elongation increase with density. Tensile specimens with density greater than 99% approach the values reported in the literature for fully dense DED austenitic stainless steels [4], exhibiting S_u exceeding 650 MPa and total elongation greater than 60%. Since the LoF defects

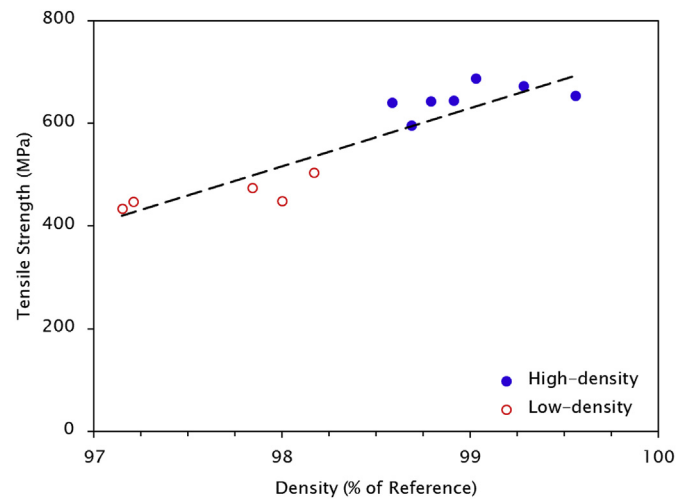


Fig. 4. Tensile strength measured from the T, L, and S tensile specimens as a function of measured density. Closed and open symbols indicate, respectively, higher ($> 98.5\%$) and lower ($< 98.5\%$) density tensile specimens.

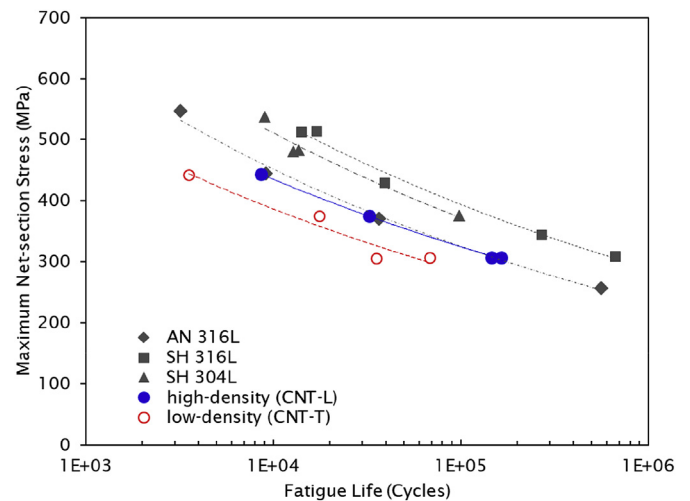


Fig. 5. Fatigue life data plotted with respect to the maximum net-section stress at the notch root (S_{Nmax}) for the high-density (CNT-L) and low-density (CNT-T) specimens. Notched fatigue life data from annealed wrought 316L (AN 316L) [13], strain-hardened 316L (SH 316L) [15], and strain-hardened 304L (SH 304L) [13] provided for reference.

contribute to the lower density, these defects are strongly linked to the diminished strength and ductility of the low-density materials. The strong correlation of S_u with density (Fig. 4) suggests that uniform

Table 2
Basquin fitting parameters for the notched fatigue life data in Fig. 5.

Material	s	b
high-density (CNT-L)	1410	-0.13
low-density (CNT-T)	1330	-0.13
SH 316L [15]	1840	-0.13
SH 304L [13]	1810	-0.14
AN 316L [13]	1690	-0.14

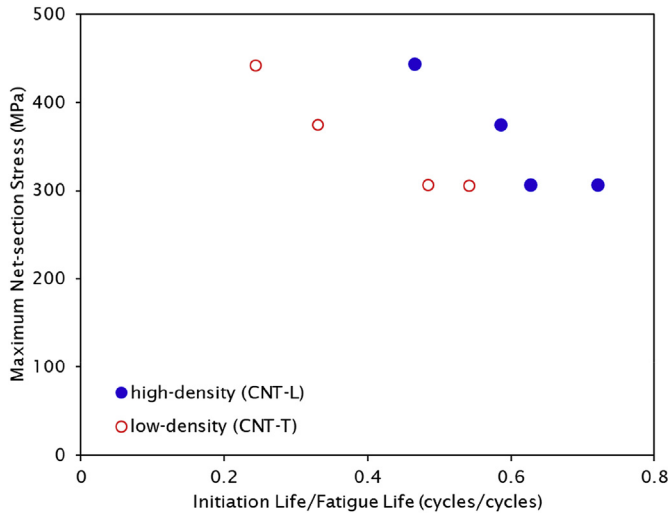


Fig. 6. The fraction of the total fatigue life spent initiating a fatigue crack determined by the direct current potential difference (DCPD) technique as a function of the maximum net-section stress at the notch root (S_{Nmax}) for the high-density (CNT-L) and low-density (CNT-T) specimens.

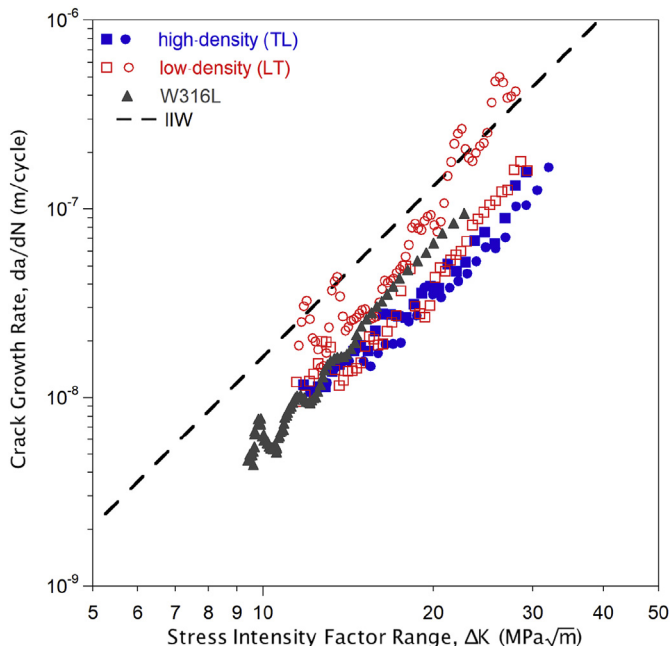


Fig. 7. Fatigue crack growth rate (da/dN) as a function of stress intensity factor range (ΔK) for the TL and LT compact tension specimens. Closed blue symbols and open red symbols indicate, respectively, higher and lower density of the compact tension specimens. Circle and square symbols differentiate the side-by-side specimens. The crack growth rate measured for a wrought 316L material [18] and a generic crack growth rate curve for steel from the International Institute of Welding (IIW) [19] are given for reference. (For interpretation of the references to colour in this figure legend, the reader is referred to the Web version of this article.)

elongation is limited by the presence of LoF defects, since the impact of LoF defects likely scales with the inverse of the density. While the spherical pores tend to deform and blunt in relative isolation, the microcrack-like LoF defects are more likely to cause stress concentrations that can drive extension and coalescence of the microcrack-like features. Other aspects of the process (such as the successive layering of material) did not have an obvious effect on the fracture surfaces of the DED materials in this study. Therefore, the fracture evolves under conditions promoting local crack extension in the low-density material, which limits uniform elongation and results in lower S_u . As described below, microcracking induced by LoF defects strongly influences fatigue crack initiation and fatigue life, while the effect on crack propagation is modest in comparison.

4.2. Crack initiation

The role of LoF defects as microcracks is particularly apparent in the crack initiation behavior of the DED 304L (Fig. 6). Profiles of the fracture surface near the notch root in Fig. 8 do not reveal LoF defects in the high-density material (Fig. 8a) and the fracture surface is nominally on a plane that extends through the notch root. The character of the fatigue fracture surface in the high-density (CNT-L) specimens is consistent with the relatively flat (fatigue fracture) surface morphology observed in fully-dense wrought austenitic stainless steels [10], and suggests cracks initiate near the surface in the notch root. In contrast, the fatigue crack in the low-density (CNT-T) specimens (Fig. 8b) follows a tortuous path, often extending well beyond the plane that defines the notch root. The presence of partially melted powder particles on the fracture plane of the low-density material distinguishes the irregular fracture surface and the LoF defects. Furthermore, the presence of secondary fatigue cracks away from the notch root and in the vicinity of LoF defects indicates that fatigue cracks initiated distant from the notch root at these defects in the low-density material.

In essence, LoF defects act as starter notches that nucleate fatigue cracks. Thus, for the same applied stress, crack initiation is enhanced in the low-density materials compared to the high-density materials (or in the presence of a high concentration of LoF defects compared to a low concentration of LoF defects). It is clear from fracture profiles that LoF defects nucleate numerous fatigue cracks consistent with the stochastic nature of the LoF defect distribution in the material. These nucleation events eventually coalesce to initiate a propagating fatigue crack. In other words, with greater number and larger size of LoF defects, microcracks are nucleated and coalesce into a growing fatigue crack earlier in the fatigue life than in the high-density material (which is characterized by spherical pores).

4.3. Fatigue crack growth

The fatigue crack growth behavior of DED 304L is generally consistent with that of wrought austenitic stainless steels. Indeed, the overall trends in FCGR for the DED 304L specimens are mostly consistent with the fatigue behavior of conventional materials (e.g., IIW FCGR curve [19] and wrought 316L [18]) with the exception of one of the low-density (LT) compact tension specimens. Post-mortem inspection of the fracture surfaces in Fig. 9 shows that small amounts of distributed, fine spherical porosity (Fig. 9b) did not manifest an obvious influence on measured FCGR, since these do not seem to evolve into microcracks. Small ($< 100 \mu\text{m}$ diameter), spherical pores were distributed on the fracture surface of the high-density material (Fig. 9b) but these defects did not change the appearance of the crack surface. However, the non-monotonic character of the FCGR (Fig. 7) for one of the specimens seems to have occurred in the proximity of a high concentration of LoF defects (Fig. 9c). The stress concentration induced by the LoF defects locally increases the driving force (ΔK) for fatigue crack growth, thus increasing the average FCGR in the vicinity of these (local) defects. Local regions of high LoF defect populations roughly

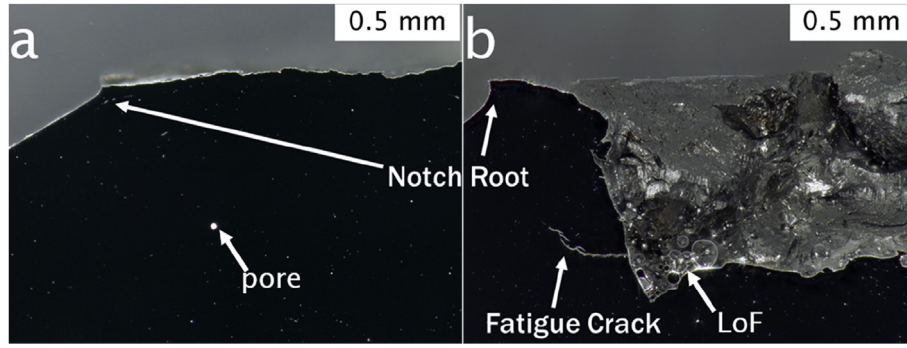


Fig. 8. Confocal micrographs of polished cross-sections of fractured (a) high-density (CNT-L) and (b) low-density (CNT-T) specimens after fatigue cycling. Arrows highlight LoF defects, pores, the notch root, and secondary fatigue cracking.

corresponded to the peaks in the FCGR, as shown in the higher magnification micrographs (Fig. 9c) of the fracture surface where $\Delta K > 20 \text{ MPa}\cdot\text{m}^{1/2}$. Once the crack has extended through the local areas of high LoF defect concentration, the FCGR drops toward the baseline behavior until the crack encounters additional clusters of LoF defects, thus creating the ratcheting behavior observed in Fig. 7 (especially for the open circles at high ΔK). The subtle and somewhat inconsistent influence of LoF defects on FCGR compared to the trends in tensile properties is likely due to the volume of material activated in the test: in the FCGR test, crack propagation is highly localized and constrained to the plane bisecting the specimen, while in tensile tests the entire gauge volume is being activated and fracture can evolve from a comparatively larger volume of material and defect population.

4.4. Fatigue life

Fatigue life measurements with notched cylindrical specimens represents the sum of the life associated with crack initiation and the life associated with crack propagation. This is evident in Fig. 6, where the initiation life represents anywhere between $\frac{1}{4}$ and $\frac{3}{4}$ of the total life (crack propagation being the balance). Although the slope of the modified Basquin relationship is similar for the DED materials in this study and wrought austenitic stainless steels from the literature (Table 2, b values), the significant variation in the stress intercept (s) of these materials suggest an important role of strength and strain

hardening characteristics on fatigue life. This idea is consistent with evidence that fatigue crack initiation life of wrought austenitic stainless steels improves with the strength of the material [11]. The greater intercept value (s) of the strain-hardened compared to the annealed materials is a clear indication of this trend, while the shape of the Basquin fit (slope, b) is not changed. The lower intercept of the low-density (CNT-T) specimens quantitatively demonstrates diminished fatigue life compared to the high-density (CNT-L) specimens, which is consistent with the lower ultimate strength of the low-density tensile specimens (Fig. 4). To generalize and compare with literature, it is necessary to account for specimen geometry (i.e., smooth versus notched) and loading conditions (i.e., load ratio, R). An equivalent stress amplitude (S_{ar}) for zero mean stress (fully-reversed loading) in a smooth specimen can be approximated by the Smith-Watson-Topper (SWT) method [20,21], which when modified for the effect of the stress concentration of the notched geometry [20,21] becomes:

$$S_{ar} = S_{Nmax} K_f \sqrt{\frac{1-R}{2}} \quad (3)$$

where K_f is the stress concentration factor associated with fatigue from Peterson's relation [12]:

$$K_f = 1 + \frac{K_t - 1}{1 + \frac{\alpha}{r}} \quad (4)$$

where r is the notch radius and α is a material constant that depends on

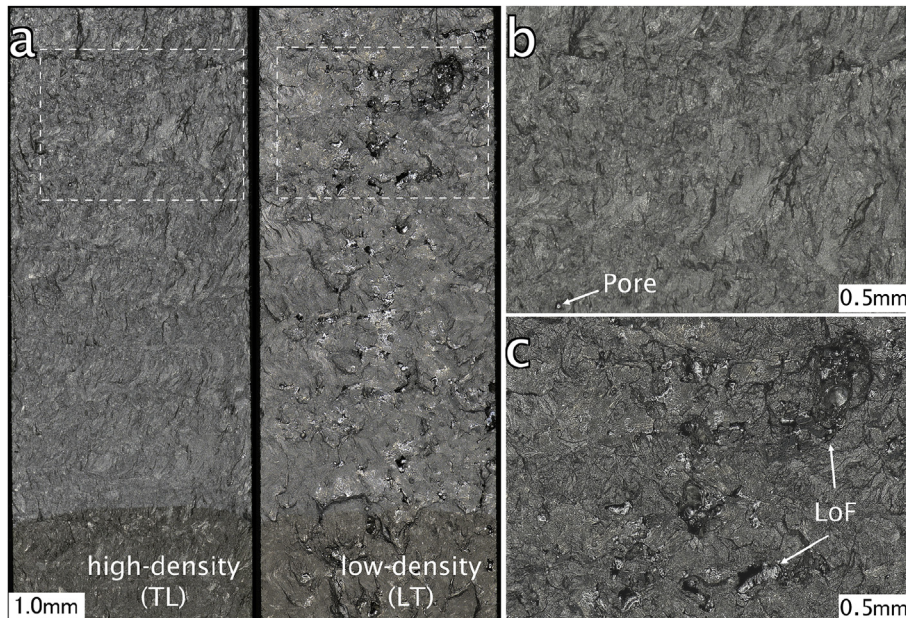


Fig. 9. (a) Fracture surfaces of the high-density (TL) and low-density (LT) compact tension specimens (crack growth from bottom to top of the image). Higher magnification micrographs of the fracture surface regions outlined in white in (a) and with arrows indicating pores and LoF defects: (b) high density (TL) and (c) low-density (LT).

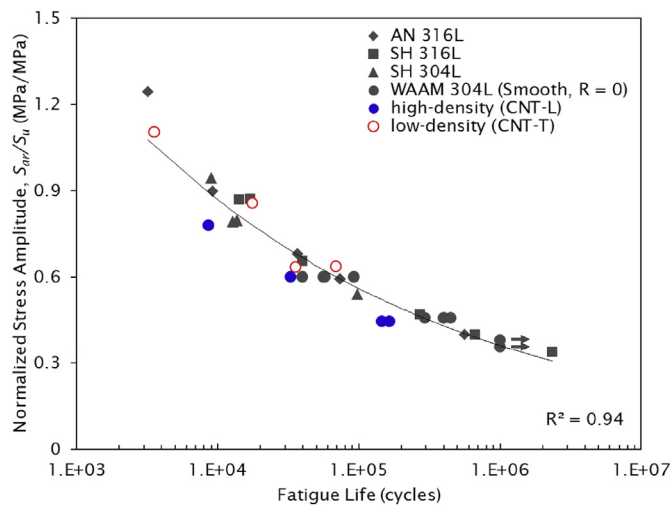


Fig. 10. The approximated smooth specimen stress amplitude (zero mean stress) normalized by ultimate tensile strength (S_{ar}/S_u) for the high-density (CNT-L) and low-density (CNT-T) specimens plotted with data from the literature for annealed wrought 316L (AN 316L) [13], strain-hardened 316L (SH 316L) [15], strain-hardened 304L (SH 304L) [13], and WAAM 304L (smooth specimen, $R = 0$) [5].

ultimate tensile strength. When S_{ar} is normalized by S_u , the fatigue life data collapse onto a single trend line (Fig. 10). This empirical normalization accounts for the observed effect of S_u on fatigue life, and in the case of the DED materials in this study indirectly accounts for the density. To emphasize the value of this normalization, data for WAAM 304L is also plotted in this figure [5]. These data suggest that the influence of defects on fatigue is reasonably captured by the change in ultimate tensile strength. In short, the measured S_u can be used to infer the effect of defects on the fatigue life of AM austenitic stainless steels, assuming baseline fatigue behavior for wrought materials (i.e., in the limit of no AM defects, fatigue behavior is equivalent to wrought materials).

5. Summary and conclusions

We have shown that DED materials reflect the fatigue behavior of wrought materials when processing defects are minimized and the materials are nominally fully dense ($> 98.5\%$). When significant defects are present, the fatigue behavior of DED austenitic stainless steel materials depends on the characteristic of these defects. In this study, two defect populations were identified (1) relatively small, equiaxed pores and (2) comparatively large, irregular-shaped two-dimensional LoF defects. These defect populations could be roughly correlated with the relative density of the as-manufactured material: high-density materials ($> 98.5\%$ dense) characterized by pores, and low-density materials ($< 98.5\%$) characterized by LoF. While low concentrations of spherical pores did not have a strong effect on tensile and fatigue properties, high concentrations of the large, two-dimensional LoF defects significantly degraded the tensile and fatigue behavior of the DED materials. In particular, the following trends were identified for DED 304L:

- Ultimate tensile strength correlated with measured material density: material with the highest density ($> 99\%$) displayed the best tensile properties with elongation greater than 60% and ultimate tensile strength over 650 MPa, while material with the lowest density ($< 98\%$) displayed elongation less than 20% and ultimate tensile strength less than 500 MPa.
- Fatigue crack initiation was facilitated by the presence LoF defects, while small, equiaxed pores had no obvious effect on crack

initiation.

- Fatigue crack propagation was consistent with wrought materials except when a high concentration of LoF defects existed in the plane of crack extension; LoF defects locally enhanced crack growth due to the stress concentration associated with the defects, resulting in locally higher fatigue crack growth rates.
- The fatigue life was consistent with wrought materials when normalized by the ultimate tensile strength.

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Data Availability

The raw/processed data required to reproduce these findings cannot be shared at this time due to technical or time limitations.

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