FATIGUE AND CREEP-FATIGUE ACCEPTANCE CRITERIA FOR AM 316 STAINLESS STEEL

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FATIGUE AND CREEP-FATIGUE ACCEPTANCE CRITERIA FOR AM 316 STAINLESS STEEL

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SUMMARY

316 stainless steel (SS) used in nuclear applications are subjected to high temperature low cycle fatigue (LCF) and creep-fatigue (CF) cycles. Conventionally manufactured 316 SS at this high stress-temperature environment have been studied by many with the use of CF damage interaction diagrams featured in various high temperature code cases: RCC-MRx, ASME, R5, etc. On the other hand, material acceptance challenges are present for additively manufactured (AM) materials. These materials can have large variation in microstructure depending on AM methodology, machine, and building specifications. This can lead to material behaviors that are different from materials manufactured through conventional methods. Currently, no material acceptance criteria for high stress-temperature environments have been placed on AM components.

This study explores the LCF and CF behavior of directed energy deposited (DED) AM 316H SS to develop a rapid test material acceptance criteria for this type of AM material. Preliminary LCF and CF studies on wrought 316L SS are conducted at a temperature range of 550 to 700°C, where a strong focus is placed on a testing temperature of 650°C. Tensile or compressive peak dwells of up to 30 minutes are used. Preliminary test results are used to establish a test matrix for DED AM 316H SS of varying build parameters. Tested samples are analyzed with the time fraction (TF) and ductility exhaustion (DE) damage life prediction models and microscopy techniques in the mesoscopic-microscopic scale to develop the material acceptance criteria.
Austenitic stainless steel (SS), such as 316 SS, have been favored in high temperature environments due to their phenomenal high temperature tensile and creep strength as well as corrosion resistance. Along with these properties, these materials are easy to form and weld, hence are favored in nuclear applications. Currently, a large portion of in-core and out-of-core components are fabricated with type 316 SS [1]. Within these reactors, long term service under high temperature, stress, and irradiation environments cause irradiation creep damage [2]. On the other hand, thermal gradients from start-up, shut-down, and stabilized power phases can inhibit cyclic thermal stress [3]. The two damage mechanism can cause interact and cause more deleterious creep-fatigue (CF) damage [3]. With the design of the new Generation IV nuclear reactors under progress, research and development of a material capable of withstanding demanding environments and accomplishing long term service are crucial.

Understanding these challenges, much attention has been brought to additive manufacturing (AM). This manufacturing method, compared to non-AM methods, are capable of producing components with higher strength and ductility [4]. Although AM may appear as a material solution when designing the Generation IV power plants, skepticisms are present. AM components have a large scatter in material property and microstructural data [4]. These scatters can arise from different printing methods and parameter used when printing [5]. With no universal standards between suppliers, AM systems, and machines for specific high temperature use, qualification of these materials are complex to say the least.
Realizing these challenges, this research examines the low cycle fatigue (LCF) and CF behavior of type 316 SS at a temperature range of 550 to 700°C to establish a technical basis for an accelerated testing method to support near-term nuclear acceptance of these components.
CHAPTER 2. BACKGROUND

2.1 Austenitic Stainless Steel

Stainless steels are used in a variety of settings from daily tools such as silverware or in environmentally demanding settings such as nuclear fusion reactors. Specific to nuclear fusion reactors, SS are used in water reactors or fast reactors which currently operate at a maximum temperature near 330°C and 550°C respectively [6, 7]. The wide use of this material is attributed to its acceptable strength in both room and elevated temperatures, corrosion resistance, high melting temperature near 1400°C, and formability [8].

Steels are categorized as stainless when the chromium content exceeds 11 wt% [9]. Corrosion resistance properties are produced from chromium. Four categories, or families, of SS are available. These include ferritic, martensitic, austenitic, and duplex. Austenitic grades, denoted as 300 series, are non-magnetic and characterized by their face-center-cubic austenite grains. They also contain a total chromium-nickel content of at least 23 wt% [10]. The various grades of some of austenitic grades are shown in Figure 1. The chemical compositions of commonly used austenitic SS are shown in Table 1. Type 304 and 316 SS are the most commonly used as they have good balance between strength and corrosion resistance. 304 SS is used in general applications such as silverware or moderately aggressive corrosive environments. 316 SS on the other hand contains more molybdenum and nickel, making it preferable for use in high temperature environments. The addition of nickel and nitrogen contribute resistance against corrosion by increasing austenite stability [9].
Figure 1: Austenitic SS [9]
Table 1: Material composition of various 316 SS [8, 9]

<table>
<thead>
<tr>
<th></th>
<th>C</th>
<th>Cr</th>
<th>Mn</th>
<th>Mo</th>
<th>Ni</th>
<th>P</th>
<th>S</th>
<th>Si</th>
<th>N</th>
</tr>
</thead>
<tbody>
<tr>
<td>201</td>
<td>&lt;0.15</td>
<td>16-18</td>
<td>5.5-7.5</td>
<td>-</td>
<td>3.5-5.5</td>
<td>0-0.06</td>
<td>&lt;0.03</td>
<td>&lt;1</td>
<td>0-0.25</td>
</tr>
<tr>
<td>304</td>
<td>&lt;0.08</td>
<td>18-20</td>
<td>0-2</td>
<td>&lt;0.3</td>
<td>8-11</td>
<td>0-0.045</td>
<td>&lt;0.03</td>
<td>&lt;1</td>
<td>-</td>
</tr>
<tr>
<td>316</td>
<td>&lt;0.08</td>
<td>16-18</td>
<td>0-2</td>
<td>2-3</td>
<td>10-14</td>
<td>0-0.045</td>
<td>&lt;0.03</td>
<td>&lt;1</td>
<td>-</td>
</tr>
<tr>
<td>316L</td>
<td>&lt;0.03</td>
<td>16-18</td>
<td>0-2</td>
<td>2-3</td>
<td>10-14</td>
<td>0-0.045</td>
<td>&lt;0.03</td>
<td>&lt;1</td>
<td>-</td>
</tr>
<tr>
<td>316H</td>
<td>0.04-0.1</td>
<td>16-18</td>
<td>0-2</td>
<td>2-3</td>
<td>10-14</td>
<td>0-0.045</td>
<td>&lt;0.03</td>
<td>&lt;0.75</td>
<td>-</td>
</tr>
<tr>
<td>316LN</td>
<td>0-0.03</td>
<td>16-18</td>
<td>0-2</td>
<td>2-3</td>
<td>10-14</td>
<td>0-0.045</td>
<td>&lt;0.03</td>
<td>&lt;1</td>
<td>0.1-0.16</td>
</tr>
<tr>
<td>316FR</td>
<td>0.008-0.012</td>
<td>16.5-17.1</td>
<td>0.83-0.95</td>
<td>2.07-2.3</td>
<td>10.6-11.4</td>
<td>0.02-0.027</td>
<td>0.003-0.005</td>
<td>0.47-0.54</td>
<td>0.068-0.09</td>
</tr>
</tbody>
</table>

The letter(s) that follow the SS grade represent chemical adjustments made to the base alloy. For example, changes can be made to carbon content to increase corrosion resistance or material strength. Relative to the base alloy, “H” is denotes an increase in while “L” signifies a reduction in carbon. The amount of carbon is connected with chromium carbide precipitation, M_{23}C_{6}, which form at high temperature. Formation of carbides cause chromium depletion along grain boundaries, reducing corrosion resistance [11-13].

Figure 2 describe the time needed for carbides to sensitize with respect to temperature and carbon content. Carbides precipitate between 450 to 900°C and precipitate more quickly with higher carbon contents [9].
The 316L alloy is expected to have stronger corrosion resistance than its base alloy, as fewer carbides will precipitate. On the other hand, chromium depleted zones can form unstable austenite which can undergo martensitic formations. With more carbon, 316H can undergo more martensitic transformations. As martensite is stronger than austenite, 316H can have higher tensile, yield, and creep strength than its base alloy, although corrosion resistance is reduced [9]. Other precipitates such as σ-phase, X-phase, laves-phase, α and δ-ferrite, can form, although the carbide precipitates will always appear first [14].

Nitrogen can also be added to further increase corrosion resistance. This change is indicated with the letter “N”. Nitrogen is increased while decreasing carbon to limit precipitation of M₂₃C₆ and hence reduce grain boundary corrosion [9].
alloying elements are indicated by a combination of letters. Low carbon composition alloys with added nitrogen for example are indicated with letters “LN”. On the other hand, “FR” is representative of a narrower alloy composition within the 316 composition range designed specifically for fast reactors. 316FR is similar in composition to the LN alloys, but with further reduction in carbon [15, 16].

2.2 Additive Manufacturing

As defined by ASTM, AM is the process of joining materials to makes parts from 3D model data, usually layer upon layer, as opposed to subtractive manufacturing and formative manufacturing methods [17]. The concept of AM was first featured in several patents during the 1980s [18]. This manufacturing process is attractive because of its robust manufacturing capabilities and potential to reduce or simplify steps taken in non-AM manufacturing methods. Technological advancements relating to the production of complex geometries and material quality has also lend to its increase in popularity within recent years [19, 20].

The general AM process is outlined in Figure 3. The process begins with the construction of a 3D CAD model. Considerations of including supporting structures may be made when structural overhangs exist. The CAD file is then converted to a desired file format, typically an STL file, then spliced via software to determine a layer-by-layer or line-by-line manufacturing path. Upon construction of a manufacturing path, appropriate manufacturing parameters are chosen relative to the AM process. Once the structure has
been printed, post-processing may take place. This may include removal of supporting structures, heat treatments, subtractive machining, and/or surface treatments [20, 21].

![Diagram](image)

**Figure 3: Basic numerical chain for AM [21]**

ASTM classifies the AM processes into seven categories [17]. These are provided in Table 2. Of these AM methods, focus will be placed on directed energy deposition (DED) and powder bed fusion (PBF), which are two common methods for fabricating 316 SS parts.
Table 2: AM process categories [17]

<table>
<thead>
<tr>
<th>Method</th>
<th>Definition by ASTM/ ISO</th>
</tr>
</thead>
<tbody>
<tr>
<td>Binder Jetting</td>
<td>Liquid bonding agent is selectively deposited to join powder materials</td>
</tr>
<tr>
<td>Directed energy deposition</td>
<td>Focused thermal energy is used to fuse material by melting as they are being deposited</td>
</tr>
<tr>
<td>Material Extrusion</td>
<td>Material is selectively dispensed through a nozzle or orifice</td>
</tr>
<tr>
<td>Material jetting</td>
<td>Droplets of build material are selectively deposited</td>
</tr>
<tr>
<td>Powder bed fusion</td>
<td>Thermal energy selectively fuses regions of a powder bed</td>
</tr>
<tr>
<td>Sheet lamination</td>
<td>Sheets of material are bonded to form a part</td>
</tr>
<tr>
<td>Vat photo polymerization</td>
<td>Liquid photopolymer in a vat is selectively cure by light-activated polymerization</td>
</tr>
</tbody>
</table>

2.2.1 Additive Manufacturing Methodologies

2.2.1.1 Directed Energy Deposition

DED is a line-by-line layering AM process where concepts from cladding and welding are utilized [22]. The printing schematic of this method is shown in Figure 4 for powder feedstock and Figure 5 for wire feedstock. The printing process begins by depositing powder or wire feedstock, shielding gas, and thermal energy simultaneously, forming a material melt pool. The shielding gas, typically argon or nitrogen, is used to
minimize oxidation [5, 23]. It is also used to control the material flow for powder feedstock. Upon deposition, the melt pool rapidly cools and solidifies onto the build plate or substrate. The process is repeated through a line-by-line layering scheme until the final component is formed.

Figure 4: Powder feeding methods for DED processes [24]

Figure 5: Wire feeding method for DED processes [24]
Process classifications within DED are based on feedstock type which then give way to the different feedstock-thermal energy combinations. These processes are summarized in Table 3. Powder feedstock can achieve finer resolution and good dimensional accuracy but is typically more expensive and will take longer to print than wire feedstock. These properties make this feedstock a suitable candidate for components with small geometry [5]. Wire feedstock on the other hand is commercially abundant, making the purchase cost lower than powder. High deposition rates and taller layer thickness contribute towards shorter print times, but with less accurate deposition geometries compared to powder feedstock. Wire feedstock is suited for components with large geometries [5]. Structures manufactured from either feedstock require secondary machining, surface grinding or polishing, if a surface roughness less than 25μm is required [23].
Table 3: DED processing parameters [5, 24]

<table>
<thead>
<tr>
<th>Process</th>
<th>Feedstock</th>
<th>Typical Layer Thickness (µm)</th>
<th>Minimum Feature Size (µm)</th>
<th>Heat Flux Density (W/mm²)</th>
<th>Scanning Speed (mm/s)</th>
<th>Dimensional Accuracy (mm)</th>
<th>Deposition Rate (g/min)</th>
</tr>
</thead>
<tbody>
<tr>
<td>Laser Additive Manufacturing (LAM)</td>
<td>Powder</td>
<td>200-500</td>
<td>380-1000</td>
<td>≈10⁶</td>
<td>5-20</td>
<td>0.5-1.0</td>
<td>&lt;8.3</td>
</tr>
<tr>
<td>Wire Laser Additive Manicuring (WLAM)</td>
<td>Wire</td>
<td>&gt;1000</td>
<td>5-15 times wire diameter</td>
<td>≈10⁶</td>
<td>-</td>
<td>-</td>
<td>1.5-48.0</td>
</tr>
<tr>
<td>Wire Electron Beam Additive Manufacturing (WEAM)</td>
<td>Wire</td>
<td>&lt;3000</td>
<td>&lt;1600</td>
<td>≈10⁸</td>
<td>1-10</td>
<td>1.0-1.5</td>
<td>&lt;330</td>
</tr>
</tbody>
</table>

2.2.1.2 Powder Bed Fusion

A schematic of the PBF process is shown in Figure 6. PBF forms structures by first spreading a powder layer across an area using a powder spreader. Then a thermal energy source is emitted along a selective path to melt the powder. Upon cooling, a solidified material layer is formed. Powder fusion can occur from various mechanism depending on the PBF process being implemented. In the case of selective laser melting, the powder is
fully melted into a melt pool, then fused upon cooling and solidification [5, 25]. The powder bed is then lowered and the process is repeated until the component is fully built [26]. It is typical for the printing process to take place in an inert gas or vacuum environment to protect the component from oxidation. Argon or nitrogen gas are commonly used, typically flowing from nozzles from one edge of the build chamber and exit on other side of build chamber [26].

Figure 6: PBF process schematic [8]
Table 4 summarize the typical processing parameters ranges in PBF. PBF production time is time consuming with restriction on thermal energy input and layer size. The geometry of the printed structure is also limited to the size of the build plate and how much this build plate can be lowered. However, complex geometries with tighter dimensional tolerances can be achieved [27]. These qualities are influenced from powder size and quality. Finer powders are capable of producing geometrically accurate structures with smoother surfaces. Downsides are seen from difficulties in the handling and spreading of powder. Coarse powders are easier to spread but can yield poor geometric accuracy and surface finishes [5, 25].

Table 4: PBF processing parameters [5, 27, 28]

<table>
<thead>
<tr>
<th>Feedstock</th>
<th>Powder</th>
</tr>
</thead>
<tbody>
<tr>
<td>Heat source</td>
<td>Laser or Electron Beam</td>
</tr>
<tr>
<td>Power (W)</td>
<td>50 – 1000</td>
</tr>
<tr>
<td>Scanning Speed (mm/s)</td>
<td>10 – 1000</td>
</tr>
<tr>
<td>Dimensional Accuracy (mm)</td>
<td>0.04 – 0.2</td>
</tr>
<tr>
<td>Typical Layer Thickness (μm)</td>
<td>20 – 100</td>
</tr>
<tr>
<td>Surface Roughness (μm)</td>
<td>7–175</td>
</tr>
</tbody>
</table>
2.2.2 PBF and DED Structures

An advantage of using AM processes is the wide structural variety. Geometry, material quality, and material performance are highly controllable. On the other hand, the process can also produce unwanted and potentially detrimental material defects unique to a given AM process. Heterogeneous microstructures may also be formed. Challenges of this technology, hence, comprises of optimizing printing parameters. Relative to powder based AM processes such as DED and PBF, these parameters include but are not limited to: material deposition rate, thermal energy power, scanning speed, printing environment, feedstock type, and hatch spacing. The material used and geometry of the printed structure will also be impactful [24, 29].

2.2.2.1 Porosity

Presence of pores in metallic material can be detrimental with respect to its mechanical properties, particularly fatigue and creep, which are controlled by local features in the microstructure. Pores create local stress concentrations that can promote fatigue crack nucleation and accelerate crack propagation. Pores can also be influential in how a material ruptures in creep [30]. Three distinct pore types may be found from parts fabricated by powder-based AM processes. These include keyhole-induced pores, lack-of-fusion pores, and gas-induced pores [5]. Images of these pores are shown in Figure 7.
For many alloys formed by DED, porosity formation result from inadequate optimization of thermal energy with powder feed rates as illustrated in Figure 8. Linear heat input is defined as the ratio between heat source energy and scanning speed [5]. The plot identifies distinct zones where pores can form more readily. Lack-of-fusion porosity occurs when deposited powders are insufficiently melted. The root cause of this defect may by excess powder deposition, weak thermal energy input, or large hatch spacing [28, 31]. Keyhole porosity forms when thermal energy is excessive and elements within the melt pool vaporize and form vapor cavities. This can occur when the heat input is high relative to powder deposition rate. In this condition, the thermal energy source is also able to affect deeper layers as seen in Figure 7(a). Repetitive formation and collapse of vapor cavities introduce these nearly spherical voids along the laser beam path [32]. Porosity may also become prevalent when non-optimized intermediate heat input and powder feed rate are used. Pores are formed from entrapment of shielding gas or alloy vapors produced by the
melt pool [5, 33]. These gas-induced pores are typically smaller in size than the lack-of-fusion or keyhole porosity and hence less detrimental to mechanical properties.

Figure 8: Optimal processing parameter map for laser beam DED (DED-LB) process of different alloys [24].

2.2.2.2 Microstructure Heterogeneity

A key quality of powder-based AM is the formation of microstructures with strong microstructural heterogeneity. These structures can have dislocation density and grain boundary strength which differ from their wrought form, producing materials with elevated
or lowered material strength. Anisotropic responses to applied loads may also be observed [24, 34-38]. Figure 8 presents electron backscatter diffraction (EBSD) and light optical microscopy (OM) images from 316L plate manufactured by DED-LB. Images of annealed wrought 316L is also included for comparison.

![Figure 9: Light OM (left) and EBSD maps (right) of annealed wrought (top) and as-built DED-LB (bottom) 316L SS [39].](image)

Comparing the OM images, the AM as-built material differ considerably from the wrought material. Overlapping “fish-scale” melt pool patterns and melt pool boundaries
highlight the direction of material deposition. The “fish-scale” features are formed from the Gaussian-distributed thermal energy source. Maximum heat transfer occurs at the center of the thermal energy source. Amount of heat transfer reduces further away from this location [40]. Viewing the microstructure at a finer scale using EBSD maps, columnar grains with varying grain sizes are found oriented along the print direction in as-built AM materials. Equiaxed microstructure is seen by the wrought material [41]. These columnar structures form from and orient along the steep localized thermal gradients [42, 43].

Control over the temperature gradient at the solid-liquid interface, G, and local solidification growth rate can yield other unique solidification structures. This is illustrated using a temperature gradient-growth rate plot shown in Figure 10. The ratio G/R defines the structural morphology while product GxR determines the size of the solidified structure. As G/R increases, the equiaxed dendritic structures can transform and become columnar dendritic, cellular, or planar. With increasing GxR, the sizes of these structures become smaller [5, 44, 45].
2.3 Mechanical behavior of 316 SS

2.3.1 Monotonic Mechanical Behavior

The popularity of 316 SS not only comes from its exceptional corrosion resistance, but also from its adequate mechanical properties. These mechanical properties are highly influenced by the manufacturing process and post manufacturing heat treatments. The variability in mechanical property with respect to different manufacturing processes and heat treatments are described in Figure 11 in the form of a rupture uniform elongation.
against yield strength plot. Material properties of as-built PBF-LB AM 316L SS are shown in this figure.

Figure 11: Uniform elongation-yield stress of 316L at room temperature. Gray, blue, and red areas are conventionally manufactured, high performance, and AM alloys, respectively. Green and yellow areas highlight material response from annealed and plastically deformed conventionally manufactured material, respectively [4].

Figure 11 can be sectioned into three zones: conventionally manufactured, high performance alloy, and AM. The non-AM, conventionally manufactured zone is shown to have the largest variability in properties and displays a strong strength-ductility tradeoff
character. This occurs from the vast combinations of plastic deformation inducing process and heat treatments. Wrought materials for example are subjected to plastic deformation inducing processes to increase material strength [9]. The processes are largely categorized into two categories, cold or hot work. Cold work produces plastic deformation at room temperature. This increases material strength at the expense of ductility. Hot work introduces plastic deformation at temperatures above recrystallization temperatures. Strength does not increase and ductility is not lost to the extent of cold work. Cold or hot work deformation processes include but are not limited to rolling, forging, extrusion, and or drawing [46]. When annealing heat treatments are performed, hardness decreases but rejuvenates material ductility by removing defects introduced from plastic deformation processes. Grain recrystallization will also occur during the process, contributing towards ductility recovery [46, 47].

Strength-ductility tradeoff can be diminished with applications of special thermo-mechanical treatments, as seen in the High Performance zone in Figure 11. These alloys can contain nano-scale twin bundles attained from dynamic plastic deformation. An image of nano-scale twin bundles is shown in Figure 12. Hardness increases and ductility is retained from grain boundaries dislocation pinning mechanisms [48].
The AM materials reported in Figure 11 had exceptional strength and ductility. The AM material was reported to have solidification cellular walls containing high density of dislocations. The solidification cells had a cell size of less than 1µm. The superior material properties were formed from the interaction of solidification cellular structures and deformation twins [4, 50]. The solidification cellular structures increase material strength by preventing dislocation motion. These structures were retained at elevated strains. Deformation twins contributed to produce a stable strain hardening behavior, leading to heightened tensile elongation [49].

Strength and ductility of 316 SS also evolve with temperature. Figure 13 and Figure 14 plots the Young’s modulus and elastic limit, labeled as yield strength, with respect to temperature for annealed 316L SS. Material strength decreases while ductility increases with increasing temperature.
Figure 13: Young’s modulus against temperature of wrought-annealed 316L SS[8]

Figure 14: Yield strength against temperature of wrought-annealed 316L SS[8]
2.3.2 Fatigue

Fatigue damage is inflicted when a component is subjected to repeated loading [51]. In strain-controlled low cycle fatigue (LCF), the yield strength is exceeded and cyclic plastic strains accumulate each cycle. Cyclic loads can appear from temperature gradient induced cyclic thermal stresses formed from startup and shutdown cycles of the power plant [52, 53].

Material behavior of 316 SS in LCF can be described using a maximum stress-cycle plot. Maximum stress-cycle plots constructed from fully reversed LCF tests with a strain ratio ($R_e$) of -1 on 316H SS at 550°C with different strain ranges is shown in Figure 15.

Figure 15: Maximum stress response from LCF testing on 316H SS at 550°C, strain rate of $1 \times 10^{-3}$ 1/s, and different strain amplitudes. Material extracted from pipe and undergone annealing at 1050 – 1100°C for 18 min. [54].
During LCF, 316 SS exhibits three distinct behaviors: initial cyclic hardening to a maximum stress, a steady decline in the maximum stress, then rapid stress decline to rupture as the crack is growing. Each behavior occurs from dislocation accumulation and annihilation interactions. During initial cyclic hardening, the rate of dislocation accumulation is higher than annihilation. Dislocation density increases and interactions between the dislocations become more prominent [55]. Structures such as dislocation tangles can form and impede dislocation motion, resulting in cyclic hardening [56]. Dislocation accumulation and annihilation are balanced during the stable or quasi-stable phase and dislocation density remains relatively constant. As the annihilation rate overcomes the accumulation rate, cross-slip and secondary slip systems are activated, promoting the rearrangement of dislocations into unique lower-energy structures [51, 57]. Dislocation walls and channels first form, then evolve into equiaxed cellular structures. Cellular structures have low resistance to plastic deformation and soften the material [54, 55]. Rapid stress decline occurs with the formation of microcracks. Microcracks form on the free surface of the material due to the activation of persistent slip bands [58]. The cracks formed on the material surface propagate inwards in a transgranular manner, leading the material to rupture [59]. The three-stage behavior is accelerated by increasing the strain range as illustrated in Figure 15. Increasing the strain range also results in an increase in the maximum peak stress.

Stress-strain hysteresis loops at different cycles can be compared to describe additional LCF trends. Hysteresis loops corresponding to Figure 15 are shown in Figure 16.
Figure 16: Hysteresis loops of annealed 316H SS at 550°C, strain rate of $1 \times 10^{-3}$ 1/s, and different strain amplitudes. Material extracted from pipe and undergone annealing at 1050 – 1100 °C for 18 min. [54].

Other test parameters such as temperature or strain rate can also influence LCF behavior. Effects of test temperature on an initially annealed 316LN is illustrated in Figure 17. The fatigue life generally decrease as the test temperature increase. When viewing the effects from temperature with respect to stress amplitude, interesting trends appear. The highest stress amplitude occur at the lowest temperature, 25°C. As the temperature increase to 300°C the stress amplitude become smaller than at 25°C. The reduction in stress...
amplitude is reconciled with decreasing yield strength with increasing temperature. Decrease in fatigue life can also be attributed towards the reduction in strength, as the material becomes more susceptible to plastic deformation. Environmental effects such as oxidation can also accelerate surface microcrack formation [60, 61]. However, the stress amplitude increases when the test temperature is further elevated to 600°C. The stress amplitude falls to a minimum at a temperature of 650°C [62]. This anomalous hardening behavior is explained by a dynamic strain aging mechanism discussed later.

Figure 17: Influence of temperature on LCF of 316LN SS solution annealed at 1373 K (1100°C) for 1 hr followed by water quenching [62].
Effects from strain rate when solution annealed 316LN is tested at 550°C is shown in Figure 18. At the elevated temperature, a reduction in strain rate increase the stress amplitude and decrease cycles to failure. Effects from dynamic strain aging become more prominent as the strain rate decreases [63]. The material has sufficient time to undergo aging at lower strain rates.

![Figure 18: Effect of strain rate on LCF of 316LN solution annealed at temperature range of 1060°C to 1110°C for 5.5 hours [63].](image)

Dynamic strain aging occurs at specific temperature and strain rate combinations. Interstitial solute atom diffuse at dislocation cores making them less mobile resulting in
higher strength [52, 64, 65]. In the case of 316 SS, the carbon solute atoms pin gliding dislocations. Tendency for cross-slip are reduced and elevated stress is required to break away from the pinning effect for further deformation [66, 67]. Dynamic strain aging appears in the form of serrations on stress-strain hysteresis plots. An example of this was shown in the 0.50% strain amplitude plot in Figure 16.

Figure 19 shows the effects of dynamic strain aging on LCF life with respect to temperature and strain rate. In Figure 19 (a), the maximum fatigue life is attained at the 573K (300°C) test temperature. The fatigue life drops severely from 673K (400°C) to 873K (600°C) where dynamic strain aging takes place. Shown in Figure 19 (b), a decline in strain rate rapidly decrease fatigue life for the three LCF test temperatures, 773, 823, and 873K (500, 550, 600°C). Comparing the 550 and 600°C test temperature, below a strain rate of $10^{-3}$ s$^{-1}$, the fatigue life of the 550°C test exceeds the fatigue life of the 600°C test, suggesting the prominence of dynamic strain aging at the 550°C temperature with lower strain rates [62]. Literature suggests dynamic strain aging effects will occur for 316 SS at temperatures as low as 250°C to 600°C when accompanied by a strain rate lower than $10^{-4}$ s$^{-1}$ [63-65, 67, 68].
2.3.2.1 Fatigue for PBF-LB AM 316 SS

DED and PBF 316 SS components may produce different fatigue life and behaviors than non-AM components. The behavioral differences are attributed towards AM defects, unique microstructure which is often anisotropic, and interactions of these features with loading orientation [44, 50, 69, 70]. Lack of fusion defects, in particular, are detrimental to fatigue strength of DED and PBF 316 SS. Figure 20 illustrates the effects of these defects on as-built PBF-LB 316L SS in LCF. Three build orientations were investigated: horizontal, vertical, and 45 degree diagonal. Horizontal specimens are loaded
perpendicular to the build direction, vertical specimens are loaded parallel to the build direction, and diagonal specimens are loaded at a 45 degree to the build direction.

Figure 20: LCF of as-built PBF-LB 316L tested at room temperature and strain rate of $10^{-3}$ 1/s. Specimens printed at parallel (V), normal (H), and 45° angle (D) with respect to loading direction [69].

The horizontal specimens have the longest fatigue life, followed by the vertical, and finally the diagonal specimen. The difference in fatigue life between different
specimen orientations in this study depended on how the lack-of-fusion defects were aligned relative to the loading direction. Lack of fusion defects form between material build layers and appear as elongated pores. These defects are aligned perpendicular to the load direction for vertical specimens and parallel to the horizontal specimens. When perpendicular to the loading direction, the crack attains a larger projected area on the loading plane, introducing a higher stress concentration, innately lowering the fatigue life [44, 69].

LCF response between as-built PBF-LB and wrought 316L are compared in Figure 21. The wrought material was extracted from a cold-drawn 316L pipe, then solution treated at 1040°C for 2 hr in air, followed by water quenching. Little strain hardening and larger stress amplitudes are observed from PBF-LB processed specimens compared to the solution annealed specimens.

Figure 21: LCF behavior of PBF-LB and wrought 316L tested at 550°C and strain rate of $10^{-3}$ 1/s [50].
The elevated stress amplitude of PBF-LB specimens is attributed to the increased yield strength. The low angle grain boundary abundant microstructure and cellular substructures such as dendrites with high dislocation density walls, both formed during the rapid cooling phase of PBF-LB, contribute towards increasing strength. Additional hardening can come from precipitates found around cell walls, also formed during the PBF-LB process. Pinning mechanisms by these precipitates enhance material strength [36]. The annihilation of these hardening mechanisms results in the extended softening period in PBF-LB materials under LCF. Although the cyclic stress-strain behaviors comparing AM and wrought are quite different, interestingly the LCF lives are similar.

2.3.3 Creep

Creep damage is formed when a material is subjected to a load for a prolonged period [47]. For typical engineering alloys, creep damage is observed when the temperature is 30 to 60% of the absolute melting temperature [71]. For 316 SS, this corresponds to a temperature range of 450 to 900°C [72, 73]. In a tensile load-high temperature environment, creep damage is observed as a time-dependent strain increase. A creep strain against time curve shown in Figure 22 presents the creep response by a hot rolled 316 SS.
Greater stress result in smaller rupture times from the 600°C temperature force-control uniaxial creep test. Three distinct regions are present. Primary creep is observed upon loading. The material will display rapid incline in strain with time. The rapid incline is caused by high dislocation climb rates contributing to material deformation. Secondary creep behavior is observed when the minimum creep strain rate is attained, and the creep strain generally increases linearly with time. As the material is crept further, void nucleation and/or necking occurs. The material enters tertiary creep where a rapid increase in strain is seen once more until rupture [47, 71]. 316 SS is dominated by secondary creep. The minimum creep strain rate, strain rate at secondary creep, are greater at elevated stresses. This is shown in Figure 23. Data shown in Figure 22 is used to plot this figure.
Temperature influence on creep of 316H is shown in Figure 24. Increasing the temperature result in a large decrease in creep rupture life. Temperature dependence on rupture life is governed through sensitization of chromium-rich carbides, $M_23C_6$, along grain boundaries and dislocations. Figure 26 presents an example image of coarse carbides formed after 22100 hrs of creep at 600°C.
Figure 24: Creep strain of 316H SS from serviced advanced gas cooled reactor boiler. Numbers next to curves denote as test temperature(°C) and stress (MPa) [75].
Figure 25: $M_{23}C_6$ type carbide along (a) grain boundary and (b) on dislocation in intergranular regions [76].

As illustrated in Figure 2, $M_{23}C_6$ form between 450 to 900°C [9]. $M_{23}C_6$ begin as fine structures and can impede grain boundary sliding or dislocation motion, acting as a
strengthen mechanism. Long exposure times in the high temperature environment coarsen the carbides and cause chromium depletion along the grain boundary. Grain boundaries become brittle and corrosion resistance is reduced. This increases the potential for grain boundary void formation [11-13].

The rupture strain dependence on stress is described in Figure 26. For 316 SS, rupture strain increase or decrease with greater stress. Similar behavior is seen from 316LN, although the decreasing rupture strain behavior begins at a higher stress, 250 MPa.

![Figure 26: Creep rupture elongation behavior for 316 and 316LN, solution treated at 1373K (1100°C) for 1 hr, tested at 600°C [13].](image)
The change in rupture strain with applied stress occur from dislocation-dislocation and dislocation-carbide interactions. For 316 SS, at the 600°C temperature, the rupture strain is small when extremely low stresses, below 100 MPa, are applied. In this condition, a large quantity of coarse carbide precipitates form and nucleate creep cavities along grain boundaries and dislocations. This leads to low ductility intergranular rupture [13, 77]. With increasing stress, up to 200 MPa, the rupture behavior gradually become ductile and transgranular. The material matrix is strengthened from fine carbides which impede grain boundary sliding and dislocation motion [13, 78]. As stress is increased further, shorter rupture times and lower ductility result from intergranular creep fracture [13, 75, 79-81]. At this load, fewer carbides precipitate. Grain boundary sliding and dislocation motion are facilitated.

The rupture elongation response for 316LN is shifted to the right along the stress axis in Figure 26. Nitrogen delays the rate of $M_{23}C_6$ formation in austenitic SS by retarding carbon and chromium diffusion. Potential for carbide coarsening is also reduced. The element can also induce solid solution strengthening [9]. The presence of fine carbides and solid solution strengthening from nitrogen increase creep strength and ductility [13, 14, 82].

The role of carbon content is described in Figure 27. Greater carbon content reduce ductility and increase creep rupture strength [82]. This trend aligns with the trends identified in Figure 2. Increase in carbon reduces time for carbide sensitization, lending to a more brittle type fracture.
Figure 27: Creep response of 308 SS weld metal with different carbon compositions [82].

Figure 28 and Figure 29 describe the creep strength and ductility with respect to bulk chemical compositions. In each figure, the mean carbon content is comparable, while difference in nitrogen, niobium, or phosphorus are high between the weakest and strongest alloy heats. Nitrogen, niobium, or phosphorus are suggested to have a more profound effect
on creep strength and ductility than carbon [82, 83]. Effects from nitrogen were addressed in Figure 26. Niobium can replace chromium in M$_{23}$C$_6$ and delay the rate of coarsening, extending the time which hardening mechanism by carbides are active. Similarly, phosphorus can enter M$_{23}$C$_6$ and promote fine matrix carbide dispersions, providing additional hardening mechanisms [9, 82, 83].

Figure 28: Creep strength comparison of annealed 304 SS for 1000 hr creep rupture test at 593°C [82].
2.3.3.1 Creep for PBF-LB and DED-LB 316 SS

With distinct microstructures from its wrought counterpart, AM materials can respond uniquely to creep loading conditions. Comparison between hot rolled 316L and as-built PBF-LB 316L is made in Figure 30.
In this study, the PBF-LB failed faster and elongated less than the HR material. The difference was produced from several microstructural sources [37, 84-86]. Dislocation cell structures with high dislocation density, had a strong influence. Figure 31 describe the cellular structures. In PBF-LB materials these substructures are formed from rapid
solidification rates during material printing [87]. The substructures limit the multiplication of dislocations which reduce the strain hardening capability of the material. Minimum strain rate is achieved quickly.

Figure 31: Dislocation cell structure in as-built PBF-LB 316L SS [84].

Accelerated failure can also occur from chemical segregation [37, 88]. The cell walls of the dislocation cell substructures are characterized with chromium and molybdenum concentrations. Figure 32 illustrates this microstructural feature. The concentrations are also formed by the rapid cooling rates during the printing process [86].
These sites become optimal location for carbide precipitation. Silicon and manganese enriched oxide particles are also seen in Figure 32. However, the particles’ contribution towards creep resistance were suggested to be negligible [89]. Other sources that can reduce creep deformation resistance include gas pores, lack of fusion pores, and or unmelted powders [35, 90, 91].
Figure 32: Energy-dispersive X-ray spectroscopy elemental map of as-built PBF-LB 316L SS [84].
Creep response change with application of annealing heat treatments. Figure 33 shows the creep behaviors of DED-LB 316L SS which was annealed for 1 hr at different temperatures.

![Graph showing creep response of as-built and annealed DED-LB 316L SS](image)

**Figure 33:** Creep response of as-built and annealed DED-LB 316L SS. Legend indicate annealing temperature [89].

The change in creep life and elongation with respect to annealing temperature is characterized with the evolution of dislocation cell structure [85, 89]. The microstructural evolution is described in Figure 34. Dislocation cell structures coarsen at an annealing
temperature range of 300 to 600°C. Comparable creep life as the as-built specimen is produced at this temperature range. At the temperature range of 600 to 800°C, the dislocation cell structure disappears, resulting in an increasing steady state strain rate and decreasing creep life. At the 1000°C temperature, the dislocation structure is removed. Creep rupture time recovers with the attainment of a more homogenous microstructure from grain recrystallization.

Figure 34: Dislocation cellular structures at different annealing temperatures for DED 316L SS [89].

Creep behavior comparisons are made between vertical and horizontal build orientation of as-built PBF-LB 316L SS in Figure 35.
Figure 35: Creep response from as-built PBF-LB 316L SS in vertical and horizontal build orientation [91].

The creep response reported in this study yielded comparable life and elongation between the two build orientations. However, the crack initiation and propagation response differed. Cracks of vertical and horizontal specimens can initiate and propagate along different melt pool boundary locations [34, 91]. These are shown in Figure 36. Vertical
specimens initiate cracks along ‘track-track’ melt pool boundaries and propagate along ‘layer-layer’ boundaries. On the other hand, horizontal specimens initiate and propagate cracks along ‘track-track’ melt pool boundaries.

Figure 36: Crack initiation and propagation paths seen on melt pool boundaries of vertically and horizontally printed as-built PBF-LB 316L SS [91].
2.3.4 Creep-Fatigue

Creep-fatigue (CF) damage is produced through repeated cycles containing a hold time. The standard CF test method is covered by ASTM E2714-13 [92]. These tests can either be in strain or force control. Figure 37 presents the stress-strain hysteresis loops for CF tests conducted on 316H SS in strain control. The dwell period was introduced at peak tensile strain. Creep deformation is exhibited during the dwell period through decreasing stress, stress relaxation. The stress relaxation magnitude generally increases with increasing dwell time.

Figure 37: Cyclic stress-strain behavior of 316H (annealed between 1050 and 1100°C for 18 min) at 550°C with different hold times at peak strain [93].
A force-controlled CF test with dwells at peak stress is shown in Figure 38. In this control mode, strain expansion occurs during the dwell period. The strain expansion magnitude increases with increasing dwell time. Force-controlled CF tests will exhibit ratchetting strain [70].

![Figure 38: Force-controlled CF cycle with dwell introduced at peak stress [92].](image)

Similar cyclic behaviors are exhibited in CF tests as in LCF tests. Evolution of peak stress for 316L SS with different strain amplitudes are shown in Figure 39. The same hardening-softening behavior is present in the strain-control CF tests: initial cyclic hardening, stable or quasi-stable maximum stress, and ending with rapid cyclic softening due to crack growth until rupture. Strain range-maximum stress and strain range-failure
life relations are also present. Relative to the hysteresis loops shown in Figure 37, dynamic strain aging mechanism, seen from the stress-strain serration, can also be present during CF loading [67, 94].

![Figure 39: Maximum stress response of 316L tested at 650°C and strain rate of 0.05%/s. Specimens aged at 650°C for 800 hours prior to testing. $\Delta \varepsilon$ is strain amplitude for this plot [55]](image)

Figure 39: Maximum stress response of 316L tested at 650°C and strain rate of 0.05%/s. Specimens aged at 650°C for 800 hours prior to testing. $\Delta \varepsilon$ is strain amplitude for this plot [55]

Effects from creep deformation are elucidated from different dwell times. A peak tensile stress-cycles plot for 316H build metal (BM) and weld joint (WJ) subjected to various CF peak tensile strain dwell periods is shown in Figure 40. Comparing CF tests of the same strain range, longer dwell periods yield smaller cycles to failure.
Figure 40: Peak tensile stress CF response from 316H build metal (BM) and weld joint (WJ) tested at 550°C, strain amplitude of 0.6%, and strain rate of 0.1%/s. BM annealed between 1050 and 1100°C for 18 min. [93].

CF responses differed between the BM and WJ material. WJ produced higher peak tensile stresses during initial cyclic hardening and failed earlier than the BM. Microstructural difference produced the differing CF response. These are shown in Figure 41. The WJ contain dendrites and δ-ferrites, as well as high-density dislocation networks. The high-density dislocation networks provide additional hardening mechanisms and produce a higher stress response during initial cyclic hardening [28, 95]. The WJ specimens fail earlier than the BM specimens primarily due to evolution of δ-ferrite evolution to carbides along δ-ferrite/austenite interfaces [93, 95, 96].
Figure 41: Microstructural of untested 316H BM and WJ. HAZ is heat-affected-zones [93].

Placements of dwells with respect to the cyclic load alter the amount creep damage. Application of dwells at peak tensile stresses are much more detrimental than at intermediate tensile stresses [97-99]. When compressive dwells are introduced, localized damage during the dwell period is minimized from crack closure and cavity sintering. However, it has been reported that transgranular crack initiation can occur more readily due to appearance of tensile mean stresses as opposed to the compressive mean stresses.
from tensile dwell tests [98]. This mechanism can slightly decrease CF life when compared to LCF tests.

Effects of multiple dwell periods on cyclic life has also explored. Figure 42 compares the cyclic lives of strain-controlled CF tests with peak tensile and peak symmetric dwell periods. A decrease in CF life is observed when dwells occur both in tension and compression. However, compared to tensile dwell tests, lives do not decrease as rapidly with increasing dwell times. Cavity sintering during the compressive hold can recover grain boundary damage formed during the tensile dwell period [98, 99].

![Figure 42: Evolution of cycles to failure with tensile and symmetrical hold times for 600°C CF test. Solid, open, upper half-filled symbol corresponded to 316, 316FR, and 316LN respectively [98].](image)
Two distinct behaviors are present during the stress relaxation responses of strain-controlled CF. This is shown in Figure 43. An initial rapid stress decline occurs within the first few seconds. This is then followed by a gradual stress descent [93, 100-102].

Figure 43: (a) Stress relaxation plots at half-life of 316H BM tested at 550°C, strain amplitude of 0.6%, and strain rate of 0.1%/s. (c) Total stress relaxation at half-life for each specimen [93].
Larger strain ranges and/or temperatures result in greater stress relaxation as illustrated in Figure 44. This increasing behavior asymptotes at long dwell times [98, 99].

Figure 44: Tensile stress relaxation response of 316 SS. Solid, upper half-filled, and lower half-filled symbol correspond to 316, 316LN and 316L SS respectively [98].

The influence of strain range and dwell period on the damage mechanism is visualized in Figure 45. Cyclic life closely follows the LCF failure curve at a high strain range and short dwell time. When strain range is low and dwell times are high, creep mechanisms control CF life, assuming a sufficiently high temperature. At intermediate strain ranges and dwell periods, an interactive form of damage termed CF interaction may
occur. A reduction in life is typically evident when this interaction occurs and the resulting CF life curve follows a-b-e-d-f in Figure 45.

Figure 45: Strain-life plot describing relations between creep and fatigue damage interactions in CF tests [103].

The microscopic damage observed in 316 SS subjected to CF vary depending on the strain range and dwell time, given an appropriate temperature [59, 101, 104]. These damage mechanisms are shown in Figure 46. Fatigue damage mechanisms are dominant when dwell periods are short. Failure is due to transgranular cracking. On the other hand, creep damage mechanisms take over when extreme hold times are implemented and rupture occurs due to intergranular cracking. Mixed mode rupture occurs at intermediate strain ranges and hold times. When mixed mode rupture happens, the material is said to have undergone a CF interaction.
Three different CF interactions can take place. Case A, competitive mode, is a weak form of CF interaction. Fatigue deformation caused transgranular and creep deformation caused intergranular damage occur independently. Case B, additive mode, sees a stronger interaction between fatigue and creep deformation mechanisms. In this case, the fatigue and creep damage initiates independently but unite with increasing deformation. Case C is an interactive mode which results in greatly accelerated crack growth. In this case, transgranular cracks grow faster driven by creep nucleated cavities, and gradually become intergranular. This form of damage is the most detrimental [59, 105].

Figure 46: Microstructural damage from CF loads in metals [59]

CF interaction “damage” diagrams can be used to assess the damage outlined in Figure 46 for a CF test [59]. This diagram is shown in Figure 47. A life fraction method is used to calculate creep and fatigue damage. Construction of this diagram is discussed in
Section 2.4. The expected failure mechanism relative to the interaction diagram’s location is summarised in Table 5. A bilinear locus is typically used when creep damage is calculated using a time-fraction (TF) approach and a linear locus is used when creep damage is calculated using a ductility (DE) approach for 316H SS [59].

Figure 47: Bilinear and linear CF interaction diagram.
Table 5: Expected failure mechanism relative to CF interaction diagram coordinate

<table>
<thead>
<tr>
<th>CF Interaction Diagram Location</th>
<th>Expected Failure Mechanism</th>
</tr>
</thead>
<tbody>
<tr>
<td>$\Phi_F &gt;\Phi_C$</td>
<td>LCF dominant failure</td>
</tr>
<tr>
<td>$\Phi_C &gt;\Phi_F$</td>
<td>Creep dominant failure</td>
</tr>
<tr>
<td>Case A</td>
<td>No CF interaction.</td>
</tr>
<tr>
<td></td>
<td>Independent CF, fatigue dominant, or creep dominant failure mode</td>
</tr>
<tr>
<td>Case B</td>
<td>Weak CF interaction</td>
</tr>
<tr>
<td></td>
<td>Additive CF failure mode</td>
</tr>
<tr>
<td>Case C</td>
<td>Strong CF interaction</td>
</tr>
<tr>
<td></td>
<td>Interactive CF failure mode</td>
</tr>
</tbody>
</table>

2.4 Life Fraction Prediction Methods

Several methods have been proposed to correlate strain-controlled CF data to serve as a life prediction model. The simplest is a life fraction approach where the fraction of damage from creep and fatigue is summed. When this damage reaches a critical value, $D$, failure occurs,

$$\Phi_C + \Phi_F = D$$  \hspace{1cm} (1)

If the critical value of damage is unity, then this is a linear damage summation model.
2.4.1 Fatigue Damage

The fatigue damage, $\Phi_F$, is determined through the application of Miners’ rule. The most simplistic form is given as,

$$\Phi_F = \sum \frac{n_i}{N_i}$$  \hspace{1cm} (2)

$n_i$ is the total cycles at a selected strain range $i$, and $N_i$ is the total cycle to failure at this strain range. The rule was proposed as a crack initiation criterion for pure-fatigue tests and assumes incremental damage, $1/N_i$, occurs per cycle at a given strain range until a value of unity or higher is achieved. The model is simplistic and can be applied to tests with multiple strain ranges. When multiple strain ranges are considered, cyclic fatigue damage can be related to a critical value below unity based on the understanding of material behavior [51].

In life fraction methods for constant amplitude strain-controlled CF tests, Miner’s rule is taken as,

$$\Phi_F = \frac{N}{N_f}$$  \hspace{1cm} (3)

$N$ and $N_f$ are number of cycles in CF and cycles to failure in pure fatigue, respectively. The number of cycles to fatigue failure is commonly correlated with strain amplitude. One common parametric relation is given by
\[ \frac{\Delta \varepsilon}{2} = \frac{\sigma_f'}{E} (2N_f)^b + \varepsilon_f' (2N_f)^c \] (4)

where \(\sigma_f'\) is the fatigue strength ductility coefficient, \(\varepsilon_f'\) is the fatigue ductility coefficient, \(b\) is the fatigue strength exponent, \(c\) is the fatigue ductility exponent, and \(E\) is the Young’s modulus.

2.4.2 Creep Damage

The creep damage is typically determined in one of two ways: (i) the time fraction (TF) approach or (ii) ductility exhaustion (DE) approach.

2.4.2.1 Time-Fraction Approach

The simplest approach of modeling creep damage is through the comparison of creep rupture time from creep tests against accumulated dwell times from CF tests using the Robinson’s rule [59]. The general form is written as,

\[ \Phi_c = \sum \frac{\Delta t}{t_c} \] (5)

where \(\Delta t\) is the dwell time per cycle in a CF test and \(t_c\) is the rupture time from a monotonic creep test at the same stress level, typically in units of hours. In strain-controlled isothermal
CF, creep damage is assumed to occur during the dwell period in which the stress is changing. Therefore, an integral form of the relation is used,

$$\Phi_C = \sum_0^{t_h} \int \frac{dt}{t_c(\sigma, T)}$$

(6)

where \(t_h\) is the total dwell time per cycle. Creep rupture time, \(t_c\), is dependent on the instantaneous stress during stress relaxation of CF tests, \(\sigma\), and test temperature, \(T\) [106].

The creep rupture time relationship is determined by regression fits of creep data. One time-temperature parametric equation commonly used to determine creep rupture time is the Larson-Miller (LM) parameter. The LM parameter is defined as,

$$P_{LM} = T(C_{LM} + log_{10}(t_c))$$

(7)

where \(P_{LM}\) is the LM parameter that depends on stress, \(T\) is the temperature in absolute units, Kelvin, and \(C_{LM}\) is the LM constant. In the derivation of Equation 7 [71], the LM parameter is also directly related to creep activation energy, \(Q\), which is assumed to depend on stress,

$$P_{LM} = log_{10}(e) \frac{Q}{R} = 0.217 Q$$

(8)

where \(R\) is the universal gas constant and \(e\) is Euler’s number which can be replaced by constant value with units on \(Q\) being cal/mole using \(R = 2.0\) cal/ (mol. K). Equation 7 is rearranged to solve for creep rupture time,
\[ t_c = 10^\left( \frac{P_{LM}(\sigma)}{T} - C_{LM} \right) \]  

(9)

The \( C_{LM} \) is commonly taken as 20 for steel with \( t_c \) in units of hours, but can vary depending on the material [71, 107, 108]. An example of a LM plot for 316LN is shown in Figure 48. A regression analysis, typically fitting to a polynomial function, is used to establish \( P_{LM}(\sigma) \) from either creep rupture or creep to a specific level of strain, e.g., 1% creep strain,

\[ P_{LM}(\sigma) = a_n \log(\sigma)^n + a_{n-1} \log(\sigma)^{n-1} + \cdots + a_1 \log(\sigma) + a_0 \]  

(10)

The parameter \( C_{LM} \) is determined by plotting the creep rupture data in a \( \log_{10}(t_c) \) against \( 1/T \) plot, Figure 49.
Figure 48: LM plot of 316LN creep rupture data found in literature, $C_{LM} = 24.56$ [109].
Another time-temperature parametric relation used to determine the creep rupture time is the Orr-Sherby-Dorn (OSD) parameter, which is derived by integrating the steady-state creep rate relationship resulting in an equation for creep strain after time $t_c$ [71],

$$\varepsilon_c = A(\sigma) \ t_c \exp \left( - \frac{Q}{RT} \right) = A(\sigma) \ \theta_c$$

(11)
\[ \theta_c = t_c \exp \left( - \frac{Q}{RT} \right) \]  

(12)

It is assumed that the activation energy \( Q \) does not depend on stress, i.e., the creep mechanism does not change for the stress and temperatures considered; therefore, Equation 12 does not directly depend on stress. The OSD parameter is defined as the base 10 log of Equation 12,

\[ P_{OSD} = \log_{10}(\theta_c) = \log_{10}(t_c) - \log_{10}(e) \frac{Q}{RT} \]  

(13)

where \( Q, R, \) and \( T \) take on the same definitions and units used by the LM parameter. Combining Equation 12 and Equation 13, the creep rupture time can be solved,

\[ t_c = 10^{P_{OSD}(\sigma)+\log_{10}(e)\frac{Q}{RT}} \]  

(14)

where \( P_{OSD} \) depends on stress for a fixed rupture strain, Equation 11. A plot showing how \( P_{OSD} \) relates to stress for 316LN SS is given in Figure 50.
Figure 50: OSD plot of 316LN creep rupture data found in literature. Temperature in legend are in units of °C [109].

Unlike the LM parameter where the activation energy will vary from stress to stress, the OSD parameter assumes a constant creep activation energy. This value can be defined as the activation energy for lattice self-diffusion or solved by plotting the creep rupture data in the form of a \( \log_{10}(t_c) \) against \( 1/T \) plot, Figure 51. Each straight line is representative of the different creep rupture stresses. All lines are parallel as the activation energy is assumed constant.
The Manson-Haferd (MH) parameter is another time-temperature parametric model that relates stress and creep rupture time. The MH parameter relation is written as,

\[ P_{MH} = \frac{\log_{10}(t_c) - \log_{10}(t_a)}{T - T_a} \]  

(15)

where \( t_a \) and \( T_a \) are constants. Solving for creep rupture time,

\[ t_c = 10^{P_{MH}(\sigma) (T - T_a) + \log_{10}(t_a)} \]  

(16)
A MH plot for 316LN SS showing the stress dependence is given in Figure 52. The two constants, $t_a$ and $T_a$, hold graphical meaning as shown in Figure 53.

**Figure 52: MH plot of 316LN creep rupture data found in literature [109].**
Figure 53: Plot of $\log_{10}(t_c)$ against $T$ plot to determine constants $t_a$ and $T_a$ [109].

Each of these time-temperature parameters have their own benefits. The LM relation is most used due to its simplicity in defining an empirical relation between stress and rupture time. The OSD relation can be beneficial when assumptions of a constant activation energy can be justified. On the other hand, the MH relation can be beneficial if the data set used can accurately determine a point corresponding to $t_a$ and $T_a$. 
2.4.2.2 Ductility Exhaustion Approach

The DE approach calculates creep damage by relating the creep strain accumulated during the CF dwell period to the creep ductility observed during monotonic creep tests. The general form for this creep damage relation is written as,

$$\phi_c = \sum \frac{\Delta \varepsilon_c}{\varepsilon_f}$$  \hspace{1cm} (17)

where $\Delta \varepsilon_c$ is the creep strain per cycle from CF tests and $\varepsilon_f$ is the creep ductility from creep rupture tests at a test temperature. Creep ductility is taken as the engineering creep strain at rupture. This can be the creep rupture elongation or reduction of area at rupture [110, 111].

Figure 54 describe how $\Delta \varepsilon_c$ is determined. $\Delta \varepsilon_{pp}$ is the plastic strain range from a fatigue cycle and $\Delta \varepsilon_{cp}$ is the creep strain range imparted by stress relaxation. The sum of being $\Delta \varepsilon_{pp}$ and $\Delta \varepsilon_{cp}$ is the total inelastic strain range. The accumulated creep strain is,

$$\Delta \varepsilon_{cp} = \Delta \varepsilon_c = \frac{\Delta \sigma}{E}$$  \hspace{1cm} (18)
As with the TF approach, the creep testing temperature must be equivalent to the CF testing temperature. Creep damage in integral form with respect to time is,

\[ \Phi_c = \sum \int_0^{t_h} \frac{\dot{\varepsilon}_c}{\varepsilon_f(\dot{\varepsilon}_c)} \, dt \]  \hspace{1cm} (19)

where \( \dot{\varepsilon}_c \) is the instantaneous creep strain rate during dwell periods of CF tests. Creep ductility depends on the instantaneous creep strain rate at a given test temperature. For cast 316H, this relationship is shown in Figure 55.
In Figure 55, the temperature dependence is found to be weak. The study used to develop the plot assumed similar creep strain at failure behaviors at the two temperatures, 570 and 600°C [113]. When constructing Figure 55, the average creep strain rate was used. This is defined as follows,

$$\dot{\varepsilon}_{avg} = \frac{\varepsilon_f}{t_c}$$  \hspace{1cm} (20)

This simplification is often used due to the lack of intermediate strain-time data for creep tests with extensive rupture times and is generally sufficient [110, 111, 113-115]. When intermediate strain-time data is available, creep ductility data may also be represented in the form of creep ductility against minimum creep strain rate and temperature [102].
As shown in Figure 55, creep ductility against creep strain rate plots are fitted to a piece-wise function when used for creep damage calculations.

\[
\begin{cases}
\dot{\epsilon}_c > \dot{\epsilon}_1, & \epsilon_f = \epsilon_{max} \\
\dot{\epsilon}_1 > \dot{\epsilon}_c > \dot{\epsilon}_2, & \epsilon_f(\dot{\epsilon}_c) \\
\dot{\epsilon}_c < \dot{\epsilon}_2, & \epsilon_f = \epsilon_{min}
\end{cases}
\]  

(21)

where \(\epsilon_{max}\) is the upper bound creep ductility and \(\epsilon_{min}\) is the lower bound creep ductility. The bounds are termed upper and lower-shelf ductility, respectively. When determining the upper-shelf ductility, rupture elongation from tensile tests have also been used [113, 116]. The intermediate relation is defined through regression analysis between the upper and lower shelves. The regression analysis can be conducted on a log-log plot, of which the regression relation then follows a power law,

\[
\epsilon_f = a \dot{\epsilon}_c^b
\]  

(22)

\(a\) and \(b\) are regression constants. A regression fit on a semi-log plot, where only the creep strain rate is normalized against logarithmic coordinates, has also been proposed,

\[
\epsilon_f = A \log_{10}(\dot{\epsilon}_c) + B
\]  

(23)

\(A\) and \(B\) are regression constants [102]. The piecewise creep ductility behavior is explained through the different cavity growth mechanisms relative to creep strain rates. At the upper-shelf ductility where creep strain rates are high, continuum cavity growth dominates. At the lower-shelf ductility where creep strain rates are low, cavities do not grow as fast. Other
creep deformation mechanisms dominate instead. At intermediate creep ductility in the transition region, cavity growth is controlled by grain boundary diffusion [77, 101].

A more conservative approach to estimating creep ductility is by assuming creep ductility is independent of creep strain rate. Creep ductility is taken as the lower-shelf ductility. This simplification yield the following relation for creep damage [115, 117],

$$\Phi_c = \sum \int_0^{t_h} \frac{\dot{\varepsilon}_c}{\varepsilon_{min}} dt$$

(24)

In contrast to the simplified creep damage relation, an in-depth approach can also be taken where creep ductility is defined as a function of instantaneous creep strain rate, stress, and temperature,

$$\Phi_c = \sum \int_0^{t_h} \frac{\dot{\varepsilon}_c}{\varepsilon_f(\dot{\varepsilon}_c, \sigma, T)} dt$$

(25)

Considerations over cavity growth mechanisms as well as cavity nucleation effects from the applied stress are made [110, 111, 113]. This form of creep ductility is termed the stress-modified DE (SMDE) approach where the creep ductility is expressed by

$$\varepsilon_f = A \exp \left( \frac{Q}{RT} \right) \dot{\varepsilon}_c^n \sigma^{-m}$$

(26)

where $A$, $Q$, $n$, and $m$ are material constants. Constants for this relation are determined by a multi-regression analysis on creep rupture data with respect to the dependent parameters:
strain rate, stress, and temperature [110, 111, 113, 118]. An engineering strain at failure against average strain rate plot for 304 SS creep rupture tests at varying temperatures using this creep ductility relations is presented in Figure 56. Similar to Figure 55, an upper-shelf ductility is identified at high strain rates.

![Figure 56: Stress modified creep ductility empirical fit to engineering strain at failure against average strain data for 304 [118].](image)

2.4.3 Strain Range Partitioning

A life fraction method that does not use a CF interaction diagram is the strain range partitioning (SRP) method. In this method, a hysteresis loop is partitioned to four distinct fully reversed cycles: tensile plastic strain reversed by compressive plastic strain, tensile
hold reversed by compressive hold, tensile hold reversed by compressive plastic strain, tensile hold reversed by compressive plastic strain [119]. These cycles are shown in Figure 57. Subscript is P stands for plastic loading and C stands for creep loading. The first letter indicates the tensile and second letter indicates the compressive strain behavior.

![Figure 57: Strain range partitioning cycles [119]](image)

Strain-life curves, typically using the Coffin-Manson relation, are formed for each SRP cycle type to determine damage contributions. An example strain-life plot for 316 SS tested in air at 705°C is shown in Figure 58.
Figure 58: SRP cycle Coffin-Manson relations for 316 tested in air at 705°C [119].

The total damage per cycle is defined as

\[
\frac{1}{N} = \frac{f_{PP}}{N_{PP}} + \frac{f_{CC}}{N_{CC}} + \frac{f_{CP}}{N_{CP}} \quad \text{(or} \quad \frac{f_{PC}}{N_{PC}} \text{)} \tag{27}
\]

where the strain range fraction is defined

\[
f_{ij} = \frac{\Delta \varepsilon_{ij}}{\Delta \varepsilon_{in}} \tag{28}
\]

and
\[ \Delta \varepsilon_{in} = \Delta \varepsilon_{pp} + \Delta \varepsilon_{cc} + \Delta \varepsilon_{cp} \quad \text{or} \quad \Delta \varepsilon_{pc} \]  

(29)

The ASTM E2714-13 plastic strain range definition, width of the hysteresis loop at zero strain, is used for \( \Delta \varepsilon_{in} \) [92]. The interaction damage rule is a modified Miner’s rule and is formed from the understanding that each SRP cycle has a different deformation mechanism. Damage contributions from each mechanism are differentiated through strain range fractions. The strain range fraction is multiplied to the respective SRP cycle damage.

Decisions for cycle partition are based on strain rate magnitudes. Portions of the cycle that produce large strain rates, such as cyclic loads, are associated with plastic strains. Portions of the cycle that produce small strain rates, such as stress or strain holds, are associated with creep loading. An example partitioning scheme from an isothermal strain-controlled CF test with a constant fast strain rate during the cyclic loading is shown in Figure 54. A \( \Delta \varepsilon_{pp} \) partition was taken along section bc’, while a \( \Delta \varepsilon_{cp} \) partition was taken along section c’e (or cd’). Strain expansion occur during force-control dwell periods, while stress relaxation occurs during displacement-control dwell periods. No \( \Delta \varepsilon_{cc} \) partitions are made as a relatively fast strain rate is assumed during the reversed loading in compression [112].

In the case of slow strain rates, the \( \Delta \varepsilon_{pp} \) cycle is partitioned to include \( \Delta \varepsilon_{cc} \) cycles. The original \( \Delta \varepsilon_{pp} \) cycle is compared to a similar fast strain rate cycle. Slower strain rate cycles, lower frequency cycling, will have larger \( \Delta \varepsilon_{cc} \) as shown in Figure 59.
The SRP method is mathematically simple. This quality is powerful when analysing hysteresis loops generated from unconventional CF tests or tests with low or varying strain rates. However, preparation for this method is cumbersome as multiple fatigue, CF, and cyclic creep tests that are unique to the SRP cycles are needed to determine the model parameters. Understanding of strain rate dependence on the material of interest are also required.

2.4.4 Stress Relaxation and Strain Rate Models

The use of stress relaxation and strain rate models are necessary for CF life fraction methods, as the creep damage calculation approaches for the models are dependent on
either or both instantaneous stress and strain rate. These models are also needed when extrapolating CF life predictions with extreme dwell periods. Some nuclear component design codes provide design curves which can be used for the dwell period. However, these curves are conservative in nature [120]. The use of stress relaxation and strain rate models can provide a more accurate representation of the stress relaxation.

Several forms of stress relaxation models have been proposed through empirical fits of stress-time data. The Feltham equation has been used with life fraction methods for 316 SS [102, 110, 111, 113],

\[
\Delta \sigma = \sigma_0 - \sigma = b \ln(at + 1)
\]  

(30)

where \(\sigma_0\) is stress at the beginning of the dwell, \(\sigma\) is the instantaneous stress, \(t\) is accumulated time during stress relaxation, and \(a\) and \(b\) are fitted constants. Figure 60 illustrates how the material constants \(a\) and \(b\) can be determined relative to the \(\Delta \sigma\) and \(\ln(t+1)\) axes.
Figure 60: Feltham equation plot illustrating response change with $b = 1$ and varying $a$ [102].

Research on copper, brass, and titanium alloys suggest that this relation is appropriate for hold times of up to 30 minutes [121]. For extreme CF hold times, $b$ can vary unpredictably and nonlinearity may occur leading to lower stress predictions [102]. Using this equation, the instantaneous stress during relaxation is

$$\sigma = \sigma_0 - b \ln(at + 1)$$  \hspace{1cm} (31)

Another relation, termed the Conway analysis, is given by [97, 106, 121],

$$\ln \left( \frac{\sigma_{max}}{\sigma_{min}} \right) = \left( -\frac{A}{1 + m} \right) t^{1+m}$$  \hspace{1cm} (32)
where $\sigma_{max}$ and $\sigma_{min}$ are peak and relaxed stresses, and $A$ and $m$ are fitted constants. Constants for this relation are derived by fitting the stress relaxation-time data in a logarithmic scale. Instantaneous stress during stress relaxation can be solved for by taking $\sigma_{min}$ to be $\sigma$ and rearranging Equation 32,

$$\sigma = \frac{\sigma_{max}}{e^{\frac{A}{1+m}e^{1+m}}}$$

(33)

Using the Conway analysis, successful stress predictions for hold times up to 100 hours have been observed [97, 106, 121].

The instantaneous creep strain rates in strain-controlled CF test dwell periods are derived mathematically using the stress relaxation models. The strain is held constant during the dwell period, therefore, the total strain rate, $\dot{\varepsilon}$, is zero. Sectioning the total strain rate into elastic and inelastic parts yields the following,

$$\dot{\varepsilon} = \dot{\varepsilon}_e + \dot{\varepsilon}_i = 0$$

(34)

Using elastic stress-strain relations, the elastic strain rate can be written as,

$$\dot{\varepsilon}_e = \frac{1}{E} \frac{d\sigma}{dt}$$

(35)

Only creep strain increasing mechanisms are at play during the dwell period. Therefore, the inelastic strain rate is entirely creep strain rate, $\dot{\varepsilon}_i = \dot{\varepsilon}_c$. Solving for the creep strain rate yields,
\[
\dot{\varepsilon}_{in} = \dot{\varepsilon}_c = -\frac{1}{E} \frac{d\sigma}{dt}
\]  

(36)

A time derivative can be taken on either Equations 31 or Equation 33 and input into this equation. The resulting strain rate relation using the Feltham is

\[
\dot{\varepsilon}_c = \frac{ab}{E} \frac{1}{at + 1}
\]  

(37)

and Conway is

\[
\dot{\varepsilon}_c = \frac{1}{E} \frac{\sigma_{max}At^m}{e^{(\frac{A}{1+m})t^{1+m}}}
\]  

(38)

The Feltham relation has been shown to behave irregularly for hold times longer than 30 minutes [121]. The Conway analysis on the other hand has seen success for extreme dwell periods [121].

2.4.5 Life Prediction Methods with Application to SS

Various nuclear component design codes exist. In each design code a CF interaction “damage” diagram, Figure 47, is constructed using the life fraction approach. The French and US design codes, RCC-MRx and ASME BPVC, Section III, respectively, use similar methods to obtain parameters necessary for CF life fraction analysis. These are outlined in Table 6.
Both design codes calculate creep damage with the TF approach. The codes expect conservative outcomes due to factors of safety used when forming the fatigue and minimum creep design curves. For a given test temperature, the fatigue design curve is established by shifting the mean strain range-cycles to failure curve by a factor of 20 along the cycles-to-failure axis (horizontally) and by a factor of 2 along the strain range axis (vertically). The minimum cycles to failure after applying these factors are used to form the design curve. The minimum creep rupture stress-time curves are formed from the lower 95% prediction band of LM plots [120]. Additional factors of safety are incorporated when determining creep rupture times. This is shown in Table 6. Although the same factors of safety are applied, different design curve and LM plots are used by the ASME and RCC-MRx codes [122].
Table 6: Methods of determining ASME BPVC and RCC-MRx design code parameters [123, 124]

<table>
<thead>
<tr>
<th></th>
<th>ASME BPVC Section III</th>
<th>RCC-MRx</th>
</tr>
</thead>
<tbody>
<tr>
<td>( N_f )</td>
<td>ASME fatigue cycles to failure design curve</td>
<td>RCC-MRx fatigue cycles to failure design curve</td>
</tr>
<tr>
<td>( t_c )</td>
<td>Minimum creep rupture stress-time curve. Stress taken as ( \frac{\sigma}{0.9} ) for elastic and ( \frac{\sigma}{0.67} ) for inelastic analysis</td>
<td>Minimum creep rupture stress-time curve. Stress taken as ( \frac{\sigma}{0.9} )</td>
</tr>
<tr>
<td>( \Delta \sigma )</td>
<td>ASME isochronous stress-strain curves</td>
<td>RCC-MRx isochronous stress-strain curves</td>
</tr>
<tr>
<td>( \Phi_F )</td>
<td>( \frac{N}{N_f} ) (Equation 3)</td>
<td>( \frac{N}{N_f} ) (Equation 3)</td>
</tr>
<tr>
<td>( \Phi_C )</td>
<td>( \sum \int_0^{t_h} \frac{dt}{t_c(\sigma,T)} ) (Equation 6)</td>
<td>( \sum \int_0^{t_h} \frac{dt}{t_c(\sigma,T)} ) (Equation 6)</td>
</tr>
</tbody>
</table>

The ASME and RCC-MRx codes require the use of a bilinear failure envelop when the CF interaction diagram is made using the TF approach for 304 or 316 SS. The intersection point of (0.3, 0.3) is used [120]. This point was selected from observations made from experimental outcomes [106]. The fatigue interaction diagram from this study, plotted in logarithmic axes, is shown in Figure 61. The CF damage is compared against the linear and bilinear envelope. Design failure values were calculated using the methods listed in Table 6’s ASME column. Actual failure values were determined using methods summarized in Table 7.
Figure 61: CF interaction diagram of 304 and 316 for 593°C CF tests for strain ranges from 0.5 to 2.0%, strain rate of $4 \times 10^{-3}$ 1/s, and hold times of up to 10 hours [106].
Table 7: Methods of determining actual failure values for CF interaction diagram shown in Figure 61 [106]

<table>
<thead>
<tr>
<th>Parameter</th>
<th>Description</th>
</tr>
</thead>
<tbody>
<tr>
<td>$N_f$</td>
<td>LCF cycles to failure</td>
</tr>
<tr>
<td>$t_c$</td>
<td>Creep stress-rupture curve from experimental data</td>
</tr>
<tr>
<td>$\Delta \sigma$</td>
<td>$\ln \left( \frac{\sigma_{\text{max}}}{\sigma_{\text{min}}} \right) = \left( \frac{A}{1+m} \right) t^{1+m}$ (Equation 32, Conway analysis)</td>
</tr>
<tr>
<td>$\Phi_F$</td>
<td>$\frac{N}{N_f}$ (Equation 3)</td>
</tr>
<tr>
<td>$\Phi_C$</td>
<td>$\sum \int_0^{t_h} \frac{dt}{t_c(\sigma,T)}$ (Equation 6)</td>
</tr>
</tbody>
</table>

With respect to the actual failure values, the bilinear (0.3, 0.3) intersection envelope better characterized the safe lower bound for CF damage [59, 106]. Comparing the design and actual failure values, the design failure values were largely outside the linear locus while much of the actual failure values were well within the linear envelope. The design curves used to determine design failure CF damage values produce lower LCF cycle to failure, $N_f$, and shorter creep rupture times, $t_c$, resulting in larger fatigue and creep damage.

A TF based CF interaction diagram taken from a different study is shown in Figure 62. A different set of data than shown in Figure 61 is used. The interaction diagram is plotted in a linear scale and points are labeled with CF strain range and dwell periods. Methods used to identify the life fraction input parameters are given in Table 8. Cycles to failure in pure fatigue was defined by the number of cycles to growth a fatigue crack from length $a_0$, to length $a_f$ using
\[ N_f = \frac{1}{\Delta \varepsilon_p \left[ \sec \left( \frac{\pi \sigma_{max}}{2P} \right) - 1 \right]} \sum_{i=0}^{n-1} \ln \left( \frac{a_{i+1}}{a_i} \right) \frac{1}{Y_i^2} \]

(39)

where \( \Delta \varepsilon_p \) is the plastic strain range, \( P \) is the material strength parameter, and \( Y \) is the dimensionless compliance function related to crack shape and specimen geometry. \( a \) is current crack length, where \( a_i \) to \( a_{i+1} \) is the range of crack length for which \( Y_i \) is constant. \( i \) is the index for increment of crack advancement and \( n \) is the number of discrete crack growth increments [97].

Figure 62: TF based CF interaction diagram for cast 316 SS, including data for strain range from 0.6 to 2\%, dwell periods up to 48 hours, strain rate \( 1 \times 10^{-3} \) 1/s, and temperature of 570°C [59, 97].
Table 8: Methods of determining actual failure values for CF interaction diagram shown in Figure 62 [59, 97]

<table>
<thead>
<tr>
<th>$N_f$</th>
<th>$N_f = \frac{1}{\Delta \varepsilon_p \sec \left( \frac{\pi \sigma}{2T} \right) - 1} \sum_{i=0}^{n-1} \ln \left( \frac{a_{i+1}}{a_i} \right) \frac{1}{Y_i^2}$ (Equation 39)</th>
</tr>
</thead>
<tbody>
<tr>
<td>$t_c$</td>
<td>Creep stress-rupture curve from experimental data</td>
</tr>
<tr>
<td>$\Delta \sigma$</td>
<td>$\ln \left( \frac{\sigma_{\text{max}}}{\sigma_{\text{min}}} \right) = \left( \frac{A}{1+m} \right) t^{1+m}$ (Equation 32, Conway analysis)</td>
</tr>
<tr>
<td>$\Phi_F$</td>
<td>$\frac{N}{N_f}$ (Equation 3)</td>
</tr>
<tr>
<td>$\Phi_c$</td>
<td>$\sum \frac{\Delta t}{t_c}$ (Equation 5)</td>
</tr>
</tbody>
</table>

In Figure 62, fatigue damage decreased when dwell periods were introduced. Strain ranges higher than 0.8% had less of a reduction in fatigue damage than lower strain ranges. No trends were seen in creep damage relative to strain range and dwell periods. Large scatter is seen along the creep damage (labeled time based damage) axis. Much of the points exceed unity creep damage, regardless of short or long dwell periods and strain range size. As no trends are observed from creep damage, CF interactions are difficult to identify [59, 97].

The R5 nuclear design code used in the UK determines creep damage through the DE approach. Methods used to determine CF damage parameters from a strain-controlled isothermal CF test for this design code are summarized in Table 9.
Table 9: Relationships used in the R5 design code [117, 120]

<table>
<thead>
<tr>
<th>$N_f$</th>
<th>Appropriate fatigue endurance-strain range curve</th>
</tr>
</thead>
</table>
| $\varepsilon_f$ | \[
\begin{cases}
\dot{\varepsilon}_c > \dot{\varepsilon}_1, & \varepsilon_f = \varepsilon_{max} \\
\dot{\varepsilon}_1 > \dot{\varepsilon}_c > \dot{\varepsilon}_2, & \varepsilon_f(\dot{\varepsilon}_c) \\
\dot{\varepsilon}_c < \dot{\varepsilon}_2, & \varepsilon_f = \varepsilon_{min}
\end{cases}
\] (Equation 21) |
| $\dot{\varepsilon}_c$ | $\dot{\varepsilon}_c = -\frac{1}{E} \frac{d\sigma}{dt}$ From stress relaxation data |
| $\Phi_F$ | $\frac{N}{N_f}$ (Equation 3) |
| $\Phi_C$ | $\sum \int_0^{t_h} \frac{\dot{\varepsilon}_c}{\varepsilon_f(\dot{\varepsilon}_c)} dt$ (Equation 19) or $\sum \int_0^{t_h} \frac{\dot{\varepsilon}_c}{\varepsilon_{min}} dt$ (Equation 24) |

A CF interaction diagram formed using R5 design code principles is shown in Figure 63. The same dataset used in Figure 62 are used to form this plot. The equation used to determine the life fraction input parameters are listed in Table 10.
Figure 63: DE based CF interaction diagram for cast 316 SS, including data for strain ranges of 0.6 to 2%, dwell period of up to 48 hours, strain rate $1 \times 10^{-3} \text{ 1/s}$, and temperature of $570^\circ \text{C}$ [59, 97].

Table 10: Relationships used for calculating failure values for CF interaction diagram shown in Figure 63 [59, 97]

<table>
<thead>
<tr>
<th>$N_f$</th>
<th>$N_f = \frac{1}{\Delta \epsilon_p \sec (\pi \sigma / 2T)^{-1}} \sum_{i=0}^{n-1} \ln \left( \frac{a_{i+1}}{a_i} \right) \frac{1}{Y_i^2}$ (Equation 39)</th>
</tr>
</thead>
<tbody>
<tr>
<td>$\epsilon_f$</td>
<td>$\begin{cases} \dot{\epsilon}_f = \dot{\epsilon}<em>1, &amp; \epsilon_f = \epsilon</em>{\text{max}} \ \dot{\epsilon}_1 &gt; \dot{\epsilon}_c &gt; \dot{\epsilon}_2, &amp; \epsilon_f(\dot{\epsilon}_c) \quad \text{(Equation 21)} \ \dot{\epsilon}_c &lt; \dot{\epsilon}<em>2, &amp; \epsilon_f = \epsilon</em>{\text{min}} \end{cases}$</td>
</tr>
<tr>
<td>$\Delta \sigma$</td>
<td>$\ln \left( \frac{\sigma_{\text{max}}}{\sigma_{\text{min}}} \right) = \left( \frac{A}{1+m} \right) t^{1+m}$ (Equation 32, Conway analysis)</td>
</tr>
<tr>
<td>$\Phi_F$</td>
<td>$\frac{N}{N_f}$ (Equation 3)</td>
</tr>
<tr>
<td>$\Phi_C$</td>
<td>$\sum \int_{0}^{t_h} \frac{\dot{\epsilon}_c}{\epsilon_f(\dot{\epsilon}_c)} , dt$ (Equation 19)</td>
</tr>
</tbody>
</table>
In Figure 63, a strong graphical trend exists between strain ranges and dwell periods relative to creep and fatigue damage. The presence of these trends suggest an L-shaped damage envelope with an intersection point at (0.1, 0.1), rather than a linear envelope used by the R5 code case. Low creep and high fatigue damage is seen from CF tests with strain ranges larger than 0.8%. Increase in dwell periods largely reduced the fatigue damage at these strain ranges with minimal increase in creep damage. At lower strain ranges, high creep and low fatigue damage was observed. Creep damage was near unity for all tests and the influence from dwell periods on fatigue damage were minimal. These graphical trends aligned well with expected CF rupture modes [59, 97].

The SMDE creep damage approach was proposed to characterize CF interactions better than the R5 DE approach [110, 111, 113, 118]. Comparison of the SMDE and R5 DE approaches are made for cast 316H SS in Figure 64. The interaction diagram is plotted in logarithmic axes. Factor of three and one-third lines are presented along with the linear locus. The relationships used to plot the data are summarized in Table 11. Creep ductility was obtained from a creep ductility-average creep strain rate plot. The data used to construct this plot was presented earlier in Figure 55.
Figure 64: CF interaction diagram formed using R5 DE and SMDE approaches on 316H tested at 570°C, strain range of 0.35% to 1.85%, undefined strain rate, and dwell period of up to 47.5 hours [113].
Table 11: Relationships used to plot the data in the CF interaction diagram shown in Figure 64 [113]

<table>
<thead>
<tr>
<th></th>
<th>R5 DE</th>
<th>SMDE</th>
</tr>
</thead>
<tbody>
<tr>
<td>$N_f$</td>
<td>Total strain range-cycles to failure plot for continuous cycling</td>
<td>$\varepsilon_f = A \exp \left( \frac{Q}{RT} \right) \dot{\varepsilon}_c^n \sigma^{-m}$ (Equation 26)</td>
</tr>
<tr>
<td>$\varepsilon_f$</td>
<td>$\begin{cases} \dot{\varepsilon}_c &gt; \dot{\varepsilon}<em>1, &amp; \varepsilon_f = \varepsilon</em>{\text{max}} \ \dot{\varepsilon}_1 &gt; \dot{\varepsilon}_c &gt; \dot{\varepsilon}_2, &amp; \varepsilon_f = \varepsilon(f_c) \ \dot{\varepsilon}_c &lt; \dot{\varepsilon}<em>2, &amp; \varepsilon_f = \varepsilon</em>{\text{min}} \end{cases}$ (Equation 21)</td>
<td>$\varepsilon_f$</td>
</tr>
<tr>
<td>$\Delta \sigma$</td>
<td>$\ln \left( \frac{\sigma_{\text{max}}}{\sigma_{\text{min}}} \right) = \left( \frac{A}{1+m} \right) \varepsilon^{1+m}$ (Equation 32, Conway analysis)</td>
<td>$\varepsilon_f$</td>
</tr>
<tr>
<td>$\Phi_F$</td>
<td>$\frac{N}{N_f}$ (Equation 3)</td>
<td>$\varepsilon_f$</td>
</tr>
<tr>
<td>$\Phi_C$</td>
<td>$\sum \int_0^{t_n} \frac{\dot{\varepsilon}_c}{\varepsilon_f(\dot{\varepsilon}_c)} dt$ (Equation 19)</td>
<td>$\sum \int_0^{t_n} \frac{\dot{\varepsilon}_c}{\varepsilon_f(\dot{\varepsilon}_c, \sigma, T)} dt$ (Equation 25)</td>
</tr>
</tbody>
</table>

The R5 DE and SMDE approach both yield consistent CF damage relative to the linear locus. Points plotted by the SMDE approach, however, are more congregated near the linear summation envelope. Figure 65 formed from a combination of constant strain rate uniaxial tensile tests, force-controlled tensile tests, and stress relaxation from CF tests is used to reconcile the differences between these methods. Initial creep strain rates for CF tests are higher than average creep strain rates from creep tests. This characteristic is retained for CF tests with low initial stress during stress relaxation. However, at high initial stresses, the CF creep strain rates become lower than the creep strain rates from creep tests.
Constants \( n \) and \( m \) in the SMDE equation, Equation 26, account for these behaviors. The empirical relations produce high creep ductility for low stress-high strain rate combinations, reducing creep damage per cycle. Whereas low creep ductility occurs for high stress-low strain rate combinations, increasing creep damage per cycle [113].

![Figure 65: Creep strain rates for relaxation dwells and creep rupture data for cast 316H. Average creep rates and rupture stress are plotted for creep tests [113].](image)

The life fraction methods can be modified to predict cycles to failure from a CF test. Assuming fatigue damage is calculated from Miner’s rule, Equation 3, and equal amounts of creep damage is accumulated per cycle, \( \Phi_{cycle} \), the life fraction relation can be written as,
\[ N \left( \frac{1}{N_f} + \Phi_{cycle} \right) = D \]  \hspace{1cm} (40)

Solving for CF life,

\[ N = \frac{D}{\left( \frac{1}{N_f} + \Phi_{cycle} \right)} \]  \hspace{1cm} (41)

CF life predictions for two temperatures, 550°C and 600°C, are shown in Figure 66. TF and DE approach for 316FR SS are compared. Equations used to construct these plots are summarized in Table 14. Two DE approaches were used: rate-independent and rate-dependent DE approach. The rate-independent approach simplifies the prediction process by neglecting effects from creep strain rates and uses the lower shelf creep ductility. The lower shelf ductility results in higher creep damage and yields less conservative predictions. The creep rupture time and creep ductility relations were formed using creep rupture time data [115].
Figure 66: Predicted cycles to failure plots for SA 316FR SS tested at (a) 550°C and (b) 600°C using (TF), strain-rate-independent DE (DE1), and strain-rate-dependent DE (DE2) approaches. Data points are from CF tests with strain rate of 0.1%/s [115].
Table 12: Relationships used to plot the data in the CF interaction diagrams shown in Figure 66 [115].

<table>
<thead>
<tr>
<th></th>
<th>TF</th>
<th>Strain-rate-independent DE (DE1)</th>
<th>Strain-rate-dependent DE (DE2)</th>
</tr>
</thead>
<tbody>
<tr>
<td>D</td>
<td>Bilinear envelope, inflection at (0.3, 0.3)</td>
<td>Linear envelope</td>
<td>Linear envelope</td>
</tr>
<tr>
<td>( \Phi_F )</td>
<td>( \frac{N}{N_f} )</td>
<td></td>
<td></td>
</tr>
<tr>
<td></td>
<td>(Equation 3)</td>
<td></td>
<td></td>
</tr>
<tr>
<td>( \Phi_C )</td>
<td>( \sum \int_0^{t_h} \frac{dt}{t_c(\sigma, T)} dt )</td>
<td>( \sum \int_0^{t_h} \frac{\dot{\varepsilon}<em>c}{\varepsilon</em>{min}} dt )</td>
<td>( \sum \int_0^{t_h} \frac{\dot{\varepsilon}_c}{\varepsilon_f(\dot{\varepsilon}_c)} dt )</td>
</tr>
<tr>
<td></td>
<td>(Equation 6)</td>
<td>(Equation 24)</td>
<td>(Equation 19)</td>
</tr>
<tr>
<td>( N_f )</td>
<td>Continuous cycling fatigue test data</td>
<td></td>
<td></td>
</tr>
<tr>
<td>( t_c )</td>
<td>Second degree fit to stress-creep rupture time data</td>
<td></td>
<td></td>
</tr>
<tr>
<td>( \varepsilon_f )</td>
<td>( \varepsilon_f = \varepsilon_{min} )</td>
<td>( \varepsilon_f = a\varepsilon_c^b )</td>
<td>(Equation 22)</td>
</tr>
<tr>
<td>( \Delta\sigma )</td>
<td>Analytical fit to stress relaxation data using primary and secondary creep strain equations [125]</td>
<td></td>
<td></td>
</tr>
</tbody>
</table>

In Figure 66, relative to the failure cycles from CF tests, the two DE approaches were clearly a better fit compared to the TF approach. The TF approach overpredicts cycles to
failure and is highly non-conservative with increasing hold time, where a maximum error exceeded one-magnitude for the shortest strain range at each test temperature. Predictions made by the two DE approaches were similar. The simplification made to develop the strain rate independent approach appears to be reasonable and does not significantly affect the prediction [115].
CHAPTER 3. EXPERIMENTAL METHODS

3.1 Material Information

3.1.1 Wrought 316L SS

Wrought 316L SS test specimens were machined from bar stocks purchased from McMaster-Carr. These bars were cold worked then annealed prior to acquisition. The bar stocks comply with ASTM A276-13a [126]. The chemical composition of the 316L bar used is shown in Table 13.

Table 13: Wrought 316L chemical composition

<table>
<thead>
<tr>
<th>C</th>
<th>Cr</th>
<th>Mn</th>
<th>Mo</th>
<th>Ni</th>
<th>P</th>
<th>S</th>
</tr>
</thead>
<tbody>
<tr>
<td>0.020</td>
<td>16.620</td>
<td>1.700</td>
<td>2.050</td>
<td>10.06</td>
<td>0.035</td>
<td>0.021</td>
</tr>
</tbody>
</table>

3.1.2 Hot Rolled 316H SS

A hot rolled 316H plate complying under ASTM A240/A240M-19 standard was obtained from EPRI [127]. Image of this plate is shown in Figure 67. The plate underwent a solution anneal heat treatment at 1052°C for 1 hour then water quenched. The material
had an ASTM 6.0 grain diameter (40µm). The chemical composition of this plate is shown in Table 14.

Figure 67: Hot rolled 316H plate provided by EPRI. Yellow arrow indicates rolling direction.
Table 14: Chemical composition of wrought 316H SS

<table>
<thead>
<tr>
<th>C</th>
<th>Cr</th>
<th>Mn</th>
<th>Mo</th>
<th>Ni</th>
<th>P</th>
<th>S</th>
</tr>
</thead>
<tbody>
<tr>
<td>0.050</td>
<td>16.91</td>
<td>1.420</td>
<td>2.000</td>
<td>10.06</td>
<td>0.031</td>
<td>0.0005</td>
</tr>
</tbody>
</table>

<table>
<thead>
<tr>
<th>Si</th>
<th>Cu</th>
<th>N</th>
<th>Ti</th>
<th>Nb</th>
<th>Co</th>
</tr>
</thead>
<tbody>
<tr>
<td>0.403</td>
<td>0.412</td>
<td>0.0436</td>
<td>0.004</td>
<td>0.017</td>
<td>0.201</td>
</tr>
</tbody>
</table>

3.1.3 DED AM 316H SS

AM 316H SS specimens were built at Auburn University using the Optomec Lens 500 DED system. Powder characteristics and printing parameters are shown in Table 15 through Table 17. Chemical compositions of the as-built bars are given in Table 18. The DED printed bars are shown in Figure 68. Some of the specimens were solution annealed at 1121°C for 1 hour followed by water quench. Vertical specimens were loaded along the build direction, whereas the horizontal specimens were loaded perpendicular to the build direction.
Table 15: 316H SS powder characteristics

<table>
<thead>
<tr>
<th></th>
<th>10th Percentile</th>
<th>50th Percentile</th>
<th>90th Percentile</th>
</tr>
</thead>
<tbody>
<tr>
<td>Powder Diameter</td>
<td>57.2 µm</td>
<td>89.7 µm</td>
<td>145.9 µm</td>
</tr>
<tr>
<td>Powder Apparent Density</td>
<td></td>
<td>4.38 g/cm³</td>
<td></td>
</tr>
</tbody>
</table>

Table 16: Printing parameters used for all DED 316H SS specimens

<p>| | |</p>
<table>
<thead>
<tr>
<th></th>
<th></th>
</tr>
</thead>
<tbody>
<tr>
<td>Scanning Speed</td>
<td>838 mm/min. (33 in/min)</td>
</tr>
<tr>
<td>Layer Thickness</td>
<td>0.38 mm (0.015 in)</td>
</tr>
<tr>
<td>Hatch Spacing</td>
<td>0.38 mm (0.015 in)</td>
</tr>
<tr>
<td>Printing scheme</td>
<td>Print material edge first then fill material center</td>
</tr>
</tbody>
</table>
Table 17: DED 316H SS specimen-by-specimen printing parameters and heat treatments

<table>
<thead>
<tr>
<th>Specimen Name</th>
<th>Build Direction</th>
<th>Heat Treatment</th>
<th>Laser Power</th>
<th>Gas Condition</th>
</tr>
</thead>
<tbody>
<tr>
<td>V-AB-400</td>
<td>Vertical</td>
<td>As-built</td>
<td>400W</td>
<td>Argon</td>
</tr>
<tr>
<td>V-AB-550</td>
<td>Vertical</td>
<td>As-built</td>
<td>550W</td>
<td>Argon</td>
</tr>
<tr>
<td>V-SA-400</td>
<td>Vertical</td>
<td>Solution Anneal</td>
<td>400W</td>
<td>Argon</td>
</tr>
<tr>
<td>V-SA-550</td>
<td>Vertical</td>
<td>Solution Anneal</td>
<td>550W</td>
<td>Argon</td>
</tr>
<tr>
<td>H-AB-550</td>
<td>Horizontal</td>
<td>As-built</td>
<td>550W</td>
<td>Argon</td>
</tr>
<tr>
<td>H-SA-550</td>
<td>Horizontal</td>
<td>Solution Anneal</td>
<td>550W</td>
<td>Argon</td>
</tr>
<tr>
<td>V-HI-400</td>
<td>Vertical</td>
<td>Hot Isostatic Press</td>
<td>400W</td>
<td>Argon</td>
</tr>
<tr>
<td>V-SA-400-A</td>
<td>Vertical</td>
<td>Solution Anneal</td>
<td>400W</td>
<td>Air</td>
</tr>
</tbody>
</table>
Table 18: Chemical compositions of DED 316H bars built in argon gas and air environment. Only nitrogen and oxygen content were analyzed for specimens built in air.

<table>
<thead>
<tr>
<th></th>
<th>C</th>
<th>Cr</th>
<th>Mn</th>
<th>Mo</th>
<th>Ni</th>
<th>P</th>
<th>S</th>
<th>Si</th>
<th>Cu</th>
<th>N</th>
<th>O</th>
</tr>
</thead>
<tbody>
<tr>
<td>Argon</td>
<td>0.04</td>
<td>16.60</td>
<td>1.38</td>
<td>2.28</td>
<td>10.96</td>
<td>0.006</td>
<td>0.005</td>
<td>0.97</td>
<td>0.14</td>
<td>0.09</td>
<td>0.001</td>
</tr>
<tr>
<td>Air</td>
<td>-</td>
<td>-</td>
<td>-</td>
<td>-</td>
<td>-</td>
<td>-</td>
<td>-</td>
<td>-</td>
<td>-</td>
<td>0.07</td>
<td>0.11</td>
</tr>
</tbody>
</table>

Figure 68: DED 316H bars. Print direction, a) longitude direction and b) out of page.

3.2 Specimen Machining and Polishing

316L and 316H cylindrical dog-bone specimens were machined in lathe for fatigue and CF testing. Specimen dimensions are shown in Figure 69. All specimens were
machined at the Montgomery Machining Mall located at the Georgia Institute of Technology. Specific to the 316H plate provided by EPRI, specimens were extracted parallel and perpendicular to the rolling direction. After machining, the specimen gauge sections were hand polished along the loading direction to remove any circumferential machining marks that may introduce stress concentrations consistent with ASTM E606-21 [128]. A succession of finer grit SiC abrasive paper was used: 800, 1200, 2000, and finally 4000 grit SiC.

![Figure 69: Cylindrical dog-bone test specimen. Dimensions in inches.](image)

### 3.3 Experiment Setup

The LCF and CF tests were conducted on either MTS servohydraulic test system (Test frame 5) and MTS dual servohydraulic test system (Test frame 13) located in the Mechanical Properties Characterization Facility (MPCF) at the Georgia Institute of
Technology. Test frame 5 was equipped with a MTS FlexTest40 and Test frame 13 was equipped with a MTS TestStar IIs digital controller. The axial load capacity of each system was 100 kN. Both test frames were equipped with water-cooled MTS model 646 hydraulic collet grips. The 0.5 in. diameter ends of the specimen were inserted to depth of 17.8 mm (0.7 in.) of these grips.

Specimens were heated by induction using an Ameritherm 2kW induction work head. Two different coil designs were implemented during this investigation. Figure 70 is the initial coil (IC1) used on 316L SS and Figure 71 is the revised coil design (IC2) used on 316H SS. ASTM E606-21 and ASTM E2714-13 require the gauge section temperature to be within 1.0% of the testing temperature [92, 128]. The initial coil design maintained a gauge section temperature to within 2.5%, while the revised coil design improved the uniformity of temperature along the gauge section within the 1.0% temperature deviation. This was verified by spot welding five thermocouples along the length of the gauge section.
Figure 70: Specimen gripped in test frame 5 in MPCF showing early induction coil design.
Figure 71: Specimen gripped in test frame 13 in MPCF showing induction coils and MTS model 632.52E-14 high temperature extensometer.

K-type thermocouples and pyrometers were used for temperature monitoring and control. K-type thermocouples were spot welded on the test specimen using a Unitek model 1-163-02 spot welder. When spot welding, the thermocouple wires were pressed against the material surface with the electrode at selected locations with the leads spaced about 1 mm apart in the circumferential direction.

Two different thermocouple placements were used. Figure 72 shows the earlier configuration for thermocouple locations (TC1). The central thermocouple was used for primary specimen temperature monitoring. This thermocouple was connected to a Eurotherm 3204PID temperature controller, of which the temperature controller was
connected to the respective MTS controllers. The other thermocouples on each test specimen were used to confirm temperature uniformity throughout the gauge section. These were monitored using an Omega CL3512A thermocouple calibrator. When analyzing the fracture behavior of tests using this configuration in early testing, some of the fatigue cracks nucleated at the spot weld of the middle thermocouple.

![Diagram of a specimen with thermocouple placements](image)

**Figure 72: TC1 where the red dot indicate where thermocouples leads were spot welded.**

A second improved method for monitoring and controlling temperature was implemented to eliminate the spot-welded thermocouple in the gauge section, and instead a pyrometer was used to measure the mid-specimen temperature. Figure 73 shows the thermocouple and pyrometer spot locations on these specimens. A Williamson 91-15-C pyrometer was used for primary temperature monitoring along the gauge section. The K-type thermocouples were placed just outside the gauge section. The bottom and top thermocouples were connected to the Eurotherm 3204PID temperature controller and
Omega CL3512A thermocouple calibrator, respectively. The temperature at the gauge section was controlled from the bottom thermocouple with reference to the initial temperature reading by the pyrometer. The pyrometer displayed large drift in temperature, up to 100°C, for tests with extreme failure times. However, using the thermocouple placements shown in Figure 72 it was confirmed that no such drift had occurred. The temperature drift by the pyrometer is likely due to oxidation, which changes material surface reflectivity [129]. The deviation in temperature from the primary thermocouples were ±1°C of the designated testing temperature. In both setups, all thermocouple wires were passed through the bottom of the coil.

![Diagram](image)

Figure 73: TC2 showing thermocouple and pyrometer measurement locations where the red dots indicate thermocouple leads and the blue dot indicates the pyrometer measurement location.

Axial displacement of the specimen gauge section was measured using the MTS model 632.52E-14 high-temperature extensometer shown in Figure 70 and Figure 71. The
nominal gauge section between the tips was 12.7 mm (1/2 in.). The extensometer rods were made of alumina and the tips were conical point. Prior to gripping the specimen in the test frame, dimpled features were placed using an alignment tool and 17-4 PH SS hole punch, Figure 74 and Figure 75, respectively. The conical tips of the extensometer were inserted into the dimples to prevent the extensometer from slipping during testing. The extensometer was set prior to heating the specimen to test temperature.

Figure 74: Dimple alignment tool with specimen
Before initiating LCF and CF tests, specimens were put in force control, held at zero force, heated to testing temperature in four minutes, and held at this temperature for 30 min. ASTM E2714-13 places the 30 min. temperature hold requirement to achieve temperature homogeneity of the specimen gauge section [92]. The specimens were then subjected to a triangular wave form with a stress ratio of -1 and force amplitude of 2.7 kN, roughly 85.5 MPa, for five cycles at 0.333 hz, Figure 76. This is roughly one-half the yield strength at the 650°C temperature. This waveform was used to determine the Young’s modulus and to verify test setup before starting the LCF or CF test. Upon completion of the modulus check, strain accumulated from thermal expansion were zeroed using the test frame controller.
3.4 LCF and CF Testing

All cyclic tests were initiated immediately after the modulus check was completed. Several types of cyclic tests were conducted. LCF tests in strain-control were conducted with guidance from ASTM E606-21 [128]. Three different CF tests were conducted: (i) strain-controlled isothermal CF test, (ii) force-dwell (FD) ratchetting CF test, and (iii) FD non-ratchetting CF test. Strain-controlled isothermal CF tests were conducted with the guidance of ASTM E2714-13 [92]. Figure 77 illustrates the CF wave form used for the
case of the tensile dwell. Dwell periods are introduced at peak tensile or peak compressive strains.

Figure 77: Standard CF test waveform of 0.6% strain range and 1 min. dwell.

The FD ratchetting CF test is a non-standard test in which cycling is performed in strain-control and the dwell in force control. An example FD ratchetting CF test waveform is shown in Figure 78. The strain range is fixed, resulting in ratchetting of the strain at each cycle.
Figure 78: FD ratchetting CF test waveform of 0.6% strain range and 1 min. dwell.

Figure 79 describe the wave form produced from a FD non-ratchetting CF test. The waveform is similar to the ratchetting test. Cycling is performed in strain-control and the dwell in force control. The target strains during the loading and unloading sequences are kept fixed and the strain is allowed to increase during the force-controlled dwell period.
Figure 79: FD non-ratchetting CF test waveform of 0.6% strain range and 1 min. dwell.

The failure criterion used to define cycles to failure is illustrated in Figure 80. The failure criterion is defined as a 20% reduction in maximum stress from a linear extension of the maximum stress evolution during the quasi-stable regime on linear-log axes. All tests were stopped prior to complete specimen separation. Specifically for the FD ratchetting tests, the continually increasing strain resulting in the extensometer reaching its upper limit and hence that test was stopped prior to reaching the failure criterion.
3.5 Test Matrix and Motivation

The test matrix used in this study is summarized in Table 19. All tests were conducted with a strain rate of $2 \times 10^{-3}$ s$^{-1}$. LCF and isothermal strain-controlled CF tests used a strain ratio of -1. The force dwell tests had varying strain ratios due to the additional creep strain during the dwell.
Table 19: LCF and CF test matrix

<table>
<thead>
<tr>
<th>T (°C)</th>
<th>$\Delta \varepsilon$ (%)</th>
<th>Dwell Time (min)</th>
<th>Dwell Type</th>
<th>Test Frame</th>
<th>TC Location</th>
<th>Material</th>
</tr>
</thead>
<tbody>
<tr>
<td>550</td>
<td>0.6</td>
<td>0</td>
<td>-</td>
<td>5</td>
<td>TC1</td>
<td>316L</td>
</tr>
<tr>
<td>600</td>
<td>1.0</td>
<td>0</td>
<td>-</td>
<td>5</td>
<td>TC1</td>
<td>316L</td>
</tr>
<tr>
<td>600</td>
<td>0.6</td>
<td>1</td>
<td>Tensile</td>
<td>5</td>
<td>TC1</td>
<td>316L</td>
</tr>
<tr>
<td>600</td>
<td>0.6</td>
<td>0</td>
<td>-</td>
<td>5</td>
<td>TC1</td>
<td>316L</td>
</tr>
<tr>
<td>650</td>
<td>1.0</td>
<td>0</td>
<td>-</td>
<td>5</td>
<td>TC1</td>
<td>316L</td>
</tr>
<tr>
<td>650</td>
<td>1.0</td>
<td>0</td>
<td>-</td>
<td>13</td>
<td>TC2</td>
<td>Wrought 316H</td>
</tr>
<tr>
<td>650</td>
<td>1.0</td>
<td>0</td>
<td>-</td>
<td>5</td>
<td>TC1</td>
<td>DED 316H</td>
</tr>
<tr>
<td>650</td>
<td>1.0</td>
<td>1</td>
<td>Tensile</td>
<td>5</td>
<td>TC1</td>
<td>316L</td>
</tr>
<tr>
<td>650</td>
<td>1.0</td>
<td>1</td>
<td>FD, non-ratchet</td>
<td>5</td>
<td>TC1</td>
<td>316L</td>
</tr>
<tr>
<td>650</td>
<td>1.0</td>
<td>1</td>
<td>FD, ratchet</td>
<td>5</td>
<td>TC1</td>
<td>316L</td>
</tr>
<tr>
<td>650</td>
<td>1.0</td>
<td>3</td>
<td>Tensile</td>
<td>5</td>
<td>TC1</td>
<td>316L</td>
</tr>
<tr>
<td>650</td>
<td>1.0</td>
<td>3</td>
<td>Compressive</td>
<td>5</td>
<td>TC1</td>
<td>316L</td>
</tr>
<tr>
<td>650</td>
<td>1.0</td>
<td>5</td>
<td>Tensile</td>
<td>5</td>
<td>TC1</td>
<td>316L</td>
</tr>
<tr>
<td>650</td>
<td>1.0</td>
<td>10</td>
<td>Tensile</td>
<td>5</td>
<td>TC1</td>
<td>316L</td>
</tr>
</tbody>
</table>
Table 19: LCF and CF test matrix

<table>
<thead>
<tr>
<th>Temp (°C)</th>
<th>Temp Factor</th>
<th>Time (h)</th>
<th>Mode</th>
<th>Cycles</th>
<th>Type</th>
<th>Material</th>
</tr>
</thead>
<tbody>
<tr>
<td>650</td>
<td>1.0</td>
<td>30</td>
<td>Tensile</td>
<td>13</td>
<td>TC1</td>
<td>Wrought 316H</td>
</tr>
<tr>
<td>650</td>
<td>0.6</td>
<td>0</td>
<td>-</td>
<td>5</td>
<td>TC1</td>
<td>316L</td>
</tr>
<tr>
<td>650</td>
<td>0.6</td>
<td>0</td>
<td>-</td>
<td>13</td>
<td>TC2</td>
<td>Wrought 316H</td>
</tr>
<tr>
<td>650</td>
<td>0.6</td>
<td>0</td>
<td>-</td>
<td>5</td>
<td>TC1</td>
<td>H-SA-550</td>
</tr>
<tr>
<td>650</td>
<td>0.6</td>
<td>0</td>
<td>-</td>
<td>5</td>
<td>TC1</td>
<td>V-SA-550</td>
</tr>
<tr>
<td>650</td>
<td>0.6</td>
<td>0</td>
<td>-</td>
<td>5</td>
<td>TC1</td>
<td>V-SA-400</td>
</tr>
<tr>
<td>650</td>
<td>0.6</td>
<td>1</td>
<td>Tensile</td>
<td>5</td>
<td>TC1</td>
<td>316L</td>
</tr>
<tr>
<td>650</td>
<td>0.6</td>
<td>1</td>
<td>FD, non-ratchet</td>
<td>5</td>
<td>TC1</td>
<td>316L</td>
</tr>
<tr>
<td>650</td>
<td>0.6</td>
<td>30</td>
<td>Tensile</td>
<td>13</td>
<td>TC2</td>
<td>Wrought 316H</td>
</tr>
<tr>
<td>650</td>
<td>1.0</td>
<td>1</td>
<td>FD, ratchet</td>
<td>5</td>
<td>TC1</td>
<td>316L</td>
</tr>
<tr>
<td>650</td>
<td>0.4</td>
<td>0</td>
<td>-</td>
<td>5</td>
<td>TC1</td>
<td>316L</td>
</tr>
<tr>
<td>700</td>
<td>0.6</td>
<td>0</td>
<td>-</td>
<td>5</td>
<td>TC1</td>
<td>316L</td>
</tr>
<tr>
<td>700</td>
<td>0.6</td>
<td>1</td>
<td>Tensile</td>
<td>5</td>
<td>TC1</td>
<td>316L</td>
</tr>
</tbody>
</table>

Variety of temperature, strain range, and dwell times is investigated to identify the parametric responses that the material may have relative to its cyclic life and damage mechanisms. FD CF tests were conducted to evaluate any change in cycle life and damage.
mechanisms when compared to the isothermal strain-controlled CF tests. Different LCF and CF tests also were necessary to calibrate the CF life fraction life prediction models.

3.6 Microscopy Methods

Several microstructural analyses techniques were used to characterize material microstructure of the wrought 316L, wrought 316H, and DED 316H alloys. Table 20 summarize the characterization methods.

Table 20: Microstructure characterization methods

<table>
<thead>
<tr>
<th>Technique</th>
<th>Where Conducted</th>
<th>Alloy Analyzed</th>
</tr>
</thead>
<tbody>
<tr>
<td>Digital Optical Microscopy (OM)</td>
<td>Georgia Institute of Technology</td>
<td>Wrought 316L</td>
</tr>
<tr>
<td>Scanning Electron Microscopy (SEM)</td>
<td>University of Nebraska- Lincoln</td>
<td>DED 316H</td>
</tr>
<tr>
<td>Electron Backscatter Diffraction (EBSD)</td>
<td>Georgia Institute of Technology</td>
<td>Wrought 316L</td>
</tr>
<tr>
<td>X-ray Computed Tomography (CT)</td>
<td>Auburn University</td>
<td>DED 316H</td>
</tr>
</tbody>
</table>
3.6.1 Digital Optical Microscopy

Digital OM using the Keyence VHX-7000 equipment was performed before and after testing on the cross section of the gauge section. Crack type, transgranular or intergranular, as well the frequency of these cracks were characterized. LCF and CF crack responses were compared to identify if CF interactions had occurred from CF tests.

Tested specimens were sectioned as illustrated in Figure 81. The Struers Labotom-3 high speed table-top cutting machine with an alumina cut-off wheel was used first to remove the specimen ends. The Buehler Isomet low speed saw with a cubic boron nitride blade was then used to cut the gauge section into two pieces. The cut was made along the middle of the largest crack found on the material surface. The separated gauge sections were cold mounted with epoxy resin to form the one-inch diameter microscopy samples shown in Figure 82. After the epoxy resin cured, the cold mounted samples were polished with the recipe shown in Table 21, using the Struers RotoPol-15 polisher. V2A etchant was used when a detailed image of the microstructural grains was needed. The recipe of the etchant is shown in Table 22. The microscopy samples were immersed in the etchant for 3 min. at room temperature.
Figure 81: Specimen sectioning lines. High speed cut made first, followed by low speed cut.

Figure 82: Polished cross section within the gauge section to investigate cracking behavior.
Table 21: 316 SS polishing method for microscopy

<table>
<thead>
<tr>
<th>Polishing Cloth</th>
<th>SiC Grinding Paper</th>
<th>Streuers MD-Largo</th>
<th>Struers MD-Dac</th>
<th>Struers MD-Nap</th>
<th>Struers MD-Chem</th>
</tr>
</thead>
<tbody>
<tr>
<td>Abrasive Type</td>
<td>Diamond</td>
<td>Diamond</td>
<td>Diamond</td>
<td>Diamond</td>
<td>Silica</td>
</tr>
<tr>
<td>Size</td>
<td>180 grit 9 μm</td>
<td>3 μm</td>
<td>1 μm</td>
<td>0.25 μm</td>
<td></td>
</tr>
<tr>
<td>Lubricant</td>
<td>Water</td>
<td>Struers DiaPro</td>
<td>Struers DiaPro Dac 3</td>
<td>Struers DiaPro Nap B 1</td>
<td>Struers OP-S NonDry</td>
</tr>
<tr>
<td>RPM</td>
<td>300 150 150 150 150</td>
<td></td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td>Force (N)</td>
<td>25 40 20 20 15</td>
<td></td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td>Time (min)</td>
<td>As needed 5 4 4 3</td>
<td></td>
<td></td>
<td></td>
<td></td>
</tr>
</tbody>
</table>

Table 22: V2A etchant recipe

<table>
<thead>
<tr>
<th>Chemical</th>
<th>Volume (ml)</th>
</tr>
</thead>
<tbody>
<tr>
<td>Hydrochloric acid</td>
<td>50</td>
</tr>
<tr>
<td>Nitric acid</td>
<td>5</td>
</tr>
<tr>
<td>Distilled water</td>
<td>50</td>
</tr>
</tbody>
</table>
3.6.2 Scanning Electron Microscopy

SEM was conducted on as-built and solution annealed vertical DED 316H specimens shown in Figure 68. The microstructural differences relative to build height and solution annealing heat treatment were of interest. These analyses were conducted on FEI Helios NanoLab 660 equipment. SEM images were taken on the specimen cross section along the gauge section.

3.6.3 Electron Backscatter Diffraction

EBSD was conducted on the cross section along the gauge section of untested 316L SS specimen to obtain the grain size and identify any notable microstructural features that may contribute towards LCF and CF failure. The untested specimen, sectioned as shown in Figure 81, underwent a polishing recipe outlined in Table 21. The linear intercept method was used to obtain the grain size of the material. A random straight line was drawn through the EBSD image quality map. The average grain size was calculated as the number of grain boundaries intersecting with the line divided by the actual line length. This process was repeated ten times. The average and standard deviation grain size are reported.

3.6.4 X-ray CT

X-ray CT images of the DED 316H SS specimens were taken before and after testing to determine if any large AM pore defects influenced material failure. All X-ray CT images were taken at Auburn University using their Pinnacle PXS-500/90 equipment. The minimum porosity detection limit was 20µm.
Two different images were taken for the failed specimens: as failed and open crack.

An axial force of 1.5 kN was applied to the failed specimen to open cracks for the open crack state. This was one-half of the 0.02% yield strength recorded from the DED 316H specimen with the smallest 0.02% yield strength.
CHAPTER 4. RESULTS AND DISCUSSION ON 316L SS

4.1 Monotonic Behavior

The Young's modulus and 0.02% yield strength of wrought 316L SS determined from the modulus check performed prior to testing and the first cycle of each cyclic test, respectively, is shown in Figure 83 and Figure 84. The average values and standard deviations, error bars, taken from LCF and CF tests are summarized in Table 23. The decrease in Young’s modulus and yield strength with increasing temperature is consistent with what is reported [8].
Figure 83: Average Young’s modulus for wrought 316L tested at different temperatures.
Figure 84: Average 0.02% yield strength for wrought 316L tested at different temperatures.
Table 23: Average wrought 316L monotonic properties from first cycle

<table>
<thead>
<tr>
<th>Temperature (°C)</th>
<th>Young’s Modulus (GPa)</th>
<th>0.02% Yield Strength (MPa)</th>
</tr>
</thead>
<tbody>
<tr>
<td>550</td>
<td>156.7</td>
<td>168.1</td>
</tr>
<tr>
<td>600</td>
<td>151.7 ± 1.3</td>
<td>161.7 ± 3.3</td>
</tr>
<tr>
<td>650</td>
<td>143.6 ± 4.7</td>
<td>153.3 ± 14.4</td>
</tr>
<tr>
<td>700</td>
<td>142.8 ± 2.0</td>
<td>140.6 ± 1.7</td>
</tr>
</tbody>
</table>

4.2 Strain-Cycles to Failure Outcomes

The strain-life plots for LCF and CF of 316L SS are shown in Figure 85. The LCF and CF data were fit to a Coffin-Manson relation, Equation 4, to determine if the life data could be related through the total and inelastic strain range. The fitting parameters and coefficient of determination, $R^2$, are shown in Table 24. $R^2$ is determined from procedure outlined in APPENDIX A.1. The LCF and CF data is summarized in Table 25 which includes the maximum stress and stress relaxation at half-life. The total and inelastic strain range at taken at half-life. The ASTM E2714-13 definition of inelastic strain range, width of hysteresis loop at zero stress, was used [92].
Table 24: Fitting parameters to strain-life relations

<table>
<thead>
<tr>
<th></th>
<th>$\sigma' f/E$</th>
<th>$\varepsilon'_f$</th>
<th>b</th>
<th>c</th>
<th>$R^2$ on $\Delta \varepsilon_t/2$</th>
<th>$R^2$ on $\Delta \varepsilon_{in}/2$</th>
</tr>
</thead>
<tbody>
<tr>
<td><strong>Fit to 650°C</strong></td>
<td>0.00332</td>
<td>0.186</td>
<td>-0.073</td>
<td>-0.562</td>
<td>0.94</td>
<td>0.95</td>
</tr>
<tr>
<td><strong>Fit to all</strong></td>
<td>0.00320</td>
<td>0.184</td>
<td>-0.069</td>
<td>-0.560</td>
<td>0.90</td>
<td>0.93</td>
</tr>
</tbody>
</table>
Figure 85: Strain-life data fit to (a) Coffin-Manson and (b) total strain-life relations.
Table 25: LCF and CF tests on 316L SS

<table>
<thead>
<tr>
<th>T (°C)</th>
<th>$\Delta \varepsilon_t$ (%)</th>
<th>Dwell Time (min)</th>
<th>Dwell Type</th>
<th>$N_f$</th>
<th>$\sigma_{max}$ (MPa)</th>
<th>$\Delta \sigma$ (MPa)</th>
<th>$\Delta \varepsilon_{in}$ (%)</th>
</tr>
</thead>
<tbody>
<tr>
<td>550</td>
<td>0.6</td>
<td>0</td>
<td>-</td>
<td>4740</td>
<td>263.80</td>
<td>0</td>
<td>0.22</td>
</tr>
<tr>
<td>600</td>
<td>1.0</td>
<td>0</td>
<td>-</td>
<td>970</td>
<td>327.05</td>
<td>0</td>
<td>0.53</td>
</tr>
<tr>
<td>600</td>
<td>1.0</td>
<td>1</td>
<td>Tensile</td>
<td>1285</td>
<td>321.26</td>
<td>45.26</td>
<td>0.56</td>
</tr>
<tr>
<td>600</td>
<td>0.6</td>
<td>0</td>
<td>-</td>
<td>3690</td>
<td>265.48</td>
<td>0</td>
<td>0.23</td>
</tr>
<tr>
<td>650</td>
<td>1.0</td>
<td>0</td>
<td>-</td>
<td>880</td>
<td>299.28</td>
<td>0</td>
<td>0.56</td>
</tr>
<tr>
<td>650</td>
<td>1.0</td>
<td>1</td>
<td>Tensile</td>
<td>1040</td>
<td>295.98</td>
<td>53.98</td>
<td>0.60</td>
</tr>
<tr>
<td>650</td>
<td>1.18</td>
<td>1</td>
<td>FD, non-ratchet</td>
<td>580</td>
<td>297.40</td>
<td>0</td>
<td>0.74</td>
</tr>
<tr>
<td>650</td>
<td>1.18</td>
<td>1</td>
<td>FD, non-ratchet</td>
<td>470</td>
<td>289.26</td>
<td>0</td>
<td>0.74</td>
</tr>
<tr>
<td>650</td>
<td>1.0</td>
<td>3</td>
<td>Tensile</td>
<td>690</td>
<td>285.45</td>
<td>79.05</td>
<td>0.64</td>
</tr>
<tr>
<td>650</td>
<td>1.0</td>
<td>3</td>
<td>Compressive</td>
<td>720</td>
<td>277.18</td>
<td>-57</td>
<td>0.64</td>
</tr>
<tr>
<td>650</td>
<td>1.0</td>
<td>3</td>
<td>Compressive</td>
<td>640</td>
<td>283.49</td>
<td>-67</td>
<td>0.63</td>
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<tr>
<td>650</td>
<td>1.0</td>
<td>5</td>
<td>Tensile</td>
<td>790</td>
<td>284.74</td>
<td>83.74</td>
<td>0.63</td>
</tr>
<tr>
<td>650</td>
<td>1.0</td>
<td>5</td>
<td>Tensile</td>
<td>805</td>
<td>281.62</td>
<td>87.62</td>
<td>0.64</td>
</tr>
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<td>650</td>
<td>1.0</td>
<td>5</td>
<td>Tensile</td>
<td>865</td>
<td>285.43</td>
<td>69.43</td>
<td>0.63</td>
</tr>
<tr>
<td>650</td>
<td>1.0</td>
<td>10</td>
<td>Tensile</td>
<td>774$^1$</td>
<td>279.45</td>
<td>202.31</td>
<td>0.64</td>
</tr>
<tr>
<td>650</td>
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<td>0</td>
<td>-</td>
<td>2980</td>
<td>238.10</td>
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<td>0.23</td>
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<tr>
<td>650</td>
<td>0.6</td>
<td>1</td>
<td>Tensile</td>
<td>2860</td>
<td>240.83</td>
<td>38.83</td>
<td>0.28</td>
</tr>
<tr>
<td>650</td>
<td>0.6</td>
<td>1</td>
<td>FD</td>
<td>1930</td>
<td>234.72</td>
<td>0</td>
<td>0.30</td>
</tr>
<tr>
<td>650</td>
<td>0.4</td>
<td>0</td>
<td>-</td>
<td>23421</td>
<td>212.81</td>
<td>0</td>
<td>0.08</td>
</tr>
<tr>
<td>650</td>
<td>0.4</td>
<td>0</td>
<td>-</td>
<td>26370</td>
<td>205.42</td>
<td>0</td>
<td>0.10</td>
</tr>
<tr>
<td>700</td>
<td>0.6</td>
<td>0</td>
<td>-</td>
<td>2600</td>
<td>231.62</td>
<td>0</td>
<td>0.28</td>
</tr>
<tr>
<td>700</td>
<td>0.6</td>
<td>1</td>
<td>Tensile</td>
<td>1600</td>
<td>220.35</td>
<td>61.65</td>
<td>0.30</td>
</tr>
</tbody>
</table>

$^1$ CF test was not tested to failure.
For a fixed total strain range, the inelastic strain range increases as temperature increases due to the reduction in strength with temperature. Introduction of dwell periods results in a decrease in cyclic life if dwell is sufficiently long. Even with these differences in temperatures and dwell periods, the LCF and CF data points are within a factor of 2 of the strain-life relation curves. The decrease in life with increasing temperature or dwells is simply captured by the increase in the inelastic strain range. It can be concluded that the decreased cycles to failure is not due to a CF interaction. These results show that the failure is dominated by the LCF behavior over the test conditions considered here, LCF and CF strain ranges from 0.6% and 1.0%, and CF tests with hold times of up to 5 minutes at temperatures of 550 to 700°C.

4.3 Cyclic Stress-Strain Behavior

The stress-strain behavior for the first 10 cycles and at half-life at 650°C is shown in Figure 86 (a) and (b), respectively. The initial cyclic hardening is much more rapid at larger total strain ranges. The inelastic strain range decreased during these initial transient cycles due to the increase in the elastic strain range with hardening. The evolution of the maximum stress is shown in a semi-log plot in Figure 86 (c) and linear axes in Figure 86 (d). Three distinct behaviors are identified as observed in the linear-log plot: initial cyclic hardening, followed by stable or quasi-stable maximum stress, and finally rapid stress decline. Initial cyclic hardening reached saturation in 20 cycles when strain range is 1.0% and in 300 cycles when the strain range is smaller, 0.6%. During the quasi-stable cycles,
the material cyclically softened. This behavior was dominant for a large part of the cyclic life and is clearly observed when linear-linear axes.

Figure 86: Cyclic behaviors of different strain range 650°C LCF tests. (a) Hysteresis loop from first to tenth cycle, (b) half-life hysteresis loop, (c) semi-log axes maximum stress-cycles, (d) linear axes maximum stress-cycles.
Figure 86: Cyclic behavior of 650°C LCF tests (b).
Figure 86: Cyclic behavior of 650°C LCF tests (c).
The influence of temperature on the cyclic stress-strain behavior for strain range of 0.6% under LCF is shown in Figure 87. In general, the peak tensile stress is reduced at greater temperatures, though the peak tensile stress at half-life for the 550°C and 600°C is similar. The initial cyclic hardening saturated more quickly as the temperature increased. At a temperature of 550°C quasi-stability was reached at 200 cycles, while at 700°C, quasi-stability was reached in just 20 cycles. The maximum stress during the quasi-stable and rapid stress decline region decreased with increasing temperature.
Figure 87: Cyclic behaviors of different temperature 0.6% strain range LCF tests (a) Half-life hysteresis loop and (b) maximum stress-cycles.
The LCF and CF responses of 316L SS tested at 650°C are shown in Figure 88. The strain-controlled isothermal creep-fatigue test with a tensile or compressive dwell experienced tensile or compressive stress relaxation, respectively. The FD non-ratchetting CF resulting in both an increase in the total and inelastic strain ranges.

Stress relaxation responses during the dwell at half-life are shown in Figure 88 (b) and Figure 88 (c). All CF tests exhibited a rapid stress decline followed by a gradual stress
decline. This is clearly shown in the stress-normalized time plot where the stress and time axes are normalized with the respective maximum stress at half-life and dwell times of each CF test.

Figure 88: Cyclic behaviors of 650°C CF tests. (a) Half-life hysteresis loop, (b) stress-time plot during stress relaxation at half-life, (c) stress-normalized time plot during stress relaxation at half-life, (d) linear-log axes maximum stress-cycles.
Figure 88: Cyclic behavior of 650°C CF tests (b).
Figure 88: Cyclic behavior of 650°C CF tests (c).
When the strain range was 1.0\%, the amount of the stress relaxation was similar for each of the dwell times, 3 min., 5 min., and 10 min. and consequently, the inelastic strain ranges and lives were similar. The CF lives were all within a factor 2 of the LCF life, Figure 85 (a). The FD non-ratchetting CF test on the other hand produced a much larger inelastic strain range than all other tests. A factor of 2 decrease in cyclic life was seen from these tests with respect to LCF tests. But even in this case, it aligns with the Coffin-Manson relation suggesting it is also primarily controlled by LCF with minimal creep contribution.
The evolutions of the maximum stress for different dwell times, dwell periods, and dwell control modes are shown in Figure 88 (d). The evolution of the cyclic stress-strain response in CF was similar to that observed in LCF: initial cyclic hardening, followed by quasi-stable maximum stress, and then rapid maximum stress decline near end of life. As with the LCF tests, the maximum stress declined gradually during the quasi-stable regime. However, all CF tests had a lower maximum stress than the LCF tests after the initial cyclic hardening behavior saturated.

The influence of temperature on the 0.6% strain-controlled CF tests with 1 min. dwell is shown in Figure 89. As temperature increases, the maximum stress is reduced and the amount of stress relaxation increases. This in turn resulted in an increase in inelastic strain range at half-life as temperature increases, corresponding to the shorter life by the higher temperature CF test which still followed the Coffin-Manson plot of the 650°C data shown in Figure 85 (a). From the maximum stress-cycles evolution plot, the initial hardening behavior under CF saturated more quickly at 700°C than at 650°C. The 700°C test reached quasi-stability within the first 10 cycles, while the 650°C test reached quasi-stability after 20 cycles. Under CF, the saturation of the maximum stress during initial hardening is reached more quickly than in LCF. The LCF tests saturated after 100 and 20 cycles for the 650 and 700°C temperatures, respectively.
Figure 89: Cyclic behaviors of different temperature CF tests. (a) Half-life hysteresis loop, (b) stress-time plot during stress relaxation at half-life, (c) linear-log axes maximum stress-cycles.
Figure 89: Cyclic behavior of different temperature CF tests (b).
4.4 Microstructural Analysis Prior to Testing

The microstructure of the 316L bar before testing is shown in Figure 90 through Figure 93. The average grain size was $14.1 \pm 2.78 \mu$m. The line artifacts observed in these images aligned parallel to the axial loading direction. The microstructure also contained some grain boundary twins, typical of annealed 316L stainless steels. The texture pole plot in Figure 92 suggests a homogeneous microstructure.
Figure 90: Microscopy image of untested 316L SS specimen after etching with V2A etchant.
Figure 91: Image quality map from EBSD on untested wrought 316L SS. Specimen was not etched. Loading direction correspond to axial direction of rod.
4.5 Microstructure and Crack Behavior after Testing

LCF and CF crack paths were investigated to identify whether any CF interactions were possibly active. As a baseline the crack initiation and propagation paths for the LCF tested at 650°C are shown in Figure 93 and Figure 94. The cracks formed at the surface and propagated inwards. Both crack initiation and propagation were transgranular. In some of the tests such as shown in Figure 94, the cracks formed at the thermocouple weld-material surface interface when thermocouples were placed within the gauge section.
Figure 93: Crack path of 650°C 0.6% strain range LCF test.

Figure 94: Crack path of 650°C 1.0% strain range LCF test. Dotted circle is thermocouple spot weld.
Crack initiation and propagation responses did not change with change in temperature, Figure 95 and Figure 96. Cracks formed at the surface and propagated towards the center of the specimen and crack propagation was transgranular.

Figure 95: Crack path of 600°C 0.6% strain range LCF test.
The crack behavior for strain-controlled CF conducted at 650°C is shown in Figure 97 and Figure 98. Regardless of strain range or hold times, the cracks formed from the surface and propagated inwards similar to the LCF tests. Crack initiation and propagation were primarily transgranular. Secondary cracks were observed near the crack tips at the end of life, Figure 98, but no clear creep voids along grain boundaries was observed near the primary crack. Specific to Figure 98, multiple cracks initiated from the surface of the specimen. Some of these cracks merged with the primary crack. Although multiple small length scale surface cracks were seen from all LCF and CF tests, the merging behavior was only present in the 650°C, 1.0% strain range, 3 min. dwell strain-controlled isothermal CF test.
Figure 97: Crack path of 650°C, 0.6% strain range, 1 min. dwell standard isothermal CF test. Etched with V2A solution.
Intergranular cracks were observed at the crack tips of the 650°C strain-controlled isothermal creep-fatigue tests under the 0.6% strain range 1 min. dwell and 1.0% strain range 3 min. dwell conditions, as shown in Figure 99 and Figure 100. The presence of these cracks suggest some creep damage had occurred. However, the absence of grain boundary damage around the primary crack and the primarily transgranular crack propagation suggests a weak, competitive creep-fatigue interaction mode, or no creep-fatigue interaction. As the intergranular cracks are only found at the crack tips, it is likely that these cracks formed near the failure cycle. Near the end of life, the stress intensity range is considerably reduced under strain-controlled conditions, resulting in nucleation of
the secondary cracks during the dwell. The vertical black streaks are the line artifacts shown in Figure 90. These artifacts are more visible after etching.

Figure 99: Crack tip of 650°C, 0.6% strain range, 1 min. dwell strain-controlled isothermal CF test. Etched with V2A solution.
The cracking behavior of a force-dwell non-ratchetting CF test is shown in Figure 101. Similar to LCF and strain-controlled CF tests, the force-dwell non-ratchetting CF test initiated a crack at the surface which propagated inwards. Crack initiation and propagation were transgranular. At the crack tip near the end of the test, Figure 102, crack blunting is observed. This feature was not observed in the LCF or conventional isothermal strain-control CF tests suggesting plastic deformation at the crack tip is larger due to force-control dwell periods.
Figure 101: Crack path of 650°C, 1.0% strain range, 1 min. dwell force-dwell non-ratchetting CF test.
Figure 102: Crack tip of 650°C, 1.0% strain range, 1 min. dwell force-dwell non-ratchetting CF test.

FD ratcheting tests were not tested to failure due to the extensometer saturating at its upper travel limit. No cracks were present in these specimens at the time the tests were interrupted. However, the high strains attained by these tests suggest the specimens would fail similar to a slow-strain rate tensile test.
CHAPTER 5. RESULTS AND DISCUSSION ON DED 316H SS

5.1 Monotonic Behavior

A comparison of the wrought and solution annealed (SA) DED 316H Young's modulus and 0.02% yield strength are shown in Figure 103 and Figure 104, respectively. The values are summarized in Table 26. The DED specimens had a greater Young’s modulus than the wrought, with the Young's modulus of the horizontal specimen being up to 22% greater than the vertical specimens and 25% greater than wrought. The yield strength of the horizontal specimens was greater than wrought by 10%, while the yield strength of the vertical specimens was lower than wrought specimen by up to 35%. The difference in Young’s modulus between the horizontal and vertical specimens suggest differing crystallographic textures [130]. Similar difference in Young’s modulus and yield strength were observed in literature [69].
Figure 103: Average Young’s modulus for wrought 316H tested at 650°C. Errors bars are standard deviations from multiple LCF and CF tests.
Figure 104: Average 0.02% yield strength for wrought 316H tested at 650°C.
Table 26: Wrought and DED 316H SS material properties at 650°C obtained from first cycle

<table>
<thead>
<tr>
<th>Specimen</th>
<th>Young’s Modulus (GPa)</th>
<th>0.02% yield strength (MPa)</th>
</tr>
</thead>
<tbody>
<tr>
<td>Wrought</td>
<td>136.3 ± 2.6</td>
<td>121.0</td>
</tr>
<tr>
<td>H-SA-550</td>
<td>175.8</td>
<td>132.8</td>
</tr>
<tr>
<td>V-SA-550</td>
<td>143.6</td>
<td>98.7</td>
</tr>
<tr>
<td>V-SA-400</td>
<td>151.7</td>
<td>113.3</td>
</tr>
</tbody>
</table>

5.2 Cyclic Stress-Strain Behavior

A comparison of the strain-life behavior for wrought and DED 316H cyclic tests at 650°C is shown in Figure 105 and summarized in Table 27. The CF test with a 30 min. dwell period was not tested to failure. Relative to the LCF tests, the wrought specimen had the largest cyclic life. Comparing the wrought and DED behaviors, the difference in life was at most one-magnitude with the H-SA-550 being the lowest life. Interestingly, the wrought specimen had comparable inelastic strain range to V-SA-400, but yielded higher cyclic life.

Comparing the DED specimens, the H-SA-550 specimen had the lowest life. This specimen had the largest inelastic strain range. The high Young’s modulus and 0.02% yield strength contributed towards expanding the inelastic strain range. The V-SA-550 specimen had the longest life out of the three DED specimens. Interestingly, it had the second largest
inelastic strain range, although having the lowest Young’s modulus and 0.02% yield strength. The largest stress at half-life was attained by the V-SA-400 specimen. This specimen also exhibited the greatest difference between the 0.02% yield strength and maximum stress at half-life.

The cyclic stress-strain behavior of wrought and DED 316H are shown in Figure 106.

Figure 105: Strain-life plots for 316HSS cyclic tests at 650°C. (a) Total strain range-cycles to failure, (b) inelastic strain range-cycles to failure, (c) cycles to failure bar chart.
Figure 105: Inelastic strain range-life plot of 316H SS cyclic tests at 650° C (b).
Figure 105: Strain-life plots of 316H SS LCF tests at 650° C (c).
Table 27: LCF and CF test outcomes of 316H SS

<table>
<thead>
<tr>
<th>Specimen</th>
<th>T</th>
<th>$\Delta \varepsilon_t$</th>
<th>Dwell Time</th>
<th>$\Delta \varepsilon_{in}$</th>
<th>$\sigma_{\text{max}}$</th>
<th>$N_f$</th>
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</thead>
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<td>0.57</td>
<td>272.6</td>
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<td>0.26</td>
<td>215.0</td>
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<td>0.29</td>
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<td>0</td>
<td>0.3</td>
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<tr>
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<td>0</td>
<td>0.28</td>
<td>231.4</td>
<td>4805</td>
</tr>
<tr>
<td>V-SA-400</td>
<td>650</td>
<td>0.6</td>
<td>0</td>
<td>0.25</td>
<td>262.6</td>
<td>3180</td>
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</table>

^2 CF test was not cycled to failure.
Figure 106: Cyclic behaviors of 650°C tests on 316H. (a) Hysteresis loop from first ten cycles, (b) half-life hysteresis loops, (c) linear-log axes maximum stress-cycles.
Figure 106: Cyclic behavior of 650°C tests on 316H (b).
During the initial ten cycles, all 0.6% strain range hysteresis loops display stress serrations. The wrought specimen hysteresis loop is stable at half-life, while the serrations remain for the DED specimens. The appearance of serrations suggests possible dynamic strain aging.

The evolution of the maximum stress displayed three distinct cyclic behaviors similar to what was observed in 316L: initial cyclic hardening, stable or quasi-stable maximum stress, and rapid stress decline. The 0.6% strain range tests exhibited a gradual
reduction in maximum stress after reaching the peak value at the completion of initial cyclic hardening, while the 1.0\% strain range test appeared to be stable and not show softening until end of life when crack was growing fast. Examining the initial cyclic hardening behaviors, the H-SA-550 and V-SA-400 specimens hardened to similar maximum stress values upon initial hardening saturation. The DED specimen had greater maximum stress than the wrought specimen after initial hardening behavior saturated. The initial hardening regime saturated near cycle 100 for the wrought specimen, while the DED specimens saturated near cycle 300.

The large difference in LCF life between the wrought and DED specimens are pointed towards the different temperature controlling apparatus used, summarized in Table 19, where the thermocouple spot welds nucleated surface cracks.

### 5.3 Microstructural Analysis Prior to Testing

Figure 107 and Figure 108 show the microstructure of the as-built vertical DED specimens (V-AB-400 and V-AB-550) prior to solution anneal treatment. In each figure, relative to the dog-bone specimens, the top and bottom image are taken at the grip sections and the middle image is taken at the gauge section.
Figure 107: OM images of top, middle and bottom sections of vertical DED 316H prior to SA treatment, built at 400W laser power. Images taken at University of Nebraska-Lincoln.

Figure 108: OM images of top, middle and bottom sections of vertical DED 316H prior to SA treatment, built at 550W laser power. Images taken at University of Nebraska-Lincoln.
Both 400W and 550W laser specimens contain overlapping fish scale like melt pool layers. This feature is uniform throughout the 400W specimen. The fish scale structure evolves with increasing height for the 550W specimen. As the height of the specimen increase, heterogeneity between the center and edge positions emerge. The differences between specimen center and edge appear from the combination of printing scheme, and high laser power. The high power laser produces more heat. Heat accumulates at the edge of specimen because the specimen contour is printed first. These sections are machined off during fatigue specimen preparation.

Higher magnifications of the as-built (AB) microstructure in the middle-center section are shown in Figure 109 and Figure 110.
Figure 109: SEM of AB DED 316H SS printed with 400W laser at different magnifications. Images taken at University of Nebraska-Lincoln.
Figure 109: SEM of AB DED 316H SS with 400W laser (b).
Figure 110: SEM of AB DED 316H SS printed with 550W laser at different magnifications. Images taken at University of Nebraska-Lincoln.
Figure 110: SEM of AB DED 316H SS with 550W laser (b).

There is no obvious microstructure heterogeneity in the 400W and 550W AB specimens. Cellular structures with length scale of 2.8 µm and 5.5 µm were present throughout the microstructure in the 400W and 550W specimens, respectively. Both specimens contained inclusion particles, black specks in SEM images, enriched with silicon, manganese, and/or oxygen, likely produced from the precursor powder.

Partially recrystallized grains were found in both specimens after solution annealing. Cellular structures and inclusion particles remained in both specimens. The solution annealed microstructure in the middle-enter section is shown in Figure 111 and Figure 112.
Figure 111: SEM of SA DED 316H SS printed with 400W laser at different magnifications. Images taken at University of Nebraska-Lincoln.
Figure 111: SEM of SA DED 316H SS with 400W laser (b).
Figure 112: SEM of SA DED 316H SS printed with 550W laser at different magnifications.
Figure 112: SEM of SA DED 316H SS with 550W laser (b).

X-ray CT images of the gauge section of the SA DED 316H specimens are shown in Figure 113 through Figure 115. Table 28 summarizes the void volume fractions in the gauge section from each DED specimen. All pores were spherical. Porosity was uniformly scattered throughout each specimen. The largest pore found on each specimen was around 150 µm in diameter. The H-SA-550 specimen had the largest void volume fraction of 0.00378%. However, void volume fractions for all specimens were lower than 0.1%, implying very high density.
Figure 113: X-ray CT image of gauge section of H-SA-550 specimen prior to testing.
Figure 114: X-ray CT image of gauge section of V-SA-400 specimen prior to testing.
Figure 115: X-ray CT image of gauge section of V-SA-550 specimen prior to testing.
Table 28: Void volume fraction of DED specimens

<table>
<thead>
<tr>
<th>Specimen</th>
<th>Void Volume Fraction (%)</th>
</tr>
</thead>
<tbody>
<tr>
<td>H-SA-550</td>
<td>0.00378</td>
</tr>
<tr>
<td>V-SA-550</td>
<td>0.00021</td>
</tr>
<tr>
<td>V-SA-400</td>
<td>0.00034</td>
</tr>
</tbody>
</table>

5.4 Microstructure and Crack Behavior After Testing

The fatigue cracks that formed in the three DED LCF tests are shown in Figure 116 through Figure 118. All cracks were found along the middle section of the specimen. These cracks formed from the surface of the specimen and propagated inwards. The flat cracks suggest that the cracks propagated transgranularly. Additional microstructural analyses will be required to make a conclusion on crack paths.
Figure 116: Cracks formed on H-SA-550 specimen from X-ray CT.
Figure 117: Cracks formed on V-SA-550 specimen from X-ray CT.
Figure 118: Cracks formed on V-SA-400 specimen from X-ray CT.
CHAPTER 6.  CF LIFE FRACTION MODELING

The different CF interaction life fraction models were evaluated using LCF data for fatigue damage and pure creep data for creep damage. The LCF data was generated in this study while the creep data for 316H SS was compiled by EPRI from multiple sources and various heats, tested at a temperatures in the range of 538 to 850°C and stresses in the range of 18 to 355 MPa [131-137]. A ±95% confidence interval (CI) band was determined when numerical models were fitted to creep rupture data. Coefficient of determination, $R^2$, was also determined when least-square or multiple-linear regression techniques are used. These statistical methods are summarized in APPENDIX A.

6.1 Creep Rupture Time Modelling

For the time-fraction (TF) model, the relationship for creep time to rupture, $t_c$, is determined from LM, OSD, or MH relations (Equation 9, Equation 14, or Equation 16). Here the LM creep time to rupture equation was used. A LM plot was constructed with a LM constant of 16.42 using the data compiled by EPRI [131-137]. The LM-stress data was fitted to a 4$^{th}$ order polynomial (Equation 10),

\[ P_{LM}(\sigma) = -5097\log(\sigma)^4 + 36045\log(\sigma)^3 - 94993\log(\sigma)^2 \\
+ 105858\log(\sigma) - 19539 \]

with stress having units of MPa. The $R^2$ is 0.95. The ±95% CI were determined from APPENDIX A.1. Figure 119 shows the polynomial fit when plotted with 316H SS LM
parameter data. The mean relation characterized the data well. The majority of the data points are captured within the ±95% CI bands. Equation 42 and LM constant of 16.42 was substituted to the LM creep time to rupture equation, to determine creep rupture time, $t_c$.

![Figure 119: LM plot developed from 316H SS creep-rupture data.](image)

6.2 Creep Ductility Modelling

The full creep ductility-average creep strain rate data extracted from multiple sources and heats of 316H SS is shown in Figure 120. The creep ductility-average creep strain rate
data for tests conducted only at 650°C is shown in Figure 121. Creep ductility was taken as creep rupture strain and creep strain rate was taken as average creep strain rate (Equation 20). Creep ductility increases with increasing creep strain rate.

Figure 120: Creep ductility response from 316H SS tested at 538 – 850°C.
Regression fits to creep ductility-creep strain rate data were used when calculating creep damage using the ductility exhaustion (DE) and stress modified ductility exhaustion (SMDE) approaches. Creep ductility-average creep strain rate data was fitted to the power law and logarithmic creep ductility equations (Equation 22 and Equation 23) using linear regression, and the SMDE creep ductility equation (Equation 26) using multiple linear regression. The power law and logarithmic equations were fitted to creep rupture data at 650°C. SMDE equation was fitted to creep rupture data at 650°C and 538 – 850°C.
Methods outlined in APPENDIX A were used to determine the ±95% CI and $R^2$. All regression fits and $R^2$ values are taken prior to applying an upper or lower-shelf ductility.

Upper and lower-shelf ductility for the piece-wise function, Equation 21, were defined using 316H creep rupture and tensile data compiled by EPRI. The minimum creep rupture elongation found in the EPRI dataset was used as the lower-shelf ductility. All fitted curves, including the ±95% CI, incorporated a lower-shelf ductility to prevent calculations of negative creep ductility. Upper-shelf ductility was defined differently for the mean and ±95% CI. The mean curve used the maximum rupture elongation found from tensile tests conducted at 650°C. The +95% and -95% CI’s upper-shelf is the 95th percentile creep rupture elongation and mean creep rupture elongation, respectively. Different upper-shelf ductility were used for the ±95% CI depending on the temperature range used to fit the equations.

Figure 122 describe creep ductility fitted to the logarithmic equation. Table 29 summarizes the fitting parameters used to construct this figure and accompanying $R^2$ value. Relative to the equations that were explored, the lowest $R^2$ value is computed from this equation. Shown in Figure 122 (b) the upper-shelf ductility is attained near $10^{-8}$ 1/s for the +95% CI. The mean curve attained upper-shelf ductility at a higher creep strain rate, $5 \times 10^{-8}$ 1/s.
Figure 122: Creep ductility at 650°C fitted to logarithmic equation. (a) Without (b) with upper-shelf ductility.
Figure 122: Creep ductility at 650°C fitted to logarithmic equation (b).
Table 29: Fitting parameters to piece-wise function and logarithmic equation

<table>
<thead>
<tr>
<th>T (°C)</th>
<th>A</th>
<th>B</th>
<th>Upper-Shelf (mm/mm)</th>
<th>Lower-Shelf (mm/mm)</th>
<th>R²</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td></td>
<td></td>
<td>+95%</td>
<td>Mean</td>
<td>-95%</td>
</tr>
<tr>
<td>650</td>
<td>0.14</td>
<td>1.04</td>
<td>-</td>
<td>-</td>
<td>-</td>
</tr>
<tr>
<td>650</td>
<td>0.14</td>
<td>1.04</td>
<td>0.76</td>
<td>0.58</td>
<td>0.40</td>
</tr>
</tbody>
</table>

Figure 123 describe creep ductility fitted to the power law equation. Table 30 summarizes the fitting parameters and R² value to construct this figure. The low R² value suggests a poor fit to data, although a higher value is computed than the previous equation and its fitting parameters. The lower-shelf ductility is not attained by the mean curve or ±95% CI within the plotted creep strain rates. The mean curve overpredict creep ductility beyond a creep strain rate of 10⁻⁶ 1/s.
Figure 123: Creep ductility at 650°C fitted to power law equation. (a) Without (b) with upper-shelf ductility.
Figure 123: Creep ductility at 650°C fitted to power law equation (b).
Table 30: Fitting parameters to piece-wise function and power law equation

<table>
<thead>
<tr>
<th>T (°C)</th>
<th>a</th>
<th>b</th>
<th>Upper-Shelf (mm/mm)</th>
<th>Lower-Shelf (mm/mm)</th>
<th>R²</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td></td>
<td></td>
<td>+95% Mean -95%</td>
<td></td>
<td></td>
</tr>
<tr>
<td>650</td>
<td>14.4</td>
<td>0.217</td>
<td>- - -</td>
<td>0.03</td>
<td>0.48</td>
</tr>
<tr>
<td>650</td>
<td>14.4</td>
<td>0.217</td>
<td>0.76 0.58 0.40</td>
<td>0.03</td>
<td>0.48</td>
</tr>
</tbody>
</table>

Figure 124 and Figure 125 describe creep ductility fitted to the SMDE equation. Figure 124 is fitted to creep ductility data for a temperature range of 538 – 850°C (811 – 1123K) and Figure 125 is fitted to a temperature of 650°C (923K). Fitting parameters and R² values are summarized in Table 31. In each figure, the SMDE equation is plotted for the temperature of 650°C and stress of 200 and 300 MPa. These stresses correspond to the lowest and highest stress found from CF stress relaxation data at 650°C. The 650°C CF tests will later be used when compiling CF interaction and life prediction plots. A greater stress reduces the predicted creep ductility.

Shown in Table 31, the R² values were different between the parameter sets derived from the two temperature ranges. The 650°C temperature contained only a third of the data points of the 538 – 850°C, full data set. The larger quantity of data points increased variability in data resulting in the lowest R². Regardless, both R² values are low.
Figure 124: Creep ductility at 538 – 850°C fitted to SMDE equation. (a) Without (b) with upper-shelf ductility.
Figure 124: Creep ductility at 538 – 850°C fitted to SMDE equation (b).
Figure 125: Creep ductility at 650°C fitted to SMDE equation. (a) Without (b) with upper-shelf ductility.
Figure 125: Creep ductility at 650°C fitted to SMDE equation (b).
Table 31: Fitting parameters to piece-wise function and SMDE equation

<table>
<thead>
<tr>
<th>T (K)</th>
<th>A</th>
<th>$\frac{Q}{R}$</th>
<th>n</th>
<th>m</th>
<th>Upper-Shelf (mm/mm)</th>
<th>Lower-Shelf (mm/mm)</th>
<th>R²</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td>+95%</td>
<td>Mean</td>
<td>-95%</td>
</tr>
<tr>
<td>811-1123</td>
<td>4.32</td>
<td>4686.60</td>
<td>0.25</td>
<td>0.68</td>
<td>-</td>
<td>-</td>
<td>-</td>
</tr>
<tr>
<td>811-1123</td>
<td>4.32</td>
<td>4686.60</td>
<td>0.25</td>
<td>0.68</td>
<td>0.85</td>
<td>0.58</td>
<td>0.40</td>
</tr>
<tr>
<td>923</td>
<td>282420</td>
<td>-</td>
<td>0.38</td>
<td>1.44</td>
<td>-</td>
<td>-</td>
<td>-</td>
</tr>
<tr>
<td>923</td>
<td>282420</td>
<td>-</td>
<td>0.38</td>
<td>1.44</td>
<td>0.76</td>
<td>0.58</td>
<td>0.40</td>
</tr>
</tbody>
</table>

All mean regression curves struggled to fit the data at average creep strain rates lower than $10^{-9}$ 1/s and higher than $10^{-6}$ 1/s. The large scatter in data formed CI creep ductility range of up to 0.8. Both the mean curve and +95% CI overpredict creep ductility above a creep strain rate of $10^{-6}$ 1/s when an upper-shelf ductility was not present. The power law and SMDE equations estimates a larger creep ductility at these strain rates than the logarithmic equation. The R² values were low for all fitted equations. The largest R² was attained from SMDE equation with Table 31 parameters, fitted to 650°C creep ductility data.
6.3 CF Stress Relaxation Modelling

The Feltham and Conway analysis stress relaxation models, Equation 31 and Equation 33, are explored. These models are used in both the DE and TF models. The fitting parameters of each model from various CF stress relaxation data at half-life are shown in Table 32 and Table 33. Stress is in units MPa and time is in units of s. All equations computed near unity $R^2$ values. The Conway analysis computed higher $R^2$ values.

Table 32: Feltham equation fitting parameters

<table>
<thead>
<tr>
<th>Material</th>
<th>CF Test</th>
<th>$\sigma_0$</th>
<th>a</th>
<th>b</th>
<th>$R^2$</th>
</tr>
</thead>
<tbody>
<tr>
<td>316L</td>
<td>1.0%, 1 min, 650°C</td>
<td>295.82</td>
<td>3.70</td>
<td>9.28</td>
<td>0.95</td>
</tr>
<tr>
<td>316L</td>
<td>1.0%, 3 min, 650°C</td>
<td>283.36</td>
<td>2.14</td>
<td>13.00</td>
<td>0.98</td>
</tr>
<tr>
<td>316L</td>
<td>1.0%, 5 min, 650°C</td>
<td>281.36</td>
<td>1.61</td>
<td>12.56</td>
<td>0.97</td>
</tr>
<tr>
<td>316L</td>
<td>0.6%, 1 min, 650°C</td>
<td>240.19</td>
<td>8.52</td>
<td>5.54</td>
<td>0.98</td>
</tr>
<tr>
<td>316H</td>
<td>0.6%, 30 min, 650°C</td>
<td>199.76</td>
<td>0.40</td>
<td>5.36</td>
<td>0.99</td>
</tr>
<tr>
<td>316L</td>
<td>0.6%, 1 min, 700°C</td>
<td>219.49</td>
<td>2.26</td>
<td>10.65</td>
<td>0.98</td>
</tr>
</tbody>
</table>
Table 33: Conway analysis fitting parameters

<table>
<thead>
<tr>
<th>Material</th>
<th>CF Test</th>
<th>$\sigma_{\text{max}}$</th>
<th>A</th>
<th>m</th>
<th>$R^2$</th>
</tr>
</thead>
<tbody>
<tr>
<td>316L</td>
<td>1.0%, 1 min, 650°C</td>
<td>295.82</td>
<td>0.017</td>
<td>-0.78</td>
<td>0.99</td>
</tr>
<tr>
<td>316L</td>
<td>1.0%, 3 min, 650°C</td>
<td>283.36</td>
<td>0.022</td>
<td>-0.75</td>
<td>0.99</td>
</tr>
<tr>
<td>316L</td>
<td>1.0%, 5 min, 650°C</td>
<td>281.36</td>
<td>0.020</td>
<td>-0.76</td>
<td>0.99</td>
</tr>
<tr>
<td>316L</td>
<td>0.6%, 1 min, 650°C</td>
<td>240.19</td>
<td>0.014</td>
<td>-0.81</td>
<td>0.99</td>
</tr>
<tr>
<td>316H</td>
<td>0.6%, 30 min, 650°C</td>
<td>199.76</td>
<td>0.0083</td>
<td>-0.78</td>
<td>0.99</td>
</tr>
<tr>
<td>316L</td>
<td>0.6%, 1 min, 700°C</td>
<td>219.49</td>
<td>0.025</td>
<td>-0.74</td>
<td>0.99</td>
</tr>
</tbody>
</table>

Graphical comparison are made between fitted stress relaxation models in Figure 126 and Figure 127. Stress relaxation data is also plotted. Both the Feltham and Conway models correlated the CF test data well, regardless of strain range and or temperature. The Feltham equation fit more poorly along the initial rapid stress relaxation segment. This is clearly observed in Figure 127.
Figure 126: Feltham equation and Conway analysis fitted to CF stress relaxation data.
Figure 127: Feltham equation and Conway analysis fitted to CF stress relaxation data. Time axis normalized by dwell period.

Identifying the minimum creep strain rate is helpful when analyzing creep damage calculated using the life fraction method. The creep ductility regions used when calculating creep ductility can be addressed. Table 34 summarizes the minimum creep strain rate calculated at half-life. The creep strain rate equations formed from the Feltham and Conway stress relaxation models (Equation 37 and Equation 38) and the Young’s modulus shown in Table 23 were used to determine the minimum creep strain rates. Comparable minimum creep strain rates were calculated from the two equations. The computed
minimum creep strain rates for CF tests with dwell times shorter than 30 min. are high relative to the creep ductility-average creep strain rate data shown in Figure 120. Excluding the 30 min. dwell period CF test, all minimum creep strain rates correspond with the upper-shelf ductility.

Table 34: Minimum creep strain rate using Feltham equation

<table>
<thead>
<tr>
<th>Material</th>
<th>CF Test</th>
<th>Minimum Creep Strain Rate During Dwell (1/s)</th>
</tr>
</thead>
<tbody>
<tr>
<td>316L</td>
<td>1.0%, 1 min., 650°C</td>
<td>1.07 × 10^{-6}</td>
</tr>
<tr>
<td>316L</td>
<td>1.0%, 3 min., 650°C</td>
<td>5.01 × 10^{-7}</td>
</tr>
<tr>
<td>316L</td>
<td>1.0%, 5 min., 650°C</td>
<td>2.90 × 10^{-7}</td>
</tr>
<tr>
<td>316L</td>
<td>0.6%, 1 min., 650°C</td>
<td>6.41 × 10^{-7}</td>
</tr>
<tr>
<td>316H</td>
<td>0.6%, 30 min., 650°C</td>
<td>2.07 × 10^{-8}</td>
</tr>
<tr>
<td>316L</td>
<td>0.6%, 1 min., 700°C</td>
<td>1.23 × 10^{-6}</td>
</tr>
</tbody>
</table>
Table 35: Minimum creep strain rate using Conway analysis

<table>
<thead>
<tr>
<th>Material</th>
<th>CF Test</th>
<th>Minimum Creep Strain Rate During Dwell (1/s)</th>
</tr>
</thead>
<tbody>
<tr>
<td>316L</td>
<td>1.0%, 1 min., 650°C</td>
<td>$1.17 \times 10^{-6}$</td>
</tr>
<tr>
<td>316L</td>
<td>1.0%, 3 min., 650°C</td>
<td>$6.35 \times 10^{-7}$</td>
</tr>
<tr>
<td>316L</td>
<td>1.0%, 5 min., 650°C</td>
<td>$3.77 \times 10^{-7}$</td>
</tr>
<tr>
<td>316L</td>
<td>0.6%, 1 min., 650°C</td>
<td>$7.13 \times 10^{-7}$</td>
</tr>
<tr>
<td>316H</td>
<td>0.6%, 30 min., 650°C</td>
<td>$2.82 \times 10^{-8}$</td>
</tr>
<tr>
<td>316L</td>
<td>0.6%, 1 min., 700°C</td>
<td>$1.43 \times 10^{-6}$</td>
</tr>
</tbody>
</table>

6.4 CF Interaction Diagrams

CF interaction diagrams for 650°C CF tests results were constructed using LCF, CF, and creep data using the relationships shown in Table 40 found in APPENDIX C. The TF CF interaction diagrams is shown in Figure 128. Relative to the bilinear envelope, the 1.0% strain range points are found far outside of the bilinear locus, indicating no CF interactions occurred. The 0.6% strain range point are found inside the bilinear locus with low fatigue and creep damage. Although a CF interaction is implied, the metallographic findings suggest otherwise. Lower fatigue damage was calculated from early CF failure times due to thermocouple weld influences.
The DE CF interaction diagrams are shown in Figure 129 and Figure 130. Each figure includes creep damage determined with and without an upper-shelf ductility piece-wise function. When the CF damage is calculated using a piece-wise function with upper-shelf ductility, the point lies further right of the linear locus. No CF interactions are implied from all CF tests when creep damage was calculated with an upper-shelf ductility is used. CF interactions are suggested to have occurred from the 1.0% 3 min. dwell and 0.6% 1 min. dwell tests when an upper-shelf ductility was not used.
Figure 129: CF interaction diagram for 650°C CF tests. Creep ductility calculated from logarithmic equation and Table 29. CF test parameter labelled as ($\Delta \varepsilon$, $t_h$, T).
The power law creep ductility equation computed lower creep damage than the logarithmic equation when creep damage was calculated without an upper-shelf ductility. Minimum creep rates were high for the CF tests plotted in the interaction diagrams as shown in Table 34 and Table 35 relative to the creep ductility data shown in Figure 120. The power law equation compute larger creep ductility at these strain rate, innately forming lower creep damage.

Figure 130: CF interaction diagram for 650°C CF tests. Creep ductility calculated from power law equation and Table 30. CF test parameter labelled as ($\Delta\varepsilon_t$, $t_h$, T).
The SMDE CF interaction diagrams are shown in Figure 131 and Figure 132. As with the previous two figures, greater creep damage was calculated when an upper-shelf ductility was included. No CF is implied for these points. When an upper-shelf ductility is not used, greater creep damage is calculated from the 650 °C creep ductility fitted data.

Figure 131: CF interaction diagram for 650°C CF tests. Creep ductility determined from SMDE equation with Table 31, 538 – 850°C fitting parameters. CF test parameter labelled as (Δε₀, t₀, T).
Creep and fatigue damage are summarized in APPENDIX B Table 39. Comparable creep damage was calculated from TF, SMDE with upper-shelf ductility, and DE with upper-shelf ductility columns. Relative to DE and SMDE approaches, high minimum creep strain rates resulted in creep damage being calculated only along the upper-shelf ductility. This produced equivalent creep damage from these tests.
Figure 133 and Figure 134 compare the creep damage calculated by the mean and ±95% CI curves. Creep damage was larger and smaller than those calculated from the mean curve when using the -95% and +95% CI, respectively. In both plots, creep damage calculated by the -95% CI exceed unity creep damage. The significant variance in creep damage is not helpful when interpreting the CF interaction diagram. A CF interaction is inferred from the -95% CI, while no CF interactions are inferred from the +95% CI.

Figure 133: Comparison of TF creep damage from mean and ±95% CI curve.
Figure 134: Comparison of SMDE creep damage from mean and ±95% CI curve. Creep ductility determined from SMDE equation with Table 31, 650°C fitting parameters without upper-shelf ductility.

There is considerable variability in creep damage calculated from each life fraction method. Several factors largely impact the creep damage computed from an individual test: scatter in creep failure time or creep ductility data, creep damage calculation approach, and goodness of fit from LM or creep ductility equations. The use of CF interaction diagrams to examine or estimate CF damage mechanism is not appropriate for 316 SS.
6.5 Influence of Dwell Times on LCF Life

The influence of dwell times on 0.6% and 1.0% strain range CF tests were examined through CF life prediction plots generated using the wrought 316H data. The predictions are compared to experimental data on both wrought 316H and 316L. The equations and fitting parameters used to construct these plots are shown in Table 41 located in APPENDIX C.

The TF predictions for strain ranges of 0.6% and 1.0% are shown in Figure 135. Each curve is characterized with an inflection point. Formation of this point is related to the CF interaction diagram’s bilinear failure envelope and is illustrated in Figure 136. As dwell time increases, the computed creep damage transition from one slope of the bilinear envelope to another. The inflection point is formed when the failure locus’ intersection point is crossed.
Figure 135: TF predictions for 650°C 0.6% and 1.0% strain-controlled with experimental data from wrought 316H and 316L SS.
Figure 136: Inflection behavior of TF life prediction plots for 650°C 0.6% strain controlled CF test.

The DE prediction plots for 0.6% and 1.0% strain ranges are shown in Figure 137 and Figure 138, respectively. LCF lives are predicted poorly by the logarithmic fitted models, especially for the 0.6% strain range plot. A factor of 2 difference in life is predicted when compared to the LCF data point. However, the logarithmic creep ductility equation produced comparable predictions as the power law equations for dwell periods between 1 to 100 hrs. At these dwell times, similar creep ductility is computed by the logarithmic and
power law equations, while smaller creep ductility is predicted for shorter dwell times resulting in lower life predictions.

Figure 137: DE predictions for 650°C 0.6% strain-controlled with experimental data from wrought 316H and 316L SS.
Figure 138: DE predictions for 650°C 1.0% strain-controlled with experimental data from wrought 316H and 316L SS.

Interestingly, DE models with an upper-shelf ductility predict the same LCF life. Only the upper-shelf ductility was used when life prediction was made for LCF tests. However, very poor LCF life predictions were made when an upper-shelf ductility was used. As illustrated in Section 6.4, inclusion of an upper-shelf ductility increases creep damage. This inevitably results in low life predictions. When using the DE model, a power law creep ductility equation fitted without an upper-shelf ductility is expected to produce better predictions for dwell periods shorter than 1 hr.
Although large differences have been observed between several prediction curves and LCF data, it is to note that the plotted prediction curves begin at 0.0001 hr (0.36 s) dwell period per cycle and does not necessarily represent LCF life predictions. However, small creep damage is expected from these short dwell periods, therefore, it is expected that the predictions curves yield similar lives as LCF data. The inaccurate predictions by the prediction curves at the short dwell periods entails that the DE model is highly sensitive to very short dwell periods, which is not observed experimentally.

The SMDE prediction plots for 0.6% and 1.0% strain range are shown in Figure 139 and Figure 140, respectively. Similar life predictions are made from the two SMDE equation parameters. Like the DE prediction plots, the LCF life is predicted poorly when an upper-shelf ductility included model is used. Use of an upper-shelf ductility is not appropriate when constructing DE and SMDE CF life prediction plots.
Figure 139: SMDE predictions for 650°C 0.6% strain-controlled with experimental data from wrought 316H and 316L SS.
Similar to the TF prediction plots, inflection points are found in the DE and SMDE models. These points form when predicted creep ductility transition along the different creep ductility piece-wise function regions. In this investigation, predictions were driven by the upper-shelf ductility when dwell periods were shorter than 0.1 hr for the 0.6% and 1.0% strain range, regardless of the DE or SMDE models used. Predictions are driven by the intermediate creep ductility relation when an upper-ductility is not used or for
intermediate dwell periods. The lower-shelf ductility drive the life predictions of the 0.6% and 1.0% strain range for dwell periods longer than 100 hrs.

Figure 141 and Figure 142 compare the TF model, DE model formed from the power law equation, and SMDE model formed from 650°C creep ductility data at the 0.6% and 1.0% strain range. The DE and SMDE models are formed without an upper-shelf ductility. All curves include a ±95% CI band. Other life prediction curves with its ±95% CI band are shown in APPENDIX D.
Figure 141: Life predictions with ±95% CI band for 650°C 0.6% strain-controlled with experimental data from wrought 316H and 316L SS.
Figure 142: Life predictions with ±95% CI band for 650°C 1.0% strain-controlled with experimental data from wrought 316H and 316L SS.

All mean life prediction curves shown in Figure 141 and Figure 142 predict experimental lives to within about a factor of 2. However, the ±95% CI suggest large variability in life predictions. The large variation appears from highly scattered creep rupture time and creep ductility data. The variability in life prediction increases with increasing dwell times since the uncertainty in the creep data has great impact. At the 100 hr dwell period, a difference of nearly one-magnitude is observed between the upper and lower CI from all 0.6% strain range life prediction plots. The difference in predictions between the
CI made at the 100 hr dwell period for the 1.0% strain range is about a half-a-magnitude. An inclusion of a ±95% CI along with the mean prediction curve is essential to characterize the large variability in predictions.

The TF life prediction is characterized by a sharp decline in predicted life with increasing dwell times when compared to the DE or SMDE life predictions. Data from literature suggest the cyclic life is reduced by up to a magnitude at a 1 hr dwell test [98, 103, 106, 113]. The reduction in life with increasing dwell time is more closely followed by the DE and SMDE approaches with power law creep ductility [125].

In the current study, the DE model formed from the power law creep ductility equation and SMDE models produce similar CF life predictions. Relative to LCF life, the SMDE equation fitted to 650°C creep ductility data produced the best life prediction. Although the SMDE equation can be fitted to a range of temperatures, the great scatter in creep ductility inevitably lowers the goodness of fit. Comparing all life fraction models, the SMDE model fitted to a single test temperature is likely to produce the most accurate life predictions. Table 42 located in APPENDIX C summarizes the set of equations which is expected to produce the best life predictions.

Recognizing the SMDE model fitted to a single test temperature may produce the best life predictions, assessments for an accelerated qualification test are made. Table 36 and Table 37 summarize the cyclic life predictions and corresponding test times of 316H SS for strain ranges of 0.6% and 1.0%. A strain loading rate of $2 \times 10^{-3}$ 1/s is used to make test time predictions.
Table 36: Cyclic life and test time predictions using the SMDE model for 316H 0.6% strain range 650°C strain-controlled CF tests with varying dwell times

<table>
<thead>
<tr>
<th>Dwell Time (hr)</th>
<th>Cycles to Failure +95%</th>
<th>Mean</th>
<th>-95%</th>
<th>Time to Failure +95%</th>
<th>Mean</th>
<th>-95%</th>
</tr>
</thead>
<tbody>
<tr>
<td>LCF (no dwell)</td>
<td>-</td>
<td>20921</td>
<td>-</td>
<td>-</td>
<td>1.45</td>
<td>-</td>
</tr>
<tr>
<td>0.0001 (0.36 s)</td>
<td>19761</td>
<td>18270</td>
<td>15399</td>
<td>1.45</td>
<td>1.34</td>
<td>1.13</td>
</tr>
<tr>
<td>0.0167 (1 min.)</td>
<td>11750</td>
<td>7142</td>
<td>3627</td>
<td>8.99</td>
<td>5.47</td>
<td>2.78</td>
</tr>
<tr>
<td>0.167 (10 min.)</td>
<td>6282</td>
<td>3028</td>
<td>1340</td>
<td>44.2</td>
<td>21.3</td>
<td>9.4</td>
</tr>
<tr>
<td>0.5 (30 min.)</td>
<td>4250</td>
<td>1956</td>
<td>838</td>
<td>88.8</td>
<td>40.9</td>
<td>17.5</td>
</tr>
<tr>
<td>1.0</td>
<td>3281</td>
<td>1464</td>
<td>618</td>
<td>136.9</td>
<td>61.1</td>
<td>25.8</td>
</tr>
<tr>
<td>3.0</td>
<td>2153</td>
<td>928</td>
<td>385</td>
<td>269</td>
<td>116</td>
<td>48.2</td>
</tr>
<tr>
<td>5.0</td>
<td>1768</td>
<td>753</td>
<td>311</td>
<td>368</td>
<td>157</td>
<td>64.8</td>
</tr>
<tr>
<td>10.0</td>
<td>1357</td>
<td>571</td>
<td>235</td>
<td>566</td>
<td>238</td>
<td>97.9</td>
</tr>
</tbody>
</table>
Table 37: Cyclic life and test time predictions using the SMDE model for 316H 1.0% strain range 650°C strain-controlled CF tests with varying dwell times

<table>
<thead>
<tr>
<th>Dwell Time (hr)</th>
<th>Cycles to Failure</th>
<th>Time to Failure</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td>+95%</td>
<td>Mean</td>
</tr>
<tr>
<td>LCF (no dwell)</td>
<td>-</td>
<td>1176</td>
</tr>
<tr>
<td>0.0001 (0.36 s)</td>
<td>1163</td>
<td>1144</td>
</tr>
<tr>
<td>0.0167 (1 min.)</td>
<td>1033</td>
<td>876</td>
</tr>
<tr>
<td>0.167 (10 min.)</td>
<td>858</td>
<td>613</td>
</tr>
<tr>
<td>0.5 (30 min.)</td>
<td>758</td>
<td>498</td>
</tr>
<tr>
<td>1.0</td>
<td>692</td>
<td>431</td>
</tr>
<tr>
<td>3.0</td>
<td>596</td>
<td>346</td>
</tr>
<tr>
<td>5.0</td>
<td>557</td>
<td>314</td>
</tr>
<tr>
<td>10.0</td>
<td>512</td>
<td>279</td>
</tr>
</tbody>
</table>

The 0.6% strain range has a larger decline in life than the 1.0% strain range with increasing dwell times. Comparing the predicted life for LCF tests and a 0.5 hr dwell time, the 0.6% strain range has over an order of magnitude decline in life, whereas the 1.0% strain range has just over a factor of 2 decline in life. The life prediction plots suggest dwell periods have a stronger impact on lower strain range CF tests. This finding follows the
trend shown in Figure 45. A stronger CF interactions is expected from the 0.6% strain range CF tests with long dwell periods.

Test times increase with increasing dwell times to promote a definite creep-fatigue interaction. Although significant life reduction is seen from dwell periods of at least 30 min. for each strain range, these tests can take 3 months or longer to complete. Hence, the challenge of promoting a creep-fatigue interaction in an accelerated qualification test still remains. These results to date suggest that the qualification testing to evaluate creep-fatigue behavior may be most expeditiously determined by evaluation of LCF (no dwell) and creep independently with the interaction being predicted within reason using interaction diagrams since tests so far have not shown significant interaction (i.e., the promotion of Case C in Figure 46 and Figure 47). Additional tests on additively manufactured specimens at lower strain range and long dwell times (30 min. or greater) need to be completed to confirm this hypothesis.
CHAPTER 7. CONCLUSIONS

Wrought 316L, wrought 316H, and DED 316H SS cyclic responses were explored with respect to various LCF and CF testing parameters. These are: test temperature, strain range, dwell periods, control mode during dwell, and dwell locations within a load-unload cycle. The follow key findings were made:

- Comparable cyclic life was attained from 316L LCF tests and strain-controlled CF tests when dwell times were shorter than 10 min. at a 1.0% strain range and dwell periods shorter than 1 min. at a 0.6% strain range.

- The combined 316L LCF and CF data correlated to a single Coffin-Manson relation for 1.0% strain range CF tests with dwell times less than 10 min. and 0.6% strain range CF tests with dwell times less than 1 min. for temperatures ranging from 550°C to 700°C. These results indicate an absence of a strong CF interaction from these test conditions.

- The primary cracks leading to failure are transgranular for 316L LCF and CF tests with a dwell time shorter than 10 min. Intergranular cracks were found along the crack tips of several CF tests. The results suggest that dwell times longer than 10 min. are required to produce a possibly significant CF interaction.

- Wrought and DED 316H SS had similar cyclic responses. However, the wrought specimen had a one-magnitude larger LCF life than the DED specimens. The lower life observed in the 316L SS was due to the thermocouple welds placed within the gauge section. Surface cracks nucleated quicker near the welds and led to early failure.
The following findings were made from the CF interaction and life prediction plots:

- Creep damage is largely affected by scatter in creep rupture time or creep ductility data, method of creep damage calculation, and goodness of fit from LM or creep ductility equations. Diagrams showing the ±95% confidence intervals (CI) were developed. The largest difference in creep damage between the upper and lower band was 2.9 and 1 for the time-fraction and stress modified ductility exhaustion diagrams, respectively. The wide attainable creep damage values result in significant uncertainty in the use of CF interaction diagrams.

- Three different life fraction models were evaluated. Among these, the stress modified ductility exhaustion (SMDE) model with creep ductility fitted to a single temperature without an upper-shelf ductility provides the best predictions. The ±95% CI predictions curves should also be included to characterize the large scatter in creep ductility.

CF tests with dwell periods longer than 10 min. must be conducted to better understand the evolution of CF damage. These tests are also critical to assess the prediction capability of the life fraction models. However, tests with long dwell times can result in failure times which are impractical to use as an accelerated qualification test for AM alloys. Rather than conducting multiple CF tests with long dwell periods, it may be more practical to conduct multiple independent LCF and creep for qualification acceptance since these tests will highlight to two main damage mechanisms. Certainly, this is only acceptable after a test program consisting of creep-fatigue tests with low strain range and tensile dwells 30 min. or longer has been conducted on the relevant AM alloys to verify that the
interactions are indeed captured conservatively on a CF interaction diagram as the current results tend to point towards.
APPENDIX A. STATISTICAL ANALYSIS METHODS

A.1 Coefficient of Determination

Coefficient of determination, $R^2$, was calculated when linear regression was used. The value is determined as,

$$R^2 = 1 - \frac{\sum((y_i - f_i)^2)}{\sum((y_i - \bar{y})^2)} = 1 - \frac{SSR}{SST} \tag{43}$$

$\bar{y}_i$ is the mean of the data set $y$. $SSR$ is the sum of square of residual and quantifies the model’s deviation from data. $SST$ is the total sum of squares and quantifies the deviation in data relative to the data’s mean. An adjusted $R^2$ was determined when multiple-linear regression was used,

$$R^2 = 1 - \frac{(1 - R^2)(k - 1)}{(k - m - 1)} \tag{44}$$

$R^2$ generally range from 0 to 1. A value near 0 indicate a poor fit, while a value near 1 indicates a perfect fit to data. Values outside the 0 to 1 range suggest the fitted model is an inappropriate representation of the mean of the data. Negative values may be calculated models are not fitted using least-square methods. In these case, the fitted equations do not statically represent the data appropriately.
A.2 Confidence Interval Bands for Creep Data

A ±95% CI band was determined when numerical models were fitted to creep rupture data. The intervals were determined by first determining the standard of error or estimation, $SEE$,

$$
SEE = \sqrt{\frac{\sum(y_i - f_i)^2}{k - m - 1}} \tag{45}
$$

$y_i$ are data values in data set $y$ and $f_i$ are the predicted values from a model. $k$ is the number of data points in the data set and $m$ is the number of independent variables. $SEE$ is then multiplied to the inverse of the two-tailed Student's t-distribution, $t_{inv}$, which was found from Microsoft Excel software. The product was added or subtracted to the fitted numerical model, $f(x_1, x_2, x_3 \ldots )$, to determine the ±95% CI band, $f_{±95\%}(x_1, x_2, x_3 \ldots )$,

$$
f_{±95\%}(x_1, x_2, x_3 \ldots ) = f(x_1, x_2, x_3 \ldots ) \pm (SEE \times t_{inv}) \tag{46}
$$
APPENDIX B. CREEP AND FATIGUE DAMAGE FROM CREEP-FATIGUE TESTS

Table 38 and Table 39 describe the creep and fatigue damage computed from TF, DE, and SMDE approaches. Equations used are summarized in Table 40. These values are used to plot the CF interaction diagrams shown in Section 6.4.

Table 38: Fatigue damage from strain-controlled CF tests

<table>
<thead>
<tr>
<th>Strain range, dwell time</th>
<th>1.0%, 1min</th>
<th>1.0%, 3min</th>
<th>1.0%, 5min</th>
<th>0.6%, 1min</th>
</tr>
</thead>
<tbody>
<tr>
<td>$\Phi_F$</td>
<td>0.88</td>
<td>0.59</td>
<td>0.70</td>
<td>0.14</td>
</tr>
</tbody>
</table>
Table 39: Creep damage from strain-controlled CF tests

<table>
<thead>
<tr>
<th>Model</th>
<th>Equation Used</th>
<th>Strain Range, Dwell Time</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td></td>
<td>1.0%, 1min</td>
</tr>
<tr>
<td>TF</td>
<td>LM polynomial fit (Equation 42) and $C_{LM} = 16.42$ substituted to LM creep time to rupture equation (Equation 9)</td>
<td>0.65</td>
</tr>
<tr>
<td>DE</td>
<td>Power law equation and Table 30, with upper-shelf ductility</td>
<td>0.63</td>
</tr>
<tr>
<td></td>
<td>Power law equation and Table 30, no upper-shelf ductility</td>
<td>0.45</td>
</tr>
<tr>
<td></td>
<td>Logarithmic equation and Table 29, with upper-shelf ductility</td>
<td>0.63</td>
</tr>
<tr>
<td></td>
<td>Logarithmic equation and Table 29, no upper-shelf ductility</td>
<td>0.27</td>
</tr>
<tr>
<td>SMDE</td>
<td>SMDE equation and Table 31, 538 – 850 °C, with upper-shelf ductility</td>
<td>0.63</td>
</tr>
<tr>
<td></td>
<td>SMDE equation and Table 31, 538 – 850 °C, no upper-shelf ductility</td>
<td>0.37</td>
</tr>
<tr>
<td></td>
<td>SMDE equation and Table 31, 650°C, with upper-shelf ductility</td>
<td>0.63</td>
</tr>
<tr>
<td></td>
<td>SMDE equation and Table 31, 650°C, no upper-shelf ductility</td>
<td>0.29</td>
</tr>
</tbody>
</table>
## APPENDIX C. EQUATIONS USED FOR CF LIFE FRACTION MODELING

Table 40: Summary of equations used for interaction diagrams

<table>
<thead>
<tr>
<th></th>
<th>TF</th>
<th>DE</th>
<th>SMDE</th>
</tr>
</thead>
<tbody>
<tr>
<td>(\Phi_F)</td>
<td>(\frac{N}{N_f}) (Equation 3)</td>
<td></td>
<td></td>
</tr>
<tr>
<td>(\Phi_C)</td>
<td>(\sum \int_0^{t_h} \frac{dt}{t_c(\sigma,T)}) (Equation 6)</td>
<td>(\sum \int_0^{t_h} \frac{\dot{\varepsilon}_c}{\varepsilon_f(\dot{\varepsilon}_c)} dt) (Equation 19)</td>
<td>(\sum \int_0^{t_h} \frac{\dot{\varepsilon}_c}{\varepsilon_f(\dot{\varepsilon}_c, \sigma,T)} dt) (Equation 25)</td>
</tr>
<tr>
<td>(N_f)</td>
<td>(316) \text{H SS: } \Delta \varepsilon = 0.6% \quad N_f = 20921 )</td>
<td>(316) \text{H SS: } \Delta \varepsilon = 1.0% \quad N_f = 1176 )</td>
<td></td>
</tr>
<tr>
<td>(t_c)</td>
<td>(\text{LM polynomial fit (Equation 42) and } C_{ LM} = 16.42 \text{ substituted to LM creep time to rupture equation (Equation 9)})</td>
<td></td>
<td></td>
</tr>
</tbody>
</table>
Table 40: (Continued)

<table>
<thead>
<tr>
<th>$\varepsilon_f$</th>
<th>Piece wise function (Equation 21), intermediate equation defined as: Logarithm equation (Equation 23) and Table 29 Power law equation (Equation 22) and Table 30</th>
<th>Piece wise function (Equation 21), intermediate equation defined as: SMDE equation (Equation 26) and Table 31</th>
</tr>
</thead>
<tbody>
<tr>
<td>$\sigma$</td>
<td>Conway analysis (Equation 33) and Table 33: 316H SS: $\Delta \varepsilon_t = 0.6%$, 30 min. 316L SS: $\Delta \varepsilon_t = 1.0%$, 3 min.</td>
<td></td>
</tr>
<tr>
<td>$\dot{\varepsilon}_c$</td>
<td>Conway analysis (Equation 38) and Table 33: 316H SS: $\Delta \varepsilon_t = 0.6%$, 30 min. 316L SS: $\Delta \varepsilon_t = 1.0%$, 3 min.</td>
<td></td>
</tr>
</tbody>
</table>
Table 41: Summary of equations used for life predictions

<table>
<thead>
<tr>
<th></th>
<th>TF</th>
<th>DE</th>
<th>SMDE</th>
</tr>
</thead>
<tbody>
<tr>
<td>(D)</td>
<td>Set to unity</td>
<td></td>
<td></td>
</tr>
<tr>
<td>(N)</td>
<td>(N = \frac{D}{\left(\frac{1}{N_f} + \Phi_{\text{cycle}}\right)})</td>
<td></td>
<td></td>
</tr>
<tr>
<td>(\Phi_F)</td>
<td>(\frac{N}{N_f})</td>
<td></td>
<td></td>
</tr>
<tr>
<td>(\Phi_C)</td>
<td>(\sum_0^{t_h} \frac{dt}{t_c(\sigma,T)})</td>
<td>(\sum_0^{t_h} \frac{\dot{\varepsilon}_c}{\varepsilon_f(\dot{\varepsilon}_c)} dt)</td>
<td>(\sum_0^{t_h} \frac{\dot{\varepsilon}_c}{\varepsilon_f(\dot{\varepsilon}_c, \sigma,T)} dt)</td>
</tr>
<tr>
<td>(N_f)</td>
<td>316H SS: (\Delta \varepsilon_t = 0.6%, N_f = 20921)</td>
<td>316H SS: (\Delta \varepsilon_t = 1.0%, N_f = 1176)</td>
<td></td>
</tr>
<tr>
<td>(t_c)</td>
<td>LM polynomial fit (Equation 42) and (C_{LM} = 16.42) substituted to LM creep time to rupture equation (Equation 9)</td>
<td></td>
<td></td>
</tr>
</tbody>
</table>
Table 41: (Continued)

<table>
<thead>
<tr>
<th>$\varepsilon_f$</th>
<th>Piece wise function (Equation 21), intermediate equation defined as:</th>
<th>Piece wise function (Equation 21), intermediate equation defined as:</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td>Logarithm equation (Equation 23) and Table 29</td>
<td>SMDE equation (Equation 26) and Table 31</td>
</tr>
<tr>
<td></td>
<td>Power law equation (Equation 22) and Table 30</td>
<td></td>
</tr>
</tbody>
</table>

<table>
<thead>
<tr>
<th>$\sigma$</th>
<th>Conway analysis (Equation 33) and Table 33:</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td>316H SS: $\Delta \varepsilon_t = 0.6%$, 30 min.</td>
</tr>
<tr>
<td></td>
<td>316L SS: $\Delta \varepsilon_t = 1.0%$, 3 min.</td>
</tr>
</tbody>
</table>

<table>
<thead>
<tr>
<th>$\dot{\varepsilon}_c$</th>
<th>Conway analysis (Equation 38) and Table 33:</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td>316H SS: $\Delta \varepsilon_t = 0.6%$, 30 min.</td>
</tr>
<tr>
<td></td>
<td>316L SS: $\Delta \varepsilon_t = 1.0%$, 3 min.</td>
</tr>
<tr>
<td>Model</td>
<td>SMDE</td>
</tr>
<tr>
<td>-------</td>
<td>------</td>
</tr>
<tr>
<td>(D)</td>
<td>Set to unity</td>
</tr>
<tr>
<td>(N)</td>
<td>(N = \frac{D}{\left(\frac{1}{N_f} + \Phi_{c,\text{cycle}}\right)}) (Equation 41)</td>
</tr>
<tr>
<td>(\Phi_F)</td>
<td>(\frac{N}{N_f}) (Equation 3)</td>
</tr>
<tr>
<td>(\Phi_C)</td>
<td>(\sum \int_{0}^{t_h} \frac{\dot{\varepsilon}_c}{\varepsilon_f(\dot{\varepsilon}_c, \sigma, T)} , dt) (Equation 25)</td>
</tr>
<tr>
<td>(N_f)</td>
<td>From LCF data</td>
</tr>
<tr>
<td>(\varepsilon_f)</td>
<td>(\varepsilon_f(\dot{\varepsilon}_c) = A \exp \left(\frac{Q}{RT}\right) \dot{\varepsilon}_c^n \sigma^{-m}) (Equation 26 and lower-shelf ductility. No upper-shelf ductility)</td>
</tr>
<tr>
<td>(\sigma)</td>
<td>(\sigma = \frac{\sigma_{\text{max}}}{e^{\left(\frac{A}{\sigma_{\text{max}}} t\right)^{1+m}}}) (Equation 33, Conway analysis)</td>
</tr>
</tbody>
</table>
Table 42: (Continued)

<table>
<thead>
<tr>
<th>( \dot{\varepsilon}_c )</th>
<th>( \dot{\varepsilon}<em>c = \frac{1}{E} \frac{\sigma</em>{\text{max}} A t^m}{e^{\left(\frac{A}{1+m}\right)t^{1+m}}} )</th>
</tr>
</thead>
<tbody>
<tr>
<td>±95% CI</td>
<td>APPENDIX A.2</td>
</tr>
</tbody>
</table>

(Equation 38, Conway analysis)
Figure 143: TF life predictions with ±95% CI band for 650°C 0.6% and 1.0% strain-controlled with experimental data from wrought 316H and 316L SS.
Figure 144: DE life predictions with ±95% CI band for 650°C 0.6% and 1.0% strain-controlled with experimental data from wrought 316H and 316L SS. Creep ductility fitted to power law equation without upper-shelf ductility.
Figure 145: DE life predictions with ±95% CI band for 650°C 0.6% and 1.0% strain-controlled with experimental data from wrought 316H and 316L SS. Creep ductility fitted to power law equation with upper-shelf ductility.
Figure 146: DE life predictions with ±95% CI band for 650°C 0.6% and 1.0% strain-controlled with experimental data from wrought 316H and 316L SS. Creep ductility fitted to logarithmic equation without upper-shelf ductility.
Figure 147: DE life predictions with ±95% CI band for 650°C 0.6% and 1.0% strain-controlled with experimental data from wrought 316H and 316L SS. Creep ductility fitted to logarithmic equation with upper-shelf ductility.
Figure 148: SMDE life predictions with ±95% CI band for 650°C 0.6% and 1.0% strain-controlled with experimental data from wrought 316H and 316L SS. Creep ductility fitted to SMDE equation for temperature range of 538 – 850°C without upper-shelf ductility.
Figure 149: SMDE life predictions with ±95% CI band for 650°C 0.6% and 1.0% strain-controlled with experimental data from wrought 316H and 316L SS. Creep ductility fitted to SMDE equation for temperature range of 538 – 850°C with upper-shelf ductility.
Figure 150: SMDE life predictions with ±95% CI band for 650°C 0.6% and 1.0% strain-controlled with experimental data from wrought 316H and 316L SS. Creep ductility fitted to SMDE equation for temperature range of 650°C without upper-shelf ductility.
Figure 151: SMDE life predictions with ±95% CI band for 650°C 0.6% and 1.0% strain-controlled with experimental data from wrought 316H and 316L SS. Creep ductility fitted to SMDE equation for temperature range of 650°C with upper-shelf ductility.
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