Low-Cycle Fatigue Behavior of Fiber-Laser Welded, Corrosion-Resistant, High-Strength Low Alloy Sheet Steel^a

J.W. Sowards, E.A. Pfeif^{**}, M. Connolly, J.D. McColskey, S.L. Miller, B. Simonds, J.R. Fekete
National Institute of Standards and Technology, 325 Broadway, Boulder, CO 80305, USA.

^{**}Corresponding Author e-mail: erik.pfeif@nist.gov telephone: +1 303 497 5101

Abstract

Advanced laser welding techniques such as fiber laser welding are desirable for joining high-strength steels in ground vehicle applications since they can greatly increase productivity. Weld fatigue is a primary design consideration for implementation of laser-welded sheet into ground vehicle structures. However, much of the previous fatigue work on fiber laser welded sheet steel has utilized stress-controlled fatigue testing. Therefore, this paper presents an optimized fiber laser welding procedure for butt welds of high-strength low-alloy steel sheet, and subsequent fully-reversed, strain-controlled low-cycle fatigue testing on weld material. Existing low cycle fatigue models adequately describe fatigue behavior of the laser welds. The results show that weld low cycle fatigue strength is significantly increased relative to base metal due to formation of martensite in the weld caused by rapid cooling inherent to laser welding. Weld fatigue life was shorter than base metal when plastic strains were high, but was close to base metal at lower plastic strains. Fatigue specimens containing weld failed at both base metal and in the weld indicating that there is some competition between weld cyclic fatigue strength and stress concentration effect at the weld fusion zone.

Keywords: fiber laser beam welding; high strength low alloy steel; low cycle fatigue; microstructure-property relationships; residual stress; sheet steel

1 Introduction

Deployment of fiber lasers for welding in industrial settings has increased tremendously in recent years. This laser technology offers several benefits over previous generations of technology including high power, low beam divergence, high wall plug efficiency, and compact size coupled with flexible beam delivery [1]. Applications of fiber lasers are varied and include welding of steel pipes [1, 2], hybrid laser arc welding of titanium [3] and high strength low alloy steel [4], and welding of various steels used in production of ground vehicle components [5-8]. Improvements in process productivity is a compelling reason for utilizing fiber lasers, and thus the process has seen particular emphasis in welding of ground vehicle components, where high weld productivity can greatly decrease manufacturing costs.

Metal fatigue is one of the most important considerations for durability design of ground vehicles and the presence of a weld in structural components typically alters fatigue response. Two methods are currently used to design for weld fatigue including the nominal stress approach (where S-N curves are used), and the local initiation-based approach [9]. The latter is the strain-controlled low-cycle fatigue approach (also called strain-life approach), which considers elastic behavior and the plastic behavior in localized regions where cracks initiate, and is frequently used in the durability design of ground vehicles [10]. This strain-life approach has been difficult to apply to weld materials since cyclic material properties vary depending on the particular laser weld region such as in the fusion zone (FZ) or heat-affected zone (HAZ) [11]. Thus, recent mechanical studies for ground vehicle fatigue design have investigated the fatigue behavior of steels joined with the fiber laser welding process using the former testing approach, where stress (applied force) is the controlled test variable.

^a Work performed by NIST, an agency of the U.S. Government. Not subject to copyright in the United States.

High cycle fatigue susceptibility of fiber laser welded advanced high-strength dual phase (DP) and high-strength low-alloy (HSLA) sheet steels have been a topic of numerous recent investigations [6-8, 12-14] because of their increased use in ground vehicles. The DP steel base metals, which contain a mixture of ferrite and martensite for a mixture of formability and strength, are susceptible to softening in the HAZ following welding. The softened region has increased susceptibility to fatigue damage, where fatigue cracks tend to propagate through those regions after initiation at FZ regions containing high concavity. The HSLA steels have exhibited transformation hardening due to high cooling rates resulting in welds that are strengthened with respect to surrounding base metal. Westerbaan, *et al.* showed that HSLA (415 MPa yield strength) fiber laser welds had lower fatigue resistance than base metal under force controlled high cycle fatigue testing, and that changes in weld concavity had little effect on fatigue resistance [8]. The fatigue life of HSLA measured in other work has also shown little dependence on weld geometry [12]. HSLA welds were also shown to fail in base metal at high stress amplitudes, where weld strengthening was more resistant to fatigue, and at fusion zone weld concavity at lower stress amplitudes due to an apparent reduction in fatigue strength [8, 12]. Further work has looked at dissimilar weld combinations of HSLA and DP steels showing that HSLA base metal tended to have the highest fatigue susceptibility at high stress amplitudes where yielding occurred, whereas failures occurred at weld fusion zone at intermediate and low stress amplitudes where fatigue susceptibility was apparently sensitive to weld concavity [15]. Based on these high cycle fatigue studies, it can be summarized that HSLA fiber laser welds are fatigue susceptible when low cyclic stresses are present and surrounding base metal is susceptible when higher cyclic stresses are present. Also, the HSLA laser weld fatigue properties tend to be insensitive to the amount of concavity typically found at the weld cap and root. Lastly, the fatigue cracks tend to initiate at ripples within the weld FZ under low cyclic stresses.

The goal of the present study is to determine a fiber laser weld processing window to achieve optimized butt weld geometry, and to evaluate microstructure, and low-cycle fatigue properties of optimized welds made on HSLA marine-resistant steel. This particular HSLA steel grade (ASTM A690) has application in ground vehicle chassis components where fatigue design is critical, although it is also used in offshore environments where seawater corrosion resistance is desired and cyclic wave loading may occur. The material was acquired in thicker sheet form than some of the past fatigue studies discussed above. The thicker sheet enables low cycle fatigue testing where applied strain is fully reversed, i.e., the minimum strain is fully compressive, since a typical problem encountered during low cycle fatigue testing is that specimens of insufficient thickness will buckle under compressive load without critical test machine alignment.

2 Experimental Procedures

2.1 Material

HSLA steel sheet (4.62 mm nominal thickness) conforming to ASTM A690 Grade 60 was used for this study. As-received sheet had nominal dimensions of 0.66 m x 1.23 m x 4.62 mm. The chemical composition of the plate measured by spark Atomic Emission spectroscopy according to ASTM E415 is shown in Table 1 along with IIW carbon equivalent (CE_{IIW}) according to Appendix A. The steel grade is microalloyed and also contains copper and nickel additions for improved corrosion resistance in marine environments. Rectangular blanks were waterjet cut from the sheet to a dimension of 100 mm x 300 mm where the longer (300 mm) direction was oriented transverse to the rolling direction of the sheet.

Table 1. Chemical composition (wt. %) of HSLA steel with balance being iron. Equations for carbon equivalent (CE_{IIW}), Cr_{eq} , and Ni_{eq} are given in the Appendix [16].

C	Si	Mn	Ni	Cr	Mo	V	Cu	Sb	Ti
0.036	0.34	0.64	0.432	0.069	0.019	0.002	0.515	0.003	0.007
N	Sn	Al	B	Nb	S	P	CE _{IIW}	Cr _{eq}	Ni _{eq}
0.01	0.012	0.02	0.001	0.027	0.003	0.087	0.224	0.61	2.08

2.2 Laser Beam Welding Process

Plates were degreased with acetone, cleaned with ethanol, then clamped and held in place with weld fixturing to achieve a tight fitup at the mating surfaces of the two plates. No special edge preparation was performed prior to welding. Two 100 mm x 300 mm sheets were butt welded using a 1070 nm continuous wave diode pumped Yb-fiber laser with a focused spot size of 0.6 mm. The optics of the system were mounted in the welding head consisting of a 100 mm focal length collimator with a 200 mm focal length final focus lens. The laser beam was delivered via a 300 μ m diameter multimode process fiber resulting in a measured spot size of 0.6 mm oriented normal to the workpiece with a nominally flat top profile. Laboratory grade nitrogen gas (> 99.99 % purity) was used to shield the welds with a nozzle that directed gas flow applied at a 30° angle with respect to the base plate flowing at 0.9 m³/s parallel to the welding direction. The gas flow rate was not optimized to provide optimal surface finish. The feed fiber from the laser to the beam switch is 100 μ m diameter. Trial welds were made where travel speed and laser power were adjusted to find welding parameters that produced the optimal weld profile and cross-section (e.g., achieve full penetration, minimize porosity, spatter, and undercut). Figure 1(a) shows the weld configuration.

Figure 1. Schematic diagrams of (a) butt weld configuration and specimen layout, and (b) low-cycle fatigue specimen dimensions from [17].

2.3 Characterization

Welds were cross-sectioned, mounted in conductive bakelite, then ground progressively with 120 to 1200 grit SiC paper. Welds were polished with 3 μ m and 1 μ m diamond suspension then given a final polish in 0.05 μ m alumina suspension. Polished samples were etched in 2 % Nital solution (2 % Nitric – 98 % Ethanol). Microstructure was imaged with scanning electron microscopy at accelerated voltages up to 10 kV.

Weld contours were characterized with a commercially available optical profilometry system to determine concavity depth during weld optimization trials. Short lengths of weld coupon (about 5 mm long and 25 mm wide) were placed in the profilometer, which was used to record a ~150 μm wide 3D image across 6 mm of weld. Surface roughness values of the base metal adjacent to the weld fusion zone were also measured. The technique has a surface height sensitivity of approximately 10 nm.

2.4 Mechanical Testing Methods

2.4.1 Hardness Measurements

Vickers microhardness tests were performed according to ASTM E384 on several of the polished weld cross-sections removed from welds made at the optimal welding parameters. A commercially available automated microhardness mapper was used to place and measure indents along a 2-dimensional grid with x - and y -spacing between indentations set to 300 μm . A 200 gram force was placed on the indenter for a 13 second dwell time at each indent location. The calibration of the mapping systems was verified on a reference test block where measured values were within 2 % of the certified hardness value of the test block.

2.4.2 Uniaxial Tension and Fatigue Testing

Uniaxial test specimens were sectioned from the welded plates with waterjet cutting according to Figure 1(a). Monotonic tension test specimens were fabricated according to ASTM E8/E8M (50 mm gage length) and cyclic fatigue specimens were fabricated according to ASTM E606 standard test methods as shown in Figure 1(b). All base-metal and weld-containing specimens were sectioned in this manner and the weld was always located along the centerline of the reduced section. The waterjet process resulted in roughness on the cut faces and potentially could have induced residual compressive stresses (although these were not characterized here). The waterjet surfaces were ground with 180, 240, 400, 600, 800, and 1200 grit SiC paper and polished to a near mirror finish (2400 grit) prior to testing. All grinding and polishing was performed in the axial direction to orient surface scratches in the direction of loading per ASTM E606 recommendations [17]. The weld cap and root faces, and adjacent base metal, were not altered from the as-welded condition (i.e., no post-weld grinding) so as to provide the most severe (and realistic) stress risers for the uniaxial tests. All uniaxial tests were performed under ambient conditions (~21 °C, ~20 – 40 % relative humidity).

A closed-loop servohydraulic universal test frame (245 kN maximum force capacity) was used to perform uniaxial testing. Hydraulic wedge grips were carefully aligned prior to the test program so that the top and bottom grip centers were within 80 μm from a common center. This alignment was particularly important for the fully-reversed fatigue testing described below. Quasistatic tension testing was performed according to ASTM E8/E8M Standard Test Method for Tension Testing of Metallic Materials [18] at a constant actuator displacement rate of 0.0508 mm/s using a 50.8 mm extensometer that was attached to the reduced section of the test specimens. Tension tests of base metal and weld specimens were performed in triplicate.

Strain-controlled fatigue testing was performed on the same test frame as the tension testing according to ASTM E606 Standard Test Method for Strain-Controlled Fatigue Testing [17]. A clip-on extensometer (with 12.7 mm gage length) was attached to the face of the specimen with the weld placed at the center of the gage section (in weld tests). A triangular waveform was used to vary from the minimum to maximum strain under the fully-reversed condition, i.e., $R = \epsilon_{\min}/\epsilon_{\max} = -1$, where R is the ratio of minimum strain ($\epsilon_{\min}$) to maximum strain ($\epsilon_{\max}$) applied during each cycle. A constant strain rate of 0.01 s^{-1} was achieved within the extensometer gage section; therefore, test cyclic frequency decreased with increasing strain amplitude and varied between 0.33 Hz and 2.5 Hz over the ranges of applied strain amplitude. The

failure condition selected for the current work was when the number of reversals ($2 \cdot N_f$, where N_f is the number of cycles to failure) where applied force decreased by 50 % of the maximum force applied to the specimen on the first loading cycle due to the presence of a fatigue crack within the gage length. Fatigue specimens that failed outside the extensometer gage section were not included in the data analysis.

Stochastic speckle patterns were applied to several tension specimens containing welds for additional studies with digital image correlation (DIC) analysis. Pattern application was a two-step process that consisted of applying a white matte enamel base layer and then a matte black speckle layer using aerosol applicators. 3D DIC was performed using a stereo digital camera setup and commercially available image capture software that was synchronized with force, displacement, and extensometer signals originating from the universal test frame controller. Strain post processing was performed with commercially available 3D DIC processing software.

2.4.3 Residual Stress Measurements

Weld residual stresses were measured in the as-welded sheet and in an E606 type specimen machined from the as-welded sheet using technique described above. Residual stress measurements were performed at the High Flux Isotope Reactor HB-2B beam line at Oak Ridge National Laboratory. Residual stress determination by neutron diffraction utilizes Bragg's law [19]:

$$\lambda = 2d \sin \theta \quad , \quad (1)$$

where λ is the incident neutron wavelength, d is the lattice spacing, and θ is half of the diffraction angle. Each crystallite in the illuminated gage volume will scatter incident radiation at an angle corresponding to Equation (1). The strain ε is determined from the measured lattice spacing as:

$$\varepsilon = (d - d_0) / d \quad , \quad (2)$$

where d_0 is the lattice spacing of the unstressed material. The stress is then calculated by:

$$\sigma_{ij} = (E / (1 + \nu)) [\varepsilon_{ij} + (\nu / (1 - 2\nu)) (\varepsilon_{xx} + \varepsilon_{yy} + \varepsilon_{zz})] \quad , \quad (3)$$

where $i, j = x, y, z$, $E = 224.8$ GPa, and $\nu = 0.2827$ are the diffraction elastic constants determined by the Kroner model for the Fe (211) lattice plane [20]. The unstressed lattice spacing was determined at a spot far from the weld on the E606 type specimen, and at a spot far from the FZ and HAZ area ($y = 40$ mm) on the as-welded sheet (coordinates are defined in Figure 1(a)). Because the stress normal to a free surface is zero at the free surface, the z-direction lattice spacing at the surface is the unstrained lattice spacing. Since a 8 mm^3 neutron interaction gage volume was used (where gage volume is defined by x , y , and $z = 2$ mm), the center of the gage volume closest to the free surface achievable during this experiment was 1.4 mm from the edge. Lattice spacings were measured as a function of z , at $y = 40$ mm from the weld center, and $x = 0$ mm from the weld center. Lattice spacing varied linearly through the thickness to the free surface ($z = 2.15$ mm), where the unstrained lattice spacing was determined to be $d_0 = 1.170742 \text{ \AA}$. Within the weld, the spacing between each measurement was 1 mm. In the base metal, the spacing between each measurement was 2 mm.

3 Results and Discussion

3.1 Welding Process Optimization

The initial work of this study was focused on optimizing geometry of the fiber laser welds since prior work had shown that weld profile optimization was critical to fatigue performance due to the notch effects present at weld cap and root locations [8]. Weld profile optimization was accomplished here by

following the past approach where welding travel speed and power were adjusted, followed by destructive weld-cross section analysis to minimize weld fusion zone concavity. The two primary laser welding parameters, travel speed and laser power, were varied in between butt welds followed by destructive evaluation of weld cross-sections. Laser power was varied between 1 and 8.5 kW and travel speed was varied between 1000 and 6000 mm/min. Figure 2 shows the resulting process window for this particular material and welding process variable combinations. Three characteristic weld profiles were observed including incomplete penetration, full penetration with excessive concavity, and full penetration with minimized concavity. Incomplete penetration was due to laser power being too low to sufficiently melt the entire thickness. Only minimal porosity was observed on a few of the weld cross-sections made at the lower power levels. Convexity of the weld cap was not observed within the parameter space studied although it has been reported when welding thinner sheet steels with a fiber laser [8]. The inset shows weld macrographs of the various weld profiles observed within the process envelope. Based on the results presented in this section, it was concluded that an optimized weld could be made at a laser power level of 5 kW and a travel speed of 2.5 m/min. This parameter set was used to perform all welds that were used for the characterization and mechanical testing work described below.

Figure 2. Fiber laser welding process envelope for butt welds.

In addition to destructive testing, further non-destructive evaluation was performed on weld surfaces using a non-contact optical profilometry technique. The benefits of a nondestructive technique include the avoidance of costlier destructive testing, and the possibility of adoption into production scenarios for near-real-time quality monitoring. Traces of the optical profilometer across the top of several locations of two laser welds made with optimized parameters are shown in Figure 3. Note that the axes are scaled differently where the scale on the y-axis is greatly expanded to emphasize sensitivity of the measurement. Numeric labels indicate profiles adjacent to where various mechanical specimens were water-jet cut from the welded plates. Variation in surface height is plotted across the weld and approximately 2 mm on either side of the fusion zone. The measurements show several scans across the weld on two different plates that were used for the subsequent machining of mechanical test specimens for strength and fatigue

studies. Optimized welds typically had less than 0.2 mm concavity depth across a fusion zone width of over about 2 mm. Base metal surface roughness was also determined by calculating a weighted average of the roughness for the total scan length of each specimen location (Average Ra = 2.4 μm , Average Rrms = 3.4 μm). Sheet fitup is also evident. Note that at some locations (e.g., K11 13 and K12 25) scanned, the sheet surface on one side is higher or lower by approximately 0.1 mm. This mismatch, which is within typical tolerances, is likely attributed to deviations in nominal sheet thickness along the weld length and distribution of clamping force by the welding fixture.

Figure 3. Contour of weld surfaces determined with optical profilometry. Approximate regions across two welded plates are indicated.

Optical profilometry is a very sensitive surface measurement technique; however, it is currently time-intensive and requires precise motorized stages to calibrate the optical measurements. Other approaches such as Light Detection and Ranging (LIDAR) have shown promise for such measurements and may provide a more robust method to analyze weld contour real time. Work on this area is currently ongoing as part of this program, although preliminary results have been demonstrated [21].

3.2 Microstructure

Scanning electron micrographs of the base metal microstructure are shown in Figure 4, where the rolling direction is indicated on the figure. The base metal steel has an equiaxed ferrite matrix, a uniform distribution of secondary carbide precipitates along the grain boundaries, and a small fraction of pearlite colonies interspersed between the ferrite phase. No porosity was found in any of the cross sections of welds during polishing or microscopy—while microporosity may exist and could influence the mechanical properties.

Figure 4. Scanning electron micrographs of base metal microstructure. Rolling direction (RD) and short transverse (ST) directions are indicated.

The welds exhibited fairly uniform profile along their length (as shown in Figure 3), and heat flow conditions are not expected to vary significantly since constant welding parameters and weld geometry were used; thus, the following characterization is representative of the two welded plates used for subsequent mechanical testing. Figure 5(a) and (b) show a weld macroscopic cross-section taken from one of the welded plates adjacent to where a fatigue test specimen (K11 23) was removed for fatigue testing, and a hardness contour map that shows variations in hardness across the weld profile. The welds exhibited a narrow aspect ratio where the weld depth was approximately four times the weld width. A macroscopic columnar grain structure (shown in Figure 5(a)) was evident in the FZ region of the weld due to the solidification as delta ferrite. Such deep and narrow weld structure can be problematic for weld solidification cracking; however, no cracking was observed in any of the welds, made with optimized parameters. Some solidification cracks were observed in the welds made at the lower laser powers (< 5 kW) and higher travel speeds where incomplete penetration was observed. The composition of this alloy is prone to solidification cracking based on analysis of weld constitution with Ni and Cr equivalencies proposed in Ref. [16]. According to that methodology, when $Cr_{eq} < 2.0$ and Ni_{eq} is < 2.3, fully ferritic solidification in the fusion zone results in a structure susceptible to solidification cracking when $(P + S) > 0.04^b$. Equivalency formulae are provided in the appendix. Therefore, according to Table 1 this alloy may exhibit solidification cracking when welded under certain conditions outside the processing envelope described above. The presence of such cracks in the fusion zone would be expected to have a significant reduction on fatigue performance.

^b With current steelmaking practice, this level of P+S is only observed in steels with an intentional P addition.

Figure 5. Cross-sections of optimized butt welds showing (a) macrostructure and weld regions, and (b) microhardness distribution. Points on microhardness distribution indicate locations of indentations.

Figure 6 shows scanning electron micrographs of the various weld regions (fusion zone is referred to as FZ and heat-affected zone is referred to as (HAZ)) on the macro image. A martensitic plate structure was observed within the FZ as reported in other laser beam welds made on low carbon HSLA steel [12]. The narrow coarse-grained region of the HAZ adjacent to the fusion boundary, which was heated above the A_3 temperature, contained mostly martensitic structure and also some ferrite; prior austenite grains were coarser than those in the FZ, although the prior austenite grain size of the base metal was not determined for comparison. The remaining HAZ, i.e., the intercritical region, contained a mixture of ferrite and martensite, where the martensite fraction was quite low. This region was only partially austenitized, and a (qualitative) decreasing area fraction of martensite was visible in the micrographs when moving away from the weld and into the base metal. The subcritical region of the weld presented no evidence of martensite as the base metal contains only ferrite and carbide. The time interval necessary to cool through the transformation temperature range (800 °C to 500 °C) of steel, the $t_{8/5}$, was estimated with the Rosenthal heat conduction analytical solution for 2D heat flow [22]. The cooling time through this temperature range was less than 0.8 s in both the FZ and HAZ. Based on a phase transformation model [23] using the composition shown in Table 1, and the calculated cooling time through the $t_{8/5}$, the structure should be about 90 % martensite, and 10 % bainite + ferrite (with predicted hardness of 292 HV) upon cooling from locations in the weld where austenization occurred. This matches qualitatively with observations of microstructure in both the FZ and the HAZ.

Figure 6. Scanning electron micrographs of weld microstructure showing fusion zone (FZ), heat-affected zone (HAZ), coarse-grained heat-affected zone (CG-HAZ), and intercritical heat-affected zone (IC-HAZ).

The hardness map (Figure 5(b)) reveals no softening in the base metal as reported in other high strength steels such as the dual phase variety. The peak hardness of the weld, which occurs in the fusion zone, near the centerline and close to the top of the weld is less than two times the base metal hardness, which is another measure of weld quality when joints are subjected to subsequent forming operations [24]. The FZ hardness was typically greater than approximately 270 HV. This HSLA steel is not highly hardenable, and carbon content is quite low, thus the hardness of the martensite is also relatively low; 310 HV_{0.2} was the median hardness observed in the FZ. Krauss demonstrated that hardness of iron-carbon martensite containing about 0.4 wt. % carbon (0.036 wt. % carbon in the present HSLA steel) would have a hardness in the range of approximately 250 to 300 HV [25], a value in line with observations reported here. Fusion zone hardness can be used to estimate FZ yield strength, which is over 900 MPa according to a formula proposed by Pavlina, *et al.* [26] (where yield strength = $15.3 + 2.86 \cdot \text{HV}$ with $R^2 = 0.879$).

3.3 Weld Constitutive Behavior

Four tensile tests were performed on weld specimens and each of them failed in the base metal at a distance of approximately 20 to 25 mm from the HAZ. Constitutive stress-strain relationships of a welded specimen and a base metal specimen are shown in Figure 7(a). Since only base metal failures occurred, this indicates that if embrittling weld defects such as porosity or weld solidification cracking were present in the FZ, they were not sufficient to induce weld failure. Yield strength and ultimate tensile strength exceeded ASTM A690 Grade 50 minimum values ($\sigma_{ys} > 345 \text{ MPa}$, $\sigma_{uts} > 485 \text{ MPa}$). The elongation at failure of welded specimens was approximately 4 to 6 % lower than all base metal

specimens which is likely a result of the transformation strengthening in the FZ and HAZ, which were visually more resistant to deformation than base metal.

DIC was performed on a weld tensile specimen to evaluate localized property variations around the weld and to determine the extent of plastic deformation imparted into the weld region. Figure 7(b) shows the relationship between applied force and measured axial strain of the specimen. Strain was measured across the entire gage section (Labeled “total”), which includes both BM and weld, and it was measured across the weld region only (Labeled “weld”), which includes HAZ and FZ. The inset figures indicate overall distribution at the axial strain levels indicated. Note that initially, the weld appeared to accumulate more strain than surrounding base metal, which is likely a result of the elastic stress concentration effect at the weld. As applied force increased, most of the strain was imparted into the base metal on either side of the weld, although the region including the FZ and HAZ accumulated over 3 % strain before the base metal failed. The weld was plastically strained after the surrounding base metal achieved sufficient strain hardening, although the weld region exhibited significantly greater strain hardening rate.

Figure 7. (a) Engineering stress-strain behavior of base metal and weld tensile samples. (b) Results of digital image correlation showing axial strain distribution of welded specimen.

3.4 Residual Stresses

The magnitude and distribution of residual stress, which results from the localized heating and cooling during welding, plays a critical role in weld service performance [27, 28]. Fatigue in particular is influenced strongly by residual stress since the mean stress of a load cycle can be affected by the residual stress state. Residual stress distribution in the three plate directions of an as-welded plate made with optimized weld parameters is shown in Figure 8(a), where plate direction and neutron analysis locations are indicated in Figure 1(a). The stress distribution was assumed to be symmetric about the weld centerline, since weld plate geometry, fixturing and heat input was symmetric for this weld, thus only half the plate was characterized. Maximum residual stress in the FZ (~450 MPa) was similar in magnitude to the base metal yield strength (450 MPa) in the welding direction. Stresses were quite low, less than about 150 MPa in magnitude, in the transverse and through-thickness directions. However, the FZ was significantly strengthened relative to the base metal by the transformation hardening, and maximum residual stress in the FZ is about half of the estimated FZ yield strength of 900 MPa (based on hardness).

Martensitic steels are known to exhibit transformation plasticity where residual stress is relieved somewhat upon transformation of austenite to martensite, and most residual stresses in martensite accumulate after cooling from the completion of the transformation (as reviewed by [28]). Therefore, the peak residual stress in the laser weld may be somewhat lower than if this material were welded in higher heat input process that would preclude rapid cooling and subsequent martensite formation.

Figure 8. Residual stress distribution of (a) as-welded plate prior to sectioning fatigue specimens, and of (b) reduced section of low-cycle fatigue specimen after machining to final dimensions.

Residual stress distribution was also determined in the reduced section of an E606-type fatigue test specimen after it was machined from the full welded plate (Figure X8(b)). Note that the stresses in the FZ and HAZ were less than 100 MPa in magnitude (over 80 % reduction) due to the stress-relieving effect of cutting the specimen from the as-welded plate. The transverse residual stresses changed from a compressive to tensile state due to stress redistribution, and all residual stresses were compressive in the base metal several mm away from the HAZ. The tensile mean stress (σ_m) at the weld of a structure containing the full residual stress field would be considerably higher than in the test specimens as shown here. An increase in the tensile mean stress can cause a significant reduction in fatigue life, whereas a compressive mean stress can be beneficial [29]. These findings bear consideration when interpreting fatigue data presented below, as the full effect of residual stress is not captured in a standard fatigue test specimen configuration due to stress relaxation effect when the welded plate is sectioned.

3.5 Low Cycle Fatigue Behavior

Base metal fatigue tests were conducted to establish baseline low-cycle fatigue behavior and then welded specimens were subsequently tested. Two failure criteria were used including the fatigue crack damage must be within the extensometer knife edges, and a 50 % reduction in force-carrying capacity on a plot of force (P) vs. number of cycles to failure (N_f) signaled the test to end (as recommended in ASTM E606). Figure 9 shows the raw data of base metal and weld fatigue failures where the points represent the number of reversals to failure ($2 \cdot N_f$) for a given applied total strain amplitude ($\Delta \varepsilon/2$). The runout condition was defined as a specimen that did not exhibit fatigue damage with a minimum of 10^6 reversals. Note that the weld was more susceptible to fatigue at the higher strain amplitudes, but at lower strain amplitudes, the fatigue susceptibility of welded specimens was similar to that of base metal. This is unlike S-N data generated for HSLA fiber laser welds, which generally shows that welds exhibit similar behavior to base metal at high stress amplitude and are more fatigue susceptible than base metal at low stress amplitudes [8, 12].

Figure 9. Low-cycle fatigue data showing applied total strain amplitude versus the number of strain reversals until failure indicated by a 50 % reduction in applied force.

3.5.1 Analysis for Strain-Life Prediction

The base metal and welded specimens exhibited a cyclic softening response, i.e., the resulting force-carrying capacity decreased initially with cycle count during a shakedown period before reaching a steady-state region where force did not decrease considerably with cycle count. This behavior is typical of a material that exhibits Luders-band formation, a mechanism where dislocations are pinned and released precipitously in a cascade during tensile fatigue loading [30], which was evident in this material when examining stress-strain response. Following the initial softening was a region of little force change until crack initiation. After sufficient damage accumulation to initiate a fatigue crack, the force began to decrease until failure rapidly progressed near the test end. Plots of calculated stress (assuming uniform thickness through the gage section in weld samples) vs. number of cycles are shown in Figure 10 for base metal and weld specimens that exhibited typical material response observed during the test program. The fatigue flaw leading to failure of the weld specimen occurred in the base metal. In general, it was observed that when failure of weld specimens occurred in the base metal, the fatigue data matched more closely to that of tests containing only base metal. Mechanical hysteresis loops were evaluated at 50 % of life (i.e., $0.5 \cdot N_f$). Luders behavior was observed in base metal hysteresis loops measured at the highest strains (0.75 %); however, no other test conditions exhibited such behavior. The hysteresis loops were used to separate the strain into plastic and elastic components for analysis with the Basquin and Coffin-Manson equations [10]. Elastic strain life is described by Basquin's equation:

$$\frac{\Delta\sigma}{2} = \sigma'_f (2N_f)^b, \quad (4)$$

which can be solved for strain,

$$\frac{\Delta\epsilon_e}{2} = \frac{\sigma'_f}{E} (2N_f)^b, \quad (4a)$$

where σ denotes stress (and $\Delta\sigma/2$ is the stress amplitude), σ'_f is the fatigue strength, b is the fatigue strength exponent, E is Young's Modulus, and ε_e is the elastic strain (and $\Delta\varepsilon_e/2$ is the elastic strain amplitude). Plastic strain life is described by the Coffin-Manson equation:

$$\frac{\Delta\varepsilon_p}{2} = \varepsilon'_f (2N_f)^c \quad , \quad (5)$$

where ε_p is the plastic strain (and $\Delta\varepsilon_p/2$ is the plastic strain amplitude), ε'_f is the fatigue ductility coefficient, and c is the fatigue ductility exponent. The total strain life is then given by the sum of the elastic and plastic components:

$$\frac{\Delta\varepsilon_t}{2} = \frac{\Delta\varepsilon_e}{2} + \frac{\Delta\varepsilon_p}{2} \quad , \quad (6)$$

where ε_t is the total strain. Combining Eqs. 4 through 6 results in:

$$\frac{\Delta\varepsilon_t}{2} = \frac{\sigma'_f}{E} (2N_f)^b + \varepsilon'_f (2N_f)^c \quad , \quad (7)$$

which describes the fatigue strain-life curve as a function of the number of reversals to failure. This equation was fit to experimental fatigue data of base metal and weld as shown in Figure 11(a) and Figure 11(b), respectively. Empirically-derived material coefficients and coefficient of determination for the curve fits are shown in Table 3. The table also contains strain-controlled fatigue data (conducted at $R = -1$) from several steels of similar strength and microstructure [10]. Note that the total strain life (Equation 7) fits well to base metal fatigue data ($R^2 = 0.93$) and to a lesser extent to the laser welds ($R^2 = 0.60$). Fatigue specimens containing weld material failed in both the base metal and at the weld as discussed below. Therefore, a third regression was performed on data containing only failures at the weld showing an improved fit of total strain life ($R^2 = 0.80$).

Figure 10. Plots showing stress versus cycle count for weld and base metal (BM) specimen tested at 0.3 % strain amplitude. Note that the weld specimen failed in the base metal.

Figure 11. Plots of strain amplitude versus reversals to failure for (a) base metal and (b) weld.

The elastic strain equation (Equation 4) fit to the elastic data fairly well in all three cases. The fatigue strength coefficient (σ'_f) and the fatigue strength exponent (b) of the base metal are lower in magnitude than the coefficients determined in prior work [10]. The fatigue strength coefficient (σ'_f) is the stress intercept in Equation 4, and the curve fit quality performed here for BM and weld likely suffers from lack of data at higher strain amplitudes. Fatigue data presented in [10] included results from tests conducted at

strain amplitudes up to 1.5 %. Also, the exponent, b is weighted more heavily to elastic strains at the higher cycles. Collection of more data at very low and very high cycles could potentially increase both of the elastic input coefficients accordingly. However, the b -exponent and fatigue strength for weld materials is similar to the martensitic steel (SAE 1141) and the cold-worked steel (SAE 1541) indicating more elasticity due to increased strength. Weld modulus was measured on each hysteresis curve using the unloading method described in the ASTM E606 standard [17]. The welds tended to increase stiffness of the specimen relative to base metal as shown by the average value in Table 3.

Table 3: Summary of Strain-Controlled Fatigue Properties

Fatigue Parameter	Base Metal	Weld (All failure locations)	Weld (Only weld failure locations)	SAE 1141 Q+T [10]	SAE 1038 Normalized [10]	SAE 1541 Cold Worked [10]
E (GPa)	211	220*	220*	227	201	205
ε'_f (mm/mm)	0.296	0.070	0.157	0.309	0.309	0.513
σ'_f (MPa)	692	744	825	1127	1043	1044
b	-0.059	-0.069	-0.079	-0.066	-0.107	-0.083
c	-0.566	-0.539	-0.618	-0.514	-0.481	-0.557
N_t (cycles)	3606	316	516	5034	27248	8378
Average σ_m (MPa)	-0.6 ± 8.0	-13.1 ± 28.0	-	-	-	-
R^2 (Total Strain Life)	0.93	0.60	0.80	-	-	-
R^2 (Elastic Life)	0.96	0.82	0.92	-	-	-
R^2 (Plastic Life)	0.88	0.52	0.48	-	-	-
K'	1080	1140	-	1277	1340	950
n'	0.153	0.141	-	0.124	0.220	0.114
σ_{YS}' (MPa)	422	464	-	591	342	469
R^2 (Cyclic Ramberg Osgood)	0.94	0.82	-	-	-	-
σ_{YS} (MPa)	450	450 min (900 estimated in weld)	-	814	331	475
Hardness (HV)	175	310	310	292 (**277 HB)	171 (**163 HB)	204 (**195 HB)
Microstructure***	F + P	M + F + P	M	M	F + P	F + P
*E represents overall stiffness of reduced section in weld containing composite of BM, HAZ, and FZ						
**Converted to HV from Brinell Hardness (HB)						
***F = Ferrite, P = Pearlite, M = Martensite						

The weld samples exhibited less cyclic plastic deformation than base metal, likely due to the strengthening effect shown in monotonic tension tests discussed above. Thus, Equation 5 fits well to plastic data in only the case of the base metal tests. The fatigue ductility exponent, c of the weld data is

significantly higher than base metal and literature values. The fatigue ductility coefficient (ϵ'_f), a value typically close to the true fracture strain [31], is significantly lower than in the other materials presented. This value is defined by the strain intercept in Equation 5, i.e., $2 \cdot N_f = 1$. Higher applied strain amplitudes than those applied in this work are needed to determine this intercept with greater accuracy. The test setup used here was unable to reliably achieve strain amplitudes over 0.3 % without buckling the welded samples, which contained inherent features of axial misalignment due to the plate fit up during welding and weld concavity, which was minimized by the optimized welding procedure. The test setup could reliably achieve strain amplitudes of 0.75 % in base metal samples, which contained parallel faces on all sides of the fatigue specimen; thus the base metal fatigue ductility coefficient is close to the literature values for SAE 1141 and SAE 1541 steels.

The transition life (N_t) provides a measure for identifying regimes where fatigue life is dominated by plastic or elastic strain. Transition life of the base metal is lower than SAE 1038 and SAE 1541 shown in Table 3, yet is comparable with other steels such as SAE 1141, and SAE 1090 (not shown in the table), a higher strength (735 MPa to 950 MPa) martensite/bainite grade [10]. The weld transition life is quite low compared to the other cases, further showing that the welds do not accommodate cyclic plastic deformation.

Other work has incorporated terms into Equation 7 to correct for mean stress (σ_m) effects, which can significantly alter fatigue behavior. Mean stress of base metal and weld samples was calculated for each test at midlife, and then averaged (Table 3). At midlife, stresses may have relaxed somewhat due to cyclic softening effect observed on plots of force versus cycle count, but σ_m provides a consistent point of comparison among all the specimens in the group of base metal and weld. The welds show a slightly compressive mean stress (-13.1 MPa) and the base metal essentially shows a zero mean stress. This is not unexpected since the welds contain some residual stress, even after removal from the parent plate. Note on Figure 8b (σ_{yy}) that weld residual stress, parallel to loading direction in the machined specimen containing a weld, have a tensile component of about 80 MPa at the weld center and -80 MPa at the edge of the reduced section. A simple moment analysis on the residual stress data shows that the mean stress within the section measured for residual stresses would be about -32 MPa, a value not far from the measured mean stress after cycling to midlife when some stress could be relieved. Compressive mean stresses have been shown to have some beneficial effect on fatigue of high-strength steels, particularly at lower strain amplitude [29, 32], thus their presence here is not detrimental to performance. One issue that is not examined here is that it is unclear what magnitude and sign the mean stress would have if a welded structure were subjected to fatigue, since residual stresses are greatly relieved when cutting the specimens.

3.5.2 Cyclic Stress-Strain Response

Steady-state cyclic deformation behavior was determined by analysis of hysteresis at midlife for base metal and weld material. A Ramberg-Osgood relationship was fit to plots of stress amplitude versus strain amplitude as shown in Figure 12 and are described by:

$$\frac{\Delta \epsilon}{2} = \frac{\Delta \sigma}{2E} + \left(\frac{\Delta \sigma}{2K'} \right)^{\frac{1}{n'}}, \quad (8)$$

where K' is the cyclic strength coefficient and n' is the cyclic strain hardening exponent. Modulus values in Table 3 were used for the fit. Note that the weld specimen had an effective modulus that was stiffer than that of the base metal due presumably to weld strengthening. This effective modulus is only an estimate since the specimen cross-section containing the weld did not have the typical uniform geometry of a standard tensile specimen. Therefore, the Ramberg-Osgood coefficients and cyclic yield strength

(σ_{ys}) shown in Table 3, which were determined using a 0.2 % offset method, are estimates for the weld material. Base metal cyclic yield strength is comparable to the other ferritic grade shown in the table (SAE 1038 [10]) and weld strength is increased relative to base metal, which was also observed during monotonic testing. However, the cyclic yield strength of base metal and weld are less than monotonic yield strength due to the observed cyclic softening. The cyclic strength coefficient and cyclic strain hardening exponents of weld and base metal measured here are comparable to other steels reported in the literature [10].

Figure 12. Cyclic stress strain curves for base metal and weld and corresponding Ramberg-Osgood fits.

3.6 Fatigue Failures

Fatigue cracks initiated and grew in both the base metal and at the fusion zone in fatigue specimens containing the optimized welds as discussed previously and in most all cases grew in a direction normal to direction of axial force. Location of weld specimen failure did not depend on strain amplitude. Also, failure initiation occurred at both the top and root faces. Three cases of fatigue initiation and growth, which are shown in Figure 13, can be summarized as follows:

- Base Metal – Cracks initiated and propagated entirely within base metal, with some cracks apparently initiating at defects that were likely associated with material rolling or handling.
- Fusion Zone – Initiation and growth of cracks associated with the weld fusion zone occurred at weld ripple marks at the top face of the weld. Ripples are a result of the weld solidification process.
- Weld Root – The root was an initiation site for cracks which propagated along the boundary of the HAZ.

Type (a) is most likely associated with a harder weld region being more resistant to deformation than the softer base metal where surface defects act as stress concentrations. Type (b) and (c) are likely associated with a stress concentration effect at the interface between the harder fusion zone and the softer base metal in the heat affected zone, which overcomes any beneficial weld strengthening effect seen in Type (a).

Figure 13. Weld failure locations in (a) base metal, (b) fusion zone, and at (c) weld root. Strain amplitudes are shown for each specimen and arrows indicate location of fatigue cracks.

Scanning electron microscope images of the fatigued fracture surfaces are shown in Figures 14 through 16 for the three cases of base metal failure, weld toe failure, and heat affected zone failure. Since fatigue testing was halted at 50 % decrease in force, the specimens still contained some ligament of material, which was broken in liquid nitrogen to reveal the full fracture surfaces. In all cases fracture surface features common in fatigue were found including: ratchet marks, striations, and featureless-smeared regions. An example of the base metal failure of the weld can be seen in Figure 14, a distinct brittle fracture region was found in the fracture profile due to the low temperature breaking, along with multiple initiation sites (labeled A and B), which were found by tracing the radial flow marks toward a surface defect (Figure 14(b)). The ductile fracture surface had a smeared, featureless surface with some cracking as can be seen in the final portion of Figure 14(c). Figure 15 illustrates failure occurring from the weld toe, in this case cracks initiated from the top surface of the weld from A, B, and C as indicated in Figure 15(a). Initiation occurred from irregularities on the top surface of the weld as can be seen in Figure 15(b) and (c), which was likely due to ripples on the top surface of the weld or other particle formation similar to other work [8]. Figure 16 shows the failure that occurred in the HAZ at the root of the weld. The specimen shown is from Figure 13(c) (bottom left) where a surface defect can be seen, which is indicated by B in the macrograph in Figure 16. This defect acts as a stress concentrator for the crack initiation at the point labeled A. Failure of this type did not always appear to be initiated by a stress concentrator but could be due to subsurface defects like small pores in the root material, or intermetallic particles found in the lower portion of the weld, as shown in Figure 16(c).

Figure 14: Base metal fatigue failure from initiation points A and B. Initiation site B in lower left. Smearing and fatigue striations are shown at location C in the bottom right.

Figure 15. Failure at the weld metal toe region of the weld. Initiation points of A, B, and C are indicated each at surface defects. (Bottom left) Initiation point A. (Bottom right) Initiation point B.

Figure 16. Failure in heat-affected zone at root of weld. Cracks initiated at A and B then propagated toward C. Inclusions or second phase found at initiation point near A as shown in the bottom right.

4 Conclusions

- A procedure was optimized for fiber laser butt welding of corrosion-resistant HSLA steel. Welding parameters including power and travel speed were adjusted to achieve optimal geometrical weld profile for fatigue resistance.
- Laser weld fusion zones were martensitic and weld HAZ contained martensite and ferrite resulting in higher hardness and strength in the weld with respect to base metal. The rapid cooling rates of the laser welding process were responsible for the transformation strengthening.
- Laser welding resulted in significant tensile residual stresses in and around the fusion zone. The magnitude of these stresses were reduced appreciably when the fatigue test specimens were machined from as-welded plate. Thus, mean stress in the fatigue specimen would be different than mean stress in a welded structure under cyclic loading conditions. This should be carefully considered when applying lab-scale fatigue test data to modeling structural fatigue of laser welded joints.
- Low cycle fatigue coefficients (for Basquin and Coffin Manson relationships) and cyclic stress-strain behavior were determined for base metal and laser welds. Fatigue coefficients of welds have greater uncertainty than base metal since curve fit quality was not as high. Fatigue coefficients are similar in value to those of other high strength steels.
- Weld fatigue susceptibility was higher than base metal at higher strain amplitudes, although weld cyclic strength was greater than that of base metal resulting in lower plastic strains.
- Fatigue damage of weld specimens accumulated in three characteristic locations depending on local strength variations and strain, and included: 1) in base metal remote from weld, 2) at ripple marks within the weld fusion zone, and 3) at the weld root. Cracks associated with the weld typically followed a path within the heat-affected zone after initiation at the fusion zone.

Acknowledgements

The authors wish to acknowledge the entire NIST “SERI” welding team for their valuable contributions that made this work possible, and Oak Ridge National Laboratory for providing the neutron beam time for residual stress experiments

Appendix

$$CE_{IIW} = C + Mn/6 + (Cr + Mo + V)/5 + (Ni + Cu)/15 \quad (A1)$$

$$Cr_{eq} = Cr + 1.37 Mo + 1.50 Si \quad [16] \quad (A2)$$

$$Ni_{eq} = Ni + 22 C + 0.31 Mn + 14.2 N + Cu \quad [16] \quad (A3)$$

References

- [1] Quintino L, Costa A, Miranda R, Yapp D, Kumar V, Kong CJ. Welding with high power fiber lasers – A preliminary study. *Materials & Design*. 2007;28:1231-7.
- [2] Miranda R, Costa A, Quintino L, Yapp D, Iordachescu D. Characterization of fiber laser welds in X100 pipeline steel. *Materials & Design*. 2009;30:2701-7.
- [3] Li C, Muneharua K, Takao S, Kouji H. Fiber laser-GMA hybrid welding of commercially pure titanium. *Materials & Design*. 2009;30:109-14.
- [4] Cao X, Wanjara P, Huang J, Munro C, Nolting A. Hybrid fiber laser - Arc welding of thick section high strength low alloy steel. *Materials & Design*. 2011;32:3399-413.

- [5] Assunção E, Quintino L, Miranda R. Comparative study of laser welding in tailor blanks for the automotive industry. *The International Journal of Advanced Manufacturing Technology*. 2010;49:123-31.
- [6] Xu W, Westerbaan D, Nayak SS, Chen DL, Goodwin F, Biro E, et al. Microstructure and fatigue performance of single and multiple linear fiber laser welded DP980 dual-phase steel. *Mater Sci Eng A-Struct Mater Prop Microstruct Process*. 2012;553:51-8.
- [7] Parkes D, Xu W, Westerbaan D, Nayak SS, Zhou Y, Goodwin F, et al. Microstructure and fatigue properties of fiber laser welded dissimilar joints between high strength low alloy and dual-phase steels. *Materials & Design*. 2013;51:665-75.
- [8] Westerbaan D, Parkes D, Nayak SS, Chen DL, Biro E, Goodwin F, et al. Effects of concavity on tensile and fatigue properties in fibre laser welding of automotive steels. *Sci Technol Weld Join*. 2014;19:60-8.
- [9] Dong P. A structural stress definition and numerical implementation for fatigue analysis of welded joints. *International Journal of Fatigue*. 2001;23:865-76.
- [10] Roessle ML, Fatemi A. Strain-controlled fatigue properties of steels and some simple approximations. *International Journal of Fatigue*. 2000;22:495-511.
- [11] Boronski D. Cyclic material properties distribution in laser-welded joints. *International Journal of Fatigue*. 2006;28:346-54.
- [12] Xu W, Westerbaan D, Nayak SS, Chen DL, Goodwin F, Zhou Y. Tensile and fatigue properties of fiber laser welded high strength low alloy and DP980 dual-phase steel joints. *Materials & Design*. 2013;43:373-83.
- [13] Jia J, Yang SL, Ni WY, Bai JY. Microstructure and mechanical properties of fiber laser welded joints of ultrahigh-strength steel 22MnB5 and dual-phase steels. *J Mater Res*. 2014;29:2565-75.
- [14] Correard GCC, Miranda GP, Lima MSF. Development of laser beam welding of advanced high-strength steels. *Int J Adv Manuf Technol*. 2016;83:1967-77.
- [15] Parkes D, Westerbaan D, Nayak SS, Zhou Y, Goodwin F, Bhole S, et al. Tensile properties of fiber laser welded joints of high strength low alloy and dual-phase steels at warm and low temperatures. *Materials & Design*. 2014;56:193-9.
- [16] Karjalainen LP, Kujanpää V, Suutala N. Hot Cracking in Iron Base Alloys Effect of Solidification Mode. *Advances in Welding Science and Technology*, (Gatlinburg, TN). 1986:145-9.
- [17] ASTM E606 - 12 Standard Test Method for Strain-Controlled Fatigue Testing. West Conshohocken, PA: ASTM International; 2012.
- [18] ASTM E8 / E8M - 11 Standard Test Methods for Tension Testing of Metallic Materials. West Conshohocken, PA: ASTM International; 2011.
- [19] Hutchings MT, Withers PJ, Holden TM, Lorentzen T. Introduction to the characterization of residual stress by neutron diffraction: CRC press; 2005.
- [20] Simmons G, Wang H. Single crystal elastic constants and calculated aggregate properties. 2nd ed. Cambridge: MIT Press; 1971.
- [21] Miller S, Pfeif E, Kazakov A, Baumann E, Dowell M. Fiber laser welding of dual-phase galvanized sheet steel (DP590): traditional analysis and new quality assessment techniques. *SPIE LASE: International Society for Optics and Photonics*; 2016. p. 97410I-I-12.
- [22] Grong Ø. Metallurgical modelling of welding: Institute of Materials; 1997.
- [23] Ion JC, Easterling KE, Ashby MF. A second report on diagrams of microstructure and hardness for heat-affected zones in welds. *Acta Metallurgica*. 1984;32:1949-62.
- [24] GM. GM4485M - Weld Specifications for Laser Welds - Butt Joints. General Motors Corporation; 2002.
- [25] Krauss G. Principles of Heat Treatment of Steel. Metals Park, Ohio: American Society for Metals; 1980.
- [26] Pavlina EJ, Van Tyne CJ. Correlation of Yield Strength and Tensile Strength with Hardness for Steels. *J Mater Eng Perform*. 2008;17:888-93.
- [27] Withers PJ, Bhadeshia HKDH. Residual stress. Part 1 – Measurement techniques. *Materials Science and Technology*. 2001;17:355-65.

- 756 [28] Withers PJ, Bhadeshia HKDH. Residual stress. Part 2 – Nature and origins. Materials Science and
757 Technology. 2001;17:366-75.
- 758 [29] Wehner T, Fatemi A. Effects of mean stress on fatigue behaviour of a hardened carbon steel.
759 International Journal of Fatigue. 1991;13:241-8.
- 760 [30] Krupp U. Fatigue crack propagation in metals and alloys: microstructural aspects and modelling
761 concepts: John Wiley & Sons; 2007.
- 762 [31] Campbell FC. Elements of metallurgy and engineering alloys: ASM International; 2008.
- 763 [32] Koh SK, Stephens RI. Mean Stress Effects on Low Cycle Fatigue for a High Strength Steel. Fatigue
764 & Fracture of Engineering Materials & Structures. 1991;14:413-28.

765