Corrosion-fatigue cracking behavior of weld clad tubes

Shane A. (Allen) Para
Lehigh University

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Corrosion-Fatigue Cracking Behavior of Weld Clad Tubes

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CORROSION-FATIGUE CRACKING BEHAVIOR OF WELD CLAD TUBES

by:

Shane A. Para

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Arnold Marder (Thesis Advisor)

(John) DuPont (Thesis Co-advisor)

Slade Cargill (Chairperson of Department)
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ABSTRACT

Circumferential Cracking has been a problem in supercritical coal-fired boilers for several years. The rapid cracking reduces the effectiveness of the corrosion resistant claddings that put on boiler tube surfaces for corrosion protection. This research was conducted to gain a better understanding of the circumferential cracking mechanism and ultimately perform a remaining life analysis based on a laboratory simulation. The laboratory simulation performed in this study successfully reproduced circumferential corrosion fatigue cracks on weld clad tubes. Many similarities were noted between simulation and service induced corrosion-fatigue cracking. Cracks initiated in the alloy 309 and alloy 625 weld claddings most readily. The crack propagation behavior in alloy 309 clad tubes was found to be the most severe circumferential cracking in the simulation, while alloy 622 and 625 experienced similar crack propagation behavior. Finite element modeling revealed that the alloy 309 coating was in a cyclic plastic strain condition, while cracks in both alloy 622 and 625 were propagating under cyclic elastic stresses. A remaining life assessment was performed based on the crack propagation results from the laboratory simulation and it was found to predict a non-conservative estimate of remaining boiler tube life.
1. INTRODUCTION

Weld clad tubes used in the waterwalls of supercritical coal-fired boilers have recently been experiencing circumferential cracking after a few years of operation. A similar circumferential cracking problem was found in low-alloy steel tubes many years before weld claddings were used to combat a change in operating environment. The underlying mechanism for circumferential crack initiation and propagation was previously studied and determined for both low-alloy steel and weld clad boiler tubes. The general mechanism of circumferential cracking, regardless of the environment or materials used, can be summarized. Crack initiation occurs at preferentially corroded sites in the microstructure which serve as stress concentrators. Upon fracture of the corrosion-product due to some service applied stress, fresh corrosive species attacks the underlying unprotected metal causing further corrosion, which results in material that is susceptible to fracture. This process continues and cracks propagate farther into the boiler tubes.

While the underlying mechanism of circumferential cracking has been well documented, little work has been done to understand the differences in crack initiation and propagation in different weld claddings. The cracking behavior of several commercially available weld claddings will be understood by considering a two-part corrosion fatigue simulation. The objective of this research is to gain a better understanding of crack initiation and propagation behavior in three different weld claddings. The differences in crack initiation and growth behavior between the different coatings studied will be analyzed. Based upon the results of the thermal
fatigue simulation a remaining life estimate will be made for tubes operating in service conditions.
2. BACKGROUND

A review of the current literature that relates to corrosion fatigue, thermal fatigue, and circumferential cracking in boiler tubes has been conducted to determine the important factors relating to evaluation of the remaining life of circumferentially cracked weld overlay coatings.

2.1. CORROSION FATIGUE CRACKING

Corrosion fatigue cracking (CFC) is a mechanism that describes both the initiation and propagation of a crack due to the combined effects of a corrosive environment and a cyclic stress. It has been determined that CFC is governed by interactions between loading, metallurgical, and environmental factors. Unlike environmentally assisted cracking (such as stress corrosion cracking), corrosion fatigue will generally occur in any alloy exposed to both a corrosive environment and a cyclic stress.

The loading factors, such as stress intensity range (ΔK), loading frequency, and stress ratio, all have significant effects on the CFC behavior of a material. The stress intensity (K) serves to define the magnitude of the crack-tip stress field. The stress intensity range, defined as: ΔK=K_{Max} - K_{Min}, traditionally controls the rate of crack propagation for fatigue controlled primarily by elastic stresses. Generally, as the stress intensity range decreases, the rate of crack propagation decreases; as seen on a da/dn vs. ΔK curve (Figure 1). The cyclic stress intensity range can determine whether environment has a significant effect on the cracking behavior. Classically,
at low stress intensity ranges, relatively slow crack propagation is expected which permits more time for interaction between the corrosive environment and the material. At high stress intensity ranges, the mechanical driving force for the crack becomes dominant, and environment may not have time to interact with the material and thus may not play as important of a role. Stress ratio, another important loading parameter, is defined as the ratio of minimum applied stress to the maximum applied stress. The stress ratio provides a measure of the magnitude of the stresses in fatigue. While the stress ratio defines the magnitude of the stresses, it does not affect the stress intensity range. Rates of corrosion fatigue tend to increase with higher stress ratios, as there are higher stresses driving crack propagation. Increased stress ratios cause an increase in crack tip strain that accelerates film/corrosion product rupture.

The loading frequency tends to have the most dominant effect in determining the importance of the environment in corrosion fatigue cracking. As loading frequency has no effect on the stress intensity range or stress ratio, its primary significance is that it controls the time which the environment can interact with the material. When loading frequency becomes sufficiently high, crack growth is dominated by fatigue and the environment has little time to interact with the material. In comparison, low frequency loading will substantially increase the material’s exposure time to the corrosive environment between mechanical cycles and the thus the environment will have a significant effect on the fatigue behavior.

Metallurgical features such as compositional variation, surface finish, and microstructure can have dramatic effects on crack initiation and propagation.
behavior. Any variation, such as second phase or composition variation can corrode preferentially and serve as a crack nucleation site. Surface features, such as weld ripples, can serve as stress concentrators that may act as a crack nucleation site.

Environmental effects, similar to microstructural effects are very dependent on the particular environment/material combination, although, any corrosive species can cause corrosion fatigue. Highly corrosive atmospheres tend to reduce the time to crack initiation, because of the potential to form stress concentrators (i.e. corrosion pit, preferential corrosion, etc.). While corrosive environments reduce crack initiation time, these environments can have a variable effect on the crack propagation. The combined effect of corrosion and fatigue will typically increase fatigue crack growth rates by damaging material ahead of the crack tip and thus, making it more susceptible to fracture by the applied cyclic stresses. Corrosion can have the opposite effect on the crack growth rate. If the corrosion product blunts the growing crack front, the stress intensity at the crack tip can be decreased. This decrease in stress intensity at the crack tip will tend to decrease the fatigue crack growth rate. It is obvious then, that while the aggressive species has a detrimental effect on crack initiation, the crack propagation behavior can be affected in a non-predictable way.

A mechanism to describe the influence of the environment and crack closure on near-threshold crack propagation behavior has been described by Suresh et al. According to this model, near threshold crack growth rates in inert atmospheres are accelerated as compared to crack growth rates in moist or corrosive environments.
Moist or corrosive atmospheres form an oxide layer at the crack tip, which can lead to premature crack closure that reduces the effective stress intensity range. Because of the reduced stress intensity ranges due to crack closure, crack growth rates are decreased. This premature crack closure leads to threshold stress intensity values in corrosive environments that are higher than the threshold stress intensity values in inert atmospheres\(^5\). While this mechanism may be active, the corrosion may also damage the material ahead of the crack tip leading to accelerated crack growth. This suggests again that even in the threshold regime of crack growth the corrosive environment will have a non-predictable effect on the crack propagation behavior.

### 2.2. THERMAL FATIGUE

Thermal fatigue has been defined as the gradual deterioration and eventual cracking of a material by alternate heating and cooling during which free thermal expansion is partially or completely constrained\(^6\). Severity of damage accumulation due to thermal fatigue is dependent upon many operational and materials parameters\(^7\). Operational parameters of prime importance include heating/cooling cycle time, maximum temperature, and maximum temperature hold-time. The heating/cooling rates have a direct effect on the temperature gradients that exist in the material. For materials that are constrained due to adjoining metal of different temperature, the temperature gradients in the material have a large effect on the stresses that develop. Typically, a steeper temperature gradient will lead to higher stresses. The maximum temperature reached during thermal cycling is often considered the most important parameter, as yielding commonly occurs most readily at higher temperatures.
resulting in more plastic deformation. The time at maximum temperature is also important, as the time is increased, the possibility of temperature induced changes (phase transformations, corrosion, creep) in the material increases.

Key physical properties include thermal conductivity and coefficient of thermal expansion. Thermal fatigue resistance of a material tends to increase as the thermal conductivity increases. This behavior is seen because materials with a higher thermal conductivity have the ability to reduce temperature gradients that exist in the material. The reduced thermal gradients tend to reduce internal constraints and subsequently reduce stresses. The thermal fatigue resistance also tends to decrease as the coefficient of thermal expansion increases. For any non-uniform temperature distribution in a single phase material, a higher thermal expansion will result in higher stresses due to increased constraints. It is important to note though, that thermal fatigue resistance of any material depends on many factors not discussed here and cannot be determined by considering a few physical properties.

2.3. BOILER ENVIRONMENTS

The gaseous environment in coal-fired boilers has recently been an area of concern. Coal fired boilers once operated primarily in an oxidizing environment, where major emissions included SO₂ and NOₓ. When the boiler operated in a primarily oxidizing environment, low-alloy boiler tubes were typically used with no coatings. Due to emission regulations, the combustion environment has changed operating conditions to lessen toxic emissions. By using a staged combustion process, furnaces were able to cut the amount of SO₂ and NOₓ emissions. The staged
combustion relies on using an oxygen starved, reducing environment which leads to
the production of high amounts of CO and H₂S in the boiler. It is well known that
the reducing boiler environments cause much more harmful corrosion than the
oxidizing boiler environments. During more recent boiler operation in the reducing
environment low-alloy boiler tubes are typically used with some type of corrosion-
resistant coating (such as weld claddings or thermal spray coatings).

The coal-fired boilers not only operate in highly corrosive conditions, but also
under conditions of severe thermal cycling. Sources of thermal cycling in coal-fired
boilers include startup, rapid shutdown, load changes, slagging, natural deslagging,
and sootblowing. The sources of thermal cycling of primary interest are natural
deslagging and sootblowing. During boiler operation, slagging occurs when
unburned coal and flyash deposit onto the waterwalls of the boiler. As these deposits
form on the surface of the waterwall, the boiler tubes are shielded from the
combustion zone of the furnace. The slag acts as a thermal barrier coating thus
causing the boiler tubes to operate at a lower than normal temperature.

The natural deslagging process occurs when the slag that formed on the surface of
the tube spalls off of the waterwall. Figure 2 shows a typical thermal response due to
the natural deslagging typically causes an increase in the surface temperature of the
boiler tube¹. Figure 2 shows the measured temperature at several different locations
on a boiler tube including the fireside surface, the fireside webbing, the fireside inner
diameter, cold side webbing and the cold side surface. Typical increases in the
fireside surface temperature of the tube due the natural slagging/deslagging process
are approximately 22°C to 100°C above typical operating temperatures. This increase in hot side surface temperature typically occurs in less than one minute and during this time the temperature of the inner diameter of the hot side of the tube usually remains constant. This deslagging results in a high temperature gradient through the thickness of the boiler tube.

Occasionally, the slag-buildup does not naturally fall off of the waterwalls and it must be manually removed using soot blowers. Sootblowers spray a water, mist, or steam spray onto the tubes causing the slag to spall off. It has been reported that water sootblower operation causes the surface of the tubes to rapidly decrease in temperature as much as 300°C from the reported steady state operating temperatures of 490 to 555°C\textsuperscript{11}.

2.4. CIRCUMFERENTIAL CRACKING

2.4.1. SVCIED INDUCED CIRCUMFERENTIAL CRACKING IN LOW-ALLOY STEEL BOILER TUBES

As previously discussed, the corrosive atmosphere and the thermal cycling in a supercritical coal-fired boiler are very aggressive. It has been well documented that service exposed low-alloy boiler tubes (0.5Cr-0.5Mo and 2.25Cr-1Mo) in supercritical coal fired boilers experience both wastage and circumferential cracking. In a study by Smith et al\textsuperscript{12}, the mechanism of circumferential crack initiation and propagation was determined by characterization of service exposed boiler tube microstructures and analysis of service conditions. Circumferential cracks on low-alloy boiler tubes (Figure 3) are usually classified as being corrosion product filled
with a sharp dagger-like or wedge-shaped appearance\textsuperscript{1}. The corrosion products formed on the surface of the boiler tubes as well as the corrosion products formed within the crack both appeared to be a dual layer product. Corrosion products in the crack tip were analyzed using Wavelength Dispersive Spectroscopy (WDS) mapping (Figure 4). The elemental analysis revealed that an iron-rich hematite scale fills the bulk of the crack, while a thin sulfur rich corrosion-spine fills the center of the crack.

In addition to the cracking, Smith et al. noted intergranular corrosion occurring at the surface of the boiler tubes (Figure 5).

Smith et al\textsuperscript{1-12} suggest that intergranular corrosion in specifically oriented grain boundaries is the key to crack initiation in low-alloy boiler tubes. The crack initiation mechanism (Figure 6) proposed by Smith states that only grain boundaries oriented normal to the surface will lead to cracking, as only these grain boundaries coincide with maximum stress relief. These favorably oriented corroded grain boundaries act as stress concentrators, which upon thermal cycling, eventually fracture. Corrosion at grain boundaries that are not orthogonal to the surface does not serve as crack initiation sites and can result in material loss, called wastage.

Crack propagation, as proposed by Smith et al\textsuperscript{1}, occurs by a mechanism similar to crack initiation. A combined effect of cyclic stresses and corrosion lead to a brittle fracture of the corrosion scale. The characteristic corrosion spine is present along the entire length of the crack, extending to the fresh uncorroded metal. As the sulfur from the environment transports down the sulfur rich spine, reactive sulfur corrodes the metal. Upon application of some thermal stress cycle, the crack
propagates a small increment through the corrosion product. This sequence of events repeats with each thermal stress application, leading to continued crack propagation. It was noted that crescent-shaped, sulfide filled formations develop along some cracks (Figure 4). These formations may be locations where crack growth had temporarily arrested, suggesting that crack growth is a discontinuous process. Smith states that the initiation and propagation of circumferential cracks is from the combined effect of both corrosion and fatigue. Without the corrosion component, the problem becomes a simple fatigue problem and by removing the thermal stresses, the problem becomes solely a corrosion problem. It is emphasized that to completely model this system, both corrosion and fatigue must be taken into account.

2.4.2. SIMULATION INDUCED CIRCUMFERENTIAL CRACKING IN LOW-ALLOY STEEL TUBES

In related work, Smith et al conducted a laboratory simulation to model the circumferential cracking seen in service. During this simulation a low alloy boiler tube (0.5%Cr-0.5%Mo) was exposed to a synthetic flue gas at 600°C for 100 hours. The tubes were then thermally cycled between 310°C and 525°C +/- 10°C for a total of 14,282 cycles. The tubes were removed from thermal cycling after 7,000 cycles to undergo another exposure to the corrosive atmosphere, to corrode any freshly exposed metal surfaces. After exposure, the tubes were returned to thermal cycling. Finite element analysis showed that the stress cycle during the simulation led to compressive yielding at the surface and subsequent residual tensile stresses.
Comparison of the service induced cracks and simulation induced cracks showed remarkable similarities (Figure 7). Cracks that formed during the simulation propagated perpendicular to the tube surface, possess a transgranular crack path, are corrosion product filled, and possess the iron sulfide spine through the crack center. Minor differences between service induced cracks and simulation induced cracks were attributed to the fact that the corrosive atmosphere and the thermal cycles were not combined. Smith suggested that this simulation might prove valuable as an accelerated test for ranking materials.$^{14}$
2.4.3. SERVICE INDUCED CRACKING IN ALLOY 625 WELD CLADDINGS

Since the changes in the combustion environment, alloy 625 and several other weld overlays have been used as a coating to protect the boiler tubes from accelerated corrosion. While the 625 weld clad tubes protect well against general corrosion and wastage, the tubes seem especially prone to circumferential cracking. Luer et al\textsuperscript{2}, characterized service exposed alloy 625 weld clad boiler tubes that had operated in a supercritical coal-fired boiler. Visual examination of the boiler tube showed evidence of a weld ripple, suggesting negligible general corrosion had occurred. While insignificant general corrosion was observed, severe circumferential cracks were found on the tubes as shown in Figure 8. Circumferential cracks in the weld overlay coatings were similar to circumferential cracks in the low-alloy steel tubes in that the cracks tended to grow perpendicular to the surface, cracks were wedge-shaped and corrosion-product filled with a second phase corrosion-product present through the center of the crack. Figure 9 shows the crack propagation tended to follow the main dendrite axis. Compositional mapping (Figure 10) revealed a chromium rich corrosion product occupying the bulk of the crack, with a thin nickel sulfide-type spine present through the center of the crack. All cracks that were examined were found to have both the primary chromium rich corrosion product as well as the nickel sulfide-type spine extending to the crack tip. Luer et al\textsuperscript{2} proposed a model for circumferential cracking of alloy 625 weld clad boiler tubes that is similar to the aforementioned model proposed by Smith et al\textsuperscript{1,12,15}. The model (Figure 11) proposes
a corrosion-fatigue cracking of a multiphase corrosion scale that forms at alloy depleted dendrite cores. This model predicts a two-layer corrosion scale will form on the surface of the metal, where molybdenum and niobium depleted dendrite cores will corrode preferentially. At some critical time, when the stresses reach a critical value, the corrosion scale will fracture, causing a NiS type spine to form through the center of the Cr$_2$O$_3$ type scale (Figure 10). This non-protective sulfide spine phase allows corrosive species to transport to the crack tip and attack fresh weld metal. As this process repeats, cracks will incrementally increase in length. It has been noted that cracks tend to grow the furthest in regions of macroscopic stress concentrators, such as weld ripples and weld defects.
3. LABORATORY SIMULATION OF CORROSION-FATIGUE CRACKING

3.1. INTRODUCTION

To better understand the circumferential cracking or corrosion fatigue cracking found in weld clad boiler tubes in supercritical coal-fired boilers, a laboratory simulation was developed to study corrosion-fatigue crack growth behavior in three commercially available weld claddings. The laboratory simulation developed for this study is a two-part simulation which includes a corrosion exposure followed by a thermal fatigue exposure. Due to testing limitations, application of a simulated combustion environment (corrosive atmosphere) and thermal fatigue could not be superimposed. Two separate thermal fatigue simulations were performed to consider the effect of different sources of thermal cycling in service.

3.2. EXPERIMENTAL PROCEDURE

3.2.1. MATERIALS SELECTION AND WELD DEPOSITION

The weld overlays examined in this study were deposited on SAE213-T12 low alloy steel (2.25Cr-1Mo) tubes. The overlays used were nickel based Inconel 625 and 622 and iron based 309 stainless steel. The weld clad tubes were made by Welding Services Inc. by depositing a gas metal arc weld (GMAW) cladding on the surface of a single SAE213-T12 tube (Figure 12). A series of overlapping weld beads were deposited longitudinally along the length of the tube, covering the entire tube surface. The welds were deposited in the vertical position, using downhill progression in 12-14” lengths (Figure 13). Water backing was used to reduce
distortions. The shielding gas used for depositing the nickel base alloys and the iron base alloy was 100% argon and 98% argon-2% carbon dioxide, respectively. Typical welding parameters used for depositing these alloys were current of 250-280 Amps, voltage of 28-33 Volts, and travel speed of 25-35 inches per minute. The final dimensions of the tube are approximately 1 5/8” outside diameter x 1 1/16” inside diameter.

The compositions of the weld overlay coatings were measured using Optical Emission Spectroscopy (OES). Prior to measuring the composition, tubes surfaces were ground flat. OES was performed on the flat surface which contained only weld metal. The compositions of the weld overlay coatings are listed in (Table I).

3.2.2. CORROSION SIMULATION

The corrosion simulation was designed to produce corrosion products comparable to that seen in service, including the microscopic stress concentration sites that may form from preferential corrosion. The corrosion simulation was performed on all samples prior to running both the natural slagging/deslagging simulation and the soot-blowing thermal fatigue simulations. The gaseous corrosion simulation was performed in an environment similar to that seen in coal-fired boilers. The corrosive gas composition used for testing can be found in (Table II).

Corrosion exposures were performed using a Lindberg Horizontal Tube furnace that is continuously supplied with the corrosive gas at a controlled rate. The gas was supplied without water vapor. Water is directly injected into the hot zone of the furnace at a controlled rate during testing at which point it vaporizes and mixes
with the corrosive gas. A schema of the experimental setup is shown in (Figure 14). Prior to exposure in the corrosive gas, all samples were cleaned ultrasonically in acetone for at least 15 minutes to ensure a clean-oil free surface. Before testing, the tube furnaces are repeatedly purged and backfilled with ultrahigh purity nitrogen six times to ensure an impurity free test environment. After cascade pumping, the corrosive gas is passed through the furnace for one hour prior to heating. During the initial stage of testing, the furnace is heated at a rate of 50°C/min. Once the testing temperature of 500°C is reached, the temperature is held constant for 100 hours. Following the 100 hours corrosion exposure, the furnace is cooled at a rate of less than 5°C/min.

During the pre-corrosion for the natural slagging/deslagging simulation, the ends of the weld clad tubes were masked to prevent corrosion of the low-alloy steel inner tube surface using a high temperature metallic adhesive and 304L stainless steel sheet. During the pre-corrosion for the sootblowing simulation, the thermal fatigue samples were completely exposed to the corrosive atmosphere.

### 3.2.3. THERMAL FATIGUE SIMULATION

#### 3.2.3.1. NATURAL SLAGGING/DESLAGGING SIMULATION

The natural slagging/deslagging thermal fatigue simulation was designed to simulate thermal stresses that exist due to the rapid heating of the tubes due to a natural slag fall. The simulation was modeled after similar work performed by Smith et al\textsuperscript{13}. The thermal fatigue simulation was performed on a 20kW Lepel-Solid State induction furnace that was modified for rapid thermal cycling. A schematic diagram
of the experimental set-up can be seen in Figure 15. The surfaces of two pre-corroded tubes were simultaneously thermally cycled between 400°C +/- 10°C and 30°C +/- 5°C. Heating to 400°C was accomplished in 12 seconds with no hold time at 400°C, after which the tubes were cooled to 30°C in 30 seconds (Figure 16). During thermal cycling, the temperature was measured on the external tube surface using a non-contact infrared pyrometer and a thermocouple in intimate contact with the surface. The cooling was facilitated by forced convection of water flowing through the centers of the tubes. Water was supplied to the tubes at a constant rate of 7 gallons per minute and maintained at an inlet temperature of 15.5°C during both heating and cooling.

Due to the nature of high-frequency induction heating, a thin layer is heated at the surface of the weld overlay which causes the surface to increase in temperature more rapidly than the underlying metal. This induction heating effect, coupled with the forced water cooling through the centers of the tubes maximizes the temperature gradient through thickness.

After thermal cycling the weld clad tubes were inspected using stereomicroscopy and light optical microscopy.

3.2.3.2. SOOT BLOWING SIMULATION

The sootblowing simulation was designed to simulate thermal stresses that exist due to the rapid surface quench that occurs during soot-blower operation. The thermal fatigue simulation is performed on a Gleeble 1000 HAZ Thermo-Mechanical Simulator. The Gleeble is ideal for simultaneous application of thermal and
mechanical loads to a test coupon. The sample is heated using electrical current passing through the sample, while a hydraulic system is used to provide mechanical loading. The heating and cooling of the sample are controlled using a continuous feedback system, precisely following the programmed thermal and mechanical loading. A typical sample used for the soot-blowing simulation is shown in Figure 17. The samples were longitudinally sectioned from the weld-clad tubes and machined flat on the inner diameter surface. The samples were made such that the final weld pass deposited on the tube was used as the center of the Gleeble samples. The ends of the sample were machined to suitable dimensions to be gripped in the Gleeble Thermo-Mechanical Simulator. The experimental setup showing the loading grips, the sample location, and the quench heads are shown in Figure 18.

The thermal cycle used for the thermal fatigue simulation was determined based on operating conditions from supercritical coal-fired boilers. In service, water or steam deslagging results in a rapid surface quench (0.1-10 seconds)\textsuperscript{11,17} on the surface of the weld cladding on the boiler tubes. Soot blowing conditions in a boiler can cause the surface of the boiler tubes to rapidly decrease temperature by up to 100°C to 300°C\textsuperscript{11,17} from the steady state operating conditions. For the simulation, a more severe quench was used where the surface of the weld cladding was quenched from a uniform steady-state temperature of 600°C +/- 2°C to 120°C +/-20°C in 2 seconds. The temperature is measured using a thermocouple that is percussion welded directly to the weld clad surface of the sample. The sample is ramped to 600°C +/-2°C in approximately 25 seconds, after which it is held at 600°C +/-2°C for a
period of 10 seconds. Immediately following the 10-second hold time, the weld-clad face of the sample is sprayed with a water mist for one second. The quench causes the weld-clad face of the sample to decrease in temperature to 120°C +/-20°C in approximately 2 seconds. A typical measured heating and cooling response during the simulation is shown in Figure 19. During thermal cycling, the temperature of the backside face decreases approximately 50°C in the 5 seconds after the quench. Based on the rapid quench, it is speculated that the decrease in temperature on the backface of the sample is most likely due to mist which contacts the backside due to the quench, not from heat loss through the sample. The thermal cycling process is repeated for a predetermined number of cycles: 1000, 2000, 5000, and 10000. During the entire thermal loading cycle the sample was allowed to contract or expand, but a constant tensile load of 500kg is applied to the sample.

Following the thermal fatigue simulation, crack depth and spacing was measured by using light optical microscopy. Each sample was inspected on five longitudinal planes through the final weld pass of each sample. The orientation and location of the inspected longitudinal planes are shown in Figure 20.

3.2.4. MICROSTRUCTURE AND CRACK CHARACTERIZATION

Samples were examined prior to any simulation, after the corrosion simulation, and after the thermal fatigue simulation using Light Optical Microscopy (LOM) and Scanning Electron Microscopy (SEM). The nickel based alloys were electrolytically etched in 10% chromic acid – 90% water (2 volts for 2-10 seconds) and the iron based alloy was etched in 60% nitric acid-40% water (1.5 volts for 2-10
The underlying low-alloy steel base metal was etched using 4% picral. SEM was performed using a JEOL 6300F scanning electron microscope equipped with a light element detector which alloys for detection of elements as light as oxygen. The uncorroded alloys were imaged at an accelerating voltage of 5 KeV. Imaging and EDS was performed on all corroded and cracked samples at an accelerating voltage of 15KeV.

3.3. RESULTS AND DISCUSSION

3.3.1. PRE-SIMULATION CHARACTERIZATION

The microstructure of the as-received tubes was studied to provide a baseline to understand any changes that occur during the pre-corrosion or thermal fatigue simulation. The general microstructure of the SAE231-T12 base metal was found to consist of a ferritic matrix with a pearlitic/bainitic second phase as shown in Figure 21. In the SAE231-T12 base metal, near the fusion line, a heat affected zone exists where rapid heating and cooling due to welding led to the formation of lath martensite or bainite (Figure 22). Microhardness traces were taken across the fusion line interface for all alloys and a dramatic increase in hardness was apparent (Figure 23). The increase in hardness in the base metal near the fusion line interface was similar for all three weld overlay coatings.

The weld substructures of all of the weld claddings are cellular/columnar dendritic, where the cells/dendrites tend to be aligned perpendicular to the base metal. The alloy 625 overlay microstructure is shown in Figure 24. The primary phase is $\gamma$-austenite, with terminal solidification products in the interdendritic region. The
terminal solidification products (Figure 25) were not studied in detail, but likely contain carbides and the Laves phase\textsuperscript{18,19}. Previous EPMA characterization of alloy 625 welds\textsuperscript{2} revealed that the dendrite cores are depleted in niobium and molybdenum, and slightly enriched in Fe, Ni, and Cr (Figure 26). The 622 overlay microstructure is shown in Figure 27. The primary phase is $\gamma$-austenite, with ternary solidification products in the interdendritic region (Figure 28). The ternary solidification products in alloy 622 welds, were characterized in previous work\textsuperscript{20,21} and determined to consist of several different Mo-Ni-Cr rich intermetallics ($\mu$, $\rho$, $\sigma$). The microsegregation behavior of alloy 622 has been evaluated\textsuperscript{22}, and it has been shown that dendrite cores are depleted in Mo and Cr, while the indendritic regions are enriched with Fe and Ni (Figure 29). The 309 stainless steel coating microstructure has a $\gamma$-austenite matrix with possible traces of ferrite (Figure 30). The composition of the 309 weld cladding was plotted on the WRC-1992 diagram\textsuperscript{23}, and the primary solidification mode was predicted to be austenitic with less than 2% ferrite present in the final weld structure. Previous analysis of 309 stainless steel welds with a two-phase structure has shown that the austenite in the dendrite core is depleted in Cr, while the ferrite rich interdendritic region is Cr rich (Figure 31)\textsuperscript{16}.

3.3.2. CORROSION SIMULATION

The resultant microstructures from the corrosion simulation were characterized to understand any changes that occur during thermal cycling. After the corrosion simulation, macroscopic examination of all samples revealed a non-uniform
gray corrosion product on the surface (Figure 32). The external scale was adherent and did not spall off during handling or during the thermal fatigue simulation. When the samples were viewed in cross section, all of the weld claddings were found to form a two-layer corrosion product consisting of both an internal and external corrosion layer, where the internal corrosion layer penetrates into the weld cladding and the external corrosion layer grows above the original metal surface of the cladding (Figure 33 – Figure 35). Elemental analysis was performed on each of the corrosion product layers using Energy Dispersive Spectroscopy (EDS). It is important to note to some limitations of using EDS to characterize corrosion scales. First, EDS cannot be used to determine the stoicheometry of the corrosion products, but rather it can only be used for identification of the elements present. Additionally, if chromium is potentially present in the corrosion scale, it is impossible to positively identify oxygen in the scale due the overlap of the chromium L\textit{alpha} and oxygen K\textit{alpha} peaks at 0.573 and 0.525 keV, respectively. Similarly, the molybdenum L\textit{alpha} and the sulfur K\textit{alpha} peak overlap at 2.239 and 2.306 keV, respectively. Despite these limitations, EDS is a useful tool for understanding the corrosion scales.

The external corrosion layer on both alloy 625 and 622 weld claddings is composed of primarily nickel, sulfur (or molybdenum), and iron rich corrosion products, while the internal corrosion layer is primarily rich in chromium and sulfur (or molybdenum). Figure 36 and Figure 37 show typical EDS results for the internal and external corrosion scales formed on alloy 625. The internal and external layers in alloy 625 and 622 may also contain oxygen, but positive identification was not
possible because the chromium and oxygen peak overlap. Based on previous research\textsuperscript{16}, it is most probably that the external layer on both alloy 625 and 622 are Ni-Fe-S type with only trace amounts of molybdenum while the internal layer is Cr-S-O type. Type 309 stainless steel has an external corrosion layer which is primarily iron and sulfur rich with traces of nickel and chromium. The internal layer on 309 stainless steel is chromium, iron, and sulfur rich. Oxygen may also be present in both the internal and external layer, but positive identification could not be made. Previous corrosion scale characterization on similar alloys exposed to the same corrosive environment revealed comparable results to those obtained in this study\textsuperscript{16}. Table III summarizes the elemental analysis performed on each corrosion layer and a comparison to the analysis performed by Luer\textsuperscript{16}.

An important feature of the internal corrosion scale is the morphology of the scale-metal interface. The morphology may have direct influence on the susceptibility of an alloy to initiate corrosion-fatigue cracks. For example, a uniform interface between the internal corrosion layer and the metal may not provide sufficient stress concentration sites for crack initiation while a non-planar interface may provide many more initiation sites, and lead to cracking under less severe stress conditions. The non-planar interface that develops in alloy 625 and to a lesser extent on alloy 622 is a result of preferential corrosion at alloy depleted regions of the microstructure. The microsegregation profiles (from previous studies)\textsuperscript{19} measured on alloy 625 and 622 gas tungsten arc weld claddings illustrates the microsegregation behavior (Figure 26 and Figure 29)\textsuperscript{16}. Alloy 625 has molybdenum and niobium
depleted dendrite cores while alloy 622 has only molybdenum depleted dendrite cores. It has been shown in previous research\textsuperscript{24} that the corrosion rate increases with decreasing molybdenum and niobium concentration in nickel based alloys exposed to sulfur bearing gases. The preferential corrosion in alloy 625 weld claddings was more severe than that in the alloy 622 claddings (Figure 33 and Figure 34). A highly non-planar internal corrosion layer-metal interface was present in the alloy 309 coating, but the corrosion rate did not appear to be enhanced at any specific microstructural site (Figure 38). Previous research though\textsuperscript{16}, has shown segregation behavior in alloy 309 gas metal arc welds and preferential corrosion in the Cr depleted austenite the when exposed to the same corrosive environments used in this study. The severity of the non-planar interface was most severe in alloy 309 and 625. The depth of internal corrosion penetration was also most severe in alloy 309 and 625.

3.3.3. NATURAL SLAGGING AND DESLAGGING SIMULATION

The natural slagging/deslagging simulation was performed on alloy 625 and 309 weld clad tubes for exposures as long as approximately 20,000 thermal cycles. The actual number of thermal cycles performed on each alloy was slightly different and is summarized in Table IV. After macroscopic and microscopic inspection of the weld clad tubes exposed to the thermal cycling, no evidence of cracking was observed. The lack of cracking implies that the thermal cycling used in this simulation did not generate severe enough stress to simulate service conditions.
While no stress analysis was performed for this simulation, it was speculated that the thermal stresses arising during the simulation were not severe enough to lead to the initiation of cracking. Previous work\textsuperscript{13} performed using a similar simulation resulted in circumferential cracking on a 0.5Cr-0.5Mo low alloy tubes in less than 14,000 cycles. The simulation in the previous study was performed using more severe thermal cycling conditions (heat to 525°C in less than 1 second, cool to 310°C in less than 1 minute). These loading conditions could not be reproduced in the current study due to equipment limitations. It is speculated that this difference in thermal loading resulted in less severe stresses in the current simulation, and accounted for the absence of crack initiation in the weld clad tubes.

### 3.3.4. SOOTBLOWING SIMULATION

The soot-blowing simulation was performed on 625, 622, and 309 weld clad samples for several different thermal fatigue exposures, which are summarized in Table V. After macroscopic and microscopic examination, cracking was observed in all coatings for all thermal fatigue exposures. The crack morphology, corrosion product composition, and the crack growth data were all characterized.

Developed circumferential cracks were inspected by viewing the cracks in cross section using light optical microscopy. The circumferential cracks in each of the three different weld claddings (Figure 39 – Figure 41) appear as wedge shaped cracks filled with corrosion product. An obvious feature of the cracking behavior is the seemingly random distribution of crack depth and spacing along the length of any given sample (Figure 42) which is similar to that which has been found in the
field. While the crack dimensions varied as a function of the number of applied thermal cycles and location along the length of any given sample, many observations can be made that were applicable to the cracks in all of the coatings. Cracks tended to propagate perpendicular to the surface of the weld overlay coatings (Figure 42), which would coincide with crack propagation normal to the maximum stress. Some cracks deviated from this surface normal growth direction to follow microstructural features such as the main axis of the dendrites, particularly in the 309 stainless steel weld cladding (Figure 41). This deviation from the surface normal is due to preferential corrosion in alloy depleted regions ahead of the crack tip which damage the material making it more susceptible to fracture.

The corrosion products that exist in the cracks that formed during the simulation were mapped using EDS. The EDS mapping provides insight into the changes that occur in the corrosion scale from the corrosion simulation during thermal cycling as well as the development of the corrosion product in the cracks. As mentioned previously, EDS is limited in its ability to distinguish between molybdenum and sulfur. While positive identification cannot be made between these two elements, the elemental information can be useful in understanding the corrosion products. Also, speculation about the elements present can be made based upon the distribution throughout the structure. EDS mapping was performed on the cracks in the alloy 625 weld cladding exposed to thermal cycles of 2000, 5000, and 10000 thermal cycles as shown Figure 43-Figure 45.
During the early stages of crack development (i.e. after 2000 thermal cycles), the corrosion products from the corrosion simulation are still intact (Figure 43). That is, the external corrosion layer is still nickel, iron, and sulfur (or molybdenum) rich while the internal corrosion layer is chromium and sulfur (or molybdenum) rich. The compositional analysis of the corrosion products on the thermally fatigued 625 weld cladding compare well with the point analysis performed on the as corroded samples prior to any thermal fatigue exposure. Resolution of the elemental distribution in the cracks from the shorter duration thermal fatigue exposures was difficult.

After the 5000 and 10000 cycle thermal fatigue exposures (Figure 44 and Figure 45), the distribution of alloying elements in the corrosion scales appears to be similar to the 2000 cycle thermal fatigue exposure. After 5000 and 10000 thermal cycles (Figure 44 and Figure 45), the external corrosion scales appear to contain less sulfur than after the 2000 cycle thermal fatigue exposure. After the longer thermal fatigue exposures a two-phase corrosion product was present in nearly all of the cracks. The bulk of the crack is occupied by a chromium and sulfur rich corrosion product which is depleted in nickel and iron, while the center of the crack contains corrosion spine which is a nickel and iron rich corrosion product which is depleted in chromium and sulfur. The cracks that formed in service conditions had a similar distribution of elements except for the sulfur. Cracks that formed in the boilers had sulfur present in both the bulk and spine corrosion phases. The difference in the elemental distributions can be explained by the lack of simultaneous corrosion and...
thermal fatigue exposure during the simulation which means that sulfur is not continuously replenished at the crack front.

The corrosion product occupying the bulk of the crack appears to be similar to the internal corrosion product layer, while the spine phase is similar to the external corrosion layer formed on the alloys. The formation of the two-phase corrosion product on the alloy surface as well as in the advancing crack is consistent with the circumferential cracking model proposed by Luer et al² for alloy 625. In the model proposed by Luer et al², it is suggested that when the crack opens due to some applied tensile stress, the crack acts as a free surface and a two-phase corrosion scale equilibrates on the surface. Figure 46 illustrates the mechanism where an external scale forms on the free surface created by the corroded-open crack. When the crack closes, the external scale on the crack face becomes the spine phase.

EDS mapping was also performed on several thermal fatigue cracks in the alloy 622 and 309 claddings after 10000 cycles. The elemental distributions in the cracks in the alloy 622 weld claddings are very similar to the elemental distribution in the cracks formed in the alloy 625 claddings (Figure 47). A chromium and sulfur rich corrosion product is present in the bulk of the crack with a thin nickel rich spine growing through the crack center. The elemental distribution in the cracks formed in the alloy 309 though is notably different (Figure 48). A chromium-nickel rich corrosion product exists around the edges of the crack while the bulk and center of the crack is occupied by an iron rich corrosion product. Sulfur is present in many of the cracks in the 309 stainless steel coating, but it is not clear which corrosion layer
contains the sulfur. Although the ‘spine’ phase in the cracks formed in the 309 coating are substantially different than that seen in alloy 625 or alloy 622, the speculation that the corrosion products in the crack are the same as the corrosion products on the surface of the alloy is still consistent.

It is important to reiterate the fact that during the thermal fatigue exposures, there was no superposition of thermal fatigue and a corrosive atmosphere. The cracks which formed though had many sulfur rich corrosion products present through the crack. During the 100 hour pre-corrosion simulation, a two-phase corrosion scale formed on the surface of samples. The two-phase corrosion product that formed on alloy 625 included an external corrosion layer which is primarily nickel, sulfur and iron rich while the underlying internal corrosion layer was primarily chromium and sulfur rich. Upon thermal fatigue exposure, the corrosion products are exposed to air at 600°C which is a highly oxidizing atmosphere. During thermal fatigue testing, the Ni-Fe-S and Cr-S rich corrosion products begin to transform into Ni-Fe-O and Cr-O rich corrosion products. As this is occurring, gaseous sulfur is released which is then free to continue to attack the metal at the advancing crack front. Similar behavior has been seen by Luer\textsuperscript{16} for FeS particles in contact with a steel substrate in oxidizing atmospheres, where FeS particles oxidize and release sulfur which attacks the steel substrate. This process of cyclic sulfidation followed by oxidation can repeat, thus sustaining a sulfidation type corrosion reaction resulting in cracks with sulfur present in the corrosion product.
3.3.5. COMPARISON OF SERVICE INDUCED AND SIMULATION INDUCED CORROSION FATIGUE CRACKING

The corrosion simulation was designed to simulate corrosive attack that occurs in supercritical coal fired boilers. As shown in Figure 49, the corrosion that occurs in alloy 625 weld clad tubes in service and in the simulation is remarkably similar. Preferential corrosion occurs in the alloy depleted dendrite cores in both service and simulation conditions. Both the severity of the preferential corrosion and depth of penetration of corrosion during the corrosion simulation are comparable to that seen in service exposed boiler tubes.

The cracks resulting from the thermal fatigue simulation also show many similarities to service induced circumferential cracks on alloy 625. As shown in Figure 50, cracks formed both in service and in the simulation tend to have a wedge-shaped morphology and propagate down the main dendrite axis. While the distribution of elements in the corrosion products was not identical, some similarities did exist. The bulk of the cracks form a chromium and sulfur rich product, while a nickel rich spine forms through the center of the crack (Figure 51).
Primary differences between the service induced and simulation induced cracks are the width of the cracks and the lack of sulfur in spine of the simulation induced cracks. These differences in the crack morphology and elemental distribution can both be explained by the singularity of the simulations. The lack of simultaneous corrosion and the differential in exposure time due to the accelerated testing prevents crack widening due to corrosion and eliminates the constant supply of sulfur to the corrosion products in the crack.

3.3.6. ANALYSIS OF THERMAL FATIGUE CRACK PROPAGATION BEHAVIOR

The crack depths, spacing, and number of cracks were considered to gain an understanding of crack initiation and growth behavior. Crack distributions were not measured in-situ, and as such crack propagation behavior is extrapolated from several ‘snapshots’ of the cracking, where each thermal fatigue exposure was performed on a different sample. The total number of cracks increases as a function of cycles for both the alloy 625 and 309 coatings (Figure 52). An anomaly exists for alloy 622, where the total number of cracks in the sample exposed to 5000 cycles has less cracks than the sample exposed for 2000 cycles. This anomaly in number of cracks likely occurred due to some unknown difference in the sample (weld ripple, dilution, etc.) relative to the other 622 samples prior to the thermal fatigue exposure. After 1000 thermal cycles, alloy 625, 622, and 309 have formed 18, 4, and 41 cracks respectively. This suggests that the initiation of cracks in alloy 622 occurs less rapidly than in either alloy 625 or alloy 309. The lower number of cracks formed in
alloy 622 at 1000 cycles (relative to 625 and 309) is most likely due to the less severe non-uniform internal corrosion interface, which reduced the severity of the stress concentration sites leading to crack initiation. While crack initiation in the 309 weld cladding apparently occurred most readily, the total number of cracks formed in the 309 coating for longer thermal cycling exposures was significantly lower than that in the 625 and 622 coatings.

To understand the cracking, the maximum (absolute deepest crack) and the average crack length (the average of the deepest 5% of the cracks) for each sample were also considered and are shown plotted as a function of the number of thermal cycles (Figure 53 and Figure 54). Both the maximum and the average crack length were found to increase monotonically as a function of thermal cycles. Although the total number of cracks was less in the 309 coating (Figure 52), the severity of cracking (crack depth) was substantially worse than the cracking in both the 625 and 622 coatings (Figure 53 and Figure 54). Both alloy 625 and alloy 622 experienced very similar crack propagation behavior simply as a function of thermal cycles.

Due to the fact that crack length could not be measured continuously, crack extension per cycle could not be directly determined. The crack growth rate appears to remain constant for all crack propagation through 100000 cycles for each alloy. A linear fit trend line was plotted through the data points for each alloy for both the maximum crack length and the average crack length on the depth versus cycle curve (Figure 55 and Figure 56). The slope of the linear fit line represents the crack growth rate. The R^2 value, which can be described as a measure of the variability explained
by a regression model (where $R^2=1$ implies the regression is a perfect fit), was
determined for each linear fit. The $R^2$ value was greater than 0.9 for the linear fit on
all of the alloys for both the maximum and average crack length versus cycle plots
implying that the linear fit is a good fit for crack growth. The crack growth rates
that were extracted from the crack growth curves for the maximum and average crack
depths (Figure 53 and Figure 56) are listed in Table VI.

The development of the crack depth distributions were considered by plotting
histograms of the crack depth for the different thermal fatigue exposures of 1000,
2000, 5000 and 10000 cycles (Figure 57-Figure 59). While it is difficult to follow
trends for cracks of a particular size for any of the coatings, it is obvious that the size
and number of large cracks continues to increase as a function of thermal cycles for
all of the alloys. For the thermal fatigue exposures considered, no maximum crack
depth was reached where the rate of crack propagation decreased or where crack
growth ceased. No clear and consistent trends were obvious for the smaller cracks in
any of the alloys.
4. STRESS AND FRACTURE FINITE ELEMENT ANALYSIS OF THE THERMAL FATIGUE SIMULATION

4.1. INTRODUCTION

To better understand the cracking phenomena during the thermal fatigue laboratory simulation, a 2-D finite element analysis was performed. The modeling was performed at Lehigh University by Citirik and Nied\textsuperscript{25} to gain an understanding of the transient thermal stresses induced during thermal cycling. FEM stress analysis results are used to generate cyclic elastic stress intensity parameters for the cracks which formed during the simulation using FEM fracture analysis software. Modeling will be combined with crack distributions used to explain crack propagation behavior.

4.2. EXPERIMENTAL PROCEDURE

4.2.1. STRESS ANALYSIS

The 2D steady state and transient thermal and stress analyses were performed using ANSYS. The elements considered during the 2D FEM had the capacity for both elastic and plastic yielding. The extent of plastic yielding was determined based upon the Von Mises yielding criterion. The thermal stress solutions are obtained from ANSYS in sequence. The first analysis determines the temperature distributions within the model based upon input thermal boundary constraints. Subsequently, the nodal temperatures from the thermal analysis are directly input into the ANSYS structural analysis to determine the stresses and strains for each value of time based upon prescribed mechanical constraints.
The thermal and stress analyses that were performed considered the temperature dependence of many of the properties for each of the alloys. Values for each physical or mechanical property are specified at various temperatures and ANSYS interpolates between the specified values. The temperature dependence on the thermal conductivity, thermal expansion, and specific heat were considered, but the density was assumed to remain constant. The variation of the physical properties with temperature can be seen in Figure 60, Figure 61, and Figure 62\textsuperscript{26-30}. The tensile response of the coating and substrate materials was modeled using a bilinear stress strain curve in both tension and compression. The bilinear stress-strain curves used for modeling at various temperatures for each alloy can be found in Figure 63 - Figure 66. The tensile response for alloys 625 and was approximated by using the yield stress for alloy 625 welds\textsuperscript{30} combined with the strain hardening behavior for wrought alloy 625 at various temperatures. The tensile response for wrought alloy 625, 622, and 309 (vs. temperature) were determined from tensile tests performed at Allegheny Ludlum\textsuperscript{31} (except for the 625 yield stress). The tensile response used for the SAE213-T12 substrate was also approximated by a bilinear stress strain curve (Figure 66)\textsuperscript{12}.

The stress analysis performed on each coating-substrate combination can be considered as a two-part analysis. The two parts involved in the analysis are the calculation of welding residual stresses and the initial heat-up and steady-state thermal cycling.
Determination of Welding Residual Stresses: Welding residual stresses were approximated by considering the overlay at 800°C placed on the cool substrate at 25°C. The upper temperature of 800°C was chosen as the temperature above which the stress carried by the coating should be insignificant; above this temperature the yield strength is sufficiently low and strain hardening behavior is negligible. The coating and substrate are allowed to thermally equilibrate and the stresses are determined during this process. It is not necessary to consider the coating during the actual welding process, as the coating is liquid and unable to carry any load. Thermal stress solutions are obtained from two separate ANSYS analyses. The first analysis determines the temperature distributions in the model at various times. The calculated temperatures are then input in an ANSYS structural analysis to calculate the stresses and strains at any given times.

Determination of Stresses during Heating and Steady State Thermal Cycling: During the thermal analysis, the back surface (T12 side) of the model was held at 600°C during the entire thermal cycle based on the measured surface temperatures during cycling (Figure 19). The surface of the weld clad side of the model followed a programmed thermal cycle (Figure 67). The thermal cycle used for the FEM modeling was taken directly from thermal fatigue simulation data (Figure 16). Thermal stress solutions are obtained from two separate ANSYS analyses using the same method used to calculate welding residual stresses.
4.2.2. FRACTURE ANALYSIS

A two-dimensional linear elastic fracture analysis was performed on both the alloy 625 and alloy 622 coatings to determine the cyclic elastic stress intensity factors at the advancing crack front for the stress cycles previously calculated. The 2D fracture analysis model calculates the stress intensity parameter by superposition of the stresses calculated from the 2D FEM stress analysis with a specified crack distribution. The fracture analysis model considers the effects of neighboring cracks which may contribute to stress relief by making the coating more compliant. The cyclic elastic stress intensity parameter \( \Delta K = K_{\text{max}} - K_{\text{min}} \) is determined by calculating \( K_{\text{max}} \) at the peak of the quench when the tensile stress is at a maximum and \( K_{\text{min}} \) is calculated upon reheating when the stress is a minimum. If \( K_{\text{min}} \) is negative, then \( K_{\text{effective}} \) is 0, since compressive stresses or negative stress intensities do not contribute to crack propagation.

The crack distributions used for the finite element were taken directly from the crack distributions measured from the samples used for the laboratory simulation of the sootblowing thermal fatigue. The crack distributions modeled in the fracture analysis were different for each thermally fatigued sample, where the deepest crack and the two surrounding cracks on either side were considered and superimposed with the stress determined from the FEM stress analysis. The fracture analysis performed determines a \( \Delta K \) value that corresponds directly to each ‘snapshot’ of crack growth.
4.3. RESULTS AND DISCUSSION

4.3.1. STRESS ANALYSIS

The stress analysis performed considers several sources of stress such as welding, thermal expansion mismatch, and the water mist quench. The character and magnitude of stresses arising from each source varies depending on the particular coating-substrate combination. The stress analysis performed on the alloy 625, 622, and 309 weld claddings deposited on a SAE213-T12 will be described.

Alloy 622 Weld Cladding:

The residual stresses present through the thickness of the sample due to the arc welding process prior to any reheating or thermal cycling are shown in Figure 68. The coating yields in compression and the substrate yields in tension during the welding process resulting in nearly a uniform residual tensile stress of approximately 380 MPa in the coating and a compressive stress of -210 MPa in the substrate. After heating to 600°C, immediately prior to any water mist quench (Figure 68), the residual stresses from welding have decreased in magnitude to approximately 250 MPa and -150 MPa, respectively, due to the decrease in yield stress at higher temperature (Figure 64 and Figure 66). The residual stresses present in the coating are greater than the yield stress at each of the corresponding temperatures suggesting the yielding and strain hardening has occurred.

At the peak of the quench when the temperature of the surface has decreased to 120°C, the coating is in a tensile state of stress which is at maximum at the surface and decreases monotonically as a function of depth (Figure 69). The surface of the
coating reaches a maximum tensile stress of approximately 540 MPa. The tensile stress in the coating is significantly higher than the yield stress at 20°C due to the strain hardening that occurred during the first quench cycle. The substrate compressively yields during the quench cycle. After reheating to 600°C the coating is in residual compressive stress due to the tensile yielding that occurred during the quench and the substrate is under residual tensile stress due to the compressive yielding that occurred (Figure 69). Modeling of the thermal fatigue simulation was repeated for five cycles (Figure 70) and the stress range was found to remain constant. By considering the plastic strain at the surface (Figure 71) it was found that significant plastic yielding occurs only during the first cycle and all subsequent cycles remain primarily in the elastic regime.

Alloy 625 Weld Cladding:

While the magnitudes of the stresses in alloy 625 are slightly different than alloy 622, the stress distributions through the thickness of the sample are similar. Figure 72 shows the surface stresses during the thermal fatigue simulation on alloy 625. The maximum tensile stress on the surface during thermal cycling is approximately 520 MPa, which is significantly greater than the yield stress at 20°C due to strain hardening that occurred during the first thermal quench. Similar to alloy 622, significant plastic yielding occurred only during the first quench cycle.

309 Stainless Steel Coating: The residual stresses present in from the arc welding process through the thickness of the sample prior to any reheating or thermal cycling are shown in Figure 73. The coating yields in compression and the substrate yields in
tension during the welding process, resulting in a nearly uniform residual tensile stress of approximately 240 MPa in the coating and a nearly uniform compressive stress of -165 MPa in the substrate. After heating to 600°C, immediately prior to any water mist quench, the resulting stress distribution (Figure 73) is a nearly uniform compressive stress of -10 MPa at the surfaces of both the coating and substrate, while a high tensile stress of 130 MPa exists near the coating-substrate interface. This resulting stress distribution after reheating is significantly affected by the high mismatch in thermal expansion between the coating and substrate.

At the peak of the quench when the temperature of the surface has decreased to 120°C, the coating is in a tensile state of stress which is at maximum at the surface and decreases as a function of depth (Figure 74). The tensile stress at the surface, after the first thermal cycle is approximately 214 MPa. After reheating to 600°C the coating is in compression due to the tensile yielding that occurred during the quench and the substrate is under tension stress due to the compressive yielding that occurred (Figure 74). Due to the thermal expansion mismatch, the coating is put under additional compressive stress during the reheat which leads to compressive yielding upon reheating. The thermal fatigue was repeated for five cycles and the stress range during thermal cycling was found to monotonically increase (Figure 75). This continuing increase in stress after each quench was attributed to the compressive yielding that occurred at 600°C upon reheating. The compressive yielding occurs because of the combination of the compressive stress generated due to the mismatch in coefficient of thermal expansion and because of the significant decrease in the
yield strength of 309 stainless steel at elevated temperatures (Figure 65). The compressive yielding that occurs at 600°C results in a high plastic strain during each quench cycle (Figure 76). Due to the severe cyclic plastic strain in alloy 309, the crack propagation behavior could not be considered as a function of the cyclic elastic stress intensity parameter.

4.3.2. FRACTURE ANALYSIS

The crack distributions used for the fracture analysis were the same as the crack distributions measured directly from the thermal fatigue samples. The deepest crack from each thermal fatigue sample was the focus of the fracture analysis. The two adjacent cracks on either side of the deepest crack were also considered. Figure 77 schematically shows the distributions used for the fracture analysis. The actual crack distributions that were used are listed in Table VII and Table VIII for alloy 622 and 625, respectively. The cracks adjacent to the deepest crack were considered because they may contribute to stress relief, thus reducing the stress at the advancing front of the deepest crack.

The stress intensity values at the crack tip for all of the cracks considered during the fracture analysis are shown for alloy 622 in Figure 78 and for alloy 625 in Figure 79. The stress intensity values for the deepest cracks are listed in Table IX and Table X for alloy 622 and 625 respectively. The calculated stress intensities are for the peak of the quench when the tensile stress in the coating is a maximum. These calculated stress intensities correspond to $K_{\max}$. During the thermal cycling for both alloy 622 and 625, the coating is under compression during reheating and as such, the
The calculated $K$ values at the peak of the quench are therefore equal to the $\Delta K$ values for thermal cycling.

The results of the finite element fracture analysis performed on alloy 622 and 625 can be coupled with the crack growth rates to understand the range of stress intensities for which the crack growth rate remains constant. Due to the fact that the crack growth rates remain constant throughout the duration of the laboratory simulation tests, $da/dn$ versus $\Delta K$ curves provide little insight into the crack growth behavior. For alloy 622, the crack growth rate (for the maximum crack) was 0.03 $\mu$m/cycle which remained constant for all testing. The corresponding $\Delta K$ values where the crack growth rate was constant ranged from 5 MPa(m)$^{-1/2}$ to 10 MPa(m)$^{1/2}$. For alloy 625, the crack growth rate (for the maximum crack) was 0.03 $\mu$m/cycle which remained constant for all testing. The corresponding $\Delta K$ values where the crack growth rate was constant ranged from 6 MPa(m)$^{-1/2}$ to 9 MPa(m)$^{1/2}$.

The results of this study, including the crack growth rates from the corrosion-thermal fatigue simulation and the finite element modeling can be compared against data from isothermal fatigue crack propagation tests. Figure 81 shows a $da/dn$ versus $\Delta K$ curve for an isothermal fatigue test performed on wrought alloy 625 at 649°C. Also plotted on Figure 81 is the $da/dn$ versus $\Delta K$ determined for the thermal fatigue simulation on alloy 625. The range of $\Delta K$ which the isothermal test was performed did not extend over the range of $\Delta K$ that existed during the thermal fatigue. Figure 81 shows two potential extrapolations of the isothermal fatigue curve, where the crack propagation either remains constant (as in the Paris Regime) and also where threshold
ΔK value is reached and crack propagation rate rapidly decreases. These two logical extrapolations of the da/dn curve bound the crack propagation behavior for the thermal fatigue test. This suggests that a high temperature isothermal fatigue test may provide a simple way to evaluate differences in crack propagation behavior. No data on high temperature isothermal fatigue of alloy 622 could be found and as such, no direct comparison could be made.
5. REMAINING LIFE ASSESSMENT

The corrosion-thermal fatigue simulation used in this study to determine crack propagation behavior for different commercially available weld claddings was designed to simulate circumferential cracking. By the singular nature of the simulation, that is the fact that corrosion and thermal fatigue were not superimposed, the simulation is not able to completely replicate the cracking phenomenon. The simulation though does provide a worst case scenario in terms of the stresses that the tubes are exposed to. The quench cycle used during the simulation was much more severe than that experienced in service conditions. Corrosion was experienced during the thermal fatigue simulation, but due to the accelerated nature of the test and the lack of the superimposed sulfidizing environment, the corrosion was not as severe as that seen in service.

For the approximation of remaining life, it is assumed that the simulation provides a crack growth rate that compares reasonably well with crack growth rates seen in service for circumferentially cracked boiler tubes exposed to the sootblowing process. Both alloy 625 and 622 experienced similar crack growth rates of 0.03 microns/cycle during the thermal fatigue simulation. It has been reported that sootblowing occurs approximately 24 times per day\(^1\). Typical cladding thicknesses found in weld clad boiler tubes are approximately 2.5mm\(^2\).
Assuming cracks have already initiated and are growing at a constant crack growth rate, the time until the circumferential crack propagate through the coating can be described by:

\[
\text{Time to Failure} = \frac{\text{Crack Length at Failure}}{\frac{da}{dn} \times \text{thermal cycles} \times \text{time}}
\]

The resulting time for a crack to propagate through a coating of 2.5mm, based on 24 thermal cycles/day is 3472 days or 9.5 years after crack initiation. It has been reported\(^2\) that cracks in an alloy 625 weld cladding have propagated as deep as 1500µm in less than two years. It is apparent that the remaining life assessment performed, based upon cycle counting and crack growth rates from the thermal fatigue simulation does not provide a conservative estimate for remaining life. The effect of corrosion in the boiler environment must have a dramatic effect on enhancing the crack propagation rates.
6. CONCLUSIONS

The objectives of this research were to gain a better understanding of the corrosion-
thermal fatigue crack initiation and growth behavior in commercially available weld
overlay claddings.

1.) A two-part corrosion-thermal fatigue simulation was developed and used to study
the cracking resistance of various weld claddings.

   a.) The corrosion simulation successfully reproduced the corrosion seen in
service including the morphology and location of preferential corrosion in
alloy depleted regions of the weld structure for alloy 625.

   b.) The natural slagging/deslagging thermal fatigue simulation was unable to
reproduce thermal fatigue cracking due to the lack of the ability to apply
severe thermal cycling (due to equipment limitations)

   c.) The sootblowing thermal fatigue simulation successfully reproduced
corrosion-fatigue cracks that are seen in coal-fired boilers. The cracks formed
in the simulation showed remarkable similarities to service induced cracking.
Simulation and service induced cracks formed at the non-planar corrosion
scale interface, propagated along the main dendrite axis, and were filled with a
corrosion product that have similar elemental distributions to service induced

   d.) The major differences between simulation and service induced cracking
were limited to the reduced amounts of sulfur present in the crack and the
reduced width of the simulation induced cracks. These differences can both
be accounted for by the lack of simultaneous corrosion and thermal cycling.
2.) Initiation of corrosion-fatigue cracks in the laboratory simulation occurred most readily in alloy 309 and 625. During the earlier stages of thermal cycling, the number of cracks formed in alloy 622 was substantially less than both 309 and 625.

3.) The severity of corrosion fatigue cracking in alloy 309 was significantly greater than both alloy 625 and 622.

4.) FEM was successfully performed to determine the stresses induced during the laboratory simulation for all coatings. FEM was also performed to determine the elastic stress intensity factors for cracks propagating in alloy 622 and 625 which allowed for a comparison of crack growth behavior to isothermal fatigue tests.

5.) Both alloy 625 and 622 experienced a constant crack propagation rates of 0.03 μm/cycle over a stress intensity range of 6-9 MPa(m)^{1/2} and 5-10 MPa(m)^{1/2}, respectively.

6.) A remaining life assessment performed for alloy 625 and 622 based upon cycle counting and a constant crack growth rate predicts a boiler tube life of 9.5 years. Based on service data, the 9.5 year estimate of remaining life is a non-conservative prediction.
Table I. Results of OEM compositional analysis for each of the overlay coatings (weight %)

<table>
<thead>
<tr>
<th>Element</th>
<th>625</th>
<th>622</th>
<th>309</th>
</tr>
</thead>
<tbody>
<tr>
<td>Ni</td>
<td>59.67</td>
<td>56.90</td>
<td>10.67</td>
</tr>
<tr>
<td>Cr</td>
<td>20.89</td>
<td>19.46</td>
<td>17.52</td>
</tr>
<tr>
<td>Mo</td>
<td>8.00</td>
<td>14.42</td>
<td>0.22</td>
</tr>
<tr>
<td>Fe</td>
<td>7.54</td>
<td>5.23</td>
<td>69.1</td>
</tr>
<tr>
<td>Nb</td>
<td>3.35</td>
<td>0.001</td>
<td>0.010</td>
</tr>
<tr>
<td>C</td>
<td>0.014</td>
<td>0.011</td>
<td>0.051</td>
</tr>
<tr>
<td>Co</td>
<td>0.019</td>
<td>0.042</td>
<td>0.063</td>
</tr>
<tr>
<td>Cu</td>
<td>0.016</td>
<td>0.085</td>
<td>0.081</td>
</tr>
<tr>
<td>Mn</td>
<td>0.046</td>
<td>0.24</td>
<td>1.38</td>
</tr>
<tr>
<td>P</td>
<td>0.001</td>
<td>0.004</td>
<td>0.015</td>
</tr>
<tr>
<td>S</td>
<td>0.001</td>
<td>0.001</td>
<td>0.003</td>
</tr>
<tr>
<td>Si</td>
<td>0.066</td>
<td>0.099</td>
<td>0.70</td>
</tr>
<tr>
<td>Ti</td>
<td>0.16</td>
<td>0.050</td>
<td>0.006</td>
</tr>
<tr>
<td>V</td>
<td>0.007</td>
<td>0.021</td>
<td>0.069</td>
</tr>
<tr>
<td>W</td>
<td>0.016</td>
<td>3.05</td>
<td>0.006</td>
</tr>
</tbody>
</table>

Table II. Composition of the corrosive gas used for the corrosion simulation (Volume %)

<table>
<thead>
<tr>
<th>Gas Species</th>
<th></th>
</tr>
</thead>
<tbody>
<tr>
<td>CO</td>
<td>10</td>
</tr>
<tr>
<td>CO₂</td>
<td>5</td>
</tr>
<tr>
<td>H₂O</td>
<td>2</td>
</tr>
<tr>
<td>H₂S</td>
<td>0.12</td>
</tr>
<tr>
<td>N₂</td>
<td>balance</td>
</tr>
</tbody>
</table>
Table III. Summary of the compositional analysis performed on the various alloys exposed to the corrosion simulation and a comparison to a previous study on the similar alloys

<table>
<thead>
<tr>
<th>Coating</th>
<th>Internal Layer (current study)</th>
<th>Internal Layer (previous study)</th>
<th>External Layer (current study)</th>
<th>External Layer (previous study)</th>
</tr>
</thead>
<tbody>
<tr>
<td>625</td>
<td>Cr, (S,Mo), (Cr, O)</td>
<td>Not performed</td>
<td>Ni, S, Fe</td>
<td>Not performed</td>
</tr>
<tr>
<td>622</td>
<td>Cr, (S,Mo), Fe, Ni, (Cr, O)</td>
<td>Cr, Mo, O, S</td>
<td>Ni, S, Fe</td>
<td>Ni3S2-Ni7S6 +Fe, Cr, Mo, W</td>
</tr>
<tr>
<td>309</td>
<td>Cr, (S, Mo), Fe, Ni, (Cr, O)</td>
<td>Cr, O, S</td>
<td>Fe, S</td>
<td>Fe1-xS-(Fe,Ni)9S8</td>
</tr>
</tbody>
</table>

Table IV. Summary of thermal fatigue cycles and crack formation behavior during the natural slagging/deslagging simulation

<table>
<thead>
<tr>
<th>Alloy</th>
<th>Cycles</th>
<th>Crack formation</th>
</tr>
</thead>
<tbody>
<tr>
<td>625</td>
<td>11920</td>
<td>No</td>
</tr>
<tr>
<td>625</td>
<td>19329</td>
<td>No</td>
</tr>
<tr>
<td>309</td>
<td>12313</td>
<td>No</td>
</tr>
<tr>
<td>309</td>
<td>24233</td>
<td>No</td>
</tr>
</tbody>
</table>
Table V. Summary of thermal fatigue cycles and crack formation behavior during the sootblowing thermal fatigue simulation

<table>
<thead>
<tr>
<th>Alloy</th>
<th>Number of Cycles</th>
</tr>
</thead>
<tbody>
<tr>
<td>625</td>
<td>1000</td>
</tr>
<tr>
<td>625</td>
<td>2000</td>
</tr>
<tr>
<td>625</td>
<td>5000</td>
</tr>
<tr>
<td>625</td>
<td>10000</td>
</tr>
<tr>
<td>622</td>
<td>1000</td>
</tr>
<tr>
<td>622</td>
<td>2000</td>
</tr>
<tr>
<td>622</td>
<td>4300</td>
</tr>
<tr>
<td>622</td>
<td>10000</td>
</tr>
<tr>
<td>309</td>
<td>1000</td>
</tr>
<tr>
<td>309</td>
<td>2000</td>
</tr>
<tr>
<td>309</td>
<td>3500</td>
</tr>
<tr>
<td>309</td>
<td>8250</td>
</tr>
</tbody>
</table>

Table VI. Crack growth rates for each alloy based upon a linear fit for cracks of the maximum depth and for the average depth

<table>
<thead>
<tr>
<th>Sample</th>
<th>Crack Growth Rate (Maximum) (μm/ cycle)</th>
<th>Crack Growth Rate (Average) (μm/ cycle)</th>
</tr>
</thead>
<tbody>
<tr>
<td>625</td>
<td>0.03</td>
<td>.01</td>
</tr>
<tr>
<td>622</td>
<td>0.03</td>
<td>.01</td>
</tr>
<tr>
<td>309</td>
<td>0.11</td>
<td>0.07</td>
</tr>
</tbody>
</table>
Table VII. Crack distributions measured on alloy 622 thermal fatigue simulations used for the finite element analysis (see Figure 77). All units are in microns.

<table>
<thead>
<tr>
<th>Crack #</th>
<th>622 10K</th>
<th>622 5K</th>
<th>622 2K</th>
<th>622 1K</th>
</tr>
</thead>
<tbody>
<tr>
<td>1- Depth Location</td>
<td>77</td>
<td>65</td>
<td>69</td>
<td></td>
</tr>
<tr>
<td>Location</td>
<td>0</td>
<td>0</td>
<td>0</td>
<td></td>
</tr>
<tr>
<td>2- Depth Location</td>
<td>105</td>
<td>106</td>
<td>30</td>
<td></td>
</tr>
<tr>
<td>Location</td>
<td>226</td>
<td>143</td>
<td>157</td>
<td></td>
</tr>
<tr>
<td>3- Depth Location</td>
<td>289</td>
<td>112</td>
<td>81</td>
<td>23.04</td>
</tr>
<tr>
<td>Location</td>
<td>446</td>
<td>513</td>
<td>398</td>
<td></td>
</tr>
<tr>
<td>4- Depth Location</td>
<td>116</td>
<td>97</td>
<td>54</td>
<td></td>
</tr>
<tr>
<td>Location</td>
<td>855</td>
<td>600</td>
<td>794</td>
<td></td>
</tr>
<tr>
<td>5- Depth Location</td>
<td>43</td>
<td>41</td>
<td>42</td>
<td></td>
</tr>
<tr>
<td>Location</td>
<td>927</td>
<td>735</td>
<td>1412</td>
<td></td>
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</table>

Table VIII. Crack distributions measured on alloy 625 thermal fatigue simulations used for the finite element analysis (see Figure 77). All units are in microns.

<table>
<thead>
<tr>
<th>Crack #</th>
<th>625 10K</th>
<th>625 5K</th>
<th>625 2K</th>
<th>625 1K</th>
</tr>
</thead>
<tbody>
<tr>
<td>1- Depth Location</td>
<td>46</td>
<td>46</td>
<td>39</td>
<td></td>
</tr>
<tr>
<td>Location</td>
<td>0</td>
<td>0</td>
<td>0</td>
<td></td>
</tr>
<tr>
<td>2- Depth Location</td>
<td>61</td>
<td>18</td>
<td>20</td>
<td></td>
</tr>
<tr>
<td>Location</td>
<td>298</td>
<td>80</td>
<td>321</td>
<td></td>
</tr>
<tr>
<td>3- Depth Location</td>
<td>318</td>
<td>108</td>
<td>94</td>
<td>59</td>
</tr>
<tr>
<td>Location</td>
<td>392</td>
<td>407</td>
<td>367</td>
<td>0</td>
</tr>
<tr>
<td>4- Depth Location</td>
<td>20</td>
<td>63</td>
<td>49</td>
<td>54</td>
</tr>
<tr>
<td>Location</td>
<td>617</td>
<td>668</td>
<td>972</td>
<td>274</td>
</tr>
<tr>
<td>5- Depth Location</td>
<td>38</td>
<td>41</td>
<td>90</td>
<td>34</td>
</tr>
<tr>
<td>Location</td>
<td>812</td>
<td>735</td>
<td>1412</td>
<td>424</td>
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</table>
Table IX. $\Delta K$ values that correspond to the deepest cracks for the specified crack distributions for alloy 622

<table>
<thead>
<tr>
<th>Sample (# of thermal exposures)</th>
<th>Max Depth ($\mu$m)</th>
<th>$\Delta K$ (MPa$\cdot$m$^{-1/2}$)</th>
</tr>
</thead>
<tbody>
<tr>
<td>10000</td>
<td>289</td>
<td>10.0</td>
</tr>
<tr>
<td>5000</td>
<td>112</td>
<td>6.5</td>
</tr>
<tr>
<td>2000</td>
<td>81</td>
<td>8.3</td>
</tr>
<tr>
<td>1000</td>
<td>23</td>
<td>5.0</td>
</tr>
</tbody>
</table>

Table X. $\Delta K$ values that correspond to the deepest cracks for the specified crack distributions for alloy 625

<table>
<thead>
<tr>
<th>Sample (# of thermal exposures)</th>
<th>Max Depth ($\mu$m)</th>
<th>$\Delta K$ (MPa$\cdot$m$^{-1/2}$)</th>
</tr>
</thead>
<tbody>
<tr>
<td>10000</td>
<td>318</td>
<td>9.3</td>
</tr>
<tr>
<td>5000</td>
<td>104</td>
<td>8.09</td>
</tr>
<tr>
<td>2000</td>
<td>94</td>
<td>8.55</td>
</tr>
<tr>
<td>1000</td>
<td>59</td>
<td>6.12</td>
</tr>
</tbody>
</table>
Figure 1. Typical da/dn vs. ΔK curve showing how crack propagation behavior generally varies as a function of the cyclic elastic stress intensity parameter.

Figure 2. Temperature response to natural deslagging and slagging. Each plot represents different thermocouple locations on the fireside and on the cold side of the boiler tubes.
Figure 3. Cross-sectional view of Circumferential Cracks formed on low-alloy steel (T11) boiler tubes removed from service\textsuperscript{12}

Figure 4. Elemental (WDS) analysis performed at the circumferential crack tip formed in a low-alloy steel (T11) boiler tube\textsuperscript{12}
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\textbf{Figure 3.} Cross-sectional view of Circumferential Cracks formed on low-alloy steel (T11) boiler tubes removed from service\textsuperscript{12}
Figure 5. Intergranular corrosion forming crack initiation sites on low-alloy steel (T11) boiler tubes

Figure 6. Crack initiation model showing that only preferential corrosion occurring at grain boundaries oriented normal to the surface result in cracking (A) cracks form and propagate, (B & C) no cracks will form
Figure 5. Intergranular corrosion forming crack initiation sites on low-alloy steel (T11) boiler tubes.

Figure 6. Crack initiation model showing that only preferential corrosion occurring at grain boundaries oriented normal to the surface result in cracking (A) cracks form and propagate, (B & C) no cracks will form.
Figure 7. Comparison of (A) Simulation Induced and (B) Service Induced Circumferential Cracking on a low-alloy steel boiler tube showing similarities and differences between the cracks.

Figure 8. Surface appearance of circumferential cracks formed on alloy 625 weld clad boiler tubes removed from service.
Figure 7. Comparison of (A) Simulation Induced and (B) Service Induced Circumferential Cracking on a low-alloy steel boiler tube showing similarities and differences between the cracks.

Figure 8. Surface appearance of circumferential cracks formed on alloy 625 weld clad boiler tubes removed from service.
Figure 9. Cross-sectional view of a Circumferential Crack formed in an alloy 625 weld clad boiler tube removed from service.
Figure 9. Cross-sectional view of a Circumferential Crack formed in an alloy 625 weld clad boiler tube removed from service.
Figure 10. Elemental (EDS) distribution at the circumferential crack tip formed in an alloy 625 weld clad boiler tube removed from service.
Figure 10. Elemental (EDS) distribution at the circumferential crack tip formed in an alloy 625 weld clad boiler tube removed from service.
Figure 11. Circumferential crack initiation model proposed for cracking on alloy 625

(1) fresh weld metal with compositionally segregated microstructure
(2-4) corrosion occurring at alloy depleted regions of the dendritic substructure
(3) fracture of the two-phase corrosion scale
(4) two-phase corrosion scale forming through the crack center
Figure 12. Typical Surface appearance of the weld clad tubes used for the thermal fatigue simulations. (A) Alloy 625 (B) Alloy 309

Figure 13. Welding setup used to deposit multipass longitudinal weld beads along the length of low-alloy steel tubes
Figure 12. Typical Surface appearance of the weld clad tubes used for the thermal fatigue simulations. (A) Alloy 625 (B) Alloy 309

Figure 13. Welding setup used to deposit multipass longitudinal weld beads along the length of low-alloy steel tubes
Figure 14. Schematic diagram of the experimental setup used for the corrosion simulation, showing the gas mixing and distribution system, the water injection system, and the corrosion chamber.

Figure 15. Schema of experimental setup used for natural slagging/deslagging simulation showing the 2-coil setup with continuous water supply. The surface temperature is continuously monitored using a non-contact infrared pyrometer.
Figure 14. Schematic diagram of the experimental setup used for the corrosion simulation, showing the gas mixing and distribution system, the water injection system, and the corrosion chamber.

Figure 15. Schema of experimental setup used for natural slagging/deslagging simulation showing the 2-coil setup with continuous water supply. The surface temperature is continuously monitored using a non-contact infrared pyrometer.
Figure 16. Measured thermal cycle on the surface of the tubes measured during the thermal fatigue natural slagging/deslagging simulation
Figure 17. Top and side view of the sample used for the thermal fatigue sootblowing simulation.

Figure 18. Experimental setup used for performing the thermal fatigue soot blowing simulation. The mechanical grips are used to apply a mechanical and electrical load to the sample. The quench head is used to spray the weld clad face of the sample with water mist.
The mechanical grips are used to apply a mechanical and electrical load to the sample. The quench head is used to spray the weld clad face of the sample with water mist.
Figure 19. Measured thermal cycles on the surface of the samples during the thermal fatigue sootblowing simulation

Figure 20. Sketch on a thermally cycled sample showing orientation of serial sectioned planes used for crack length and distribution measurements
Figure 19. Measured thermal cycles on the surface of the samples during the thermal fatigue short-cycling simulation.

Figure 20. Sketch on a thermally cycled sample showing orientation of serial sectioned planes used for crack length and distribution measurements.
Figure 21. Representative microstructure of T12 base metal taken at 1000x. Etched in 4% picral.

Figure 22. Representative microstructure of the T12 base metal near the weld fusion line taken at 1000x. Etched in 4% picral.
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VITA

Shane Allen Para was born on August 11, 1979 in Southbridge, MA to his parents Richard and Debra. He was raised in Massachusetts until sixth grade, at which point he moved to Louisiana. Shane grew up with his six brothers and sisters, Danny, Bridgette, Eric, Sean, Meghan, and Rachel. He graduated with honors in the spring of 1997 from East St. John High School in Reserve, Louisiana. Shane began his Lehigh career in the fall of 1997 as an engineering student. In spring of 2001, Shane graduated with a Bachelors of Science Degree in Materials Science and Engineering. He entered the Materials Science and Engineering Department as a graduate student in the fall of 2001.
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