Mechanism of Fatigue Crack Initiation in Austenitic Stainless Steels in Light Water Reactor Environments

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ABSTRACT

This paper examines the mechanism of fatigue crack initiation in austenitic stainless steels (SSs) in light water reactor (LWR) coolant environments. The effects of key material and loading variables on the fatigue lives of wrought and cast austenitic SSs in air and LWR environments have been evaluated. The influence of reactor coolant environments on the formation and growth of fatigue cracks in polished smooth SS specimens is discussed. The results indicate that the fatigue lives of these steels are decreased primarily by the effects of the environment on the growth of cracks <200 µm and, to a lesser extent, on enhanced growth rates of longer cracks. The fracture morphology in the specimens has been characterized. Exploratory fatigue tests were conducted to study the effects of surface micropits or minor differences in the surface oxide on fatigue crack initiation.

KEY WORDS: Fatigue crack initiation, fatigue life, light water reactor, austenitic stainless steel, strain amplitude, strain rate, microstructurally small crack, dissolved oxygen, electrochemical potential, conductivity, slip dissolution, oxide film, crack growth rate.

INTRODUCTION

Existing fatigue strain–vs.–life (ε–N) data illustrate potentially significant effects of light water reactor (LWR) coolant environments on the fatigue resistance of carbon and low–alloy steels [1-7] as well as of austenitic stainless steels (SS) [8-15]. The key parameters that influence fatigue life in LWR environments are temperature; dissolved–oxygen (DO) level in water; strain rate; strain (or stress) amplitude; and, for carbon and low–alloy steels, sulfur content in the steel. Under certain environmental and loading conditions, fatigue lives of carbon steels can be a factor of 70 lower in coolant environments than in air [3-5].

For carbon and low–alloy steels, environmental effects on fatigue life are significant in high–DO water (>0.04 ppm DO) and only moderate (less than a factor of 2 decrease in life) in low–DO water. The reduction in fatigue life of carbon and low–alloy steels in LWR environments has been explained by the slip oxidation/dissolution mechanism for crack advance [16]. The requirements for the model are that a strain increment occur to rupture the protective surface oxide film and thereby expose the underlying matrix to the environment; once the passive oxide film is ruptured, crack extension is controlled by dissolution of freshly exposed surfaces and their oxidation characteristics.

Unlike the case of carbon and low–alloy steels, environmental effects on the fatigue lives of austenitic SSs are significant in low–DO (i.e., <0.01 ppm DO) water; in high–DO water, environmental effects appear to be either comparable [12,13] or, in some cases, smaller [8] than those in low–DO water. These results are difficult to reconcile in terms of the slip oxidation/dissolution model.

This paper examines the mechanism of fatigue crack initiation in austenitic SSs in LWR coolant environments. The influence of reactor coolant environments on the formation and growth of fatigue cracks in polished smooth specimens is discussed. The growth of fatigue crack as a function of fatigue cycles was determined in water by block loading that leaves beach marks on the fracture surface. Exploratory fatigue tests were conducted on specimens that were preexposed to either low– or high–DO water and then tested in air or water environments in an effort to understand the effects of surface micropits or minor differences in the surface oxide on fatigue crack initiation.
FATIGUE ε–N BEHAVIOR

The fatigue lives of austenitic SSs are decreased in LWR environments; the reduction in life depends on strain amplitude, strain rate, temperature, and DO level in the water [8-15]. The effects of LWR environments on fatigue lives of wrought materials are comparable for Types 304, 316, and 316NG SSs. The critical parameters that influence fatigue life and the threshold values that are required for environmental effects to be significant are summarized below.

Strain Amplitude: A minimum threshold strain is required for an environmentally–assisted decrease in fatigue lives of SSs to be significant. The threshold strain appears to be independent of material type (weld or base metal) and temperature in the range of 250–325°C, but it tends to decrease as the strain amplitude is decreased [14].

Loading Cycle: Environmental effects on fatigue life occur primarily during the tensile–loading cycle and at strain levels greater than the threshold value. Consequently, loading and environmental conditions, e.g., strain rate, temperature, and DO level, during the tensile–loading cycle are important for environmentally–assisted reduction of fatigue lives of these steels.

Dissolved Oxygen in Water: The fatigue lives of austenitic SSs are decreased significantly in low–DO (i.e., <0.01 ppm DO) water; the decrease in life is greater at low strain rates and high temperatures [8-15]. Environmental effects in high–DO water appear to be either comparable to [12,13] or, in some cases, smaller [8] than those in low–DO water. The composition or heat treatment of the steel appears to have an effect on the magnitude of environmental effects in high–DO environments [17].

Strain Rate: Fatigue life decreases with decreasing strain rate. In low–DO environments, fatigue life decreases logarithmically with decreasing strain rate below ≈0.4%/s; the effect of environment saturates at ≈0.0004%/s [8-15]. For some SSs, the effect of strain rate may be less pronounced in high–DO water than in low–DO water.

Temperature: The results suggest a threshold temperature of 150°C, above which the environment decreases fatigue life in low–DO water if the strain rate is below the threshold of 0.4%/s [15,18]. In the range of 150–325°C, the logarithm of fatigue life decreases linearly with temperature.

Sensitization Anneal: In low–DO water, a sensitization anneal has no effect on the fatigue life of Types 304 and 316 SS, whereas, in high–DO water, environmental effects are enhanced in sensitized steel [12,13]. Sensitization has little or no effect on the fatigue life of Type 316NG SS in low– and high–DO water.

Flow Rate: Limited data indicate that the water flow rate has little or no effect on the fatigue life of austenitic SSs in high–purity water at 289°C [19].

FORMATION OF FATIGUE CRACKS

Although at one time the fatigue ε–N curves may have been interpreted as specifying, for a given strain amplitude, the number of cycles required to produce gross failure in a component, they are better interpreted as specifying the number of cycles required for the formation of surface cracks and their growth as shear and tensile cracks (Stage I and II growth) to an “engineering” size (e.g., a 3–mm–deep crack). Fatigue life has conventionally been represented by two stages: (a) initiation, which represents the cycles N_i for formation of microcracks on the surface, and (b) propagation, which represents cycles N_p for propagation of the microcracks to an engineering size. Thus, fatigue life N is the sum of the two stages, N = N_i + N_p. The former is considered to be sensitive to the stress or strain amplitude. For example, at low strain amplitudes, most of the life may be spent in initiating a microcrack, whereas at high strain amplitudes, cracks initiate easily.

During cyclic loading of smooth test specimens, surface cracks 10 µm or longer form quite early in life (i.e., <10% of life) at surface irregularities or discontinuities either already in existence or produced by slip bands, grain boundaries, second–phase particles, etc. [3,20–24]. Consequently, fatigue life may be considered to be composed entirely of propagation of cracks from 10 to 3000 µm long [25]. A schematic illustration of the two stages of fatigue life, i.e., initiation and propagation, is shown in Fig. 1. The initiation stage involves growth of microstructurally small cracks (MSCs), characterized by decelerating crack growth (Region AB in Fig. 1a). The propagation stage involves growth of mechanically small cracks, characterized by accelerating crack growth (Region BC in Fig. 1a). The growth of MSCs is very sensitive to microstructure [21,22]. Fatigue cracks greater than the critical length of MSCs show little or no influence of microstructure, and are termed “mechanically small cracks.” Mechanically small cracks correspond to Stage II (tensile) cracks, which are characterized by striated crack growth, with a fracture surface normal to the maximum principal stress.
Once a microcrack forms on the surface, it continues to grow along its slip plane as a Mode II (shear) crack in Stage I growth (orientation of the crack is usually at 45° to the stress axis). At low strain amplitudes, a Stage I crack may extend across several grain diameters before the increasing stress intensity of the crack promotes slip on systems other than the primary slip system. A dislocation cell structure normally forms at the crack tip. Because slip is no longer confined to planes at 45° to the stress axis, the crack begins to propagate as a Mode I (tensile) crack, normal to the stress axis in Stage II growth. At high strain amplitudes, the stress intensity is quite large and the crack propagates entirely by the Stage II process. Stage II continues until the crack reaches engineering size (≈3 mm deep). Various criteria have been used to define the crack length for transition from MSC to mechanically small crack. These criteria indicate that the transition crack length is a function of applied stress and microstructure of the material; actual values may range from 150 to 250 μm [26].

At low stress levels, e.g., $\Delta \sigma_1$ in Fig. 1b, the transition from MSC growth to accelerating crack growth does not occur. This circumstance represents the fatigue limit for the smooth specimen. Although cracks can form below the fatigue limit, they can grow to engineering size only at stresses greater than the fatigue limit. However, cracks larger than the transition length, either preexisting, e.g., defects in welded samples, or those created by growth of MSCs at high stresses, can grow at stress levels below the fatigue limit.

GROWTH RATES OF SMALL CRACKS IN LWR ENVIRONMENTS

Studies on growth of short cracks in smooth specimens in LWR environments [17] indicate that at the same number of cycles, the crack length is longer in low–DO (PWR) water than in air (Fig. 2). For example, after 1500 cycles the crack length in air, high-DO (BWR) water, and low–DO (PWR) water is ≈40, 300, and 1100 μm, respectively. The growth of cracks during the initiation stage, i.e., growth of MSCs, is enhanced in water; fatigue cycles needed to form a 500–μm crack are a factor of ≈12 lower in low–DO water than in air. Figure 2 shows that the number of cycles required to produce a 500–μm crack is 800, 3000, and 9,000 in low–DO, high-DO, and air environments, respectively; thus, the number of cycles is more than a factor of 10 lower in low–DO water than in air.
The crack growth rates (CGRs) during the propagation stage, i.e., growth of mechanically small cracks, in air and water environments are plotted as a function of crack length in Fig. 3; they were calculated from the best fit of the data in Fig. 2. In high–DO water, the CGRs of a specimen that was soaked only for 24 h (closed diamonds in Fig. 3) instead of 120 h are also included in the figure; for this specimen, growth rates were determined from measurements of fatigue striations on the fracture surface. The CGRs are a factor of 2–6 higher in water than in air. Growth rates in PWR water or high–DO water with a 24–h soak period are higher than those in high–DO water with a 120–h soak period. At a crack length of ≈1000 μm, the CGRs in air, high–DO water, and low–DO PWR environment are 0.30, 0.64, and 1.05 μm/cycle, respectively. For the 0.75% strain range and 0.004%/s strain rate, these values correspond to growth rates of ≈1.6 x 10⁻⁹, 3.4 x 10⁻⁹, and 5.6 x 10⁻⁹ m/s in air, high–DO water, and low–DO water, respectively. Growth rates are a factor of 3.5 greater in low–DO water than in air.

**Figure 3. Crack growth rates plotted as a function of crack length for stainless steels in air and water environments (Refs. 23,27)**

The CGRs determined from the ε–N tests are consistent with the data obtained from fracture–mechanics tests [28]. However, the large reductions in fatigue life of austenitic SSs in PWR environments cannot be explained entirely on the basis of enhanced CGRs during the propagation stage, i.e., growth of mechanically small cracks. For example, the CGRs in low–DO water are a factor of 1.6 greater than those in high–DO water, but the fatigue life is a factor of ≈4 lower in low–DO water than in high–DO water. As shown in Fig. 2, the decrease in fatigue lives of austenitic SSs in PWR environments is caused predominantly by the effects of environment on the growth of cracks <200 μm.

**FRACTURE MORPHOLOGY**

The fracture morphology of austenitic SSs in air or the LWR environment does not differ significantly; during Stage II growth, well–defined fatigue striations are observed in air and water [8,9]. The striation spacings in air and water environments show very good agreement with macroscopic crack growth rates. The presence of well–defined striations suggests that mechanical factors, and not the slip dissolution/oxidation process, are important. Fatigue striations should not be observed if crack growth is enhanced by the slip dissolution/oxidation process. The crack morphology of the specimen surface, however, is somewhat different in air or high–DO water than in low–DO water. Figure 4 shows that fatigue cracks are generally straight and normal to the stress axis in low–DO water, whereas in air or high–DO water, they follow certain crystallographic features [17].

**EXPLORATORY TESTS**

The reduction of fatigue life in high–temperature water has often been attributed to the presence of surface micropits that are formed in high–temperature water and may act as stress raisers and provide preferred sites for the formation of fatigue cracks. Also, austenitic SSs exposed to LWR environments develop an oxide film that consists of two layers; a fine–grained, tightly–adherent, chromium–rich inner layer, and a crystalline, nickel–rich outer layer composed of large and intermediate–size particles. The inner layer forms by solid–state growth, whereas the crystalline...
outer layer forms by precipitation or deposition from the solution. A schematic representation of the surface oxide film is shown in Fig. 5. The characteristics of the surface oxide films that form on austenitic SSs in LWR coolant environments can influence the mechanism and kinetics of corrosion processes and thereby influence the initiation stage, i.e., the growth of MSCs.

In an effort to understand the effects of surface micropits or minor differences in the surface oxide film on fatigue crack initiation, fatigue tests were conducted on Type 316NG (Heat P91576) specimens that were preexposed to either low– or high–DO water and then tested in air or water environments. The results of these tests, along with data obtained earlier on this heat and Heat D432804 of Type 316NG SS in air and low–DO water at 288°C, are plotted in Fig. 6.

The results indicate that surface micropits have no effect on the formation of fatigue cracks; the fatigue lives of specimens preoxidized at 288°C in low–DO water and then tested in air are identical to those of unoxidized specimens (Fig. 6). If the presence of micropits was responsible for the reduction in life, the preexposed specimens should show a decrease in life. Also, the fatigue limit of these steels should be lower in water than in air; however, the fatigue limit of austenitic SSs is approximately the same in water and air environments. The presence of an oxide film is not a sufficient condition for the environmentally– assisted decrease in fatigue lives of materials in LWR environments.

The results also indicate that minor differences in the composition or structure of the surface oxide film also have no effect on the fatigue life of SSs in low–DO water. The fatigue lives of specimens preoxidized in high– or low–DO water and then tested in low–DO water are identical.

SUMMARY

This paper describes the effects of key material and loading variables, such as strain amplitude, strain rate, temperature, DO level in water, and material heat treatment, on the fatigue lives of austenitic SSs in LWR environments. Unlike carbon and low–alloy steels, environmental effects on the fatigue life of austenitic SSs are significant in low–DO water; effects on life in high–DO water are either comparable or, for some steels, less pronounced than those in low–DO water.
The mechanism of fatigue crack initiation in austenitic SSs in LWR environments has been examined. Fatigue crack initiation has been divided into two stages: an initiation stage that involves the growth of MSCs (i.e., cracks smaller than ≈200 µm), and a propagation stage that involves the growth of mechanically small cracks. Crack lengths as a function of fatigue cycles have been determined in air and LWR environments. The results indicate that decreases in the fatigue lives of these steels are caused primarily by the effects of environment on the growth of MSCs and, to a less extent, on enhanced growth rates of mechanically small cracks.

The crack morphology of the specimen surface is different in low–DO water than in air or high–DO water; cracks are always straight and normal to the stress axis in low–DO water, whereas, in air or high–DO water, they follow certain crystallographic features. However, the morphology of crack growth into the material is similar in air and water environments; during the propagation stage, well–defined fatigue striations are observed in both air and water environments. The differing crack morphology of the surface of the specimens tested in low–DO water indicates that the mechanism of crack initiation is different in the low–DO PWR environment than in air or high–DO water. The presence of well–defined striations indicates that mechanical factors are important; environmentally–assisted reduction in the fatigue life of austenitic SSs is most likely caused by mechanisms such as hydrogen–enhanced crack growth.

Austenitic SSs exposed to LWR environments develop a dark, fine–grained, tightly–adherent, chromium–rich inner layer that forms by solid–state growth, and a crystalline nickel–rich outer layer composed of large– and intermediate–size particles that form by precipitation or deposition from the solution. The characteristics of the surface oxide films can influence the mechanism and kinetics of corrosion processes and thereby influence fatigue crack initiation. Exploratory fatigue tests were conducted on austenitic SS specimens that were preexposed to either low– or high–DO water and then tested in air or water environments in an effort to understand the effects of surface micropits or minor differences in the surface oxide on fatigue crack initiation. The results indicate that the presence of a surface oxide film or any difference in the characteristics of the oxide film has no effect on fatigue crack initiation in austenitic SSs in LWR environments.

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