NUREG/CR-4667-Vol. 23

NUREG/CR-4667, Vol. 23
ANL-97/10

Environmentally Assisted Cracking in Light Water Reactors

Semiannual Report
July 1996 - December 1996

Manuscript Completed: August 1997

Date Published: October 1997

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Prepared for

Division of Engineering Technology

Office of Nuclear Regulatory Research

U.S. Nuclear Regulatory Commission

Washington, DC 20555-0001

NRC Job Code W6610

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Environmentally Assisted Cracking in Light Water Reactors Semiannual Report October 1994—March 1995, NUREG/CR-4667 Vol. 20, ANL-95/41 (January 1996).

Environmentally Assisted Cracking in Light Water Reactors Semiannual Report April—December 1995, NUREG/CR-4667 Vol. 21, ANL-96/1 (July 1996).

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by

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Abstract

This report summarizes work performed by Argonne National Laboratory on fatigue and environmentally assisted cracking (EAC) in light water reactors from July 1996 to December 1996. Topics that have been investigated include (a) fatigue of carbon, low-alloy, and austenitic stainless steels (SSs) used in reactor piping and pressure vessels, (b) irradiation-assisted stress corrosion cracking of Type 304 SS, (c) EAC of Alloy 600, and (d) characterization of residual stresses in welds of boiling water reactor (BWR) core shrouds by numerical models. Fatigue tests were conducted on ferritic and austenitic SSs in water that contained various concentrations of dissolved oxygen to determine whether a slow strain rate applied during various portions of a tensile-loading cycle are equally effective in decreasing fatigue life. Slow-strain-rate-tensile tests were conducted in simulated BWR water at 288°C on SS specimens irradiated to a low fluence in the Halden reactor and the results were compared with similar data from a control-blade sheath and neutron-absorber tubes irradiated in BWRs to the same fluence level. Crack-growth-rate tests were conducted on compact-tension specimens from a low-carbon content heat of Alloy 600 in high-purity oxygenated water at 289°C. Residual stresses and stress intensity factors were calculated for BWR core shroud welds.

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Executive Summary

Environmental Effects on Fatigue Strain-versus-Life (S-N) Behavior of Primary Pressure Boundary Materials

Plain carbon steels (CSs) and low-alloy steels (LASs) are used extensively in steam supply systems of pressurized and boiling water nuclear reactors (PWRs and BWRs) as piping and pressure vessel materials. Fatigue tests are being conducted on CSs and LASs in water and in air to establish the effects of material and loading variables on fatigue life. The results indicate a significant decrease in fatigue life when five conditions are satisfied simultaneously, viz., when applied strain range, service temperature, dissolved-oxygen (DO) in the water, and sulfur content of the steel are above a minimum threshold level, and the loading strain rate is below a threshold value. Only a moderate decrease in fatigue life is observed in light water reactor (LWR) environments when any one threshold condition is not satisfied. This report presents (a) results of fatigue tests that examine the influence of the reactor environment on crack initiation and crack growth of CSs and LASs, and (b) data on austenitic stainless steel (SS) that establish the effects of various loading and environmental variables on fatigue lives of these steels.

Crack lengths as a function of fatigue cycles were determined for A533-Gr B LAS and A106-Gr B CS in air and water environments. The results indicate that, in air, fatigue cracks that are 10 μm or longer, form quite early during fatigue (<10% of life) even at low strain ranges ($\approx 0.4\%$). The decrease in fatigue life of CS and LASs in high-DO water is primarily caused by the effects of environment on the growth of short cracks. During the initial stages of fatigue damage (for cracks <100 μm), crack growth rates are nearly two orders of magnitude greater in high-DO water than in air, and one order of magnitude greater for cracks >100 μm long. Metallographic examination of test specimens indicates that in high-DO water, the surface cracks appear to grow entirely as Mode I tensile cracks normal to the stress axis, whereas, in air or low-DO water, growth of surface cracks initially occurs as shear cracks at $\approx 45^\circ$ to the stress axis.

Fatigue tests have also been conducted on Types 316NG and 304 SS at 288°C in LWR coolant environments. The results indicate a significant decrease in fatigue life of the water relative to that in air. The decrease in life depends both on strain rate and DO content in water. Environmental effects on fatigue life are the same for Types 304 and 316NG SS. However, unlike CS and LASs, environmental effects are more pronounced in low-DO than in high-DO water. At $\approx 0.004\%/s$ strain rate, reduction in fatigue life in water that contains <10 ppb DO is greater by a factor of ≈ 2 than in water that contains ≥ 200 ppb DO. A detailed examination of test specimens indicates that the formation and growth of surface cracks appear to be different in simulated PWR water than in air. In all environments, cracks primarily form within persistent slip bands. In air, and for most cases in high-DO water, surface cracks initially grow along slip planes as shear cracks at 45° to the stress axis in Stage I growth. In simulated PWR water (<10 ppb DO), the surface cracks grow entirely as tensile cracks normal to the stress axis in Stage II growth.

Irradiation Assisted Stress Corrosion Cracking

Slow-strain-rate-tensile (SSRT) tests were conducted in simulated BWR water at 288°C on model SS alloys that were irradiated at 288°C in helium to a fluence of $\approx 0.45 \times 10^{21}$ n-cm⁻² ($E > 1$ MeV) in the Halden reactor. Initial results indicate that the ductility of irradiated commercial-purity Type 304 SS sheet specimens is higher than that of BWR neutron-absorber tubes at a comparable fluence level. High-purity (HP) Type 316 SS was more susceptible to intergranular stress corrosion cracking (IGSCC). This behavior is similar to that observed for HP heats of Type 304 SS. Type 304 SS heats high in carbon content were more susceptible to transgranular stress corrosion cracking (TGSCC) but less susceptible to IGSCC. Unusually high levels of oxygen and sulfur in SS appear to have a deleterious effect on SCC of Type 304 SS, even in the nonirradiated state, during SSRT tests in HP oxygenated water at 288°C.

Shielded-metal-arc (SMA) welds were prepared from Types 304, 304L, and 316 SS to simulate BWR core shroud welds. Significant contamination by oxygen, sulfur, and fluorine was observed in the heat-affected zone (HAZ) of the SMA welds. This contamination originates at the coating on the welding electrode and from the air atmosphere. In contrast, contamination in gas-tungsten-arc welds was insignificant.

Microhardness profiles were obtained for the fusion zone and HAZ of an SMA weld of Type 304 SS. A region of maximum hardness was in the HAZ at a distance of ≈ 0.8 -3 mm from the fusion zone. This region appears to coincide with the region of most frequent cracking in BWR core shroud welds. When the surface of the weld was smoothed with a grinding wheel, hardness near the ground surface was significantly higher, indicating that a cold-worked surface layer may facilitate crack nucleation. The local hardening characteristics, together with contamination by oxygen, fluorine, and sulfur, may be important factors that promote susceptibility of core shroud welds to SCC in BWRs.

Analysis by Auger electron spectroscopy shows that fluorine atoms segregate to grain boundaries in BWR neutron-absorber tubes and a core shroud weld. However, the exact mechanism of the segregation (i.e., thermal, irradiation-induced, or both) is not understood. Because fluorine and oxygen are virtually insoluble in steel, they are likely to segregate by a thermal process to local sites such as matrix/precipitate interfaces, grain boundaries, and stacking faults. In addition to thermal segregation, irradiation-induced segregation cannot be ruled out, and further investigation is needed.

The design, fabrication, and calibration of a facility for J-R testing of the fracture toughness irradiated compact-tension specimens in a hot cell has been completed. Several out-of-cell fracture toughness J-R tests have been conducted on annealed and nonirradiated Type 304 SS specimens according to the "Standard Test Method for J-Integral Characterization of Fracture Toughness," which is currently under consideration within the ASTM technical committee. The results indicate that both the elastic compliance method or DC potential-drop technique can be used to estimate crack length during J-R or fatigue crack growth tests. These findings will be validated by (a) conducting fracture toughness J-R curve tests at room temperature and 288°C on two heats of thermally aged CF-8M cast SS and on 50% cold-worked Type 316NG SS, and (b) performing fatigue crack growth studies on the same three materials at 288°C in air.

Environmentally Assisted Cracking of Low-Carbon Alloy 600 in Simulated LWR Water

Corrosion-fatigue experiments were conducted on compact-tension specimens of a low-carbon content (0.03 wt.%) heat of Alloy 600 in HP oxygenated water to investigate the effects of load ratio on crack growth rates (CGRs) at 289°C. The specimens were fabricated from blocks of material that were solution-annealed at 1025 and 1115°C for 2 h and heat treated at 600°C for 24 h after solution annealing at the two temperatures. Solution annealing at 1025 and 1115°C produced ASTM grain sizes of ≈ 1.0 -1.5 and 0.2, respectively, which correspond to large average grain diameters of ≈ 210 -250 and 340 μm . Based on results from four specimens, these heat treatment conditions did not have a significant effect on the CGRs in HP oxygenated water. Other heats of material with differing carbon content and heat treatment conditions could produce a wider variation in the results.

"Best-fit" correlations for the CGRs in water that contained ≈ 300 ppb DO versus ΔK were obtained for each specimen and for the combined data from the four specimens of the same material with differing heat treatments. The effect of HP oxygenated water on the CGRs relative to that of air was determined from (a) a ratio of the CGRs at 289°C in water to the CGRs of another heat of Alloy 600 in air at 289°C, and (b) a ratio of the CGRs from the combined data in water for all of the heat treatment conditions to the CGR in air at 289°C. The results indicate that the rates for Alloy 600 are higher in water than in air at ΔK values $< 18 \text{ MPa}\cdot\text{m}^{1/2}$ and are lower in water than in air at higher values of ΔK . At a ΔK of $\approx 2 \text{ MPa}\cdot\text{m}^{1/2}$, the rate is higher by a factor of ≈ 10 in water than in air. At this ΔK , CGRs for specimens from other heats of Alloy 600 were higher by a factor of ≈ 2 in HP water that contained < 5 ppb DO than in air.

The morphology of corrosion-fatigue cracks in the Alloy 600 specimens was that of transgranular cracking in HP oxygenated water because of the strong contribution of mechanical cyclic loading in tests at load ratios of 0.1-0.8 and because of the relatively small degree of crack extension that occurred in tests at load ratios > 0.8 .

Modeling of Residual Stresses in Core Shroud Structures

Under a subcontract with Argonne National Laboratory (ANL), Battelle Columbus Laboratories has used numerical models to characterize weld residual stresses and the associated stress intensity factors at BWR core shroud welds. A detailed description of the weld residual stresses has been obtained for the H4 weld. The multipass welding process was simulated with an axisymmetrical solid-element model. A shell element analysis was performed to quantify the effects of specimen removal on residual stress measurements. The residual stress results formed the basis for the calculation of stress intensity factors.

The H4 weld is a multipass submerged-arc weld that joins two Type 304 SS cylinders. The axial residual stresses for this weld show a "thick-shell" type of distribution. In the middle of the wall, the axial residual stresses are compressive, whereas tensile stresses were present at both inner and outer surfaces of the pipe except at the weld centerline, where a small-amplitude compressive stress is present on the outer surface. This compressive axial stress was primarily due to axial bending of the wall caused by radial shrinkage of the weld and the presence of a weld cap. The axial stresses are tensile in the HAZs on the inner and outer surfaces. At these cross sections, the maximum tensile axial stresses are at the inner surface of the shroud wall. The hoop stresses are tensile almost throughout these cross

sections. The maximum tensile hoop stress occurs at the weld centerline between the midthickness and outer surfaces. At the HAZs, the tensile stresses peak at the inner and outer surfaces. However, their magnitudes are much smaller than the peak value at the weld centerline.

After the residual stress distributions in the uncracked weldment were determined, stress intensity factors K for weldments that contained flaws were determined. Complete 360° cracks and a surface crack that extends over only a portion of the circumference of the cylinder were considered. For the axisymmetrical case with a complete 360° crack, the results indicate that a crack growing from the inside to the outside of the vessel surface, or vice versa, would be likely to stop growing at approximately midwall, because K becomes negative for deeper cracks. For the surface crack that extends over only a portion of the circumference, the K values at the deepest point of the crack decrease as the crack depth increases. However, the K values along the crack front near the surface remain positive, and in fact, increase as the crack grows deeper. These results suggest that cracks, if driven by corrosion mechanisms that depend on K , will tend to increase in length in the angular direction much more rapidly than they will grow throughwall. Indeed, there will be a tendency for full 360° cracks to develop; the average depth of the cracks is expected to be ≈50% of the wall thickness, although local perturbations in the stress distributions could produce localized regions of deeper cracking.

Acknowledgments

The authors thank W. F. Burke, T. M. Galvin, and D. R. Perkins for their contributions to the experimental effort. This work is sponsored by the Office of Nuclear Regulatory Research, U.S. Nuclear Regulatory Commission, under Job Code W6610; Program Manager: Dr. M. B. McNeil.

1 Introduction

Since 1967, the U.S. Nuclear Regulatory Commission (NRC) and its predecessor the U.S. Atomic Energy Commission, have conducted research programs that address aging of reactor components. The results of the research have been used to evaluate and establish regulatory guidelines to ensure acceptable levels of reliability for light water reactor (LWR) components. The products of these programs have been technical reports, methodologies for evaluating licensee submittals, and other inputs to the regulatory process. The results have led to the resolution of regulatory issues, as well as the development, validation, and improvement of regulations and regulatory guides. The current research on the effects of simulated reactor coolant environments on cracking of reactor components was initiated to resolve remaining critical technical issues related to cracking phenomena in LWR components. Initially, this project addressed cracking of boiling water reactor (BWR) pipes. Subsequently, in response to requests from the NRC Division of Nuclear Reactor Regulation (NRR) for assistance in dealing with cracking problems that were developing in aging reactors, the focus shifted to other problems in environmentally assisted cracking (EAC) of LWR components.

The overall objective of the current program is to provide data and physical models that can be used by the NRC staff to assess environmentally assisted degradation of primary pressure boundary components in LWRs. The research is divided into five tasks as follows:

(a) *Effects of environment on fatigue, crack growth, and stress corrosion cracking*

Fatigue and EAC of piping, pressure vessels, and core components in LWRs are important concerns in plant operation and for extended reactor lifetimes. The degradation processes in U.S. reactors include fatigue, intergranular stress corrosion cracking (IGSCC), and propagation of fatigue or stress corrosion cracks that initiate in the weld-sensitized heat-affected zone (HAZ) of stainless steel (SS) components. Occurrences of fatigue failures induced by mechanical vibration and thermal fluctuation in LWR plants have also been documented. The objective of this task is to improve fatigue design curves and assess the additivity of fatigue damage in piping and vessel steels under load histories typical of LWR components. The results of this work will be used to assess industry fatigue evaluations related to license renewal.

(b) *Vulnerability of components to irradiation-assisted stress corrosion cracking*

Irradiation-assisted stress corrosion cracking (IASCC) of in-core components of both BWRs and pressurized water reactors (PWRs) is becoming a more common problem as reactors age. The general pattern of the observed failures indicates that, as nuclear plants age and neutron fluence increases, many apparently nonsensitized austenitic materials become susceptible to intergranular failure by IASCC. Some of these failures have been reported for components subjected to relatively low or negligible stress levels, e.g., control-blade sheaths and handles and instrument dry tubes of BWRs. Although most failed components can be replaced, some safety-significant structural components, such as the BWR top guide, core plate, and shroud, would be very difficult or impractical to replace. The objective of this task is to provide data and models that are needed to assess industry analyses of the likelihood of degradation and failure of core internal components due to IASCC, and to evaluate licensee submissions that concern inspection and remediation.

(c) *Cracking of nickel alloy components of LWR primary systems*

Internal components of reactor vessels are made of nickel-base alloys, e.g., Alloys 600, X750, and 182, which are susceptible to IGSCC. The causes and mechanisms of the cracking of these alloys are not adequately understood and increase the uncertainty when licensee submissions are evaluated for factors such as damage accumulation and inspection intervals. The objective of this task is to provide technical data on the effects of cracks in nickel-alloy components on residual life, inspection, and repair. The results will be used to support NRR staff assessment of industry crack growth models, and potential detection and mitigation measures.

(d) *Analysis of postweld heat treatment processes and validation of flaw acceptance criteria*

The objective of this task is to evaluate the effect of postweld heat treatment on long-term resistance to environmental cracking by assessing sensitization and other microstructural changes. Emphasis is to be placed on components made of welded Alloy 600. This evaluation will provide insights to NRC that will be useful in reviewing licensee submittals.

(e) *Assessment of industry crack-growth models*

This task has two objectives. The first is to perform an independent evaluation of industry models that are used to establish inspection intervals and repair criteria. The second objective is to perform more detailed analyses of flaw acceptance criteria.

Research during the past six months has focused on fatigue of carbon steels (CSs) and low-alloy ferritic steels (LASs) and SSs used in piping and pressure vessels, IASCC during slow-strain-rate tensile (SSRT) tests in simulated BWR water of SS specimens that were irradiated to a low fluence in the Halden reactor, EAC of Alloy 600 in air and in high-purity (HP) oxygenated water, and modeling of residual stresses in core shroud structures.

2 Environmental Effects on Fatigue Strain-versus-Life (S-N) Behavior of Primary Pressure Boundary Materials

The ASME Boiler and Pressure Vessel Code Section III, Subsection NB, ¹ which contains rules for the construction of Class 1 components for nuclear power plants, recognizes fatigue as a possible mode of failure in pressure vessel steels and piping materials. Cyclic loadings on a structural component occur because of changes in the mechanical and thermal loadings as the system goes from one load set (e.g., pressure, temperature, moment, and force loading) to any other load set. For each pair of load sets, an individual fatigue usage factor is determined by the ratio of the number of cycles anticipated during the lifetime of the component to the allowable cycles. Figures I-9.1 to I-9.6 of Appendix I to Section III of the Code specify fatigue design curves that define the allowable number of cycles as a function of applied stress amplitude. The cumulative usage factor (CUF) is the sum of the individual usage factors. The ASME Code Section III requires that the CUF at each location must not exceed 1.

The Code design fatigue curves are based on strain-controlled tests of small polished specimens at room temperature in air. In most studies, the fatigue life of a test specimen is defined as the number of cycles needed for the tensile stress to drop 25% from its peak value, which corresponds to an ≈3-mm-deep crack. Consequently, fatigue life N represents the

number of cycles required to initiate a crack that is ≈ 3 mm deep. In the Code, the best-fit curves to the experimental data are expressed in terms of the Langer equation² of the form

$$\epsilon_a = B(N)^{-b} + A, \quad (1)$$

where A, B, and b are parameters of the model. Equation 1 may be written in terms of stress amplitude S_a instead of strain amplitude ϵ_a , in which case stress amplitude is the product of strain amplitude and elastic modulus, i.e., $S_a = E \cdot \epsilon_a$. The design fatigue curves were obtained by decreasing the best-fit curves by a factor of 2 on stress or 20 on cycles, whichever was more conservative, at each point on the best-fit curve. As described in the Section III criteria document, these factors are intended to account for the differences and uncertainties that are encountered when the fatigue lives of laboratory test specimens are related to those of actual reactor components. The factor of 20 on cycles is the product of three separate subfactors: 2 for scatter of data (minimum to mean), 2.5 for size effects, and 4 for surface finish, atmosphere, etc.³ "Atmosphere" was intended to reflect the effects of an industrial environment rather than the controlled environment of a laboratory. The factors of 2 and 20 are not safety margins but rather conversion factors that must be applied to the experimental data to obtain reasonable estimates of the lives of actual reactor components. They do not intend to address the effects of reactor coolant environments on fatigue life.

Subsection NB-3121, of Section III of the Code states that the data on which the fatigue design curves (Figs. I-9.1 to I-9.6) are based do not include tests in the presence of corrosive environments that might accelerate fatigue failure. Article B-2131 in Appendix B to Section III states that the owner's design specifications should provide information about any reduction to fatigue design curves that was necessitated by environmental conditions. Recent fatigue S-N data demonstrate potentially significant effects of LWR coolant environments on the fatigue resistance of CSs, and LASs,⁴⁻¹⁰ and austenitic SSs.¹¹ As shown in Fig. 1, under certain conditions of loading and environment, fatigue lives of CSs can be a factor of 70 lower in water than in air. Therefore, the margins in the ASME Code may be less conservative than originally intended.

Figure 1.
Fatigue S-N data for CSs in water

Experience with operating nuclear plants worldwide reveals that many failures may be attributed to fatigue. Examples of such failures include emergency core cooling or residual heat removal systems (USNRC Bulletin No. 88-08), pressurizer surge lines (USNRC Bulletin No. 88-11), PWR feedwater lines (USNRC Information Notice No. 79-13), BWR pressure vessels

(USNRC Information Notice No. 90-29), PWR steam generator vessels (USNRC Information Notice No. 90-04), and steam generator feedwater distribution piping (USNRC Information Notice No. 91-19 and No. 93-20). These failures may be classified into three categories: thermal fatigue caused by thermal stratification, cycling, and striping loadings; mechanical fatigue due to vibratory loading; and corrosion fatigue from exposure to a corrosive environment. Significant thermal loadings due to stratification were not included in the original design basis analysis. Some of these fatigue sensitive locations are routinely monitored in nuclear power plants worldwide to better define the transients and more accurately assess CUF. Occurrences of mechanical-vibration- and thermal-fluctuation-induced fatigue failures in LWR plants in Japan have also been documented.¹²

In 1991, the USNRC issued a draft Branch Technical Position (BTP) on fatigue evaluation of nuclear plant components for license renewal. The BTP raised a concern about the adequacy of the ASME Code in addressing the effects of environment on fatigue resistance of materials in operating PWRs and BWRs, whose primary-coolant pressure boundary components are constructed as specified in Section III of the Code. Researchers at Argonne National Laboratory (ANL) have compiled and reviewed the existing fatigue S-N data to identify key variables that influence the fatigue life of primary pressure boundary materials.^{8,9} A program was initiated at ANL to obtain fatigue data under conditions where information was lacking in the existing data base. The results from the study have been presented in previous progress reports.¹³⁻¹⁹ Researchers at ANL have developed interim design fatigue curves that are based on the existing S-N data. The curves address the effects of environment on fatigue life of CSs, LASs, and austenitic SSs.²⁰ Statistical models have also been developed for estimating the effects of various material and loading conditions on fatigue life of these materials.^{21,22} Results of the statistical analysis have been used to estimate the probability of fatigue cracking in reactor components. The Pressure Vessel Research Council has also been compiling and evaluating fatigue S-N data related to the effects of LWR coolant environments on the fatigue life of pressure boundary materials; the results have been presented by Van Der Sluys and Yukawa.²³

In 1993, the USNRC directed its staff to treat fatigue as a potential safety issue within the existing regulatory process for operating reactors. The staff developed a Fatigue Action Plan (FAP) to resolve three principal issues: (a) the adequacy of fatigue resistance of older vintage plants designed to the United States of America Standard (USAS) B31.1 Code that did not require an explicit fatigue analysis of components, (b) the effect of LWR environments on the fatigue resistance of primary pressure boundary materials, and (c) the appropriate corrective action required when the limits allowable by the Code have been exceeded, i.e., when the CUF is >1 . The Idaho National Engineering Laboratory assessed the significance of the interim fatigue design curves by performing fatigue evaluations of several components in the reactor coolant pressure boundary.²⁴ In all, six locations were evaluated from facilities designed by each of the four U.S. nuclear steam supply system vendors. Selected components from older vintage plants designed to comply to B31.1 Code were also included in the evaluation. An assessment of risk to components in the reactor coolant pressure boundary from failure due to fatigue was performed under Generic Safety Issue 78, "Monitoring of Fatigue Transient Limits for the Reactor Coolant System." On the basis of these studies, it was concluded* that no immediate action is necessary to deal with fatigue issues addressed in the FAP. The risk

* Policy Issue, SECY-95-245, Completion of the Fatigue Action Plan, Sept. 25, 1995.

study indicated that a fatigue failure of piping is not a significant contributor to core-melt frequency. Although fatigue cracks may occur, they may not propagate to failure and, even if failure did occur, safety systems, such as an emergency core cooling system, mitigate the consequences. On the basis of the risk assessment, a backfit of the environmental fatigue data to operating plants cannot be justified.

The types and extent of conservatisms present in the ASME Section III fatigue evaluations and the effects of LWR environments on fatigue margins have been documented by Structural Integrity Associates, Inc., under contract to Sandia National Laboratories for the U.S. Department of Energy and in cooperation with the Electric Power Research Institute (EPRI).²⁵ A review of numerous stress reports indicated a substantial amount of conservatism in many existing component fatigue evaluations. The sources of conservatism include design transients considerably more severe than those experienced in service, grouping of transients, simplified elastic/plastic analysis, and Code rules prior to 1979. Environmental effects on two components, the BWR feedwater nozzle/safe end and PWR steam generator feedwater nozzle/safe end, known to be affected by severe thermal transients were also investigated in the study. The report concludes that reductions in fatigue life due to environmental effects (factors of up to 40 and 22 for PWR and BWR nozzles, respectively) are more than offset by the margins in fatigue life (≈ 60 and 90, respectively, for PWR and BWR nozzles) found in typical ASME Code fatigue evaluations. The margins in fatigue life were determined from the ratio of CUFs based on the mean experimental S-N curve and the Code design fatigue curve. The margins of ≈ 60 and 90 would be true only if it was demonstrated that the fatigue S-N curve for these specific components was comparable to or better than the mean experimental curve, and that the effects of mean stress, loading history, or component size and geometry are insignificant. As discussed earlier, the factors of 2 on stress and 20 on cycles should not be considered as safety margins but rather conversion factors that are required to obtain reasonable estimates of the lives of actual reactor components.

The conservatism in ASME Code fatigue evaluation procedures have also been demonstrated in fatigue tests on piping welds and components.²⁶ In air, the margins on the number of cycles to failure for elbows and tees were 118-2500 and 123-1700, respectively, for CSs; and 47-170 and 25-322, respectively, for SSs. The margins for girth butt welds were significantly lower, e.g., 14-128 and 6-76, respectively, for CSs and SSs. However, in these tests on welds and components, the fatigue life was expressed as the number of cycles for the crack to penetrate through the wall, which ranged from 6 to 18 mm (0.237 to 0.719 in.). The fatigue design curves represent the number of cycles to form a 3-mm-deep crack. Consequently, depending on wall thickness, the actual margins to failure may be lower by more than a factor two.

In addition, fatigue tests conducted on vessels at the Southwest Research Institute for the Pressure Vessel Research Council²⁷ show that ≈ 5 -mm deep cracks can form in CSs and LASs; very close to the values predicted by the ASME Code design curve, as shown in Fig. 2. The tests were performed in room-temperature water on 0.914-m (36-in.)-diameter vessels with 19-mm (0.75-in.) walls. These results clearly demonstrate that the Code design curves do not necessarily guarantee any margin of safety. A new nonmandatory Appendix to Section XI has been developed that does account for environmental effects.

Figure 2.
Fatigue data for CS and LAS vessels
tested in room-temperature water

The objective of this task is to provide data and models for predicting the effects of environment on fatigue design curves and to assess the additivity of fatigue damage in piping and vessel steels under load histories typical of LWR components. The data will be used by the NRC staff to assess industry fatigue evaluations related to license renewal. The information, which will be helpful in assessing applicant submissions with regard to residual lifetimes of components subjected to fatigue in reactor coolant environments, will be provided to the ASME to help improve the design curves. Fatigue tests are being conducted to establish the effects of various loading and environmental variables on the fatigue S-N behavior of pressure boundary steels. Progress on this task during the current reporting period is presented below.

2.1 Crack Initiation and Crack Growth Characteristics of Carbon and Low-Alloy Steels (D. J. Gavenda, P. R. Luebbers, and O. K. Chopra)

Available fatigue S-N data for CSs and LASs in an air environment indicate significant heat-to-heat variation. At 288°C, fatigue life may vary up to a factor of 5 above or below the mean value. The results also indicate that the ASME mean curve for LASs is in good agreement with the experimental data, and that for CSs it is somewhat conservative.

In air, the fatigue life of CSs and LASs depends on steel type, temperature, orientation (rolling or transverse), and strain rate. The fatigue life of CSs is a factor of ≈ 1.5 lower than that of LASs. For both steels, fatigue life decreases with increase in temperature up to 320°C. Some heats of CSs and LASs exhibit effects of strain rate and orientation. For these heats, fatigue life decreases with decreasing strain rate. Also, based on the distribution and morphology of sulfides, the fatigue properties in the transverse orientation may be inferior to those in the rolling orientation. Both CSs and LASs exhibit dynamic strain aging at temperatures between 200 and 370°C, which leads to enhanced cyclic hardening, a secondary hardening stage, and negative strain rate sensitivity, i.e., cyclic stress increases with decreases in strain rate. The temperature range and extent of dynamic strain aging vary with compositional and structural factors. However, at the temperatures of dynamic strain aging, fatigue life of the steel may increase or decrease.

The fatigue S-N data in water indicate that environmental effects on fatigue life are significant when five conditions are satisfied simultaneously, viz., applied strain range, temperature, dissolved oxygen (DO) in water, and sulfur content in steel are above a minimum

threshold level, and strain rate is below a critical value. There is little or no difference in susceptibility to environmental degradation of fatigue life of CSs and LASs.

For both steels, the fatigue data indicate threshold values of 150°C, 0.05 ppm DO, and ≈ 0.003 wt.% sulfur, above which fatigue life in LWR environments may decrease. The effect of DO content on life saturates at 0.5 ppm. The data also indicate a threshold strain rate of 1%/s, below which fatigue life in LWR environments decreases; the effect saturates at $\approx 0.001\%/s$. Limited data suggest that the threshold strain is either equal to or slightly greater than the fatigue limit of the material. In the temperature range of 150–320°C, fatigue life decreases linearly with temperature. When the threshold conditions for all five parameters are satisfied, fatigue life decreases logarithmically with decreasing strain rate and DO level.

Only a moderate decrease in fatigue life is observed in LWR environments when any one of the threshold conditions is not satisfied, e.g., at temperatures $\leq 150^\circ\text{C}$, or in low-DO PWR environments (≤ 0.05 ppm DO), or for low-sulfur steels (≤ 0.003 wt.% sulfur). Under these conditions, life is 30–50% lower in water than in air. The fatigue life of these steels in air and LWR environments can be estimated from statistical models.^{21,22} The models have recently been updated because we determined that, in the range of 0.05–0.5 ppm, the effect of DO was more logarithmic than linear.* In air, fatigue life N , defined as the number of cycles to form of a 3-mm-deep crack, of CSs is expressed as

$$\ln(N) = 6.570 - 0.00133 T - 1.871 \ln(\epsilon_a - 0.11) \quad (2a)$$

and that of LASs, as

$$\ln(N) = 6.667 - 0.00133 T - 1.687 \ln(\epsilon_a - 0.15), \quad (2b)$$

where ϵ_a is applied strain amplitude (%) and T is temperature ($^\circ\text{C}$). For LWR environments the fatigue life N of CSs is expressed as

$$\ln(N) = 6.186 - 1.871 \ln(\epsilon_a - 0.11) + 0.1097 S^* T^* O^* \dot{\epsilon}^* \quad (3a)$$

and that of LASs, as

$$\ln(N) = 5.901 - 1.687 \ln(\epsilon_a - 0.15) + 0.1097 S^* T^* O^* \dot{\epsilon}^*, \quad (3b)$$

where ϵ_a is the applied strain amplitude (%); T is test temperature ($^\circ\text{C}$); and S^* , T^* , O^* , and $\dot{\epsilon}^*$ are transformed sulfur content (wt.%), temperature ($^\circ\text{C}$), DO (ppm), and strain rate (%/s), respectively, defined as follows:

$$\begin{aligned} S^* &= S & (0 < S \leq 0.015) \\ S^* &= 0.015 & (S > 0.015) \end{aligned} \quad (4a)$$

$$\begin{aligned} T^* &= 0 & (T < 150) \\ T^* &= T - 150 & (T = 150-350) \end{aligned} \quad (4b)$$

* O. K. Chopra, presented at the joint meeting of the working groups on Evaluation Methods and S/N Data Analysis, Pressure Vessel Research Council, April 1996, Orlando, FL.

$$\begin{aligned}
 O^* &= 0 & (DO < 0.05) \\
 O^* &= \ln(DO/0.04) & (0.05 \leq DO \leq 0.5) \\
 O^* &= \ln(12.5) & (DO > 0.5)
 \end{aligned}
 \tag{4c}$$

$$\begin{aligned}
 \dot{\epsilon}^* &= 0 & (\dot{\epsilon} > 1) \\
 \dot{\epsilon}^* &= \ln(\dot{\epsilon}) & (0.001 \leq \dot{\epsilon} \leq 1) \\
 \dot{\epsilon}^* &= \ln(0.001) & (\dot{\epsilon} < 0.001)
 \end{aligned}
 \tag{4d}$$

The functional forms for S^* , T^* , O^* , and $\dot{\epsilon}^*$ were defined on the basis of the experimental data. The last term in Eqs. 3a and 3b is zero when any one of the critical threshold conditions is not satisfied, e.g., in low-DO (DO levels <0.05 ppm) PWR environments, or at temperatures <150°C, or strain rates >1%/s.

Fatigue tests are also being conducted to determine the crack initiation and crack growth characteristics of CSs and LASs in air and LWR environments. Results of fatigue tests that examine the influence of reactor environment on formation and growth of short cracks in A533-Gr B LAS have been presented earlier.¹⁹ The results indicate that in air, fatigue cracks 10 μm or longer form quite early during fatigue (<10% of life). The largest crack is <100 μm at half life. The decrease in fatigue life in high-DO water is primarily caused by the effects of environment during early stages of fatigue damage, i.e., growth of cracks <100 μm deep. For cracks >100 μm deep, crack growth rates are an order of magnitude higher in high-DO water than in air.

During this reporting period, additional fatigue tests have been conducted to determine the crack initiation and crack growth characteristics of A533-Gr B LAS and A106-Gr B CS in water. Results of these tests and data obtained earlier are presented below and the effects of LWR environments on the formation and growth of short cracks are discussed.

2.1.1 Mechanism of Fatigue Crack Initiation

The fatigue S-N curve represents the process of fatigue crack initiation (or fatigue damage accumulation). It specifies that, for a given stress or strain amplitude, the number of cycles needed to form an "engineering" crack (e.g., a 3-mm-deep crack), is defined as the fatigue life of the material. Deformation and microstructural changes in the surface grains control fatigue cracking. During cyclic straining, irreversibility of dislocation glide leads to the development of surface roughness. Also, strain localization in persistent slip bands (PSBs) results in the formation of extrusions and intrusions. With continued cycling, microcracks ultimately form in the PSBs or at the edges of slip-band extrusions. At high strain amplitudes, microcracks form in notches that develop at grain or twin boundaries. Similarly, microcracks can also develop at phase boundaries (e.g., ferrite/pearlite boundaries) or by cracking of second-phase particles (e.g., sulfide or oxide inclusions).

Once a microcrack forms, it continues to grow along its slip plane or a PSB 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. 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 crack

propagation continues until the crack reaches engineering size (≈ 3 mm deep). In air or mildly corrosive environments, Stage II fracture is characterized by fatigue striations.

Fatigue life, or number of allowable cycles, 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 surface cracks to engineering size. Thus, fatigue life N is the sum of the two stages, i.e., $N = N_i + N_p$. The N_i stage is considered to be sensitive to the stress or strain amplitude, e.g., most of the life may be spent in initiating a crack at low strain amplitudes, whereas, at high strain amplitudes, cracks initiate easily.

An alternative approach considers fatigue life to be entirely composed of crack propagation;^{28,29} for polycrystalline materials, the period for the formation of surface cracks is zero (Fig. 3). Fatigue damage in a material is the current size of a fatigue crack, and damage accumulation is the rate of crack growth.²⁹ During fatigue loading of smooth test specimens, cracks form immediately at surface irregularities/discontinuities either already in existence or produced by slip bands, grain boundaries, second-phase particles, etc. Growth of these surface cracks may be divided into three regimes: (a) initial period which involves growth of microstructurally small cracks (MSCs) that are characterized by decelerating crack growth, as shown in region AB of Fig. 3; (b) a final period of growth characterized by accelerating crack growth, region CD; and (c) a transition period controlled by a combination of the two regimes, region BC.

Figure 3.
Growth of cracks in smooth fatigue specimens

The growth of MSCs is very sensitive to microstructure.³⁰⁻³³ The MSCs correspond to Stage I cracks and grow along slip planes as shear cracks in the early stage of growth. For MSCs, microstructural effects are strong because of the Stage I growth, i.e., crystallographic growth. Grain boundaries, triple points, and phase boundaries markedly decrease growth rates. In ferritic-pearlitic steels, fatigue cracks initiate and propagate preferentially in the ferrite phase that forms as long allotriomorphs at the prior austenite phase boundaries.^{32,33} The ferrite-pearlite phase boundaries act as strong barriers to crack propagation, and growth rates decrease significantly when small cracks grow into the pearlite from the ferrite.³² Limited data suggest that microstructural effects are more pronounced at negative stress

ratios; the compressive component of the applied load plays an important role in the formation of Stage I facets and as a driving force for cracking.³¹

Fatigue cracks greater than the critical length of MSCs show little or no influence of microstructure and are called mechanically small cracks.³¹ For a stress ratio of -1 , the transition from an MSC to a mechanically small crack for several materials has been estimated to be approximately eight times the unit size of the microstructure.³¹ Mechanically small cracks correspond to Stage II, or tensile, cracks characterized by striated crack growth, with a fracture surface normal to the maximum principal stress. Their growth rates tend to decrease as the cracks grow because crack closure becomes more significant for larger cracks. For ferritic-pearlitic steels, Stage II crack propagation occurs when stress intensity and mode of growth attain a critical level, break through the pearlite, and join other ferrite cracks.

At low stress levels, e.g., $\Delta\sigma_2$ in Fig. 3, the transition from MSC growth to accelerating crack growth does not occur and the cracks are nonpropagating. 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. Possible preexisting large cracks in the material, e.g., defects in welded samples, or those created by growth of microcracks at high stresses, can grow at stress levels below the fatigue limit.

Reduction in fatigue life in LWR coolant environments may arise from easy formation of surface microcracks and/or an increase in growth rates of cracks, during either the initial stage of MSC and shear crack growth or the transition and final stage of tensile crack growth. Carbon steel and LAS specimens tested in water show surface micropitting and cavities that were produced by either corrosion or dissolution of MnS inclusions.^{13,14} These micropits can act as sites for the formation of fatigue cracks.

Detailed metallographic studies of fatigue specimens^{13,14} and exploratory tests^{8,9} have been conducted to evaluate the contributions of environment to the formation of surface microcracks. The results indicate that the reduction in fatigue life in high-DO water is primarily due to environmental effects on crack propagation. Environment appears to have little or no effect on the formation of surface microcracks. Although all specimens tested in water show surface micropitting, there is no indication that these micropits enhance cracking. Irrespective of environment, cracks in CS and LAS nucleate either along slip bands, sulfide inclusions, carbide particles, or at the ferrite/pearlite phase boundaries.

The environmental enhancement of fatigue crack growth in pressure vessel steels in high-temperature oxygenated water and the effects of sulfur content, loading rate, and flow velocities are well known.³⁴⁻³⁹ The enhanced growth rates in LWR environments have been attributed to either slip dissolution^{40,41} or hydrogen-induced cracking⁴² mechanisms. The requirements for a slip dissolution model are that a protective oxide film is thermodynamically stable so that a crack will propagate with a high aspect ratio without degrading into a blunt pit, and that a strain increment occurs to rupture the film and thereby exposes the underlying matrix to the environment. Once the passive oxide film is ruptured, crack extension is controlled by dissolution of freshly exposed surfaces and by the oxidation characteristics. Hydrogen-induced cracking of LASs is explained as follows: hydrogen produced by the oxidation reaction at or near the crack tip is partly absorbed into the metal; the absorbed hydrogen diffuses ahead of the crack tip, interacts with MnS inclusions, and leads to the

formation of cleavage cracks at the inclusion/matrix interface; and linkage of the cleavage cracks results in discontinuous crack extension in addition to extension caused by mechanical fatigue.

Both mechanisms depend on oxide rupture rates, passivation rates, and liquid diffusion rates. For the environment to affect fatigue life, a critical level of sulfide ions must be present at the crack tip. Dissolution of MnS inclusions changes the water chemistry near the crack tip, making it more aggressive. The result is enhanced crack growth rate (CGR) because either the dissolved sulfides decrease the repassivation rate, which increases the amount of metal dissolution for a given oxide rupture rate;⁴¹ or the dissolved sulfide poisons the recombination of hydrogen atoms liberated by corrosion, which enhances hydrogen uptake by the steel at the crack tip.⁴³ Recent studies, based on a balance between sulfide supply rate due to exposure of sulfide inclusions from an advancing crack and mass transport of sulfur ions away from the crack tip by diffusion, indicate that in low-DO PWR water, the average critical crack velocity for environmental enhancement of crack growth is inversely proportional to the crack depth.³⁴ Furthermore, the critical velocity must be maintained for a minimum crack extension of ≈ 0.3 mm for enhancement to occur. The results indicate that for cracks < 0.3 mm, a low-DO environment should not effect the rate, and for cracks 0.3 – 3 mm, velocities of 3.8×10^{-9} to 4.2×10^{-10} m·s⁻¹ are required for the environment to decrease fatigue life. Although these velocities are not achieved under loading conditions typically used in fatigue tests, low-DO water can decrease fatigue life of steels that have a favorable distribution and morphology of sulfides. In these steels, a critical concentration of sulfide ions can be maintained at the crack tip because the probability of an advancing crack to intercept sulfide inclusions is higher. The existing fatigue S-N data are consistent with these results; fatigue life of CSs and LASs in low-DO water (< 0.01 ppm DO) at 288°C is lower than that in air by a factor of < 2 . However, it is plausible that, in high-DO water environments, MSCs or nonpropagating cracks could grow by anodic dissolution or other processes and thereby decrease the fatigue life of the material.

2.1.2 Experimental

Low-cycle fatigue tests have been conducted on A106-Gr B CS and A533-Gr B LAS. The A533-Gr B material was obtained from the lower head of the Midland reactor vessel, which was scrapped before the plant was completed. The A106-Gr B material was obtained from a 508-mm-diameter Schedule 140 pipe fabricated by the Cameron Iron Works, Houston, TX. The compositions of the two materials are given in Table 1, and average room-temperature tensile properties provided by the material suppliers are given in Table 2. The CS has a pearlitic structure, whereas the LAS consists of tempered bainite. Smooth cylindrical specimens with a diameter of 9.5 mm and a gage length of 19 mm were used for the fatigue tests. The specimen gage length was given a $1\text{-}\mu\text{m}$ surface finish in the axial direction to prevent circumferential scratches that might act as sites for crack initiation.

The tests in water were conducted in a small autoclave, under stroke control, where the specimen strain was controlled between two locations outside the autoclave. All tests were conducted at 288°C with fully reversed axial loading (i.e., $R = -1$) and a triangular or sawtooth waveform. The strain rate for the triangular wave and for the fast-loading half of the sawtooth wave was $0.4\%/s$. Tests in air were performed under strain control with an axial extensometer; specimen strain between the two locations that were used in the water tests was also recorded. Information from the air tests was used to determine the stroke required to

Table 1. Composition (wt.%) of ferritic steels

Material	Source	C	P	S	Si	Cr	Ni	Mn	Mo
A106-Gr B ^a	ANL	0.29	0.013	0.015	0.25	0.19	0.09	0.88	0.05
	Supplier	0.29	0.016	0.015	0.24	-	-	0.93	-
A533-Gr B ^b	ANL	0.22	0.010	0.012	0.19	0.18	0.51	1.30	0.48
	Supplier	0.20	0.014	0.016	0.17	0.19	0.50	1.28	0.47

^a Schedule 140 pipe 508-mm O.D., fabricated by Cameron Iron Works, Heat J-7201. Actual heat treatment not known.

^b Hot-pressed plate, 162 mm thick, from Midland reactor lower head. Austenitized at 871-899°C for 5.5 h and brine quenched, then tempered at 649-663°C for 5.5 h and brine quenched. The plate was machined to a final thickness of 127 mm. The inside surface was inlaid with 4.8-mm weld cladding and stress relieved at 607°C for 23.8 h.

Table 2. Average room-temperature tensile properties of CS and LAS

Material	Yield Stress (MPa)	Ultimate Stress (MPa)	Elongation (%)	Reduction in Area (%)
A106-Gr B	301	572	23.5	44.0
A533-Gr B	431	602	27.8	66.6

maintain constant strain in the specimen gage length for tests in water; the stroke was gradually increased during the test to account for cyclic hardening of the material and to maintain constant strain in the specimen gage section. Details about the test facility and procedure have been described elsewhere.^{8,13}

Surface replication by an acetate film was employed to measure surface length of cracks in specimens tested in room-temperature air. This technique provides $\approx 1\text{-}\mu\text{m}$ resolution of the surface and allows surface features such as PSBs and cracking to be identified. The tests were interrupted periodically to allow for surface replication. A detailed examination of the gage surface of the specimen was conducted to establish the morphology of final fatigue cracks. The region on surface replicas that corresponded to the final cracks was examined to determine the length of surface cracks as a function of fatigue cycles.

The crack morphology and length of the final crack for an A533-Gr B specimen tested at 0.75% strain range with a sawtooth waveform (Test No. 1779, Specimen Id. 44-67) are shown in Fig. 4. Surface crack lengths were used as reference markers to define probable crack fronts on the fracture surface. Fatigue striations on the fracture surface were used as guidelines to determine the shape of the crack fronts. The probable crack fronts for the surface crack in Fig. 4, after 5316, 5779, 5955, and 6195 cycles, are shown in Fig. 5. The results indicate that three cracks merged to form the final fracture surface. The primary crack was initiated near an inclusion and reached a surface length of $\approx 100\text{ }\mu\text{m}$ after 3062 cycles (i.e., $\approx 50\%$ of fatigue life). Two secondary cracks merged with the primary crack after ≈ 5700 and 6000 cycles. Crack depth, a , was determined from the surface crack length S from the relationship

$$\frac{a}{S} = \frac{1}{\pi}. \quad (5)$$

The calculated values agree closely with measurements of the probable crack depth on the fracture surface.

Figure 4. Morphology and length of surface crack after varying number of cycles for A533-Gr B steel tested in air at room temperature, 0.75% strain range, and slow/fast sawtooth waveform. Letters denote reference points used to locate the same regions on replicas.

Figure 5. Fracture surface and probable crack front after varying numbers of fatigue cycles for surface crack shown in Fig. 4

For the tests in high-temperature water, crack growth characteristics were determined with block loading, e.g., the slow/fast fatigue tests were interrupted each period at $\approx 10\%$ of life, and the specimen was subjected to a block of fast/fast cycles at either the same or a lower strain range. The block of fast/fast cycles leaves a beach mark on the fracture surface that can be used to characterize crack size as a function of fatigue cycles for the slow/fast test. A photomicrograph of the fracture surface of a slow/fast test at 0.75% strain range, interrupted every 50 cycles and subjected to 170 cycles of fast/fast loading at 0.4% strain range, is shown in Fig. 6. Beach marks produced by four of the five fast/fast loading sequences at the low strain range can be seen clearly on the fracture surface.

2.1.3 Fatigue Life

The results of room-temperature tests in air with surface replication and block loading tests in water at 288°C are given in Table 3. Relevant data obtained earlier^{8,9,17} on A533-Gr B and A106-Gr B steels in air and water environments are also listed in the table. The results from the present study and data obtained earlier are shown in Figs. 7-9. Estimates of fatigue lives based on statistical models, Eqs. 2-4, are also included in the figures. The recent test results are consistent with the earlier data.

Figure 6.
Beach marks produced by four of five block loading sequences on the fracture surface of A533-Gr B steel tested at 288°C in high-DO water

Table 3. Fatigue results for CSs and LASs in air and water environments

Test No.	Environ. ^a	Temp. (°C)	Strain Range (%)	Strain Rate (%/s)		Stress Range (MPa)	Life N ₂₅ (Cycles)	Number of Blocks	CUF ^b
				Ten.	Comp.				
A533-Gr B Low-Alloy Steel									
1785	Air	25	0.76	0.4	0.4	763.7	8,840	c	0.974
1779			0.76	0.004	0.4	759.8	5,960	c	0.657
1786			0.40	0.4	0.4	687.7	61,100	c	0.513
1533	PWR	288	0.77	0.004	0.4	916.0	3,416	–	0.812
1818			0.80	0.004	0.4	919.0	2,030	5 ^d	0.536
1645	High DO	288	0.72	0.4	0.4	831.1	2,736	–	0.819
1626			0.79	0.004	0.4	910.1	247	–	0.794
1715			0.81	0.004	0.4	904.1	381	–	1.310
1718			0.81	0.004	0.4	904.3	346	–	1.190
1800			0.81	0.004	0.4	897.9	303	5 ^e	1.042
1803			0.82	0.004	0.4	913.9	272	7 ^f	0.966
1641			0.39	0.4	0.4	693.0	17,367	–	0.387
1665			0.38	0.004	0.4	717.0	3,455	–	0.522
1815			0.38	0.004	0.4	735.8	2,200	5 ^g	0.332
A106-Gr B Carbon Steel									
1619	Air	288	0.40	0.4	0.4	741.7	37,142	–	0.844
1636			0.40	0.4	0.4	749.6	34,829	–	0.791
1634	High DO	288	0.40	0.4	0.4	733.2	19,318	–	0.743
1624			0.46	0.004	0.4	775.7	2,276	–	2.104
1639			0.42	0.004	0.4	751.6	2,951	–	1.940
1816			0.43	0.004	0.4	759.6	1,410	5 ^g	1.015

^a PWR = simulated PWR water that contains 2 ppm lithium and 1000 ppm boron; high-DO = deionized water with 0.6–0.8 ppm DO.

^b Fatigue usage factor based on the statistical model, i.e., usage = observed life/life estimated from the model.

^c Interrupted periodically to allow for surface replication.

^d Every 450 cycles, a block of 170 fast/fast cycles at 0.4% strain range and 0.4%/s strain rate.

^e Every 50 cycles, a block of 170 fast/fast cycles at 0.4% strain range and 0.4%/s strain rate.

^f Every 30 cycles, a block of 170 fast/fast cycles at 0.4% strain range and 0.4%/s strain rate.

^g Every 420 cycles, a block of 500 fast/fast cycles at the same strain range but 0.4%/s strain rate.

Figure 7.
Fatigue lives of A533-Gr B steel tested at room temperature in air and high-DO water

Figure 8. Fatigue lives of A533-Gr B steel tested at 288°C in simulated PWR and high-DO water

Figure 9.
Fatigue lives of A106-Gr B steel tested at 288°C in high-DO water

Figure 10. Photomicrographs of fractured specimens tested with slow/fast waveform at (a) 0.8% strain range interrupted every 30 cycles and subjected to 170 cycles at lower strain range, and (b) 0.38% strain range interrupted every 420 cycles and subjected to 170 cycles at fast strain rate

At room temperature, the fatigue life in water is $\approx 50\%$ lower than in air. The difference in fatigue lives of specimens tested at 0.4 and 0.004%/s strain rate may be attributed to experimental scatter and not to strain rate effects. Although the available S-N data indicate that, in air, strain rate may influence fatigue life of some CSs and LASs, these effects occur in the temperature range for dynamic strain aging (200–370°C).⁹ In room-temperature air, strain rate has little or no effect on the fatigue life of CSs and LASs.

The results of the block loading tests are consistent with results obtained from tests conducted at constant strain range and strain rate (Figs. 8 and 9). Both A533-Gr B LAS and A106-Gr B CS show significant reduction in fatigue life in high-DO water at 288°C and a strong dependence on strain rate. Fatigue life decreases rapidly with decreasing strain rate; however, a few differences may be observed between the results from tests with block loading and those obtained at constant strain range and rate. For both steels, the fatigue lives of specimens subjected to block loading are consistently lower than those of specimens subjected to constant strain range and rate. The fatigue lives, expressed as CUF, were estimated from a statistical model and are listed in Table 3. For a specific test condition, the fatigue usage factor is expressed as the ratio of the observed and estimated values of fatigue life; CUF is the sum of individual usage factors.

The fracture behavior during block loading tests also differs from that observed during constant strain tests. Photomicrographs of the fracture surface of A533-Gr B specimens tested at 0.38 and 0.82% strain range are shown in Fig. 10. Both fracture surfaces consist of several cracks that were initiated at different axial and circumferential locations. These cracks do not merge into a single primary crack. The fracture surface of specimens tested at constant strain range consist of a single crack, or a few cracks that merge to form the final fracture surface, as shown in Fig. 5.

2.1.4 Fatigue Crack Depth

For the tests in air and water environments, the depth of the largest crack is plotted as a function of fatigue cycles in Fig. 11, and as a function of fractional life in Fig. 12. The results

Figure 11. Depth of largest crack as a function of fatigue cycles for A533-Gr B LAS and A106-Gr B CS in air and water environments

Figure 12. Depth of largest crack as a function of fractional life for A533-Gr B LAS and A106-Gr B CS in air and water environments

indicate that, in room-temperature air, $\approx 10\text{-}\mu\text{m}$ cracks form quite early during fatigue, e.g., $<10\%$ of fatigue life. Also, depth of the largest crack is $<100\text{ }\mu\text{m}$ at half life. The results in high-DO water show that the decrease in fatigue life of both A106-Gr B CS and A533-Gr B LAS is primarily caused by the effects of environment during early stages of fatigue damage, i.e., growth of cracks $<100\text{ }\mu\text{m}$ in depth (Stage I crack growth). For the same life fraction, crack depth is greater in high- or low-DO water than in air. At $\approx 0.8\%$ strain range, only 30–50 cycles are needed to form a $100\text{-}\mu\text{m}$ crack in high-DO water (average growth rate of $2\text{--}3\text{ }\mu\text{m}/\text{cycle}$), whereas it takes more than 3000 cycles to form a $100\text{-}\mu\text{m}$ crack in air (average growth rate of $0.033\text{ }\mu\text{m}/\text{cycle}$). In low-DO PWR environment, the growth of cracks is faster than it is in air but slower than it is in high-DO water.

Figure 12 shows that in both air and water environments, fatigue crack size at different life fractions is independent of strain range, strain rate, and DO content of the water. The depth of the largest crack at different life fractions is approximately the same at 0.75 and 0.4% strain ranges. These results show good agreement with earlier studies on CSs^{32,44} and LASs⁴⁵ (see Fig. 13); crack depth was determined from the surface crack length with Eq. 5.

Figure 13. Crack depth as a function of fractional life for CSs and LASs tested in room-temperature air

These results indicate that, for strain or stress ranges above the fatigue limit, linear damage summation (Miner's rule) is valid for CS and LAS in both air and water environments.

A detailed examination of the fatigue test specimens was conducted to investigate the role of high-temperature oxygenated water on fatigue cracking. Photomicrographs of longitudinal sections of A533-Gr B and A106-Gr B specimens tested in air, PWR, and high-DO environments at 288°C, ≈0.8% strain range, and a slow/fast waveform (sawtooth) are shown in Fig. 14. In both steels, the formation and growth of surface cracks appear to be different in high-DO water than in air or simulated PWR water. In air and low-DO PWR environments, surface cracks grow as Mode II (shear) cracks in Stage I growth along planes that were usually at 45° to the stress axis. The Stage I crack extends across several grains until the increasing stress intensity of the crack promotes slip on systems other than the primary slip, and the crack begins to propagate as a Mode I (tensile) crack normal to the stress axis. Furthermore, in A106-Gr B CS, Stage I crack growth occurs entirely along the soft ferrite grains, avoiding the relatively hard pearlite regions.

In high-DO water, the surface cracks appear to grow entirely as Mode I tensile cracks normal to the stress axis. In A106-Gr B CS, cracks propagate across both ferrite and pearlite regions. The absence of Stage I crack growth suggests that factors other than mechanical fatigue are important for growth of surface cracks in high-DO water. These results are consistent with a slip dissolution model for enhanced crack growth rates in LWR environments.

2.1.5 Crack Growth Rate

The CGRs determined from the crack-depth-versus-cycles data of Fig. 11 are plotted as a function of crack depth in Fig. 15. The results indicate that, for the same crack depth, growth rates for the tests at 0.8% strain range are higher than those at 0.4% strain range. At both 0.8 and 0.4% strain range, CGRs in high-DO water are one order of magnitude higher than in air. Crack growth rates in a low-DO PWR environment are higher than those in air by a factor of 2 to 3. As discussed in the previous section, the average growth rate during the early stage of fatigue damage is ≈2.5, 0.22 and 0.033 μm/cycle, respectively, in high-DO water, low-DO water, and air environments.

Figure 14. Photomicrographs of surface cracks along longitudinal sections of A533-Gr B LAS and A106-Gr B CS in air, high-DO water, and simulated PWR environments

The results from the present study indicate that the decrease in fatigue life of CSs and LASs in high-DO water is primarily caused by the effects of environment on the growth of short cracks. Crack growth rates are nearly two orders of magnitude higher in high-DO water than in air during the initial stages of fatigue damage (for cracks $<100 \mu\text{m}$), and one order of magnitude higher for cracks $>100 \mu\text{m}$. Metallographic examination of the test specimens indicate that, in high-DO water, surface cracks grow entirely as tensile cracks normal to the stress. In air and low-DO water, surface crack growth occurs initially as shear cracks $\approx 45^\circ$ to

Figure 15. Crack growth rates as a function of crack depth for A533-Gr B LAS tested in air and water environments. Dotted lines show the probable growth rates during the early stage of fatigue damage in air and low- and high-DO.

the stress axis, and then as tensile cracks normal to the stress axis when slip is no longer confined to planes at 45° to the stress axis.

These results suggest that, in LWR environments, the growth of MSCs occurs by anodic dissolution; the growth rates depend on the DO in water and on the sulfur content of the steel. When the applied strain is greater than the threshold strain required to rupture the passive oxide film, crack extension occurs by anodic dissolution of freshly exposed surfaces and depends on the oxidation characteristics. Studies of the effects of strain rate change on fatigue life of CSs and LASs in high-DO water also show that the applied strain must exceed a threshold value to rupture the passive surface film for environmental effects to occur.^{8,9,16,17} The crack tip strain rate is used to describe oxide rupture at the crack tip during crack advance. The critical crack tip velocity depends on sulfur content in the steel and sulfide ion diffusion away from the crack tip in water. The relationship of environmentally assisted crack growth rate V (cm·s⁻¹) to the crack tip strain rate $\dot{\epsilon}_{ct}$ (s⁻¹) is

$$V = A(\dot{\epsilon}_{ct})^n, \quad (6)$$

where constants A and n depend on the material and environmental conditions at the crack tip. There is a lower limit on the crack propagation rate associated with either blunting (when the crack tip cannot keep up with general corrosion rate of the crack sides) or the fact that a critical level of sulfide ions cannot be maintained at the crack tip. The crack propagation rate at which this transition occurs will depend on DO level, flow rate, etc. Based on these factors, the maximum and minimum environmentally controlled crack propagation rates have been defined.⁴⁰ For crack-tip sulfide ion concentrations above the critical level,

$$V = 2.25 \times 10^{-4} \dot{\epsilon}_{ct}^{0.35}, \quad (7a)$$

and for crack tip sulfide ion concentrations below the critical level,

$$V = 10^{-2} \dot{\epsilon}_{ct}^{1.0}. \quad (7b)$$

Assuming that the crack tip strain rate $\dot{\epsilon}_{ct}$ is the same as the applied strain rate $\dot{\epsilon}_{app}$ and crack advance due to mechanical fatigue is insignificant during the initial stages of fatigue damage, the crack advance (in cm) per cycle is given by

$$\Delta a_r = 2.25 \times 10^{-4} (\Delta \epsilon - \epsilon_f) (\dot{\epsilon}_{app})^{-0.65}, \quad (8)$$

where ϵ_f is the threshold strain increment needed to rupture the oxide film. For A533-Gr B steel, the threshold strain range for environmentally assisted reduction in fatigue life is $\approx 0.36\%$.^{8,9} For an applied tensile strain rate of 0.004%/s, from Eq. 8, the crack growth during the initial stage of fatigue damage (for cracks $< 100 \mu\text{m}$) is ≈ 7 and $0.65 \mu\text{m/cycle}$ at 0.8 and 0.4% strain ranges, respectively. These values are higher than the measured average CGRs by a factor of 2 to 3 (Fig. 15).

In Fig. 16, the measured CGRs are presented with the current and proposed⁴⁶ ASME Section XI reference crack growth curves for CSs and LASs. Data for A333-Gr 6 CS in 288°C water with 1 or 0.005 ppm DO* are also included in the figure. For the cylindrical fatigue specimens, the stress intensity ranges ΔK were determined from the values of ΔJ , which, for a small half-circular surface crack,⁴⁵ are given by

$$J = 3.2 \left(\frac{\sigma^2}{2E} \right) a + 5.0 \left(\frac{\sigma \epsilon_p}{1+n} \right) a, \quad (9)$$

where E is the elastic modulus, ϵ_p is nominal plastic strain, and a is crack depth. The stress intensities associated with conventional cylindrical fatigue specimens were modified on the basis of rigorous finite-element models.⁴⁷ The cyclic stress σ and strain ϵ is defined as

$$\epsilon = \frac{\sigma}{E} + \left(\frac{\sigma}{A} \right)^n, \quad (10)$$

where the constant A and the exponent n were determined from experimental data.^{8,9,16,17} The results show good agreement with the proposed reference curves. The growth rates for the A333-Gr 6 steel in low-DO water seem to be significantly higher than the ANL data on A533-Gr B steel tested under similar conditions. The results also suggest that, for smooth cylindrical specimens, the threshold value of ΔK below which environment affects growth rates is either very low or does not exist.

* M. Higuchi, presented at the joint meeting of the working groups on Evaluation Methods and S/N Data Analysis, Pressure Vessel Research Council, April 1996, Orlando, FL.

Figure 16.
Crack growth rates determined from smooth cylindrical fatigue test specimens and ASME Section XI reference curves for CSs and LASS in air and water environments

2.2 Austenitic Stainless Steels (D. J. Gavenda and O. K. Chopra)

Fatigue tests have been conducted on Types 304 and 316NG SS in air and LWR environments to evaluate the effects of material and loading variables, such as steel type, strain rate, DO, and strain range, on the fatigue lives of these steels. The composition of the SSs used for the fatigue tests is listed in Table 4. Experimental results have been compared with estimates of fatigue life based on a statistical model. The formation and growth of fatigue cracks in austenitic SSs in air and LWR environments are discussed. Details about the test facility and procedure have been described in Section 2.1.2. The fatigue test data from the present study and those obtained previously¹¹ on Type 316NG SS in air and high-DO water are listed in Tables 5 and 6.

Table 4. Composition (wt.%) of austenitic SSs used for fatigue tests

Material	Heat	Source	C	P	S	Si	Cr	Ni	Mn	Mo	Cu	N
Type 316NG ^a	D432804	Vendor	0.011	0.020	0.001	0.52	17.55	13.00	1.76	2.49	0.10	0.108
		ANL	0.013	0.020	0.002	0.49	17.54	13.69	1.69	2.45	0.10	0.105
Type 304 ^b	30956	Vendor	0.060	0.019	0.007	0.48	18.99	8.00	1.54	0.44	-	0.100

^aASME SA312 seamless SS pipe (hot finished), 610 mm O.D. and 30.9 mm wall, fabricated by Sumitomo Metal Industries, Ltd. Solution annealed at 1038–1093°C for 0.5 h and water quenched.

^bSolution annealed at 1050°C for 0.5 h.

2.2.1 Air Environment

Existing fatigue S-N data indicate that the fatigue lives of austenitic SSs in air is independent of temperature in the range from room temperature to 427°C. Fatigue lives of Types 304 and 316 SS are similar and those of Type 316NG SS are superior. The effects of strain rate on fatigue life cannot be established from existing S-N data. Limited results suggest that some heats are sensitive to strain rate; fatigue life may decrease up to 30% with decreasing strain rate.

Table 5. Fatigue test results for Type 316NG SS

Temp. and Test No.	Environ. ^a	DO (ppb)	pH at RT	Conduc- tivity (μS/cm)	Strain Rate (%/s)		Stress Range (MPa)	Strain Range (%)	Life N ₂₅ (Cycles)
Room Temp.									
1394	Air	-	-	-	0.50	0.50	694.7	1.51	4,649
1391	Air	-	-	-	0.50	0.50	554.8	1.00	13,561
1390	Air	-	-	-	0.50	0.50	518.1	0.75	25,736
1396	Air	-	-	-	0.50	0.50	506.7	0.76	30,000
1420	Air	-	-	-	0.49	0.49	495.3	0.49	54,249
1392	Air	-	-	-	0.50	0.50	475.9	0.51	60,741
1393	Air	-	-	-	0.50	0.50	464.7	0.41	127,386
1395	Air	-	-	-	0.50	0.50	456.7	0.35	183,979
1397	Air	-	-	-	0.50	0.50	446.0	0.30	347,991
1398	Air	-	-	-	0.50	0.50	436.7	0.27	666,000
1399	Air	-	-	-	0.50	0.50	431.8	0.25	>1,900,000
1400	Air	-	-	-	0.50	0.50	427.4	0.25	1,775,000
288°C									
1408	Air	-	-	-	0.50	0.50	416.6	0.76	21,548
1790	Air	-	-	-	0.005	0.50	452.8	0.75	16,765
1409	Air	-	-	-	0.50	0.50	377.2	0.50	53,144
1410	Air	-	-	-	0.50	0.50	377.6	0.50	51,194
1792	Air	-	-	-	0.005	0.50	413.4	0.51	35,710
1407	Air	-	-	-	0.50	0.50	364.4	0.40	82,691
1430	Air	-	-	-	0.50	0.50	348.3	0.30	168,852
1435	Air	-	-	-	0.50	0.50	342.0	0.25	314,352
1480	Air	-	-	-	0.49	0.49	340.1	0.25	319,308
1485	Air	-	-	-	0.51	0.51	340.4	0.25	369,206
320°C									
1405	Air	-	-	-	0.50	0.50	426.0	0.75	20,425
1404	Air	-	-	-	0.50	0.50	387.4	0.50	47,011
1406	Air	-	-	-	0.50	0.50	371.6	0.40	82,691
288°C									
1426	Hi DO	>200	-	-	0.80	0.80	405.1	0.80	12,069
1427	Hi DO	>200	-	-	0.082	0.082	421.7	0.82	6,679
1428	Hi DO	>200	-	-	0.0074	0.0074	441.4	0.74	5,897
1797	Hi DO	750	5.9	0.076	0.005	0.50	437.3	0.78	4,520
1414	Hi DO	>200	-	-	0.50	0.50	375.3	0.50	26,230
1418	Hi DO	>200	-	-	0.50	0.50	375.5	0.50	25,714
1423	Hi DO	>200	-	-	0.05	0.05	378.8	0.50	17,812
1425	Hi DO	>200	-	-	0.0049	0.0049	393.2	0.49	13,684
1431	Hi DO	>200	-	-	0.29	0.29	356.5	0.29	116,754
1434	Hi DO	>200	-	-	0.029	0.029	350.0	0.29	40,643
1436	Hi DO	>200	-	-	0.025	0.025	354.0	0.25	>1,719,851
1512	Hi DO	>200	-	-	0.24	0.24	361.2	0.24	2,633,954
1796	PWR	5	6.4	20.20	0.50	0.50	403.6	0.80	12,500
1812	PWR	2	6.5	20.00	0.05	0.50	413.9	0.80	6,375
1791	PWR	4	6.5	19.23	0.005	0.50	441.9	0.77	3,040
1793	PWR	4	6.4	19.23	0.005	0.50	434.3	0.80	3,020
1794	PWR	4	6.4	20.00	0.005	0.50	390.9	0.50	7,370
1814	PWR	1	6.5	20.00	0.05	0.50	348.7	0.29	33,200

^a PWR = simulated PWR water that contains 2 ppm lithium and 1000 ppm boron; Hi DO = deionized water with 0.6-0.8 ppm DO.

Table 6. Fatigue test results for Type 304 SS

Temp. and Test No.	Environ. ^a	DO (ppb)	pH at RT	Conduc- tivity (μS/cm)	Strain Rate (%/s)		Stress Range (MPa)	Strain Range (%)	Life N ₂₅ (Cycles)
288°C									
1801	Air	–	–	–	0.4	0.4	419.2	0.76	24,500
1805	Air	–	–	–	0.004	0.4	467.9	0.76	14,410
1804	Air	–	–	–	0.4	0.4	382.8	0.51	61,680
1817	Air	–	–	–	0.004	0.4	421.7	0.51	42,180
1806	PWR	4	6.0	18.87	0.4	0.4	428.9	0.73	11,500
1810	PWR	5	6.4	18.89	0.04	0.4	447.6	0.77	5,800
1808	PWR	4	6.4	18.87	0.004	0.4	468.3	0.77	2,850
1821	PWR	2	6.5	22.22	0.004	0.4	474.3	0.76	2,420
1807	PWR	4	6.5	18.87	0.4	0.4	374.6	0.51	25,900
1823	PWR	3	6.6	23.06	0.004	0.4	408.2	0.51	6,900

^a PWR = simulated PWR water that contains 2 ppm lithium and 1000 ppm boron.

Statistical models have been developed for estimating the effects of material and loading conditions on the fatigue lives of austenitic SSs.^{21,22} These models are based on the JNUFAD* data base for "Fatigue Strength of Nuclear Plant Components" from Japan, the data compiled by Jaske and O'Donnell⁴⁸ for developing fatigue design criteria for pressure vessel alloys, and the tests conducted by Conway et al.,⁴⁹ Keller,⁵⁰ and by ANL.¹¹ In air, the fatigue life N of Types 304 and 316 SS is expressed as

$$\ln(N) = 6.690 - 1.980 \ln(\epsilon_a - 0.12) \quad (11a)$$

and that of Type 316NG SS as

$$\ln(N) = 7.072 - 1.980 \ln(\epsilon_a - 0.12), \quad (11b)$$

where ϵ_a is the applied strain amplitude (%). The fatigue lives of Types 304, 316, and 316NG SS in air at various temperatures are presented in Fig. 17, with the values estimated from Eqs. 11a and 11b. At temperatures of 25–450°C, the fatigue lives of Types 304 and 316 SS in air show no dependence on temperature. For Type 316NG SS, the results at room temperature show good agreement with estimates from the statistical model or the Jaske and O'Donnell⁴⁸ curve, but the data at 288°C are lower than the estimated values at strain amplitudes <0.2%.

Also, the ASME mean curve is not consistent with the existing fatigue S–N data for austenitic SSs. At strain amplitudes <0.5%, the mean curve predicts significantly longer fatigue lives than those observed experimentally; the fatigue lives at 288°C and ≈0.25% strain range fall very close to the ASME Code design curve. When the Code fatigue S–N curve for austenitic SSs was extended to 10⁸ cycles, this discrepancy was taken into account, but no change was made to the curve for <10⁶ cycles. The statistical model is in good agreement with the Jaske and O'Donnell⁴⁸ average curve.

The fatigue S–N data from the present study and those obtained previously on Type 316NG SS¹¹ in air are shown in Fig. 18. Also shown are the S–N curves estimated from the

* Private communication from M. Higuchi, Ishikawajima-Harima Heavy Industries Co., Japan, to M. Prager of the Pressure Vessel Research Council, 1992. The old data base "FADAL" has been revised and renamed "JNUFAD."

Figure 17.
Fatigue $S-N$ behavior for Types 304,
316, and 316NG SS in air at various
temperatures

Figure 18. Total-strain-range versus fatigue-life data for Types 316NG and 304 SS in air

statistical model, best-fit curve proposed by Jaske and O'Donnell,⁴⁸ and the ASME mean curve. The results show good agreement with estimates from the statistical model or Jaske and O'Donnell curve.⁴⁸

The cyclic strain hardening and cyclic stress-versus-strain curves for Types 316NG and 304 SS at room temperature and 288°C have been presented earlier.¹⁹ At both temperatures, the steels exhibit rapid hardening during the first 50–100 cycles of fatigue life. The extent of hardening increases with applied strain range and is greater at room temperature than at 288°C. The initial hardening is followed by softening and a saturation stage at 288°C and by continuous softening at room temperature. The results also indicate that, at 288–430°C, cyclic stresses increase with decreasing strain rate.

2.2.2 Water Environment

Fatigue Life

The fatigue S–N data on austenitic SSs indicate a significant decrease in fatigue life in water; the effect of environment on fatigue life increases with decreases in strain rate. The fatigue lives of austenitic SSs in LWR coolant environments can be estimated from statistical models.^{21,22} These models are based on the JNUFAD data base and the tests conducted by General Electric Corporation in a test loop at the Dresden 1 reactor^{51,52} and by ANL.¹¹ However, most of the data used in developing the models were obtained in water that contained 0.2 ppm or more of DO and at temperatures in the range of 288–320°C. Consequently, the models do not consider the effects of temperature or DO content on the fatigue lives of these steels. In LWR environments, the fatigue life N of Types 304 and 316 SS is expressed as

$$\ln(N) = 6.331 - 1.980 \ln(\epsilon_a - 0.12) + 0.134 \dot{\epsilon}^* \quad (12a)$$

and that of Type 316NG, as

$$\ln(N) = 6.713 - 1.980 \ln(\epsilon_a - 0.12) + 0.134 \dot{\epsilon}^*, \quad (12b)$$

where ϵ_a is applied strain amplitude (%) and $\dot{\epsilon}^*$ is transformed strain rate defined as

$$\begin{aligned} \dot{\epsilon}^* &= 0 & (\dot{\epsilon} > 1 \%/\text{s}) \\ \dot{\epsilon}^* &= \ln(\dot{\epsilon}) & (0.001 \leq \dot{\epsilon} \leq 1 \%/\text{s}) \\ \dot{\epsilon}^* &= \ln(0.001) & (\dot{\epsilon} < 0.001 \%/\text{s}). \end{aligned} \quad (12c)$$

The statistical model is recommended for predicting fatigue lives $\leq 10^6$ cycles. The lower bound value of 0.001%/s on the strain rate effect was based on the results for CSs and LASs.^{8,9}

The fatigue S–N data for Types 316NG and 304 SS in water at 288°C are shown in Fig. 19, which also shows the ASME Code fatigue design curve. The results indicate a significant decrease in fatigue life in water relative to that in air; the reduction in life depends on both strain rate and DO content of the water. The fatigue lives of Types 316NG and 304 SS in air, simulated PWR, and high-DO water are plotted as a function of strain rate in Fig. 20. In all environments, the fatigue lives of these steels decrease with decreasing strain rate. The effect of strain rate is smallest in air and largest in a low-DO PWR environment. In a simulated PWR environment, a decrease in strain rate from 0.4 to 0.004%/s decreases fatigue life by a factor of ≈ 8 . The decrease in life is lower at low strain ranges, e.g., a factor of ≈ 8 at 0.75% and ≈ 5 at 0.3% strain range.

Figure 19. Total-strain-range-versus-fatigue-life data for Types 316NG and 304 SS in air and water

Figure 20. Effect of strain rate on fatigue lives of austenitic SSs in air and in simulated PWR and high-DO water environments

The results also indicate that environmental effects on the fatigue lives of austenitic SSs are more pronounced in low-DO than in high-DO water. At slow strain rates, e.g., $\approx 0.004\%/s$, the reduction in fatigue life in a simulated PWR environment (<10 ppb DO) is greater by a factor of ≈ 2 than in high-DO water (≥ 200 ppb DO). Such a dependence of fatigue life on DO content is quite different from that of ferritic steels. For CSs and LASs, environmental effects on fatigue life increase with increasing DO content above a minimum threshold value of 0.05 ppm.^{8,9} Also, environmental effects on the fatigue life of CSs and LASs are modest at DO levels below 0.05 ppm, i.e., fatigue life is lower by a factor of <2 than it is in air. In view of these results, the statistical models for austenitic SSs (Eqs. 12a–12c) will be updated to incorporate the effects of DO and strain rate on fatigue life.

Metallographic Examination

A detailed examination of the fatigue test specimens was conducted to investigate the role of high-temperature oxygenated water on fatigue cracking. In general, the specimens tested in air show slight discoloration, whereas the specimens tested in oxygenated water developed a gray/black corrosion scale. Photomicrographs of the gage surface of Type 316NG SS

specimens tested in air, simulated PWR, and high-DO water environments are shown in Fig. 21. Crystalline oxides and a thin gray corrosion scale are present on specimens tested in water. X-ray diffraction analyses of specimens tested in water indicate that the corrosion scale primarily consists of magnetite (Fe_3O_4) or ferroferric oxide (FeFe_2O_4), chromium oxide (CrO), and maghemite ($\gamma\text{-Fe}_2\text{O}_3$). In addition to these phases, a specimen tested in high-DO water also contained hematite (ferric oxide or $\alpha\text{-Fe}_2\text{O}_3$).

Figure 21.
SEM photomicrographs of gage surface of Type 316NG SS specimens tested in (a) air, (b) simulated PWR, and (c) high-DO-water environments

Figure 22 shows photomicrographs of the fracture surface at approximately the same crack length for Type 316NG SS specimens tested in air, high-DO water, and low-DO PWR water. Fatigue striations can be seen clearly on all specimens. The spacing between striations indicates that crack growth rates increase in the following sequence: air, high-DO water, and low-DO PWR water.

The crack frequency, i.e., number of cracks per unit length of the specimen gage surface, was measured along longitudinal sections of specimens tested in various environments. Preliminary results indicate that crack frequency is lower in high-DO water than in air or the simulated PWR environment. For Type 316NG SS tested at 288°C , $\approx 0.75\%$ strain range, and $0.005\%/s$ strain rate, the number of cracks longer than $20\text{ }\mu\text{m}$ along a 7-mm gage length was 16, 14, and 8 in air, simulated PWR, and high-DO water, respectively.

Figure 22. Photomicrographs of fracture surface of Type 316NG SS specimens tested at 288°C and 0.75% strain range in (a) air, (b) high-DO water, and (c) simulated low-DO PWR water

The formation and growth of surface cracks appears to be different in simulated PWR water from that in air, although the crack frequency is the same in all of the environments (Fig. 23). Metallographic examination of the test specimens indicates that, in all of the environments, cracks primarily form within PSBs. During cyclic straining, strain localization in PSBs results in the formation of extrusions and intrusions at the surface; ultimately, with continued cycling, microcracks develop in these PSBs. Once a microcrack is formed, it continues to grow along its slip plane or a PSB as a Mode II (shear) crack in Stage I growth. The orientation of the crack is usually 45° to the stress axis. The Stage I crack may extend across several grains before the increasing stress intensity of the crack promotes slip on systems other than the primary slip. 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 as Stage II growth. This behavior was observed in all of the specimens that were tested in air and, in most instances, for specimens that were tested in high-DO water. In a simulated PWR environment (<10 ppb DO), the surface cracks appear to grow entirely as Mode I tensile cracks normal to the stress axis; this suggests that factors other than mechanical fatigue control the growth of surface microcracks in PWR water.

The formation and growth of surface cracks in PWR water are consistent with both the slip dissolution^{40,41} and hydrogen-induced cracking mechanisms.⁴² However, lower fatigue lives in low-DO water than in high-DO water can not be reconciled in terms of the slip dissolution mechanism. If slip dissolution alone was responsible for enhanced CGRs in water, similar growth behavior (growth of surface cracks entirely as Mode I tensile cracks) and reduction in fatigue lives should also be observed in high-DO water. It is likely that hydrogen-induced cracking is important in PWR water with low-DO. Fatigue tests are in progress on austenitic SSs to characterize the formation and growth of surface cracks in LWR environments.

Strain Rate: 0.4 %/s

Strain Rate: 0.004 %/s

Air

Simulated PWR Water

water with ≈ 0.7 ppm DO

Figure 23. Photomicrographs of surface cracks along longitudinal sections of Type 316NG SS in air and water environments at strain rates of 0.4 and 0.004 %/s

3 Irradiation-Assisted Stress Corrosion Cracking of Austenitic SS

In recent years, failures of reactor-core internal components in both BWRs and PWRs have increased after accumulation of relatively high fluence ($>5 \times 10^{20}$ n-cm⁻², $E > 1$ MeV). The general pattern of the observed failures indicates that, as nuclear plants age and neutron fluence increases, various apparently nonsensitized austenitic SSs become susceptible to intergranular failure. Some components (e.g., BWR core shroud, control-blade handle and sheath) are known to have cracked under minimal applied stress. Although most failed components can be replaced, some safety-significant structural components (e.g., BWR top guide, shroud, and core plate) would be very difficult or impractical to replace. Therefore, to understand this type of degradation, which is commonly known as IASCC, the structural integrity of these components after accumulation of high fluence has been a subject of concern and extensive research.

The primary effects of irradiation include alteration of microchemistry, microstructure, and mechanical properties of structural materials, which are usually fabricated from ASTM Type 304, 316, or 348 SSs. Irradiation produces defects and defect clusters in grain matrices, and alters the dislocation network, dislocation loop, and dislocation channel structures, leading to radiation-induced hardening. Irradiation also leads to changes in the stability of second-phase precipitates and the local alloy chemistry near grain boundaries, precipitates, and defect clusters. Grain-boundary microchemistry that differs significantly from that of the bulk composition can be produced in association with not only radiation-induced segregation (RIS) but also thermally driven equilibrium and nonequilibrium segregation of alloying and impurity elements. Neutron irradiation also alters the water chemistry, in particular, near the grain-boundary surfaces in a crack-tip region.

For many years, irradiation-induced grain-boundary depletion of chromium has been considered as the primary metallurgical process that causes IASCC. One of the most important factors that has been considered by many investigators to support the chromium-depletion mechanism is the observation that the dependence of IGSCC of nonirradiated thermally sensitized material and of IASCC of BWR-irradiated solution-annealed material on water chemistry (i.e., oxidizing potential) is similar. Recently, attention has also focused on the effects of fabrication-related variables, i.e., grain-boundary segregation and depletion of alloying and impurity elements by thermal processes, the effect of final thermomechanical treatment, hardening by cold work, the use of recycled scrap metals, uptake of minor impurities during iron- and steelmaking, and contamination by minor impurities during welding of field components such as BWR core shrouds.

In this reporting period, our effort has focused on: (a) SSRT testing of model SS alloys irradiated in the Halden reactor, (b) characterization of deleterious impurities from contamination during welding of core internal components, (c) hardness and grain size distribution in weld HAZs, and (d) analysis of fluorine and oxygen in the materials. Preliminary results obtained on base-metal components, such as neutron-absorber tubes and control-blade sheath irradiated in several BWRs, indicated that inadvertent contamination by impurities such as fluorine and oxygen can be deleterious. Consistent with this, significant fluorine and oxygen contamination was observed in a BWR core shroud weld specimen and in nonirradiated weld specimens of Type 304 SS prepared by a shielded-metal-arc (SMA) process.

3.1 Slow-Strain-Rate-Tensile Tests of Austenitic Stainless Steels Irradiated in the Halden Reactor (H. M. Chung, W. E. Ruther, and R. V. Strain)

Additional results from SSRT tests were obtained for model SS alloy specimens irradiated to a fluence of $\approx 0.45 \times 10^{21}$ n-cm⁻² ($E > 1$ MeV) at $\approx 288^\circ\text{C}$ in a helium environment in the Halden reactor. Companion tests on archive materials were also conducted in HP water that contained ≈ 8 ppm DO. The alloy composition and updated test results are given in Tables 7 and 8, respectively.

Heats L7 (high in oxygen and sulfur) and L2 (high in phosphorus and sulfur) exhibited unusually low ductility in water even in the nonirradiated state, i.e., only 8.4-9.1%, versus 19-47% for other steels listed in Table 8. Heat L7, which contains ≈ 274 wppm oxygen and ≈ 380 wppm sulfur, exhibited high ductility (45.2%) when tested in air, whereas ductility was only $\approx 8.4\%$ when tested in water (see Fig. 24). This indicates a significant effect of oxygen and sulfur, or both, dissolved in the material on SSRT properties in oxygenated water. Sulfur content was also high (≈ 340 wppm) in Type 304 SS Heat L2, which also exhibited unusually low ductility in water.

An unusual brittle morphology was observed on $\approx 20\%$ of the fracture surface of the nonirradiated specimen from Heat L7. The brittle morphology was associated with dense aggregates of precipitates that are tetrahedral in shape and 10-20 nm in size. It is not clear whether the precipitates formed in the material before fracture during the SSRT test, or after fracture during exposure to water. Nevertheless, no intergranular fracture morphology was observed in this specimen. Further analysis of microchemical and microstructural characteristics of the fracture surfaces of Heats L7 and L2 is being conducted by scanning electron microscopy energy-dispersive spectroscopy (SEM-EDS) and Auger electron spectroscopy (AES).

Table 7. Composition (in wt.%) of model SS alloys used for SSRT tests in simulated BWR water at $\approx 288^\circ\text{C}$

Alloy ID	Ni	Si	P	S	Mn	C	N	Cr	Mo or Nb	O	Remarks
L23	12.04	0.68	0.030	0.047	0.96	0.043	0.092	17.30	Nb 1.06	0.093	CP 348
L7	10.60	0.18	0.040	0.038	1.02	0.007	0.111	15.40	-	0.274	high O, N, S; low Si, C
L14	7.93	1.49	0.080	0.002	1.76	0.107	0.028	15.00	-	0.045	high Si, P, C; low S, N
L17	8.00	0.66	0.090	0.009	0.48	0.061	0.078	15.30	-	0.090	high P; low S
L6	10.00	1.90	0.020	0.005	1.13	0.096	0.087	17.10	-	0.058	high Si, C, Cr; low S
L27	10.30	0.96	0.040	0.002	0.97	0.057	0.019	15.30	Mo 2.01	-	CP 316, B 0.030
L26	8.09	0.79	0.004	0.002	0.91	0.070	0.089	17.20	-	0.080	low P, S
L2	10.50	0.82	0.080	0.034	1.58	0.074	0.102	17.02	-	0.066	high P, S, N
L25	8.93	0.92	0.020	0.008	1.54	0.019	0.095	17.20	-	0.085	B 0.010
C1	8.12	0.50	0.038	0.002	1.00	0.060	0.060	18.11	-	-	CP 304
L5	9.66	0.90	0.113	0.028	0.47	0.006	0.033	21.00	-	-	high P, Cr; low Si, C
L22	13.30	0.24	0.015	0.004	0.40	0.003	0.001	16.10	Mo 2.04	-	HP 316
C3	8.91	0.46	0.019	0.004	1.81	0.016	0.083	18.55	-	-	CP 304L
C16	12.90	0.38	0.014	0.002	1.66	0.020	0.011	16.92	-	-	high Ni; low S, N
L4	10.20	0.94	0.031	0.010	1.75	0.110	0.002	15.80	-	-	high Mn, C; low N
L18	8.13	0.14	0.016	0.033	1.13	0.080	0.001	18.00	-	-	low Si, N

Table 8. SSRT^a test and SEM fractography results for specimens of austenitic SSs irradiated in helium at 288°C in Halden reactor

Alloy Ident. No.	Fast-Neutron Fluence (n · cm ⁻²)	SSRT No.	Feedwater Chemistry				SSRT Parameters			Fracture Behavior	
			Oxygen Conc. (ppm)	Average ECP (mV SHE)	Cond. at 25°C (μS · cm ⁻¹)	pH at 25°C	Failure Time (h)	Max. Stress (MPa)	Total Elong. (%)	TGSCC (%)	IGSCC (%)
L23	0	CHR-1	8.6	+228	0.07	6.65	240.3	480	18.6	-	-
L7	0	CHR-2	8.0	+217	0.07	7.37	141.2	370	8.4	20 ^b	0
L7	0	CHR-7 test in air	-	-	-	-	761.0	572	45.2	0	0
L14	0	CHR-3	8.6	+208	0.07	7.37	794.4	474	47.2	0	0
L17	0	CHR-4	7.5	+262	0.06	7.09	236.1	412	14.0	-	-
L6	0	CHR-5	7.9	+286	0.08	6.85	773.4	545	45.9	0	0
L27	0	CHR-6	8.0	+247	0.08	6.96	405.0	483	24.1	0	0
L26	0	CHR-8	9.4	+223	0.07	6.65	703.9	596	41.8	0	0
L2	0	CHR-9	8.6	+292	0.06	6.55	152.6	348	9.1	-	-
L25	0	CHR-10	8.2	+239	0.06	6.42	479.5	458	28.4	-	-
C1	0.45 x 10 ²¹	HR-1	8.3	+184	0.07	7.03	299.9	680	17.8	4	0
L5	0.45 x 10 ²¹	HR-2	9.7	+208	0.07	6.89	552.9	539	32.8	2	2
L22	0.45 x 10 ²¹	HR-3	8.0	+236	0.07	6.80	160.4	596	9.5	50	15
C3	0.45 x 10 ²¹	HR-4	8.7	+161	0.07	6.68	552.0	491	32.8	5	0
C16	0.45 x 10 ²¹	HR-5	8.3	+204	0.08	6.74	368.6	527	21.9	2	0
L4	0.45 x 10 ²¹	HR-6	9.0	+202	0.08	6.70	400.4	542	23.8	46	0
L18	0.45 x 10 ²¹	HR-7	9.0	+203	0.08	6.33	156.7	572	9.3	54	0

^aStrain rate of $1.65 \times 10^{-7} \text{ s}^{-1}$ (strain rate for Test CHR-1 was $2.15 \times 10^{-7} \text{ s}^{-1}$).

^bBrittle crack morphology on ~20% of the fracture surface in association with dense aggregates of internal oxides that are tetrahedral in shape and 10-20 μm in size.

Figure 24. Stress-versus-elongation behavior of laboratory Heat L7 of Type 304 SS in air and water that contained ~8 ppm DO

Fracture surface morphology of several Halden-reactor-irradiated SSRT specimens was characterized by SEM; the results are summarized in Table 8 and the percent IGSCC is shown in Fig. 25 for specimens from seven heats that have been tested and characterized thus far. Data obtained from BWR neutron-absorber tubes and a control-blade sheath are also shown in the figure for comparison. Figures 26 and 27 show similar data in terms of percent transgranular stress corrosion cracking (TGSCC) and combined IGSCC and TGSCC, respectively.

Preliminary observations indicate that unusually high oxygen and sulfur dissolved in the material is deleterious to SSRT behavior in water, even at low fluence. In previous studies

Figure 25.

Percent IGSCC of SS heats irradiated in helium in Halden reactor to fluence of $\approx 0.45 \times 10^{21} \text{ n}\cdot\text{cm}^{-2}$ ($E > 1 \text{ MeV}$). Similar results from BWR neutron-absorber tubes and control-blade sheath irradiated to $\approx 0.6 \times 10^{21} \text{ n}\cdot\text{cm}^{-2}$ ($E > 1 \text{ MeV}$) are shown for comparison. HP = high purity; CP = commercial purity.

Figure 26.

Percent TGSCC of SS heats irradiated in Halden reactor to fluence of $\approx 0.45 \times 10^{21} \text{ n}\cdot\text{cm}^{-2}$ ($E > 1 \text{ MeV}$). Similar results from BWR neutron-absorber tubes and control-blade sheath are shown for comparison. HP = high purity; CP = commercial purity.

Figure 27.

Combined percent IGSCC and TGSCC of SS heats irradiated in Halden reactor to fluence of $\approx 0.45 \times 10^{21} \text{ n}\cdot\text{cm}^{-2}$ ($E > 1 \text{ MeV}$). Similar results from BWR neutron-absorber tubes and control-blade sheath are shown for comparison. HP = high purity; CP = commercial purity.

on BWR neutron-absorber tubes and a control blade sheath, we observed that susceptibility to IGSCC increased significantly for higher oxygen content in the irradiated materials. It is necessary, however, to obtain more conclusive evidence that is based on results from the entire test matrix of the Halden irradiation experiment for low-, medium-, and high-fluence levels, i.e., ≈ 0.45 , ≈ 1.1 , and $\approx 3.0 \times 10^{21} \text{ n}\cdot\text{cm}^{-2}$, respectively.

Results on Heats CP-A, L4, and L18 in Figs. 25 and 26 are consistent with a trend wherein susceptibility of high-carbon Type 304 SS to TGSCC is high, whereas susceptibility to IGSCC is low. However, Type 316 SS Heat L22 exhibited high susceptibility to TGSCC as well as IGSCC despite of its low carbon content of ≈ 0.003 wt.%.

3.2 Properties of Stainless Steel Welds (J.-H. Park and H. M. Chung)

Most large components, such as BWR core shrouds, are welded in the field by either SMA or submerged-arc (SA) procedures. These procedures include the use of welding electrodes that are coated with a welding flux. Consequently, contamination by impurities can occur in the fusion zones and HAZs, and the microchemistry and microstructure of these regions differ from those of the base metal; factors that can influence IGSCC and IASCC behavior. The structure and properties of weld HAZs of three commercial heats of Types 304, 304L, and 316 (Table 9) were investigated further in this reporting period.

Table 9. Composition (in wt.%) of commercial heats of Types 304, 304L, and 316 SS that were used to investigate SCC and IASCC behavior of welds with respect to BWR core shroud cracking

ANL ID	Type	Ni	Si	P	S	Mn	C	N	Cr	Mo
C1	304	8.12	0.50	0.038	0.002	1.00	0.060	0.060	18.11	-
C3	304L	8.91	0.46	0.019	0.004	1.81	0.016	0.083	18.55	-
C21	316	10.24	0.51	0.034	0.001	1.19	0.060	0.020	16.28	2.08

3.2.1 Chemical Composition of HAZ

Weldments of 12.7-mm-thick plates of the commercial-grade heats were prepared by SMA and gas-tungsten-arc (GTA) procedures. The compositions of the HAZ of the SMA weld and base metal of Type 304 SS Heat C1 and Type 316 SS Heat C21, determined by bulk chemical analysis,⁵³ are summarized in Tables 10 and 11, respectively. Similar results for GTA welds of the same heats are given in Tables 12 and 13, respectively. Significant contamination or depletion of elements in HAZ relative to those in base metal is denoted by bold face in the tables.

Only the fluorine content was analyzed for the SMA weld of Heat C1; results for other heats are not available. Based on the results in Tables 10-13, the following trends are evident:

- Chemical composition of the HAZ in the GTA weld is similar to that of the base metal.
- Welding by the SMA procedure leads to significant contamination of the HAZ by oxygen, fluorine, and sulfur. These impurities are conducive to higher susceptibility to IASCC. The major sources of contamination are the weld electrode coating and air.
- The weld HAZ of Type 304 SS picks up some molybdenum during SMA welding.
- The weld HAZ of Type 316 loses a significant amount of molybdenum during welding by either the SMA or GTA procedure.

Table 10. Composition of SMA weld HAZ and base metal of commercial heat C1 of Type 304 SS

Element	SMA HAZ	Base Metal
In wt. %		
Cr	19.91	18.11
Ni	9.36	8.12
Mn	1.18	1.00
Mo	0.27	<0.05
Cu	0.07	0.05
Si	0.56	0.50
In ppm		
C	600	600
O	640	80
P	310	380
S	50	20
F	30	0

Table 11. Composition of SMA weld HAZ and base metal of commercial heat C21 of Type 316 SS

Element	SMA HAZ	Base Metal
In wt. %		
Cr	19.19	16.28
Ni	10.10	10.24
Mn	1.21	1.19
Mo	0.81	2.08
Si	0.56	0.51
In ppm		
C	600	600
O	420	70
P	280	340
S	60	10

Table 12. Composition of GTA weld HAZ and base metal of commercial heat C1 of Type 304 SS

Element	SMA HAZ	Base Metal
In wt. %		
Cr	18.88	18.11
Ni	9.60	8.12
Mn	1.30	1.00
Mo	0.06	<0.05
Cu	0.05	0.05
Si	0.47	0.50
In ppm		
C	600	600
O	97	80
P	330	380
S	20	20

Table 13. Composition of GTA weld HAZ and base metal of commercial heat C21 of Type 316 SS

Element	SMA HAZ	Base Metal
In wt. %		
Cr	17.74	16.28
Ni	19.24	10.24
Mn	1.42	1.19
Mo	1.45	2.08
Si	0.53	0.51
In ppm		
C	500	600
O	51	70
P	310	340
S	20	10

3.2.2 Characterization of Grain Size and Hardening in HAZ

Grain size distribution in the HAZ of the SMA weld of Type 304 SS Heat C1 was measured and is shown in Fig. 28. Grain size was measured as a function of distance from the boundary between the fusion zone and the HAZ (see Arrow B in Fig. 29). Grain size decreases from $\approx 65 \mu\text{m}$ at the boundary and reaches a nearly constant value of $\approx 30 \mu\text{m}$ at $>1 \text{ mm}$ from the fusion zone. As indicated in Fig. 29, small inclusions were observed occasionally in the fusion zone. The ellipsoidal inclusion in the figure was $\approx 0.8 \text{ mm}$ long and $\approx 0.4 \text{ mm}$ wide. Analysis by SEM-EDS indicates that the inclusion was rich in oxygen, sodium, aluminum, silicon, potassium, calcium, titanium, chromium, and manganese, elements that are abundant in the coating of the E308-16 weld electrode that was used to produce the SMA

Figure 28.
Grain size of HAZ in SMA weld of Type 304 SS, Heat C1, as function of distance from fusion zone

Figure 29.
Schematic representation of SMA weld of Type 304 SS Heat C. WZ and HAZ denote, respectively, fusion and heat-affected zone. Top surface was smoothed with a grinding wheel.

weld. This finding indicates that the inclusion was a small piece from the weld electrode coating.

To characterize local hardening of the SMA weld HAZ, Vickers microhardness was measured as a function of distance from the boundary between the fusion zone and the HAZ (in the direction of Arrow B in Fig. 29); the results are shown in Fig. 30. Hardness at ≈ 0.5 mm from the fusion zone is low and close to the base metal hardness of ≈ 220 -230 VHN. However, hardness is relatively high (280-310 VHN) at 0.8-2.5 mm away from the fusion zone. Hardening in HAZ is influenced by several factors: (a) local oxygen contamination, (b) grain size, (c) dislocations and defects produced during welding, and (d) local annealing and recovery during cooling.

Most cracking in BWR core shroud welds occurs at 1-4 mm away from the fusion zone; a distance not much different from the region of relatively high microhardness shown in Fig. 30. This indicates that local hardening plays an important role in SCC of the HAZ in components such as BWR core shrouds.

Rough surfaces of welded components are frequently smoothed in the field by grinding. During this procedure, the surface region is locally heated and is subject to significant cold work. To simulate the effect of this practice, the free surface of an SMA weld of Type 304 SS Heat C1 was smoothed with a grinding wheel. The microhardness measurements in Fig. 31 show that the grinding procedure introduces a hard subsurface layer ≈ 0.4 mm thick.

Figure 30.

Microhardness of SMA weld HAZ of Type 304 SS Heat C1 as function of distance from fusion zone (Arrow B, Fig. 29). Note hardness maximum at 0.7-1.5 mm.

Figure 31.

Microhardness of HAZ in SMA weld of Type 304 SS, Heat C1, as function of distance (Arrow A, Fig. 29) from weld surface smoothed with a grinding wheel to simulate BWR core shroud weld surface finish

Figure 32 shows the hardening behavior of the ground specimen of the SMA weld of Type 304 SS Heat C1 and of the HAZs of two core shroud weld specimens, i.e., specimens of Type 304 SS from irradiated core shrouds in BWR-Q and -D. Information on the type of steel and approximate fluence level is somewhat limited and the chemical composition of the core shroud materials is not known. The hardness profile of the ground SMA weld of Heat C1 is similar to that of the H5 core shroud weld of the BWR-Q. However, the cold-worked layer of the H5 core shroud weld from BWR-D is significantly thicker, i.e., 0.4 versus 1.3 mm.

Based on the results shown in Figs. 30 and 31, a three-dimensional (3-D) local hardness profile of the ground HAZ of the SMA weld of Heat C1 was constructed (see Fig. 33). The hardness results depicted in the figure suggest that a crack mostly likely would nucleate at the ground surface and propagate parallel to and 1-3 mm away from the boundary between the fusion zone and the HAZ.

3.2.3 Secondary-Ion Mass Spectroscopy of Fluorine and Oxygen in Laboratory Welds

The detection limit of fluorine in the SMA weld of Type 304 SS, Heat C1, by the wet-chemistry method was ≈ 30 wppm when several grams of material were available for analysis, but only ≈ 120 wppm (≈ 360 appm) when the sample mass was < 1 mg. Therefore, fluorine and

Figure 32.
Microhardness profiles of cold-worked HAZs of Type 304 SS, Heat C1, and BWR core shroud welds as function of distance from ground surface

Figure 33.
Vickers microhardness of HAZ in SMA weld of Type 304 SS, Heat C1, as function of distance from ground surface and fusion zone. Note hardness maximum at ≈ 1 mm from fusion zone.

oxygen contamination in the specimen was analyzed by secondary-ion mass spectroscopy (SIMS).⁵³ The SMA weld specimen that contained the fusion zone and the HAZ was polished, etched in a dilute oxalic acid solution, and ultrasonically cleaned in alcohol before SIMS analysis. Secondary ions of fluorine that were ejected from the weld fusion zone and the HAZ were analyzed by using cesium primary ions in one case and oxygen ions in another.

Figure 34 shows distribution maps of fluorine and oxygen secondary ions obtained with cesium primary ions. The figure provides direct evidence that local regions high in oxygen concentration are also high in fluorine concentration. Although the solubility of fluorine and oxygen in steels is very low or negligible, the arc in the weld fumes may promote dissolution of atomic fluorine and oxygen in the fusion zone and the HAZ. This observation is consistent with the results of AES analyses of BWR core shrouds and neutron-absorber tubes, which showed a high fluorine content in oxide or spinel precipitates.⁵⁴ It also seems to be consistent with the observation by Ward et al. that a visible oxide film promotes fluorine-assisted IGSCC.⁵⁵

Figure 34. Fluorine (top) and oxygen (bottom) ion maps of SMA weld of Type 304 SS, Heat C1, determined by SIMS analysis

3.2.4 Summary of Characterization of Simulated Core Shroud Weld

In summary, we believe the following factors exacerbate the susceptibility of the HAZ of a core shroud weld to SCC:

- (a) Oxygen, fluorine, and sulfur contamination that occurs during welding
- (b) Hardening at a maximum distance of ≈ 1 -3 mm from the fusion zone
- (c) Hardened surface layers that are a result of postweld grinding
- (d) Hardening by neutron irradiation that is compounded by hardening effects of (b) and (c)
- (e) Thermal sensitization during welding
- (f) Irradiation-induced grain-boundary depletion of chromium, which is compounded by thermal sensitization.

These factors may be more significant in some materials than in others, e.g., Type 304 versus Type 304L SSs. Additional results after irradiation of the laboratory-produced weld specimens and microstructural and microchemical analyses of BWR core shroud specimens will provide more definitive evidence of the above factors and a more quantitative understanding of the cracking mechanism. Initial results of grain-boundary segregation of fluorine in BWR neutron-absorber tubes and a core shroud weld are given below.

3.3 Analysis of Fluorine in BWR Components (H. M. Chung and J. E. Sanecki)

To explore the role of fluorine in IASCC of SS, the segregation behavior of fluorine to grain boundaries in BWR neutron-absorber tubes and a core shroud weld was analyzed with a scanning Auger microscope (SAM). The composition and fast-neutron fluence of the BWR components are given in Table 14. Auger specimens were charged with hydrogen and fractured in situ in the ultrahigh vacuum of the SAM. Auger electron spectroscopy appeared to be superior to other techniques for determining grain-boundary compositions of most nonmetallic impurities at low concentrations. Derivative AES signals were obtained with a focused electron beam $\approx 1 \mu\text{m}$ in diameter. The grain-boundary profile of a core shroud weld specimen was determined by an automated depth-profiling technique, i.e., repeated argon ion sputtering and acquisition of an AES fluorine signal at 625 eV.⁵⁴

Table 14. Material type, composition (in wt.%), and fluence of BWR components

Heat ID	Composition (wt.%)									Source	Fluence ($10^{21} \text{ n}\cdot\text{cm}^{-2}$)
	Cr	Ni	Mn	C	N	B	Si	P	S		
HP304-A	18.50	9.45	1.53	0.018	0.100	<0.001	<0.03	0.005	0.003	BWR-Pa	0.2-1.4
HP304-B	18.30	9.75	1.32	0.015	0.080	<0.001	0.05	0.005	0.005	BWR-Pa	0.2-1.4
HP304-CD	18.58	9.44	1.22	0.017	0.037	0.001	0.02	0.002	0.003	BWR-Pa	0.2-2.0
CP304-A	16.80	8.77	1.65	0.08 ^b	0.052	-	1.55	0.045 ^b	0.030 ^b	BWR-Vc	0.2-2.0
CP304-B	18-20	8-10	2.00 ^b	0.08 ^b	-	-	1.00 ^b	0.045 ^b	0.030 ^b	BWR-Ld	0.5-2.6
CP348A	17.4	11.5	1.56	0.074	0.042	-	0.34	0.007	0.009	BWR-Ke	≈ 1.5
HP348B	17.7	11.1	1.65	0.041	0.008	-	0.19	0.002	0.007	BWR-Ke	≈ 3.5
507-B	core shroud weld, Type 304L, composition unknown									BWR-C	0.05
CSW-D	core shroud weld, Type 304, composition unknown									BWR-D	≈ 0.1
CSW-Q	core shroud weld, Type 304, composition unknown									BWR-Q	≈ 0.00003

^aHigh-purity neutron absorber tubes, OD = 4.78 mm, wall thickness = 0.63 mm, composition before service.

^bRepresents maximum value in the specification; actual value not measured.

^cCommercial-purity absorber tubes, OD = 4.78 mm, wall thickness = 0.79 mm, composition after service.

^dCommercial-purity control blade sheath, thickness 1.22 mm, actual composition not measured.

^eSwelling mandrel tube irradiated in 7 x 7 fuel assembly in BWR KKP-1.

Initial results indicated that fluorine concentration is higher on grain boundaries than in grain matrices.⁵⁴ Therefore, further analysis was conducted on specimens from neutron-absorber tubes and a core shroud weld. Figure 35 shows the relative intensities of the fluorine signal at 625 eV from ductile and intergranular fracture surfaces of the neutron-absorber tubes. The CP and HP Type 304 SS absorber tubes were fabricated from Heat CP304-A and HP304-A, respectively, and were irradiated to fluence of $\approx 2.0 \times 10^{21} \text{ n cm}^{-2}$ in a BWR. In Fig. 35, average concentrations of fluorine are significantly higher on grain boundaries than in grain matrices, however, absolute concentrations on grain boundaries could not be determined.

Auger electron spectroscopy analysis of grain-boundary distribution of fluorine was also conducted on the Type 304L SS BWR core shroud weld (Specimen 507-B, Table 14). The HAZ of the weld specimen was fractured in situ in the ultrahigh vacuum environment of the SAM, and one of the grain boundaries was selected and analyzed by means of a depth-profiling technique. The fluorine signal (at 625 eV) as a function of distance beneath the grain boundary was plotted in Fig. 36, which shows that fluorine content is significantly higher on grain boundaries than in grain matrices, probably as a result of thermal segregation during cooling of the HAZ. Because of its very low solubility in SSs, fluorine will segregate by a thermal process to local sites such as matrix/precipitate interfaces, grain boundaries, and stacking faults; however, irradiation-induced segregation cannot be ruled out and further investigation will be needed to clarify this point.

Figure 35. Fluorine 625-eV intensities from spots on ductile (grain matrix, denoted "D") and intergranular (grain boundary, denoted "I") fracture surfaces in (a)CP and (b) HP BWR neutron-absorber tubes

Figure 36.
Intensity of fluorine AES signal as function of sputter distance from grain boundary surface in Type 304L core-shroud weld HAZ, Specimen 507-B

3.4 Development of Hot-Cell J-R Test Facility (T. H. Hughes and E. E. Gruber)

3.4.1 Introduction

The equipment for fracture toughness J-R testing of irradiated compact-tension (CT) specimens in a hot cell has been assembled. The design of the test train has been modified to reduce the heat loss from the furnace and to minimize thermal gradients near the specimen. As part of the preliminary checkout of the equipment, a series of out-of-cell fracture toughness J-R tests has been conducted on annealed and nonirradiated Type 304 SS specimens according to the "Standard Test Method for J-integral Characterization of Fracture Toughness," which is currently under consideration within the ASTM technical committee. This procedure is a merger of ASTM Specifications E 813 (Standard Test Method for J_{IC} , a Measure of Fracture Toughness) and E 1152 (Standard Test Method for Determining J-R Curve). In this method, the J-R curve characterizes the resistance of the material to slow stable crack growth after initiation from a preexisting fatigue crack. The J_{IC} value characterizes the toughness of the material near the onset of stable crack extension. The parameter J_C represents a measure of fracture toughness at instability prior to the onset of significant stable crack extension.

The method involves pin loading of precracked fatigue specimens and determination of J as a function of crack growth. Load-versus-load-line displacement is recorded to calculate the J -integral. The calculated J -integral is plotted against estimated or physical crack growth Δa , within specified limits of crack growth. The crack length and crack growth are determined from either elastic compliance measurements or direct current DC electric potential (EP) technique.

3.4.2 Equipment

The test system is designed for in-cell testing with the test train, furnace, and other equipment mounted on a portable, wheeled cart that can be rolled into the cell. A schematic representation of the system is shown in Fig. 37. If required, a small autoclave can be installed inside the furnace for conducting tests in simulated LWR environments. The hydraulic actuator is mounted on top of the frame, with the test train components suspended beneath it. An enlarged drawing and photograph of the actuator, load cell, test train, and furnace are shown in Figs. 38 and 39, respectively. The 22.24 kN (5 kip) load cell is at the top of the pull rod. A "trim" heater has been fitted to the lower end of the pull rod to compensate for axial heat loss from the specimen chamber. This heater is a small 125-watt band heater clamped around an aluminum cylinder that surrounds the pull rod, just above the upper clevis holder. The furnace is mounted on a pneumatic cylinder and can be raised to enclose the load cage and the specimen during the test.

Figure 37. Schematic representation of hot-cell J-R test facility

A detailed drawing of the specimen load cage is shown in Fig. 40. The load cage that contains the specimen consists of two 12.7-mm (0.5-in.)-thick plates separated by four compression rods. The lower two-piece clevis assembly is fastened at the bottom of this rigid cage, with the two sections connected by an oxidized Zircaloy pin that is electrically insulated with mica washers. The same arrangement is used for the upper clevis assembly that is connected to the pull rod. The CT specimen is mounted in the clevises with Inconel pins. Current is passed through the specimen for resistance measurements by means of a ceramic current clip. This spring-loaded clip incorporates platinum contacts that are accurately positioned on the specimen by locating pins inserted in holes in the specimens. Potential leads are secured by short SS pins that are inserted into threaded holes in the specimens. Jigs have been designed and fabricated to facilitate this procedure.

Figure 38. Schematic representation of actuator, load cell, load train, and furnace

3.4.3 Specimen Dimensions and Configuration

The configuration of the SS specimens that are being irradiated in the Halden reactor is shown in Fig. 41. The designs for CT and disk-shaped compact-tension (DCT) specimens that have been used successfully and recommended in ASTM Standards are shown in Fig. 42. Both CT and DCT specimen designs are single-edge-notched fatigue precracked plates loaded in tension. Figure 41 shows that, although the Halden specimens are rectangular, their dimensions are similar to those of the DCT specimen. For example, the span of the load points (center of the pin holes) is $0.275 W$, where W is the specimen width given by the distance from the load line to the back edge of the specimen. However, the specimen height and thickness (side-to-side dimension) are $0.583 W$ and $0.54 W$, respectively, instead of the $0.675 W$ and $0.5 W$ recommended for the DCT specimen design.

Figure 39. Photograph of actuator, load cell, load train, and furnace

Figure 40. Detailed drawing of specimen load cage

Calculations to determine crack length and J-integral were performed with the correlations recommended for DCT specimens. The applicability of the procedure to a rectangular specimen geometry used in the present study will be evaluated.

Figure 41. Configuration of compact-tension specimen in this study

Figure 42. DCT and CT specimen designs from other investigations

3.4.4 Calculations of Crack Length

Results of crack length measurements by the potential drop method have been presented previously.⁵⁶ For a single specimen method that uses the elastic compliance technique on DCT specimens with crack opening displacements measured at the load line, the crack length a is given by

$$a_i/W = 0.998193 - 3.88087u_x + 0.187106u_x^2 + 20.3714u_x^3 - 45.2125u_x^4 + 44.527u_x^5, \quad (13)$$

where

$$u_x = \frac{1}{\left[(B_{ef} E_{ef} C_{ci})^{1/2} + 1 \right]}, \quad (14)$$

$$B_{ef} = B - (B - B_N)^2 / B, \quad (15a)$$

$$E_{ef} = E / (1 - \nu^2), \quad (15b)$$

W is specimen width, B is specimen thickness, B_N is net specimen thickness (distance between roots of the side grooves), E is elastic modulus, ν is Poisson's ratio, and C_{ci} is the corrected specimen elastic compliance. The measured compliance is corrected for rotation of the crack centerline by applying the expression

$$C_{ci} = \frac{C_i}{\left[\frac{H^*}{R} \sin \theta - \cos \theta \right] \left[\frac{D}{R} \sin \theta - \cos \theta \right]}, \quad (16)$$

where

$$\theta = \sin^{-1} \left[\frac{\left(\frac{d_m}{2} + D \right)}{(D^2 + R^2)^{1/2}} \right] - \tan^{-1} \left(\frac{D}{R} \right); \quad (17)$$

C_i is the measured specimen elastic compliance at the load line, H^* is the initial half-span of the load points (i.e., center of the pin holes), R is the radius of rotation of the crack centerline (Fig. 43), given by $(W + a_i)/2$; D is one-half the initial distance between the displacement measuring points, d_m is the total measured load-line displacement; and θ is the angle of rotation of a rigid body element about the unbroken midsection line.

Figure 43.
Elastic compliance correction for
specimen rotation

3.4.5 Calculations of J-Integral

The J-integral is calculated from the load and load-line displacement curves. The total value of J is the sum of the elastic and plastic components J_{el} and J_{pl} .

$$J = J_{el} + J_{pl}. \quad (18)$$

The total area and plastic component of the area $A_{pl(i)}$ at each recorded deflection are computed during the test by summation of the increase in areas for each increment in deflection; the elastic component of deflection is calculated from specimen load-line elastic compliance at each step and subtracted from the total deflection to obtain plastic deflection. The elastic component of J, at a point corresponding to a_i , V_i , P_i on the specimen load-versus-load-line displacement record, is given by

$$J_{el(i)} = \frac{(K_{(i)})^2 (1 - \nu^2)}{E_{ef}}, \quad (19)$$

where the stress intensity $K_{(i)}$ is expressed as

$$K_{(i)} = \left[\frac{P_i}{(BB_N W)^{1/2}} \right] f\left(\frac{a_i}{W}\right), \quad (20)$$

$$f\left(\frac{a_i}{W}\right) = \left[\frac{\left(2 + \frac{a_i}{W}\right)}{\left(1 - \frac{a_i}{W}\right)^{3/2}} \right] \left\{ 0.76 + 4.8\left(\frac{a_i}{W}\right) - 11.58\left(\frac{a_i}{W}\right)^2 + 11.43\left(\frac{a_i}{W}\right)^3 - 4.08\left(\frac{a_i}{W}\right)^4 \right\}, \quad (21)$$

and P_i is the load. The plastic component of J is given by

$$J_{pl(i)} = \left[J_{pl(i-1)} + \left(\frac{\eta_{(i-1)}}{b_{(i-1)}} \right) \frac{A_{pl(i)} - A_{pl(i-1)}}{B_N} \right] \left[1 - \gamma_{(i-1)} \frac{a_{(i)} - a_{(i-1)}}{b_{(i-1)}} \right] \quad (22)$$

where the factors that account for limited crack growth $\gamma_{(i)}$ and for the tensile component of the load $\eta_{(i)}$ are expressed as

$$\eta_{(i-1)} = 2.0 + 0.552 \frac{b_{(i-1)}}{W}, \quad (23)$$

$$\gamma_{(i-1)} = 1.0 + 0.76 \frac{b_{(i-1)}}{W}, \quad (24)$$

and $B_{(i-1)}$ is the remaining ligament (distance from the physical crack front to the back edge of the specimen) at a point $i-1$.

The quantity $A_{pl(i)} - A_{pl(i-1)}$ is the increment of plastic area under the load-versus-load-line displacement record between lines of constant displacement at points $i-1$ and i . The

quantity $J_{pl(i)}$ represents the total crack-growth-corrected plastic J at point i and is obtained by first incrementing the existing $J_{pl(i-1)}$ and then modifying the total accumulated result to account for the crack growth increment. Accurate evaluations of $J_{pl(i)}$ require small uniform crack-growth increments. The plastic area under the load-versus-load-line displacement record is given by

$$A_{pl(i)} = A_{pl(i-1)} + \frac{[P_i + P_{i-1}][V_{pl(i)} - V_{pl(i-1)}]}{2}, \quad (25)$$

where the total and plastic component of load-line displacement $V_{(i)}$ and $V_{pl(i)}$ are expressed as

$$V_{pl(i)} = V_{(i)} - P_i C_{ci}. \quad (26)$$

3.4.6 Fracture Toughness J-R Tests

Two fracture toughness J-R tests have been completed on solution-annealed Type 304 SS at room temperature and at 288°C. The specimens were precracked at loads based on the maximum allowable load P_m given by the relationship

$$P_m = (0.4 \sigma_f B b^2)/(2W + a), \quad (27)$$

where the flow stress σ_f is the mean of the yield and ultimate stress. The final fatigue precrack extension (≈ 1 mm) was carried out at loads $\leq P_m$ or at a load such that the ratio of the maximum stress intensity applied during fatigue precracking to the elastic modulus (K_{max}/E) was $\leq 1.6 \times 10^{-4} \text{ m}^{1/2}$ ($\leq 0.001 \text{ in.}^{1/2}$).

The fatigue precracked specimens were loaded at a constant extension rate and the tests were interrupted periodically to determine crack length. For crack length measurements, the specimen was either held at constant extension to measure crack length by the EP technique, or was unloaded up to $0.5 P_m$ and then reloaded to measure crack length by the elastic compliance method. For most steels, load relaxation occurs during the hold period or during unloading, which causes a time dependent nonlinearity in the unloading curve. Consequently, prior to unloading or EP measurements, the specimen was held for ≈ 1 min to allow for load relaxation. Furthermore, because specimen extension was monitored and controlled away from the specimen in the present method, the specimen was strained during the hold period because of load relaxation. Figure 44a shows that, for a load relaxation of ≈ 0.1 kN (≈ 22 lbs), the specimen will extend ≈ 0.01 mm. The load-line displacement was recorded to calculate the J-integral according to the procedure described in the previous section.

The specimen extension was monitored and controlled outside of the high-temperature zone. The displacement of load points (center of the loading pins) was determined by subtracting the machine compliance from the measured extension. The curves for machine compliance, expressed in terms of load-versus-displacement (of the loading points) at room temperature and 288°C, are shown in Fig. 44a. The displacement consists of elastic and some plastic extension. The latter most likely occurred in and around the pin holes. The plastic component of the machine compliance at 288°C is shown in Fig. 44b. Because of plastic deformation, machine compliance during unloading was different from the loading compliance.

Figure 44. (a) Machine compliance expressed in terms of load-versus-displacement of loading points at room temperature and 288°C and (b) plastic component of machine compliance at 288°C

The results indicate that, at both room temperature and 288°C, the unloading compliance curve varies with the maximum applied load. The unloading compliance curves at 288°C and maximum loads of 667 and 2891 N (150 and 650 lbs) are shown in Fig. 45.

The actual displacement of load points was measured optically and compared with the estimated load-line displacement. The measured and estimated values of load-line displacement for the room-temperature J-R tests (Fig. 46) show very good agreement. For load-line displacements up to 2 mm, the error in the estimated values is <0.02 mm.

Both DC potential-drop and elastic unloading compliance techniques were used to determine crack length and extension. The measured and estimated crack lengths during precracking of one of the Type 304 SS specimens are shown in Fig. 47. Both methods yield very good estimates of fatigue crack length.

The load-versus-load-line displacement curves and fracture toughness J-R curves from the two tests are shown in Figs. 48 and 49, respectively. The specimens were broken after the test and the initial crack length and physical crack extension were determined by the 9/8 averaging technique, i.e., the two near-surface measurements were averaged and the resultant value was averaged with the remaining seven measurements. For one of the specimens, the final crack size was marked by heat tinting and for the other, it was marked by fatigue cycling. The elastic compliance was corrected by using the measured initial crack size.

At room temperature and 288°C, the J-R curves determined by the elastic compliance method show good agreement with the data obtained from 1T CT specimens; the data points follow the $4\sigma_f$ blunting line. However, at both temperatures, the estimated final crack length is significantly lower than the optically measured crack length. The difference most likely arises from the effects of crack closure during unloading. For relatively tough materials, such as solution-annealed Type 304 SS, crack extension occurs by ductile tearing. Small misalignment in the specimen loading would cause the fracture surfaces to touch each other during unloading, which results in lower apparent crack lengths.

Figure 45. Unloading machine compliance at 288°C for maximum loads of 667 and 2981 N

Figure 46. Measured and estimated values of load-line displacement for solution-annealed and 50% cold-worked Type 304 SS tested at room temperature

Figure 47.
Measured and estimated values of crack length of solution-annealed Type 304 SS specimen precracked at room temperature

Figure 48. Load-versus-load-line displacement curves for solution-annealed Type 304 SS at room temperature and 288°C

Figure 49. Fracture toughness J-R curves for solution-annealed Type 304 SS at room temperature and 288°C

Although the final crack lengths estimated by the potential-drop technique agree with the optically measured crack lengths, the J-R curves do not follow the blunting line. The results show significant crack extension during the initial stages of loading that is typically characterized by crack blunting in SSs. Crack extensions estimated by either the elastic compliance or the EP techniques can be adjusted so they agree with the measured physical crack size. The results will be validated by conducting fracture toughness J-R curve tests at room temperature and 288°C on two heats of thermally aged CF-8M cast SS and on 50% cold-worked Type 316NG SS, and performing fatigue crack growth studies on the same three materials at 288°C in air.

4 Environmentally Assisted Cracking of Alloys 600 and 690 in Simulated LWR Water (W. E. Ruther, W. K. Soppet, and T. F. Kassner)

The objective of this work is to evaluate the resistance of Alloys 600 and 690 to EAC in simulated LWR coolant environments. High-nickel alloys have experienced general corrosion

(tube wall thinning), localized intergranular attack (IGA), and SCC in LWRs. Secondary-side IGA* and axial and circumferential SCC** have occurred in Alloy 600 tubes at tube support plates in many steam generators. Primary-water SCC of Alloy 600 steam generator tubes in PWRs at roll transitions and U-bends and in tube plugs*** is a widespread problem that has been studied intensively. Cracking has also occurred in Alloy 600 and other high-nickel alloys (e.g., Inconel-82 and -182 and Alloy X750) that are used in applications such as instrument nozzles and heater thermal sleeves in the pressurizer† and the penetrations for control-rod drive mechanisms in reactor vessel closure heads in the primary system of PWRs;†† in dissimilar-metal welds between SS piping and low-alloy steel nozzles, in jet pump hold-down beams,††† and in shroud-support-access-hole covers§ in BWRs. Alloy 600, in general, undergoes differing thermomechanical processing for applications other than those used for steam generator tubes. Because environmental degradation of the alloys in many cases is very sensitive to processing, further evaluation of even SCC is needed. In addition, experience strongly suggests that materials that are susceptible to SCC are also susceptible to environmental degradation of fatigue life and fatigue-crack growth properties. In this investigation, we have obtained information on the effect of temperature, load ratio R, and stress intensity (K) on EAC of Alloys 600 and 690 in simulated BWR and PWR water.

4.1 Experimental Methods for Measuring CGR in Alloy 600

The experimental procedures relevant to our mechanical loading systems are similar to those in ASTM E647. The tests are performed under controlled loading conditions with closed-loop servo-controlled machines (MTS™ systems). The contribution to the load that arises from the pressure difference between the inside and the outside of the autoclave is taken into consideration. Additional mechanical contributions that arise from friction between the pull rod in the load train and the autoclave pressure seal are typically <0.4% of the mechanical loading force under autoclave operating pressure and temperature.

The tests in air are performed in a system that is also used for tests in simulated reactor coolant environments. Air is circulated at a slow rate through the autoclave at atmospheric pressure to help maintain a uniform temperature within the autoclave. For tests in air, there is no mechanical contribution from pressure differences between the inside and the outside of

* USNRC Information Notice No. 91-67, "Problems with the Reliable Detection of Intergranular Attack (IGA) of Steam Generator Tubing," Oct. 1991.

** USNRC Information Notice No. 90-49, "Stress Corrosion Cracking in PWR Steam Generator Tubes," Aug. 1990; Notice No. 91-43, "Recent Incidents Involving Rapid Increases in Primary-to-Secondary Leak Rate," July 1991; Notice No. 92-80, "Operation with Steam Generator Tubes Seriously Degraded," Dec. 1992; Notice No. 94-05, "Potential Failure of Steam Generator Tubes with Kinetically Welded Sleeves," Jan. 1994.

*** USNRC Information Notice No. 89-33, "Potential Failure of Westinghouse Steam Generator Tube Mechanical Plugs," March 1989; Notice No. 89-65, "Potential for Stress Corrosion Cracking in Steam Generator Tube Plugs Supplied by Babcock and Wilcox," Sept. 1989; Notice No. 94-87, "Unanticipated Crack in a Particular Heat of Alloy 600 Used for Westinghouse Mechanical Plugs for Steam Generator Tubes," Dec. 1994.

† USNRC Information Notice No. 90-10, "Primary Water Stress Corrosion Cracking (PWSCC) of Inconel 600," Feb. 1990.

†† USNRC Generic Letter 97-01: "Degradation of Control Rod Drive Mechanism and Other Vessel Closure Head Penetrations," Apr. 1, 1997; USNRC Information Notice No. 96-11, "Ingress of Demineralizer Resins Increases Potential for Stress Corrosion Cracking of Control Rod Drive Mechanism Penetrations," Feb. 1996; INPO Document SER 20-93 "Intergranular Stress Corrosion Cracking of Control Rod Drive Mechanism Penetrations," Sept. 1993.

††† USNRC Information Notice No. 93-101, "Jet Pump Hold-Down Beam Failure," Dec. 1993.

§ USNRC Information Notice No. 92-57, "Radial Cracking of Shroud Support Access Hole Cover Welds," Aug. 1992.

the autoclave or from friction in the load train, because there is no seal between the pull rod and the autoclave.

A DC potential-drop measurement system is used for both the tests in air and those in an HP water environment. The specimens in the load train are electrically insulated from the autoclave and each other by using oxidized Zircaloy 705 pins (oxidized by heating at 550°C for 4-8 h) to attach the specimens to the load train. The resistance across the crack is measured by the potential drop for each specimen as a computer automatically controls a programmable DC power supply to direct the electrical current first in one direction and then in the reverse direction. The DC and drops in potential are measured by high-accuracy, stable, scanning multimeters that are controlled by the computer over an IEEE-488 interface. Hundreds of measurements are averaged for one reading, which is transformed to a crack length by a software algorithm that is integral to the control computer. Corrosion-fatigue specimens are held at operating conditions without mechanical loading until the resistance of the specimens stabilizes, typically one week, before the initial fatigue loading at $R = 0.4$ is applied. Resolution of the crack length varies somewhat from one specimen to another and depends on the alloy composition and uniform crack length; a typical value is ± 0.05 mm (0.002 in.) for Alloy 600 or 690.

When tests are conducted in a simulated PWR environment (boric acid, lithium hydroxide and a hydrogen overpressure), the DC potential-drop system, which indicates negative CGRs, fails shortly after the cracks begin to grow. Our best explanation at present is that hydrogen enters the metal matrix and changes the intrinsic resistance of the specimen. For CGR tests in simulated PWR water, a compliance technique that was used previously, was upgraded to perform measurements without interrupting the corrosion-fatigue test.

The compliance measurement system is composed of MTS™ Model 632.10 hermetically sealed crack-opening-displacement (COD) gages that are interfaced to a computer data acquisition system along with a conditioned load cell signal. The COD gages are calibrated to an accuracy of $\pm 1\%$ at the desired test temperature prior to each test with a micrometer fixture traceable to a National Institute Standard. The resulting typical COD gage output transfer function is 0.075 mm/V. Experience with compliance measurements with MTS™ clip gages indicates that the resolution is ± 0.025 mm (0.001 in.). The computer data acquisition system accepts the multiplexed load and COD voltage signals with a resolution of $\pm 0.05\%$ of the applied voltage range. Compliance data from the load cell and COD gages are acquired during the unloading phase of the ramp waveform. Initially, the load-versus-COD data are graphically analyzed to select the region of the waveform with the steepest linear slope to optimize the data acquisition period. The compliance value for a test cycle is computed by a least-squares linear-regression analysis of 250 data samples. The computer software monitors the fit of the linear regression curve for a minimum correlation coefficient of 0.999. During a corrosion-fatigue test, ten successive load-line compliance values are recorded at 1-h time intervals, averaged, and applied to a subroutine that calculates crack length by means of a polynomial elastic compliance expression derived for CT and wedge-opening-load specimens.⁵⁷ However, for conditions of accelerated crack growth, the measurement time interval can be reduced to 10 min.

Both the initial crack length and the crack length and shape at the end of the test are determined from the fracture surface for comparison with measurements inferred from COD or DC potential-drop measurements. The specimen is cut into two pieces along the midplane of

the thickness and the crack in one piece is extended by fracturing at a cryogenic temperature. One of these segments is used to examine the fracture surface by SEM. Excess metal in the noncracked region the other piece is removed with a cutoff wheel and a transverse section is prepared and polished for metallographic evaluation of the region that contains the corrosion-fatigue crack to confirm the nature of the crack path and determine whether crack branching had occurred.

Water for the CGR test systems is obtained from a reverse osmosis purification system and is passed through a demineralizer with mixed-bed ion-exchange resins; the conductivity is $<0.1 \mu\text{S}\cdot\text{cm}^{-1}$. To avoid contamination by nonionic organic contaminants, the water flows through a final water purification system (e.g., a Millipore Super-Q Ultrapure Water System™ coupled with an Ultra Dynamics UV Purifier™) that contains a bed of activated charcoal, an Organex bed, a resin bed, and a $0.2\text{-}\mu\text{m}$ filter. The purified water is stored in 135-liter SS feedwater tanks that are refilled weekly. The feedwater tanks are connected to the autoclave systems with SS tubing. Whenever the conductivity of the stored feedwater solution increases to $\geq 0.1 \mu\text{S}\cdot\text{cm}^{-1}$ during the storage period, the tanks are spray sterilized/passivated with a 10% hydrogen peroxide solution and thoroughly rinsed with deionized water prior to refilling. The DO level is adjusted by vacuum sparging of the filled tanks; nitrogen or hydrogen is used for low-DO tests and oxygen-nitrogen gas mixtures are used to establish predetermined DO levels. Final adjustment of the effluent DO in the latter tests is achieved by slightly varying the storage tank pressure and/or the flow rate of water in the autoclave. The DO concentration in the effluent stream from the autoclaves is monitored by Chemetrics™ ampoules and/or by an Orbisphere™ DO monitor. The electrochemical potential of platinum and SS electrodes is measured versus a room-temperature 0.1N KCl/Ag/AgCl reference electrode located in the high-temperature effluent stream (289°C) from the autoclave. The platinum potential, in particular, responds rapidly to changes in DO concentrations in simulated LWR coolant water.

4.2 Tensile Properties of Solution-Annealed and Thermally Treated Alloy 600

The composition, microstructure, and tensile properties of heats of Alloys 600 and 690 that have been used in the fatigue-crack-growth investigation were reported previously.^{58,59} A heat of Alloy 600 (NX9244G) with a low-carbon content (0.03 wt.%) was obtained for CGR measurements in simulated LWR environments. The alloy was produced by INCO Alloys International, Huntington, WV, for the Kobe Material Testing Laboratory Co. Ltd., Hyogo, Japan. The plates were heat treated at the Kobe Material Testing Laboratory and cut into small blocks ($\approx 32 \times 65 \times 140 \text{ mm}$) for distribution to various laboratories that are participating in the International Crack Growth-EAC investigation of Alloy 600 in simulated BWR and PWR environments. The composition of the material is given in Table 15. Compact-tension specimens were fabricated in which the orientation of the crack plane in the specimens corresponds to the identification code T-L for plates in ASTM Specification E399. In our previous CGR experiments on other heats of Alloys 600 and 690, the CT specimens were in the L-T orientation.

Tensile specimens were also obtained from the blocks and from several of the CT specimens after the CGR tests were completed. The tensile properties of the material at 25, 290, and 320°C in four heat-treatment conditions, hardness at room temperature, and grain size are given in Table 16. Two blocks of material were solution annealed at 1025 and 1115°C for 2 h and two blocks that had been subjected to these solution heat treatments were

thermally treated at 600°C for 24 h. Solution heat treatments at 1025 and 1115°C produced ASTM grain sizes of ≈ 1.0 -1.5 and 0.2-0.4, respectively, which correspond to average grain diameters of ≈ 250 -210 and 340-310 μm (Fig. 50). The ultimate tensile strengths at 290°C for materials that were solution annealed at 1025 and 1115°C are ≈ 550 and 513 MPa, respectively. The yield strength was 156 MPa for three heat-treatment conditions and was slightly higher (173 MPa) for material that was solution annealed at 1025°C and thermally treated. Because of the low carbon content and relatively high solution annealing temperatures, intergranular carbides were not observed in the specimens by optical microscopy and SEM.

Figure 51 shows the dependence of yield stress on grain size for Heat NX9244G (≈ 0.03 wt.% carbon) in the solution-annealed and solution-annealed and thermally treated conditions and Heat NX8844 (0.08 wt.% carbon) in the solution-annealed condition. The latter heat was used in corrosion-fatigue experiments that were reported previously.⁶⁰ Heat treatment at 600°C for 24 h has virtually no effect on the tensile properties when compared with material that was solution-annealed at the two temperatures. The data follow a Petch relationship, i.e., $\sigma_y = \sigma_i + k \cdot d^{-1/2}$, where σ_y is the yield stress; d is the grain diameter; k , an empirical constant; and σ_i , the "friction" stress, which is a measure of intrinsic resistance of the material for dislocation motion. The dependence in Fig. 51 may be somewhat fortuitous because the decrease in tensile stress for the material with the large grain size (Heat NX9244G), in part, may be caused by the low carbon content of the material.

Figure 50.
Relationship between ASTM grain size number and average grain diameter

Figure 51.
Dependence of 0.2% yield stress at 25 and 290°C on grain size of solution-annealed and thermally treated Alloy 600 specimens

Table 15. Composition of Alloy 600 used for corrosion fatigue tests

Material	Heat No.	Heat Analysis	Composition (wt.%)															
			Cr	Mo	Ni	Fe	Mn	Si	C	N	P	S	B	Cu	Ti	Al	Co	Nb+Ta
Alloy 600	NX8844	Vendor	14.97	0.15	75.21	8.26	0.26	0.24	0.069	0.01	0.009	<0.001	0.004	0.22	0.29	0.27	0.04	-
	NX8844J-26	ANL	15.00	0.16	74.94	8.14	0.23	0.32	0.060	0.0155	0.014	0.002	0.004	0.22	0.24	0.24	0.03	0.03
Alloy 600	NX9244G	Vendor	15.51	-	75.32	7.90	0.23	0.25	0.032	0.01	0.007	<0.001	0.001	0.09	0.39	0.20	-	0.02
	NX9244G-J31	ANL	15.47	0.13	75.40	7.73	0.23	0.37	0.040	0.0052	0.010	0.001	0.001	0.07	0.40	0.17	0.06	<0.01
	NX9244G-J32	ANL	16.39	0.11	75.50	7.44	0.22	0.35	0.030	0.0067	0.011	0.001	0.001	0.08	0.40	0.17	0.06	0.01

Table 16. Tensile properties of Alloy 600 under various heat-treatment conditions

Alloy 600 Heat No.	Material Condition	Test No.	Spec. No.	Temp. (°C)	σ_u (MPa)	σ_y (MPa)	ϵ_t (%)	RA (%)	Hardness ^a (VN)	Hardness (Rb)	ASTM Grain Size
NX8844B-33	Annealed 872°C/1 h	-b	-b	25	748.8	339.9	35.5	-	-	90	7.5
NX8844B-33	Annealed 872°C/1 h	T19	B33-05 ^c	25	714.9	333.2	39.6	66.9	190	91	8
NX8844B-33	Annealed 872°C/1 h	T8	B33-03 ^c	290	686.1	282.6	38.6	61.1	-	-	-
NX8844B-33	Annealed 872°C/1 h	T10	B33-04 ^c	320	680.6	282.6	39.1	55.5	-	-	-
NX8844J-26	Annealed 1038°C/1 h	-b	-b	25	694.3	298.6	41.0	-	-	86	4
NX8844J-26	Annealed 1038°C/1 h	T21	J26-05 ^c	25	653.5	245.5	49.2	61.1	173	87	4
NX8844J-26	Annealed 1038°C/1 h	T23	J26-06 ^c	290	637.8	234.0	45.2	53.3	-	-	-
NX8844J-26	Annealed 1038°C/1 h	T44	J26-08 ^c	290	626.5	218.0	48.7	57.2	-	-	-
NX8844J-26	Annealed 1038°C/1 h	T25	J26-07 ^c	320	639.4	246.8	45.8	48.9	-	-	-
NX9244G	Mill Annealed	-b	-b	25	657.0	300.0	43.9	60.6	-	83	-
NX9244G-J320	Annealed 1025°C/2 h	T43J320-04	^c	25	580.6	205.1	62.7	82.0	145	78	1.0
NX9244G-J320	Annealed 1025°C/2 h	T38J320-02	^c	290	547.9	155.7	68.4	72.2	-	-	-
NX9244G-J320	Annealed 1025°C/2 h	T39J320-03	^c	320	551.9	154.5	69.9	70.1	-	-	-
NX9244G-J321	Annealed 1025°C/2 h + 600°C/24 h	T32J321-03	^c	25	582.0	208.4	59.5	66.1	148	79	1.5
NX9244G-J321	Annealed 1025°C/2 h + 600°C/24 h	T35J321-05	^c	290	552.1	172.5	65.2	59.8	-	-	-
NX9244G-J321	Annealed 1025°C/2 h + 600°C/24 h	T34J321-04	^c	320	551.6	166.9	67.0	64.4	-	-	-
NX9244G-J310	Annealed 1115°C/2 h	T40J310-01	^c	25	551.7	188.7	58.7	77.5	144	78	0.4
NX9244G-J310	Annealed 1115°C/2 h	T41J310-02	^c	290	508.9	155.7	65.5	68.8	-	-	-
NX9244G-J310	Annealed 1115°C/2 h	T42J310-03	^c	320	507.5	146.1	67.7	69.5	-	-	-
NX9244G-J311	Annealed 1115°C/2 h + 600°C/24 h	T30J311-03	^d	25	545.7	194.5	57.2	61.5	-	-	-
NX9244G-J311	Annealed 1115°C/2 h + 600°C/24 h	T31J311-04	^c	25	553.1	196.4	56.8	63.0	142	77	0.2
NX9244G-J311	Annealed 1115°C/2 h + 600°C/24 h	T33J311-05	^c	290	517.0	155.7	66.7	65.9	-	-	-
NX9244G-J311	Annealed 1115°C/2 h + 600°C/24 h	T36J311-06	^c	320	514.9	155.7	65.3	60.6	-	-	-

^aVickers hardness at room temperature, 500 gf, 15 s.^bResults from vendor (EPRI document or certified material test reports).^cTensile tests conducted in air at a strain rate of $1.0 \times 10^{-4} \text{ s}^{-1}$.^dTensile test conducted in air at a strain rate of $1.0 \times 10^{-5} \text{ s}^{-1}$.

4.3 Crack Growth Rates of Low-Carbon Alloy 600 in HP Oxygenated Water at 289°C

Corrosion-fatigue experiments were conducted on Alloy 600 specimens in HP oxygenated (≈ 300 ppb DO) water at 289°C to determine the effects of heat treatment conditions, load ratio, and stress intensity on CGRs. Two specimens from a heat of low-carbon Alloy 600 (NX9244G), both solution-annealed at 1025°C for 2 h and one thermally treated at 600°C for 24 h, were tested in one autoclave system; another autoclave contained two specimens from this heat, both of which were solution annealed at 1115°C for 2h and one that was also thermally treated at 600°C for 24 h. The experimental conditions and results from 14 CGR tests on each set of specimens are given in Tables 17 and 18. Crack growth was determined by the DC potential-drop method. The experiments were conducted at load ratios R between 0.2 and 0.95 and maximum stress intensity values $K_{\max}$ between ≈ 30 and 39 $\text{MPa}\cdot\text{m}^{1/2}$. The frequency and rise time T_r of the positive sawtooth waveform were 8×10^{-2} Hz and 12 s, respectively. This range of R and $K_{\max}$ values produced $\Delta K = K_{\max} (1 - R)$ values of ≈ 1.6 -31 $\text{MPa}\cdot\text{m}^{1/2}$ and CGRs between $\approx 1 \times 10^{-11}$ and 2×10^{-8} $\text{m}\cdot\text{s}^{-1}$. The dependence of the CGRs on ΔK from the results in Tables 17 and 18 is shown in Figs. 52 and 53, respectively. The dependence on ΔK of the CGR for each specimen is given by a power-law relationship; the coefficient-of-fit indicated on the figures is quite good.

Figure 52.

Dependence of CGRs in oxygenated HP water at 289°C on ΔK of Alloy 600 specimens solution-annealed at 1025°C and solution-annealed at this temperature and thermally treated at 600°C for 24 h

Figure 53.

Dependence of CGRs in oxygenated HP water at 289°C on ΔK of Alloy 600 specimens solution-annealed at 1115°C and solution-annealed at this temperature and thermally treated at 600°C for 24 h

Table 17. Crack growth results at 289°C in HP water with ~300 ppb DO for Alloy 600 specimens^a annealed at 1025°C

Test No.	Test Time (h)	Water Chemistry			Electrode Potential			Alloy 600 (Annealed)			Alloy 600 (TT)			
		O ₂ ^b Conc. (ppb)	Cond. at 25°C (μS·cm ⁻¹)	pH at 25°C				Specimen No. J320-02			Specimen No. J321-02			
		304 SS	Pt	Load Ratio ^c	K _{max} ^d (MPa·m ^{1/2})	ΔK ^e (10 ⁻¹⁰ m·s ⁻¹)	Rate (10 ⁻¹⁰ m·s ⁻¹)	K _{max} ^d (MPa·m ^{1/2})	ΔK ^e (10 ⁻¹⁰ m·s ⁻¹)	Rate (10 ⁻¹⁰ m·s ⁻¹)				
					mV(SHE) @ 289°C									
1	836-868	310	0.07	6.90	118	117	0.45	31.2	17.16	23.3	31.0	17.05	12.6	
2	870-888	310	0.07	6.94	123	118	0.40	31.7	19.02	41.8	31.2	18.72	15.6	
3	890-908	230	0.07	6.94	123	118	0.35	32.3	20.99	51.6	31.3	20.37	15.3	
4	1028-1051	300	0.07	6.86	-	-	0.30	34.1	23.84	117.5	31.8	22.23	27.2	
5	1052-1075	310	0.07	6.86	-	-	0.50	34.8	17.42	44.2	32.0	16.00	14.3	
6	1218-1228	260	0.06	6.55	118	111	0.20	36.2	28.95	194.4	32.4	25.95	75.1	
7	1228-1245	280	0.06	6.55	124	118	0.60	36.5	14.60	26.7	32.6	13.03	14.0	
8	1250-1272	300	0.06	6.55	135	127	0.75	36.8	9.21	12.4	32.7	8.19	7.45	
9	1272-1417	300	0.06	6.55	146	140	0.90	37.0	3.70	1.12	32.9	3.29	1.25	
10	1417-1682	290	0.06	6.55	152	143	0.95	37.1	1.85	0.10	33.0	1.65	0.48	
11	1682-1701	260	0.06	6.55	152	140	0.70	37.3	11.19	22.4	33.2	9.95	15.5	
12	1701-1746	270	0.06	6.55	155	143	0.83	37.5	6.38	5.64	33.3	5.66	3.21	
13	1746-1755	290	0.06	6.55	158	146	0.20	38.3	30.64	140.7	33.6	26.88	49.8	
14	1756-1772	320	0.06	6.55	166	155	0.65	39.0	13.67	14.4	33.8	11.83	9.81	

^a Compact tension specimens (1TCT) of Alloy 600, Heat No. NX9244G. The specimens (Nos. J320-02 and J321-02) were tested in the solution-annealed (1025°C for 2h) and solution-annealed (1025°C for 2h) and thermally treated (600°C for 24h) conditions, respectively. Yield stresses of the materials at 289°C for these heat-treatment conditions are 156 and 173 MPa, respectively.

^b Effluent DO concentration was determined with Chemetrics ampules and an Orbisphere Oxygen meter.

^c Frequency and rise time of the positive sawtooth waveform were 8×10^{-2} Hz and 12 s, respectively.

^d Stress intensity, K_{max}, values at the end of the time period.

^e $\Delta K = K_{max}(1 - R)$, where load ratio $R = K_{min}/K_{max}$.

Table 18. Crack growth results at 289°C in HP water with ≈300 ppb DO for Alloy 600 specimens^a annealed at 1115°C

Test No.	Test Time (h)	Water Chemistry			Electrode Potential			Alloy 600 (Annealed)			Alloy 600 (TT)		
		O ₂ ^b	Cond.	pH at				Specimen No. J310-02			Specimen No. J311-02		
		Conc.	at 25°C	25°C	304 SS	Pt	Load Ratio ^c	K _{max} ^d	ΔK ^e	Rate	K _{max} ^d	ΔK ^e	Rate
		(ppb)	(μS · cm ⁻¹)		mV(SHE) @ 289°C			(MPa · m ^{1/2})		(10 ⁻¹⁰ m · s ⁻¹)	(MPa · m ^{1/2})		(10 ⁻¹⁰ m · s ⁻¹)
1	1-28	300	0.09	6.30	145	112	0.50	29.6	14.80	22.9	29.6	14.80	22.4
2	30-70	300	0.09	6.30	150	122	0.60	30.0	12.00	15.7	29.9	11.90	13.6
3	70-143	310	0.09	6.30	151	124	0.70	30.4	9.12	9.87	30.2	9.07	9.18
4	143-191	325	0.09	6.30	152	124	0.75	30.5	7.63	6.34	30.4	7.60	6.07
5	395-480	300	0.07	6.25	151	129	0.90	31.1	3.11	0.89	31.0	3.10	0.82
6	482-574	290	0.07	6.25	151	129	0.85	31.2	4.67	1.37	31.1	4.67	2.08
7	575-646	310	0.07	6.25	155	134	0.80	31.3	6.26	3.05	31.3	6.26	4.18
8	1200-1230	300	0.12	6.76	153	142	0.45	31.8	17.49	20.9	31.6	17.38	18.2
9	1367-1392	300	0.12	6.76	154	137	0.40	32.4	19.44	35.3	32.1	19.26	25.4
10	1393-1414	305	0.12	6.76	153	147	0.35	33.0	21.45	46.3	32.4	21.06	25.3
11	1415-1484	300	0.12	6.76	156	144	0.68	33.7	10.78	11.8	32.7	10.46	6.92
12	1585-2208	320	0.07	6.78	149	140	0.95	33.8	1.69	0.78	32.8	1.64	0.13
13	2208-2236	300	0.07	6.78	158	133	0.30	35.2	24.64	78.8	33.7	23.59	46.3
14	2237-2244	280	0.07	6.78	150	131	0.20	36.4	29.12	236.0	34.4	27.52	132.0

^a Compact tension specimens (1TCT) of Alloy 600, Heat No. NX9244G. The specimens (Nos. J310-02 and J311-02) were tested in the solution-annealed (1115°C for 2h) and solution-annealed (1115°C for 2h) and thermally treated (600°C for 24h) conditions, respectively. Yield stress of the material at 289°C for both heat-treatment conditions is 156 MPa.

^b Effluent dissolved -oxygen concentration was determined with Chemetrics ampules and an Orbisphere Oxygen meter.

^c Frequency and rise time of the positive sawtooth waveform were 8×10^{-2} Hz and 12 s, respectively.

^d Stress intensity, K_{max}, values at the end of the time period.

^e $\Delta K = K_{max}(1 - R)$, where load ratio $R = K_{min}/K_{max}$.

Although the crack growth behavior of the specimens appears quite similar, the results for the solution-annealed and thermally treated specimens from the two experiments are shown in Figs. 54 and 55, respectively. Figures 52-55 indicate that different solution annealing temperatures (1025 versus 1115°C) and thermal treatment at 600°C for 24 h do not produce large variations in CGRs of this heat of low-carbon, large-grain-size material in oxygenated water at 289°C. The combined data from the four specimens in the two experiments are shown in Fig. 56. Differences in the rates mainly occur at ΔK values $>20 \text{ MPa}\cdot\text{m}^{1/2}$, which correspond to tests at low load ratios ($R < 0.4$).

Figure 54.

Dependence of CGRs in oxygenated HP water at 289°C on ΔK of solution-annealed Alloy 600 specimens

Figure 55.

Dependence of CGRs in oxygenated HP water at 289°C on ΔK of thermally treated Alloy 600 specimens

Figure 56.

Combined data for dependence on ΔK of CGR of low-carbon heat of Alloy 600 in HP oxygenated water at 289°C

4.4 Comparison of CGRs of Alloy 600 in HP Oxygenated Water and Air

The CGRs in HP oxygenated water of specimens from Heat NX9244G with different heat-treatment conditions can be compared with CGRs in air from another heat of Alloy 600 (NX8197), which were reported previously (Fig. 57).⁶⁰ The most useful comparison would be for rates in water and in air, determined from the same heat of material with the same heat treatment; however, the present results preclude such a comparison. Nevertheless, a ratio of the CGRs for each specimen in water to those in air, or from the combined data from several specimens with differing heat treatments to those in air can be obtained from the "best-fit" correlations for the CGR-versus- ΔK data. The results of several such comparisons in water (Figs. 52, 53, and 56) and in air (Fig. 57) are shown in Figs. 58-60, respectively.

The ratios of the CGRs for Alloy 600 in water and air at 289°C (Figs. 58-60) indicate that the rates are higher in water than in air (ratio >1) at ΔK values <18 MPa·m^{1/2} and are lower in water than in air (ratio <1) at higher values of ΔK . For high load ratios (i.e., $R \geq 0.95$) and a K_{max} of ≈ 40 MPa·m^{1/2}, which correspond to ΔK values of ≤ 2 MPa·m^{1/2}, the rates in water are higher than in air by a factor of ≈ 10 . Extrapolation of the curves to lower ΔK values, i.e., higher load ratios that approach constant load ($R = 1.0$), implies even greater enhancement of the rates in water; however, this extrapolation leads to considerable uncertainty because the measured rates in both water and air approach the sensitivity of the crack length measurement system under these loading conditions, even in long-term tests (≈ 1000 h).

Figure 57.
Dependence on ΔK of CGR of Alloys 600 and 690 in air at 289°C

Figure 58.
Dependence on ΔK of ratio of CGRs of solution-annealed (1025°C) and thermally treated Alloy 600 in HP oxygenated water to CGRs of mill-annealed Alloy 600 in air at 289°C

Figure 59.

Dependence on ΔK of ratio of CGRs of solution-annealed (1115°C) and thermally treated Alloy 600 in HP oxygenated water to CGRs of mill-annealed Alloy 600 in air at 289°C

Figure 60.

Dependence on ΔK of ratio of CGRs in HP oxygenated water of a low-carbon heat of Alloy 600 with differing heat treatments to CGRs in air of mill-annealed Alloy 600 at 289°C

Crack growth experiments are being conducted in HP oxygenated water under constant load at $K_{\max}$ of $<40 \text{ MPa}\cdot\text{m}^{1/2}$ to determine the degree of rate enhancement as ΔK approaches zero.

In Fig. 60, as ΔK increases from $\approx 18 \text{ MPa}\cdot\text{m}^{1/2}$, the ratio of the CGRs in water and air is <1 , and at a ΔK of $40 \text{ MPa}\cdot\text{m}^{1/2}$, the rate in water is lower than in air by a factor of ≈ 3 , based on the combined data from the four specimens. This behavior was observed for Alloy 600 specimens in HP deoxygenated water ($<5 \text{ ppb DO}$) at 289°C, as shown in Fig. 61.⁶⁰ The CGRs of Alloy 600 are lower in deoxygenated water than in air at ΔK values $>6 \text{ MPa}\cdot\text{m}^{1/2}$, and at lower ΔK , the rates in water are higher than in air by, at most, a factor of ≈ 2 , in contrast to a factor of ≈ 10 in water that contains $\approx 300 \text{ ppb DO}$. The results in Figs. 60 and 61 show the interdependence of loading conditions and DO in HP water and CGRs of Alloy 600, albeit for different heats of material with differing carbon contents and heat treatments. These figures also illustrate that generalization of CGR results from a single loading condition and DO level in water to other loading and/or DO conditions can be misleading. For example, at a ΔK of $18 \text{ MPa}\cdot\text{m}^{1/2}$, the CGRs of Alloy 600 are the same in air as in water that contains 300 ppb DO at 289°C, whereas at this ΔK , the rate is higher by a factor of 2 in air than in water that contains $<5 \text{ ppb DO}$. Similarly, at a ΔK of $6 \text{ MPa}\cdot\text{m}^{1/2}$, the CGRs are the same in air and in water that contains $\approx 5 \text{ ppb DO}$ at 289°C, although the rate in water that contains $\approx 300 \text{ ppb DO}$ is higher by a factor of 5 than air at this ΔK .

Figure 61.
Dependence on ΔK of ratio of CGRs in HP deoxygenated water of Alloy 600 with 0.08 wt.% carbon and differing heat treatments to CGRs in air of mill-annealed Alloy 600 at 289°C

4.5 Morphology of Crack Path and Fracture Surface of Alloy 600 Specimens

Figures 62-65 show the fracture surface, fracture morphology, and crack path in the crack-tip region of the specimens listed in Tables 17 and 18. The 1TCT specimens were sectioned vertically, and one-half of each specimen was split in the plane of the crack in liquid nitrogen. Corrosion-product films were removed from the fracture surface by a chemical process to reveal the morphology of the underlying material. The intact portion of the specimen that encompassed the crack was polished and etched to corroborate the mode of crack propagation and also to determine if crack branching had occurred during the test. The total crack lengths at the end of the test were consistent with values obtained by the DC potential-drop technique and no branching of the cracks had occurred.

ALLOY 600	HEAT TREATMENT	LOAD CONDITIONS	ENVIRONMENT
Spec. No. J320-02 Heat No. NX9244G	Solution Annealed 1025°C/2 h	$K_{max} = 31-39 \text{ MPa}\cdot\text{m}^{1/2}$ $R = 0.20-0.95$ Freq. = 0.077 Hz	300 ppb DO $0.06-0.07 \mu\text{S}\cdot\text{cm}^{-1}$

Figure 62. Crack path, fracture surface, and fracture morphology of 1TCT solution-annealed (1025°C for 2 h) specimen of Alloy 600 (No. J320-02) after crack growth experiment in HP oxygenated water at 289°C

ALLOY 600	HEAT TREATMENT	LOAD CONDITIONS	ENVIRONMENT
Spec. No. J321-02 Heat No. NX9244G	Solution Annealed 1025°C/2 h and Thermally Treated 600°C/24 h	$K_{max} = 31-34 \text{ MPa} \cdot \text{m}^{1/2}$ $R = 0.20-0.95$ Freq. = 0.077 Hz	300 ppb DO $0.06-0.07 \mu\text{S} \cdot \text{cm}^{-1}$

Figure 63. Crack path, fracture surface, and fracture morphology of 1TCT solution-annealed and thermally treated (1025°C for 2 h and 600°C for 24 h) specimen of Alloy 600 (No. J321-02) after crack growth experiment in HP oxygenated water at 289°C

ALLOY 600	HEAT TREATMENT	LOAD CONDITIONS	ENVIRONMENT
Spec. No. J310-02 Heat No. NX9244G	Solution Annealed 1115°C/2 h	$K_{max} = 30-36 \text{ MPa} \cdot \text{m}^{1/2}$ $R = 0.20-0.95$ Freq. = 0.077 Hz	300 ppb DO $0.07-0.12 \mu\text{S} \cdot \text{cm}^{-1}$

Figure 64. Crack path, fracture surface, and fracture morphology of 1TCT solution-annealed (1115°C for 2 h) specimen of Alloy 600 (No. J310-02) after crack growth experiment in HP oxygenated water at 289°C

ALLOY 600	HEAT TREATMENT	LOAD CONDITIONS	ENVIRONMENT
Spec. No. J311-02 Heat No. NX9244G	Solution Annealed 1115°C/2 h and Thermally Treated 600°C/24 h	$K_{max} = 30-34 \text{ MPa} \cdot \text{m}^{1/2}$ $R = 0.20-0.95$ Freq. = 0.077 Hz	300 ppb DO $0.07-0.12 \mu\text{S} \cdot \text{cm}^{-1}$

Figure 65. Crack path, fracture surface, and fracture morphology of 1TCT solution-annealed and thermally treated (1115°C for 2 h and 600°C for 24 h) specimen of Alloy 600 (No. J311-02) after crack growth experiment in HP oxygenated water at 289°C

The Alloy 600 specimens from the CGR experiments in HP oxygenated water exhibited a transgranular mode of crack propagation. This mode of crack propagation can be attributed to the strong contribution of mechanical cyclic loading in tests at R of 0.2-0.8 and the relatively small degree of crack extension that occurred in tests at $R > 0.8$.

5 Modeling of Residual Stresses in Core Shroud Structures (J. Zhang, P. Dong, F. W. Brust, Battelle Columbus Laboratories; and W. J. Shack, ANL)

Under contract to ANL, Battelle Columbus Laboratories (BCL) has used numerical models to characterize weld residual stresses and the associated stress intensity factors at BWR core shroud welds. A detailed description of the weld residual stresses has been obtained for the H4 weld. The multipass welding process was simulated by an axisymmetric solid element model. A shell element analysis was also performed to quantify the effects of specimen removal on residual stress measurements. Based on the residual stress results, stress intensity factors were calculated.

5.1 Core Shroud Weld Geometry and Calculation of Residual Stress Distributions

The H4 weld is a multipass SA weld that joins two Type 304 SS cylinders. The geometry of the cylinder and the H4 weld is shown in Fig. 66. The inner radius and thickness of the cylinder are 2.23 m (88 in.) and 31.1 mm (1.5 in.), respectively. The H4 weld groove is a symmetrical "double-J" type. The sides of the groove are 22.5° from the normal and the root regions have a 6.3-mm (0.25-in.) radius. The specified welding parameters for the H4 weld are

Figure 66. Core shroud structure and H4 weld geometry

Table 19. Welding parameters for H4 weld

Passes	Current (A)	Voltage (V)	Travel Speed (mm/s)
Root	285	28	6.3
Balance	510	30	7.6

listed in Table 19. An arc efficiency of 70% is assumed. The root passes on the inner side of the cylinder were made first. Additional passes were used to fill the entire inner groove. Then, the outer groove was back-chipped, and the root passes on the outer side of the cylinder were then made. The outer groove was finally filled during additional passes.

Based on the H4 weld geometry and pass sequence, an axisymmetric finite-element model was generated, as shown in Fig. 67. The finite-element model consists of 2952 elements and 3145 nodes. A refined mesh is used in and near the weld region. In the weld region, 18 lumped weld passes are assumed, as indicated in Fig. 67b, where Passes 1 and 10 are the root passes on the inner and outer sides of the cylinder. A weld cap is assumed for the weld profile on both sides of the cylinder. Because the H4 weld was made after the H1-H3 welds, the radial and axial displacements were fixed at the H3 end of the model to simulate the restraints by the remainder of the shroud structure. The H5 end of the model was left free.

A sequentially coupled thermal mechanical analysis approach was used. A thermal analysis was performed, and the temperature solutions obtained from the thermal analysis were then used as the input for the structural analysis. The thermal analysis was performed with the TEMPER code developed at BCL. The mechanical analysis was performed with the ABAQUS finite-element code. Currently, almost all general-purpose commercial finite-element codes,

Figure 67.
Axisymmetric finite-element
model for H4 weld analysis

including ABAQUS, are incapable of directly modeling important welding phenomenon such as material melting and metal deposition. Effective modeling of these phenomena is especially important for multipass welds, and a special user-material subroutine was developed at BCL to model the material behavior during welding.

The H4 weld temperature history generated by TEMPER was read into ABAQUS as a field variable by a user subroutine. This field variable was then passed to the special material subroutine to calculate the driving force for the mechanical analysis. The strain components and material yielding status were calculated by the special user-material routine as solution-dependent state variables. The sequential metal deposition effects associated with multiple passes were simulated by assigning a negligible material stiffness (comparable to the stiffness at melting temperature) to those filler elements that are not yet deposited. Once a filler element is deposited, its virgin stress state is restored. Temperature-dependent material properties with nonlinear hardening behavior were used for the analysis.

Figure 68a shows the resulting axial residual stress distributions through the thickness of shroud wall at the weld centerline ($Y = 0$) and HAZs ($Y = -19\text{ mm}$ [-0.75 in.], 19 mm [0.75 in.]). The axial residual stresses show a "thick-shell" type of distribution. In the middle of the wall, the axial residual stresses are compressive, whereas, at both inner and outer surfaces of the pipe, tensile stresses were present, except at the weld centerline, where a small-amplitude compressive stress is present on the outer surface. This compressive axial stress was primarily due to axial bending of the wall caused by radial shrinkage of the weld and the presence of a weld cap. At the HAZs (19 mm [0.75 in.] away from the weld centerline), tensile stresses occur at both the inner and outer surfaces. The stress distributions at the two HAZ locations are very similar except near the outer surface. The difference near the outer surface was primarily due to the effects of the weld pass sequence in the last weld layer (Passes 16, 17 and 18). At these cross sections, the maximum tensile axial stresses are at the inner surface of the shroud wall.

Figure 68b shows the corresponding throughwall hoop residual stress distributions. As expected, the hoop stresses are tensile almost everywhere at these cross sections. The maximum tensile hoop stress occurs at the weld centerline, between the midthickness and outer

Figure 68. Throughwall residual stress distribution in H4 Weld (a) axial stress, (b) hoop stress

surfaces. At the HAZs, the tensile stresses peak at the inner and outer surfaces; however, their magnitudes are much smaller than the peak value at the weld centerline.

Figures 69a and 69b show contour plots of the axial and hoop residual stresses in the weld area. For clarity, zones of tensile and compressive stresses are also indicated. Fig. 70a shows the axial residual stress distributions along the outer and inner surfaces of the shroud. Again, the axial residual stress at the inner surface of the weld centerline is compressive. Rapid variations of the axial stresses occur within a region of 50-mm (2-in.) from the weld centerline. As shown in Fig. 70b, the hoop residual stress distributions at the inner and outer surfaces are very similar. The hoop stresses are tensile within the weld area and become compressive immediately outside this area. The hoop stresses decay more quickly with distance from the weld than do the axial stresses. The residual stress distributions (Fig. 70) obtained from the current analysis are quite similar to those determined experimentally by neutron diffraction for a similar H4 weld.⁶¹

To better understand the mechanism of residual stress evolution in the H4 weld, the residual stress states at the end of various weld passes were determined. These stress states are shown in Fig. 71 (axial stress) and 72 (hoop stress). Eight residual stress states were plotted, corresponding to states after a layer of weld passes was deposited.

Both axial and hoop stresses exhibited a drastic change from Pass 9 to Pass 10. After Pass 9, the inner groove was filled, and at Pass 10 the root passes in the outer groove were started. After Pass 9, the axial stress distributions are more like a "thin-shell" type of bending distribution in which the magnitudes gradually increase from Pass 1 to Pass 9. At Pass 10, the compressive axial stress zone was pushed from the midsurface toward the inner surface, but a tensile stress zone remained at the inner surface. Thus, the characteristic, doubly curved "thick-shell" type of axial stress distribution was developed. This pattern was further established at Pass 12, where the second weld layer was made in the outer groove, and maintained throughout the rest of the passes until the outer groove was filled. The same observation can also be made for the hoop stress evolution, except that tensile stresses occurred at all stages.

Figure 69. Contour plots of residual stress distributions in H4 weld

Figure 70. Residual stress distributions along inner and outer surfaces of core shroud: (a) axial stress, (b) hoop stress

Figure 71. Evolution of axial residual stress through shroud wall at H4 weld centerline

Figure 72. Evolution of hoop residual stress through shroud wall at H4 weld centerline

5.2 Effects of Specimen Removal on Residual Stress Distributions

Residual stress measurements of core shroud welds generally require that a portion of the shroud wall which contains the weld be removed from the structure. The process of specimen removal may alter the in-situ residual stress field. The degree of relaxation depends on the size of the specimen and on the characteristics of the surrounding residual stress field. To better quantify the effects of specimen removal, the removal process was analyzed with a 3-D shell element model.

Figure 73 shows the finite-element mesh for the shell element model used in this analysis. Only one-eighth of the entire core shroud was modeled due to size limitations of the finite-

Figure 73.
3-D shell element model used to
evaluate effect of specimen
removal on residual stress
distribution in H4 weld

element model. The finite-element mesh consists of 2001 shell elements and 2086 nodes. As in the axisymmetrical model, a refined mesh was used in and around the weld area. Eleven sectional integration points were used through the shell thickness to accurately capture through-wall stress variations. The 0.3-m x 0.4-m (12-in. x 14-in.) shroud specimen that was removed was located at the center of the shell model, as indicated in Fig. 73.

Symmetrical displacement boundary conditions were imposed in the circumferential direction at the nodes along the two edges at $\theta = 0^\circ$ and $\theta = 45^\circ$. As in the axisymmetric model, all of the nodal displacements at the H3 end were fixed; those at the H5 end were left free. The shell element analysis was also performed with ABAQUS, along with some utility programs developed at BCL. The detailed residual stress field obtained from the axisymmetrical model was mapped onto the 3-D shell element model as an initial stress field, and equilibrium iterations for the mapped residual stresses on the shell model were performed. The specimen was then removed from the model (i.e., stresses equal and opposite to those exerted on the specimen by the surrounding material were applied to obtain stress-free boundary conditions), and the resulting changes in the stress fields were computed.

Figures 74a and 74b show the contour plots of axial stress distributions at the inner surface of the shroud before and after specimen removal. Note that the axial stresses are plotted with the same contour scale. After removal, the compressive axial residual stresses at the inner surface were released along two cutting edges in the hoop direction to achieve stress-free surface conditions, but the tensile axial stress at the center of specimen (in the weld) was increased (a darker color). Figures 75a and 75b show the through-thickness distributions of both axial and hoop stresses at the center of specimen before and after removal. For the axial component, both the tensile stress at the inner surface and the compressive stress at the outer surface increased. For the hoop component, however, the tensile stress at the outer surface decreased, and the tensile stress at the inner surface increased slightly. The overall changes were fairly small.

Figure 74. Axial residual stress distribution at inner surface of core shroud

The effects of specimen removal on the residual stress redistribution may be explained by the free-body diagram shown in Fig. 76. The effects of cutting along the specimen boundary can be approximately represented by equivalent sectional moments or forces obtained from the residual stress field and acting in a reversed direction. At the two circumferential cutting edges, sectional bending moments are present due to tensile axial residual stresses at the outer surface and compressive axial residual stresses at the inner surface (Fig. 70a). Along the other two edges (longitudinal cuts), sectional normal forces are present in the hoop direction and their distributions are similar to those of the hoop stresses in Fig. 70b. From the free-body diagram,

Figure 75. Throughwall residual stresses at center of a specimen removed from core shroud

it is clear that additional tensile and compressive axial stresses are introduced at the inner and outer surfaces, respectively (due to the equivalent bending moments at the circumferential cutting edges). The equivalent normal forces at the longitudinal cuts also produce a resultant bending moment at the specimen center due to the specimen curvature. This resultant bending moment causes a decrease in tensile hoop residual stress at the outer surface. Because of this relaxation, the radius of curvature of the specimen is increased after removal, as was observed experimentally for a specimen from a core shroud H4 weld.⁶¹

5.3 Calculation of Stress Intensity Factors for Welds with Flaws

After the residual stress distributions in the uncracked weldment were determined, the finite-element alternating method (FEAM)⁶²⁻⁶⁶ was used to obtain stress intensity factors for weldments that contained flaws. FEAM methods have been verified for many different crack problems, loading conditions, etc. The major advantage of the method is that only a finite-element mesh of the uncracked geometry is needed to obtain stress intensity factors (or the J-Integral, displacements, stresses, etc). More important, the same mesh can be used to obtain solutions for cracks of many different sizes and geometries. Because the finite-element stiffness matrix needs to be reduced only once, regardless of the crack size, crack location, crack orientation, crack number, etc., the method is extremely efficient (mixed-mode conditions can be handled as well).

The stress intensity factors due the weld residual stresses for an internal circumferential crack on the H3 side of the weld in the HAZ obtained with the FEAM are shown in Fig. 77 as a function of crack depth. The crack location (19-mm [0.75-in.]) from the weld centerline on the H3 side is shown in Fig. 78, which also illustrates the redistribution of axial residual stresses associated with a 10.4-mm (0.41-in.) crack. The crack faces are stress-free. The stress directly ahead of the crack tip decreased to ≈ 138 Mpa (20 ksi). The tensile stresses at the outer surface decreased somewhat as well.

In Fig. 77, in addition to the FEAM solution, two other solutions are plotted. One solution was obtained using the NASA FLAGRO program,⁶⁷ which uses a weight function method with the crack plane residual stresses serving as the initial stress state. Another was calculated with

Figure 76.
Equivalent boundary
forces and moments on
removed specimen

Figure 77.
Stress intensity factor
solutions for complete
circumferential cracks in H4
weld growing from the inner
surface on H3 side HAZ (19-
mm [0.75-in.] from weld
centerline) as function of
crack depth

the original crack closure integral technique.⁶⁸ This required a reanalysis in which double nodes were placed along the crack plane and released to obtain the energy release rate, and subsequently, K . This method is quite tedious compared with the FEAM method. The FEAM solution is probably the most accurate, although the three methods give comparable results. The stress intensity factor decreases as the crack depth increases until it reaches negative values for crack depths ≥ 15 -mm. This is consistent with the results shown in Fig. 78, where the stresses at the crack tip for a crack depth of 10.4-mm (0.41-in.) are decreasing and the stresses a small distance ahead of the crack tip are negative.

The stress intensity factors for external cracks at the same axial location on the H3 side of the weld are plotted versus crack depth in Fig. 79. The variation of K with crack depth is similar to that for internal cracks both in magnitude and shape. However, the K values approach zero for a crack depth of ≈ 22.9 -mm (0.9-in.) for the external cracks versus 15.2-mm (0.6-in.) for the internal cracks.

Figure 78. Redistribution of residual stress due to crack growth

Figure 79. Stress intensity factor solutions for complete circumferential cracks in H4 weld growing from the outer surface on the H5 side HAZ (15.5-mm [0.61-in.] from weld centerline) as function of crack depth

Figure 80. Stress intensity factors for surface cracks in weld centerline of H4 weld

To analyze the case of a surface crack of length a_1 , the residual stress state from the axisymmetrical case (Fig. 67) was mapped onto a 3-D model. The 3-D mesh is shown at the top of Fig. 80. For this analysis, the stresses were also assumed to be symmetrical about the weld centerline. The stresses on the H3 side of the weld were used for the symmetrical distribution. The mesh refinement in the radial and axial directions for the axisymmetrical model was greater than for the 3-D model. Hence, the mapped stresses in the 3-D model represent averages of the stresses over several elements in the axisymmetrical model, and the 3-D solution is expected to be somewhat less accurate than the solution for the axisymmetrical model. The surface crack was then introduced into the model (at the weld centerline) as illustrated in the upper portion of Fig. 80. The crack length was fixed at 88.9-mm (3.5-in.) and the crack depth, a_2 was varied. The FEAM approach was used to obtain stress intensity factors as a function of elliptic crack

angle, as illustrated at the bottom of Fig. 80. Although the stress state in the uncracked case was axisymmetric, the stress redistribution that occurs when the surface crack is introduced is not axisymmetric. Hence, K varies with the elliptic crack angle. An elliptical angle of 90° indicates the deepest point of the crack (at a_2) and an angle of 0° corresponds to the inner surface of the pipe (at location a_1). The K values are larger here at the deepest point of the crack are larger than those in Fig. 78 (note that a very long a_1 would represent an axisymmetric crack). This is expected because the stresses at the weld centerline are larger than those at the H3-side HAZ crack.

The K values at the deepest point of the crack decrease as the crack depth increases (recall that the crack length a_1 is kept constant in these calculations). However, as the crack depth increases, K increases at elliptic angles of $\approx 40^\circ$ or less. For crack depths of 10-13 mm (0.4-0.5 in.), the K at the deepest point of the crack is markedly smaller than the K near the surface of the pipe. These results suggest that cracks, if driven by corrosion mechanisms that depend on K , will tend to increase in length in the angular direction much more rapidly than they grow throughwall. Indeed, there will be a tendency for full 360° cracks to develop. For the axisymmetrical case with a complete 360° crack, the results indicate that a crack that grows from the inside to the outside of the vessel surface, or vice versa, would be likely to stop approximately midwall, because K becomes negative for deeper cracks.

Several other issues should be addressed before definitive conclusions can be reached. The effect of elastic-plastic crack growth was not considered here. The welding process induces significant plastic deformation in the vessel, and, as stress corrosion cracks grow in the vessel, the crack-opening stresses at the tip will cause additional plasticity near the crack tip. The size of this plastic zone will determine whether nonlinear effects will be significant in the analysis. However, elastic effects should dominate as the cracks get deeper, and the crack opening stresses become small. In addition, the effect of 3-D variations in the residual stress fields, e.g., those associated with the start/stop positions, on the growth of the surface crack is unknown. Finally, the effect on crack growth of the additional constraint caused by other welds in the vessel after full fabrication should be considered.

6 Summary of Results

6.1 Environmental Effects on Fatigue S-N Behavior of Primary Pressure Boundary Materials

Fatigue tests have been conducted to determine the crack initiation and crack growth characteristics of A533-Gr B LAS and A106-Gr B CS in air and LWR environments. Crack length as a function of fatigue cycles was determined in air and water environments. The significant results are summarized below.

- (a) In air, fatigue cracks, 10 μm or longer, form quite early during fatigue ($< 10\%$ of life) even at low strain ranges ($\approx 0.4\%$). The largest crack is $< 100 \mu\text{m}$ at half life. The results suggest that linear damage summation should be valid for strain or stress ranges above the fatigue limit, e.g., the crack-depth-versus-life-fraction plots are approximately the same over a wide range of strain or stress.

- (b) The decrease in fatigue life of CSs and LASs in high-DO water is primarily caused by the effects of environment on the growth of short cracks. During the initial stages of fatigue damage (for cracks $<100\ \mu\text{m}$), crack growth rates in high-DO water are nearly two orders of magnitude higher than in air. For crack sizes $>100\ \mu\text{m}$, the growth rate is one order of magnitude higher in high-DO water than in air.
- (c) In high-DO water, the surface cracks appear to grow entirely as Mode I tensile cracks normal to the stress axis. In air or low-DO water, growth of surface cracks initially occurs as a shear crack at $\approx 45^\circ$ to the stress axis.
- (d) The results from the present study are consistent with the slip-dissolution model for enhanced crack growth rates in LWR environments.

Fatigue tests have also been conducted on Types 316NG and 304 SS to establish the effects of LWR coolant environments on fatigue lives of these steels. The results indicate significant decrease in fatigue life in water when compare with that in air; the decrease in life depends both on strain rate and DO content of the water. Environmental effects on fatigue life are the same for Types 304 and 316NG SS. However, unlike CSs and LASs, environmental effects are more pronounced in low-DO than in high-DO water. At $\approx 0.004\%/s$ strain rate, reduction in fatigue life in water that contains <10 ppb DO is greater by a factor of ≈ 2 than in water with ≥ 200 ppb DO. In a simulated PWR environment, a decrease in strain rate from 0.4 to $0.004\%/s$ decreases fatigue life by a factor of ≈ 8 . Experimental results have been compared with estimates of fatigue life that are based on a statistical model.

A detailed examination of the fatigue test specimens was conducted to investigate the role of high-temperature oxygenated water in fatigue cracking. The formation and growth of surface cracks in simulated PWR water and air differ. In a simulated PWR environment (<10 ppb DO), the surface cracks appear to grow entirely as tensile cracks in Stage II growth normal to the stress axis. In air and, for most cases, in high-DO water, surface cracks initially grow along slip planes as shear cracks in Stage I growth along planes at 45° to the stress axis. In all environments, cracks primarily formed within PSBs. Additional tests are in progress to characterize the formation and growth of surface cracks in LWR environments. Fatigue tests are also being conducted to define a threshold strain amplitude below which environmental effects on fatigue life do not occur, and the threshold strain rate below which effects of environment saturate.

6.2 Irradiation Assisted Stress Corrosion Cracking

Slow-strain-rate-tensile tests were conducted in simulated BWR water at 288°C on model SS alloys that were irradiated at 288°C in helium to a fluence of $\approx 0.45 \times 10^{21}\ \text{n}\cdot\text{cm}^{-2}$ ($E > 1\ \text{MeV}$) in the Halden reactor. Initial results indicate that the ductility of irradiated CP Type 304 SS sheet specimens is higher than that of BWR neutron-absorber tubes at a comparable fluence level. High-purity Type 316 SS was more susceptible to IGSCC. This behavior is similar to that observed for HP heats of Type 304 SS. Type 304 SS heats high in carbon content were more susceptible to TGSCC but less susceptible to IGSCC. Unusually high levels of oxygen and sulfur in SS appear to have a deleterious effect on SCC of Type 304 SS, even in the nonirradiated state, during SSRT tests in HP oxygenated water at 288°C .

Shielded-metal-arc welds were prepared from Types 304, 304L, and 316 SS to simulate BWR core-shroud welds. Significant contamination by oxygen, sulfur, and fluorine was observed in the HAZ of the SMA welds. The source of this contamination is the coating on the welding electrode and the air atmosphere. In contrast, contamination in GTA welds was insignificant.

Microhardness profiles of the fusion zone and HAZ of an SMA weld of Type 304 SS were determined. A region of maximum hardness was observed in the HAZ at ≈ 0.8 -3 mm away from the fusion zone. This region coincides with the region where cracking occurs in BWR core shroud welds. When the surface of the weld was smoothed with a grinding wheel, the hardness near the ground surface was significantly higher, indicating that a cold-worked surface layer can promote crack nucleation. The local hardening characteristics, together with contamination by oxygen, fluorine, and sulfur, may be important factors that increase susceptibility to SCC of core shroud welds in BWRs.

Auger electron spectroscopy analyses show that fluorine atoms segregate to grain boundaries in BWR neutron-absorber tubes and in a core shroud weld; however, the mechanism of the segregation (i.e., thermal, irradiation-induced, or both) is not understood. Because the solubility of both fluorine and oxygen in steel is very low, it is likely that these elements will segregate by a thermal process to local sites such as matrix/precipitate interfaces, grain boundaries, and stacking faults; however, irradiation-induced segregation cannot be ruled out and further investigation is needed.

Equipment for fracture toughness J-R testing of irradiated CT specimens in a hot cell has been assembled. Results of preliminary fracture toughness J-R tests indicate that an elastic compliance method can be used to estimate crack length during J-R tests or fatigue crack growth in LWR environments. These findings will be validated by (a) conducting fracture toughness J-R curve tests at room temperature and 288°C on two heats of thermally aged CF-8M cast SS and on 50% cold-worked Type 316NG SS, and (b) performing fatigue crack growth studies on the same three materials at 288°C in air.

6.3 Environmentally Assisted Cracking of Low-Carbon Alloy 600 in Simulated LWR Water

Corrosion-fatigue experiments were conducted on CT specimens of a low-carbon content (0.03 wt.%) heat of Alloy 600 in HP oxygenated water to investigate the effects of load ratio and heat treatment conditions on CGRs at 289°C. The specimens were fabricated from blocks of material that were solution-annealed at 1025 and 1115°C for 2 h and heat treated at 600°C for 24 h after solution annealing at the two temperatures. Solution annealing at 1025 and 1115°C produced ASTM grain sizes of ≈ 1.0 -1.5 and 0.2, respectively, which correspond to large average grain diameters of ≈ 250 -210 and 340 μm . Based on results from four specimens, these heat treatment conditions did not have a significant effect on the CGRs in HP oxygenated water. Other heats of material with differing carbon content and heat treatment conditions could produce a wider variation in the results.

"Best-fit" correlations for the CGRs in water that contained ≈ 300 ppb DO versus ΔK were obtained for each specimen and for the combined data from the four specimens of the same material with different heat treatments. The effect of HP oxygenated water on the CGRs

relative to that in air was determined from a ratio of the CGRs at 289°C in water to the CGRs of another heat of Alloy 600 in air at 289°C, and from a ratio of the CGRs from the combined data in water for all of the heat treatment conditions to that in air at 289°C. The results indicate that the rates for Alloy 600 are higher in water than in air at ΔK values $<18 \text{ MPa}\cdot\text{m}^{1/2}$ and are lower in water than in air at higher values of ΔK . At a ΔK of $\approx 2 \text{ MPa}\cdot\text{m}^{1/2}$, the rate is higher by a factor of ≈ 10 in water than in air. At this ΔK , CGRs for specimens from other heats of Alloy 600 in HP water that contained <5 ppb DO were higher by a factor of ≈ 2 than they were in air.

The morphology of corrosion-fatigue cracks in the Alloy 600 specimens was determined. Transgranular cracking occurred in air and HP oxygenated water because of the strong contribution of mechanical cyclic loading in tests at R values of 0.2-0.8 and the relatively small degree of crack extension that occurred in tests at $R > 0.8$.

6.4 Modeling of Residual Stresses in Core Shroud Structures

Under a subcontract to ANL, BCL has used numerical models to characterize weld residual stresses and the associated stress intensity factors at BWR core shroud welds. A detailed description of the weld residual stresses has been obtained for the H4 weld. Based on the residual stress results that were obtained, stress intensity factors were calculated.

The H4 weld is a multipass SA weld that joins two Type 304 SS cylinders. The axial residual stresses for this weld show a "thick-shell" type of distribution. In the middle of the wall, the axial residual stresses are compressive, whereas, at both inner and outer surfaces of the pipe, tensile stresses were present, except at the weld centerline, where a small-amplitude compressive stress is present on the outer surface. The stresses are tensile in the HAZs on the inner and outer surfaces. At these cross sections, the maximum tensile axial stresses are at the inner surface of the shroud wall.

After residual stress distributions in the uncracked weldment were determined, stress intensity factors K for weldments that contained flaws were determined. Complete 360° cracks, as well as a surface crack extending over only a portion of circumference, were considered. For the axisymmetrical case with a complete 360° crack, the results indicate that a crack growing from the inside to the outside vessel surface, or vice versa, would be likely to stop growing about midwall, because K becomes negative for deeper cracks. For the surface crack that extends over only a portion of the circumference, the K values at the deepest point of the crack decrease as the crack depth increases. However, the K values along the crack front near the surface remain positive, and in fact, increase as the crack grows deeper. These results suggest that cracks, if driven by corrosion mechanisms that depend on K , will tend to increase in length in the angular direction much more rapidly than they grow throughwall. Indeed, there will be a tendency for full 360° cracks to develop; the average depth of the cracks is expected to be $\approx 50\%$ of the wall thickness, although local perturbations in the stress distributions could produce localized regions of deeper cracking.

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BIBLIOGRAPHIC DATA SHEET

(See instructions on the reverse)

1. REPORT NUMBER
(Assigned by NRC. Add Vol., Supp., Rev.,
and Addendum Numbers, if any.)
NUREG/CR-4667, Vol. 23
ANL-97/10

2. TITLE AND SUBTITLE

Environmentally Assisted Cracking in Light Water Reactors.
Semiannual Report July 1996—December 1996

3. DATE REPORT PUBLISHED

MONTH

YEAR

October

1997

4. FIN OR GRANT NUMBER

W6610

5. AUTHOR(S)

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6. TYPE OF REPORT

Technical; Semiannual

7. PERIOD COVERED (Inclusive Dates)

July 1996—December 1996

8. PERFORMING ORGANIZATION — NAME AND ADDRESS (If NRC, provide Division, Office or Region, U.S. Nuclear Regulatory Commission, and mailing address; if contractor, provide name and mailing address.)

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9. SPONSORING ORGANIZATION — NAME AND ADDRESS (If NRC, type "Same as above"; if contractor, provide NRC Division, Office or Region, U.S. Nuclear Regulatory Commission, and mailing address.)

Division of Engineering Technology
Office of Nuclear Regulatory Research
U.S. Nuclear Regulatory Commission
Washington, DC 20555-0001

10. SUPPLEMENTARY NOTES

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11. ABSTRACT (200 words or less)

This report summarizes work performed by Argonne National Laboratory on fatigue and environmentally assisted cracking (EAC) in light water reactors from July 1996 to December 1996. Topics that have been investigated include (a) fatigue of carbon, low-alloy, and austenitic stainless steels (SSs) used in reactor piping and pressure vessels, (b) irradiation-assisted stress corrosion cracking of Type 304 SS, (c) EAC of Alloy 600, and (d) characterization of residual stresses in welds of boiling water reactor (BWR) core shrouds by numerical models. Fatigue tests were conducted on ferritic and austenitic SSs in water that contained various concentrations of dissolved oxygen to determine whether a slow strain rate applied during various portions of a tensile-loading cycle are equally effective in decreasing fatigue life. Slow-strain-rate-tensile tests were conducted in simulated BWR water at 288°C on SS specimens irradiated to a low fluence in the Halden reactor and the results were compared with similar data from a control-blade sheath and neutron-absorber tubes irradiated in BWRs to the same fluence level. Crack-growth-rate tests were conducted on compact-tension specimens from a low-carbon content heat of Alloy 600 in high-purity oxygenated water at 289°C. Residual stresses and stress intensity factors were calculated for BWR core shroud welds.

12. KEY WORDS/DESCRIPTORS (List words or phrases that will assist researchers in locating this report.)

Corrosion Fatigue
Crack Growth
Irradiation-Assisted Stress Corrosion Cracking
Radiation-Induced Segregation
Residual Stress
Stress Corrosion Cracking
A106-Gr B Steel
A533-Gr B Steel
Type s 304, 316, and 316NG Stainless Steel
Alloys 600 and 690

13. AVAILABILITY STATEMENT

Unlimited

14. SECURITY CLASSIFICATION

(This Page)

Unclassified

(This Report)

Unclassified

15. NUMBER OF PAGES

16. PRICE