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A SUMMARY OF ENVIRONMENTALLY ASSISTED CRACK-GROWTH STUDIES
PERFORMED AT WESTINGHOUSE ELECTRIC CORPORATION:
UNDER FUNDING FROM THE HEAVY-SECTION
STEEL TECHNOLOGY PROGRAM

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FOREWORD

The work reported here was performed at Westinghouse Electric Corp., Generation Technology Systems Division, under subcontract with the Heavy-Section Steel Technology (HSST) Program, C. E. Pugh, Program Manager. The program is sponsored by the Office of Nuclear Regulatory Research of the U.S. Nuclear Regulatory Commission (NRC). The technical monitor for the NRC was Milton Vagins.

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A SUMMARY OF ENVIRONMENTALLY ASSISTED CRACK-GROWTH STUDIES
PERFORMED AT WESTINGHOUSE ELECTRIC CORPORATION:
UNDER FUNDING FROM THE HEAVY-SECTION
STEEL TECHNOLOGY PROGRAM

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ABSTRACT

This report was prepared as an attempt to collect the information developed in the crack-growth studies at Westinghouse from 1969 to 1986. Although progress reports were issued regularly over the entire period, the program results have not previously been integrated in a single document.

The effect of a pressurized-water environment in the enhancement of fatigue crack growth at low frequencies of loading was discovered by Kondo in 1971 and verified in this program shortly thereafter. To characterize plates and forgings, as well as welds and heat-affected zones, a very large number of tests were done in the program. These tests were carried out in a series of matrices structured to investigate various issues, and the results are collected in this report. This work provided a data base for development of ASME Code reference crack-growth curves, as well as insights into some of the mechanisms of environmental enhancement.

Static-load crack-growth-rate tests were conducted in the program with some specimens loaded in excess of 11 years. Static-load crack growth was observed in the heat-affected zone specimens and in one base metal heat after several years of exposure.

Extensive fractography was done on the tested specimens, and the fracture surface appearance was found to vary according to the material. Base metal showed striations with areas of cleavagelike facets according to the size and distribution of the manganese sulfide inclusions. Welds primarily displayed a dimpled appearance that was found to be related to the carbides present. The fracture surface appearance was found to correlate more with the microstructure than with the level of environmental enhancement, although striation measurements matched the macroscopic growth rate in all cases.

The data obtained in this program compared well with results obtained at other laboratories (discussed in Sect. 8)

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and led to several revisions of the ASME reference crack-growth laws (documented in Sect. 9). Some of the accomplishments of the program are highlighted in Sect. 10.

1. BACKGROUND AND INTRODUCTION

Corrosion fatigue crack-growth rate testing in simulated pressurized-water reactor (PWR) environments has been under way at Westinghouse Electric since 1969. In FYs 1980 and 1981 the work was carried out under a subcontract to the Naval Research Laboratories (NRL), but in all of the other years through FY 1986, the work has been under subcontract to the Oak Ridge National Laboratory (ORNL). This program was the first in the United States to focus on this subject.

During this time, a great deal of understanding of the corrosion fatigue process in pressure vessel steels and associated welds has been gained. This report summarizes the findings of the program under a single cover although the research results have been reported in the quarterly and semiannual progress reports of the Heavy-Section Steel Technology (HSST) Program (and NRL progress reports). In addition to collecting the results obtained, this report also attempts to relate the findings to each other and to present a current summary of the "state of the art."

Testing in the early years of the program (1969-1971) concentrated on development of a data base for the reactor pressure vessel steels in an air environment. A large data base was developed, and variables, such as test temperature, frequency of loading, and R ratio, were investigated. The specimens tested throughout the 17 years of this program are listed in Sect. 3.7.

In 1970 testing began in pressurized-water environments, and a large number of tests were conducted at frequencies of 60 and 600 cycles/min. These data, along with the air data, formed the basis of the original Sect. XI curve for pressure vessel steels, which first appeared in 1972.

In the fall of 1971, following Kondos discovery¹ of the effect of environment at low cyclic frequencies, testing was refocused on low cyclic frequencies, and the investigations in this area have continued through FY 1986. These investigations and their results are the focus of this report.

The first of the major sections of the report provides, in a historical progression, a summary of all of the fatigue crack-growth data obtained in the program. A number of test matrices were completed to characterize various aspects of corrosion fatigue, and each is reviewed in detail. At the end of this section is a compilation of all of the materials and conditions studies.

The second major area of emphasis in this program has been the study of static-load cracking. This work began in 1974, and some specimens were loaded for in excess of 11 years. Included in this section (Chap. 4) are a number of attempts to develop short-term screening tests for susceptibility to environmental cracking.

Chapter 5 reviews the environmental monitoring techniques developed in the program to ensure that accurate and meaningful test environments were maintained. The state of the art in high-temperature environmental monitoring was advanced considerably in this work, which was done in close cooperation with the International Cyclic Crack Growth Rate Group (ICCGR).

As the program progressed, emphasis was placed increasingly on improved understanding of the variables affecting corrosion fatigue crack growth and the mechanisms through which they interact. Chapter 6 reviews some of the work done to understand these mechanisms.

Chapter 7 is a compendium of all the microscopic and fractographic work done in the program, which generally follows the various test matrices of Chap. 3. Chapter 8 was included to summarize the various cooperative programs carried out with other laboratories, including the work of the ICCGR, in which membership has been maintained since its founding in 1978.

Chapter 9 provides a review of the present state of the art relative to reference fatigue crack-growth laws and contains a prognosis for future revisions of the present laws. Finally, Chap. 10 summarizes some of the accomplishments of the program.

2. MATERIALS AND PROCEDURES

The water fatigue crack-growth tests were carried out in a pressurized-water environment produced in autoclaves that were heated to the desired temperature with electrical heaters embedded in the chamber wall. Complete details of the test chamber have been discussed previously,² and a typical chamber is illustrated in Fig. 2.1. Water test temperatures have ranged from 93 to 288°C (200 to 550°F), and internal pressures have ranged from 7 to 14 MPa (1000 to 2000 psi). The pressurized-water chemistry was maintained by a flowing system with carefully controlled chemistry and thermal conductivity. The water chemistry and its control and monitoring are discussed in Sect. 5.1. The test loop is shown schematically in Fig. 2.2 and pictorially in Fig. 2.3.

The tests were carried out on precracked fatigue specimens of the compact (CT) and WOL configurations according to American Society of Testing and Materials standards³ (ASTM). These specimens were tested under load control, and most tests involved sinusoidally varying loads although one matrix was carried out with saw tooth ramps and varying hold times. Test specimens ranged from 1 to 4 in. thick with the majority being 2 in. thick.

As part of the study of mechanisms of environmental enhancement, several specimens were tested in a hydrogen sulfide atmosphere and in a mixture of hydrogen sulfide and water. These tests were carried out at a pressure of 0.4 MPa (60 psi) and temperatures of 22.2 and 93°C (72 and 200°F) in a simple enclosed environmental chamber. The hydrogen sulfide was bled into the test chamber from a reagent-grade tank with no preconditioning treatment. Results of these tests are discussed in Chap. 6.

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Fig. 2.1. Schematic cutaway drawing of typical tests chamber.

Fig. 2.2 Schematic of an environmental test loop (four such loops were used in this program).

ORNL PHOTO 5559-87

Fig. 2.3 Photographs of test apparatus (cylindrical vessels are insulated test chambers).

3. FATIGUE CRACK-GROWTH CHARACTERIZATION

The fatigue crack-growth rate characterization studies carried out over the life of this program have been quite varied and have been modified as new knowledge of the behavior was gained. The majority of the testing carried out in the program was in the form of well-controlled matrices. The organization of this chapter follows the various matrices that were carried out, summarizing the findings of each ensuing matrix of tests in light of the other results now known. It is inevitable that there is some interplay among the matrices — this has often led to new findings. This chapter presents the matrices in nearly chronological order with a summary of all the specimens tested provided in Sect. 3.7.

3.1 MATERIAL-EFFECTS MATRIX

This test matrix was designed to allow an evaluation of the range of pressure vessel steels and welds found in operating nuclear reactor systems. The tests have been conducted on precracked standard fracture specimens, 2 in. (5.08 cm) thick, in autoclave chambers with environments designed to simulate the PWR environment.

The three different welds studied in the matrix were made with Linde 80, Linde 124, and Linde 0091 flux. The heat-affected zones (HAZs) from these welds were also characterized; two of these welds were in A 533 grade B class 1 plate, and the other was in A 508 class 2 forging steel. Additionally, both plate and forging base metals were studied. The chemistry for all of the materials studied is summarized in Table 3.1.

There are two major loading variables, R ratio and loading form, that were intended for study in this test matrix (Table 3.2). It has long been known that R ratio is a very important variable in this environment, and two values were studied. The second loading variable was the wave form, in which a sinusoidal loading form was compared with a ramp-reset condition. Both load forms resulted in a cyclic frequency of 17 MHz.

One major finding from the matrix was that the material chemistry plays an important part in determining its sensitivity to environmental acceleration of crack-growth rate. The sulfur content appears to be the key parameter here with environmental effects much more evident in those materials with high sulfur content. For this reason, a separate matrix, including both R ratio and orientation effects, was devised to study sulfur effects from A 533 grade B plate materials. Orientation of the crack with respect to the rolling or working direction can be important in base materials with high sulfur content, so it was included as a test variable. A detailed discussion of this series of tests is presented in Sect. 3.3. The nomenclature for specimen orientations followed the ASTM designations shown in Fig. 3.1.

Table 3.1. Materials chemistries

Material	C	Mn	P	S	Si	Ni	Cr	Mo	Cu	V	Co
<i>Weldments</i>											
Linde 124 weld "C"	0.085	1.32	0.013	0.012	0.48	0.91	0.14	0.49	0.05	0.005	
Linde 124 weld HAZ	0.23	1.33	0.009	0.014	0.22	0.58		0.55	0.12		
"C" (adjoining plate)											
Linde 124 weld HAZ	0.23	1.45	0.011	0.009	0.26	0.63		0.53	0.05		
"KH" (adjoining plate)											
Linde 80 weld "C"	0.14	1.36	0.019	0.014	0.42	0.54	0.06	0.37	0.18		
Linde 80 weld "C"	0.23	0.58	0.005	0.008	0.28	0.72	0.39	0.60	0.05	0.03	0.02
HAZ (adjoining forging)											
Linde 0091 weld "D"	0.14	1.06	0.012	0.008	0.20			0.48	0.15		
Linde 0091 weld "D"-	0.25	1.37	0.010	0.020	0.27	0.61		0.55	0.13		
HAZ											
<i>A 533 B class 1</i>											
"L83"	0.22	1.37	0.008	0.008	0.22	0.66	0.15	0.54	0.18		
"2D", "02GB"	0.22	1.45	0.011	0.019	0.22	0.69	0.12	0.53			
"04A"	0.23	1.29	0.010	0.018	0.24	0.60	0.09	0.50	0.14		
"1HT"	0.19	1.28	0.009	0.13	0.25	0.61	0.04	0.55	0.10	0.005	
"IN"	0.21	1.26	0.012	0.026	0.25	0.47		0.47	0.19		
"TW"	0.21	1.38	0.008	0.004	0.21	0.67		0.56	0.08		
"W7"	0.23	1.40	0.005	0.004	0.25	0.70		0.57			
"PN"	0.21	1.33	0.012	0.016	0.22	0.56		0.54	0.13		
"B76"	0.21	1.21	0.16	0.012	0.20	0.66	0.08	0.49	0.09		0.013
"L4"	0.17	1.39	0.003	0.004	0.24	0.60	0.07	0.46	0.02		0.009
"T"	0.17	1.32	0.008	0.013	0.23	0.59	0.05	0.55	0.10	0.004	
"CQ2"	0.21	1.28	0.006	0.014	0.24	0.56		0.53	0.12		
<i>A 508 Class 2</i>											
"R", "R2", R1"	0.19	0.64	0.010	0.009	0.23	0.69	0.42	0.60			
"F", "F1"	0.19	0.60	0.010	0.012	0.26	0.70	0.33	0.56		0.02	0.008
"V82"	0.20	0.60	0.013	0.012	0.20	0.73	0.35	0.56			
"Q71"	0.19	0.69	0.007	0.009	0.31	0.82	0.38	0.62	0.01		
"B"	0.185	1.40	0.010	0.006	0.195	0.655	0.13	0.49	0.095		

Table 3.2. Material-effects test matrix^a

Material	Specimen No.			
	R = 0.2		R = 0.7	
	1-cycle/min sine	1-min-ramp no hold	1-cycle/min sine	1-min-ramp no hold
Linde 124 weld (2TWOL)	C-2 C-3	C-7 C-9	C-1 C-6	C-8 C-10
Linde 124 weld HAZ (in A 533 B class 1 plate) (2TCT)		C-24-HAZ-1		C-23-HAZ-1
Linde 80 weld (2TCT)	C-3-WLD C-6-WLD	C-1-WLD	C-4-WLD C-7-WLD	C-2-WLD
Linde 80 weld HAZ (in A 508 class 2 forging) (2TCT)	C-3-HAZ	C-1-HAZ	C-4-HAZ	C-2-HAZ
Linde 0091 weld (2TCT)	D-1-WLD D-3-WLD	Q342W2 ^b	D-2-WLD D-4-WLD D-5-WLD	Q34W1 ^b
Linde 0091 weld HAZ (in A533 B class 1 plate) (2TCT)	D-1-HAZ		D-2-HAZ D-4-HAZ D-6-HAZ	
A 533 B class 1 plate	04A-103 IHT20 ^c IN4 2D3Q-6 2D3Q-5 2D3C-3 IHT16 ^c R2-05 ^b V82-8 ^b	W7-2C2 ^b R-8 F-10 V82-5 ^b R2-16 ^b	02GB-2 04A-102 2D1Q-2 F-1	
A 508 class 2 forging				

^aAll specimens pH = 9 at room temperature unless noted.^bpH = 6 at room temperature.^cDemineralized water only.

Fig. 3.1. ASTM nomenclature for specimen orientation.

3.1.1 Linde 124 Flux

The first weld to be studied was a submerged arc weld of the type used for many, if not most, reactor vessel welds. The results available are summarized in Figs. 3.2 and 3.3 for R ratio values of 0.2 and 0.7, and there is a very clear R ratio effect, typical of that observed on all of the materials studied. The most striking result of these tests in Fig. 3.2 is that there is considerable scatter in the results; also, the crack-growth rates show small increases and decreases along the overall advance of growth rate with applied ΔK . The growth rates observed were in general somewhat less than for base metal specimens of the same chemistry. The HAZ specimen in Fig. 3.2 behaved very similarly to the welds, and there is no definite effect of the ramp-reset loading form

Fig. 3.2. Results for submerged arc weld and HAZ (Linde 124 flux) — low R ratio.

Fig. 3.3. Results for submerged arc weld and HAZ (Linde 124 flux) — high R ratio.

compared with the sine wave. This lack of wave form effect is in marked contrast to results for other materials.

The results obtained for higher R ratio (summarized in Fig. 3.3) are less well behaved. This was true in general for the high R tests of all of the materials. It was apparently very difficult for the crack-growth acceleration as a result of the environment to sustain itself. Only one of the four weld specimens tested could produce a growth rate above 10^{-3} mm/cycle: all of the others arrested, as can be seen in Fig. 3.3. The cracks failed to advance far enough to produce applied ΔK values $>20 \text{ MPa}\cdot\sqrt{\text{m}}$. In one case, a specimen was tested for 7 months, amassing 166,000 cycles. The fact that the single specimen that grew was tested under a ramp-reset loading is probably not significant. The HAZ specimen did show a more conventional behavior — more typical of that of the A 533 grade B, class 1 plate specimens tested at high R — to be discussed later. Thus, even though it was a HAZ, the fact that it was plate material was probably the governing factor.

3.1.2 Linde 80 Flux

These welds were also made by the submerged arc process and were primarily used in vessels made entirely of forgings. Test results are summarized in Figs. 3.4 and 3.5 for the welds and in Figs. 3.6 and 3.7 for the heat-affected zones. An effect not found in most other materials was exhibited in these welds: after the initial part of the test, the crack-growth rate decreased tremendously, producing in effect a deep "V," portrayed as a growth-rate plot. The effect is particularly pronounced for the ramp-reset loading but also was produced in one sine wave specimen to a considerably smaller degree. The effect at present cannot be effectively explained, but because the sinusoidal wave form is more typical of an actual reactor system loading condition, these results must be considered more typical for in-service behavior prediction, and the ramp-reset results must be set aside for the present.

Considering the sinusoidal results, apparently the same R ratio effect is observed as for the other materials, and also note that the crack growth at high R ratio does not show the arrest phenomenon observed in the Linde 124 weld. The crack-growth rates are, in general, not as environmentally accelerated in this weld as in the Linde 124 weld even though the sulfur content is reported to be nearly the same. The available results appear to be in much better agreement in the high R case than the low R case (Fig. 3.5).

The reversal effect noted above for the ramp-reset condition is even more noticeable in the case of the HAZs (test results are reported in Figs. 3.6 and 3.7). Again, the sine loading results were much more typical of other materials although a reversal did occur in the high R sine specimen shown in Fig. 3.7.

3.1.3 SA 533 Grade B Class 1 Plate

The test results available on this material make it the best characterized in the matrix. Seven heats were investigated in this matrix with the most tests conducted at low R ratio and under a sinusoidal wave

Fig. 3.4. Results for submerged arc weld (Linde 80 flux) — low R ratio.

Fig. 3.5. Results for submerged arc weld (Linde 80 flux) — high R ratio.

Fig. 3.6. Results for submerged arc weld HAZ, SA 508 class 2 forging — low R ratio.

Fig. 3.7. Results for submerged arc weld HAZ, SA 508 class 2 forging — high R ratio.

form. The low R test results are summarized in Figs. 3.8 and 3.9. Figure 3.9 was presented separately to summarize the relatively large amount of data available on a single heat, HSST Plate 02, and to better show the behavior of individual specimens under these conditions.

Figure 3.8 contains the results of testing four heats of this material under a sine wave load, and the results follow two distinctly different patterns. The two patterns can again be separated by the sulphur content of the steels tested. Three heats showed considerable environmental enhancement and were remarkably consistent in their behaviors. A look at Table 3.1 shows that the three heats, "1HT," "IN," and "04A," had sulphur contents of 0.012, 0.026, and 0.018 wt %, respectively.

The second set of results shown in Fig. 3.8 show much lower environmental enhancement, and Table 3.1 shows that both heats "TW" and "W7" contain only 0.004 wt % sulfur. Thus, the sulfur content as a measure of environmental sensitivity is strongly supported by the plate results. It appears that once the sulphur content reaches a certain level, consistent environmental enhancement is obtained. Specimen W7-2C2 was tested under a ramp-reset loading and showed somewhat less enhancement than TW-4, but results were too sparse to produce a strong conclusion about wave form. These two heats were studied in more detail later and are discussed in Sect. 3.5.

Figure 3.9 shows a collection of results obtained in some of the earliest tests on a single heat of this material, HSST plate 02. The data display considerable scatter but again show significant environmental enhancement. From Table 3.1 the sulphur content of plate 02 is found to be 0.019 wt %. There was no obvious explanation for the large amount of scatter in the data from this heat compared with the other heats in Fig. 3.8 tested under similar conditions. The only apparent differences are the dates of the tests, which were several years earlier than those of Fig. 3.8 and the fact that these are WOL-type specimens while most of the other heats were compact specimens. Later studies, reported in Sects. 3.3 and 3.4, revealed that for high-sulfur plate materials it is common to have wide variations in sulfur content within the plate, and this probably led to the scatter in behavior.

The R ratio effect has been characterized for the high-sulfur plate materials (Fig. 3.10) and is similar to that observed for the other materials. The "04A" heat shows somewhat less enhancement than the other heat for reasons as yet unknown.

3.1.4 A 508 Class 2 Forging

In this matrix of tests, three heats of forging material were studied. Results of low R ratio tests are summarized in Figs. 3.11 and 3.12. Again, the two figures have been provided so that individual specimen results can be seen more clearly. The separation in this case was according to loading wave form. These materials were primarily used for the ramp and hold-time studies, so most comparisons involve an interaction with Sect. 3.2.

A summary of the 1-min ramp/reset results is contained in Fig. 3.11 and again shows two behavioral characteristics, which apparently separate

ORNL-DWG 87-4357 ETD

Fig. 3.8. Results for A 533 grade B class 1 plate at low ratio.

Fig. 3.9. Results for A 533 grade B class 1, HSST plate 02 at low R ratio.

ORNL-DWG 87-4359 ETD

Fig. 3.10. Results for A 533 grade B class 1, HSST plate at high R ratio.

Fig. 3.11. Results for A 508 class 2 forging steel at low R ratio — ramp loading.

Fig. 3.12. Results for A 508 class 2 forging steel at low R ratio — sine wave loading.

according to sulfur content. Heats "V82" and "F" both contain 0.012 wt % sulfur and show considerable enhancement; the other heat, "R," shows very little environmental effect. In one case, specimen R-8 displays a steep reversal at the beginning of the test, similar to the behavior of both the Linde 80 weld and the Linde 0091 welds tested under this type of wave form. Specimen R2-16 shows a consistent behavior throughout the test with no enhancement at all, but the sulfur effect may be confused in this case with a starting effect because this specimen was begun at $\Delta K = 40 \text{ MPa}\cdot\sqrt{\text{m}}$. This topic will be further discussed in Sect. 3.6 on the effects of starting conditions.

The behavior of both the high-sulfur heats is quite consistent at the beginning of the test with the V82 heat showing relatively smooth progression while the F heat underwent an arrest event part way through the test.

The results for sine wave loadings and low R ratio are summarized in Fig. 3.12, and it can be seen that two specimens have been tested, representing two sulfur levels. The relatively high-sulfur (0.012-wt %) specimen V82-8 underwent a reversal at the beginning of the test and thereafter followed the behavior of the lower-sulfur specimen R2-05 very closely. The reason for the arrest is unclear, but it is perhaps why the behavior of these two heats appears to be so similar. In comparing the ramp/reset with the sinusoidal loading forms, results are somewhat confusing. The low-sulfur heat "R" shows that the sine wave is clearly more severe, but the higher-sulfur heat V82 shows the ramp/reset results in more enhancement, perhaps because of the large arrest in the sine specimen V82-8. This apparent contradiction to results obtained on all of the other materials could not be explained.

3.1.5 Linde 0091 Weld and HAZ Behavior

The Linde 0091 weld and its associated HAZ studied were in A 533 grade B class 1 plate and have a very low sulfur content (0.008), as seen in Table 3.1. On the other hand, the plate material has a rather high sulfur level (0.020), and, thus, the HAZ specimens would be expected to have a higher crack-growth rate under the same loading conditions.

Fatigue crack-growth-rate results are summarized for the HAZ specimens in Fig. 3.13. At a relatively high value of mean stress ($R = 0.7$), it can be seen that the crack-growth rates are rather high for two of the specimens, following the predicted mean from the Sect. XI reference curve. On the other hand, one specimen did not achieve acceleration in growth, stagnating at a growth rate $< 10^{-4} \text{ mm/cycle}$ ($4 \times 10^{-6} \text{ in./cycle}$). This behavior has been observed in other welds and, even occasionally, base metal results obtained at high R ratio, as discussed earlier. There is presently no adequate explanation for this arrestlike behavior, but it is not a result of the loadings applied. Identical loadings were applied to all three specimens, and no significant test interruptions occurred on any of the three tests.

The level of crack growth that did occur on specimens D-4HAZ and D-6HAZ was well within the range of expected results for a high-sulfur material, as shown by comparison with the Sect. XI reference curve, which was based on high-sulfur materials. A similar acceleration in growth

Fig. 3.13. Summary of results, Linde 0091 HAZ specimens tested at $R = 0.7$.

rate occurred at low R ratio on specimen D-1HAZ (Fig. 3.14). Here it can be seen that the growth rate in the HAZ specimen is much higher than in the weld specimens, presumably because the sulfur level in the weld was so much lower (0.008 vs 0.020). Note that a plateau in growth rate was clearly observed for the HAZ at low R ratio at $\sim 2.5 \times 10^{-3}$ mm/cycle (4×10^{-4} in./cycle). The plateau is not as clearly defined for the high R ratio tests but appears to be occurring at the same level. This plateau is probably related to constant load crack-growth behavior and will be discussed further in Sect. 4.

Fig. 3.14. Results for submerged arc weld (Linde 0091 flux) — low R ratio.

The crack-growth behavior of the Linde 0091 weld itself shows a considerable effect of R ratio. At the lower ratio (Fig. 3.14) there is little or no enhancement of growth rates over air behavior. There is a very strong environmental effect on growth rate at high R ratio (Fig. 3.15). Although this effect of R ratio appears to be somewhat larger than average, the crack-growth rates attained are still not as great as those observed on the HAZ specimens, as shown by the band. A plateau appears to be occurring in the weld growth rate at $\sim 10^{-3}$ mm/cycle (4×10^{-5} in./cycle), which is slightly lower than the apparent plateau for the HAZ specimens.

Fig. 3.15. Summary of results for Linde 0091 weld at R = 0.7.

Note that the behavior of specimens D-4WLD and D-5WLD is somewhat more environmentally enhanced than that of D-2WLD, tested and reported earlier. All three specimens were taken from the same location in the weldment and tested at identical loadings, so there is no explanation for the observed differences. The comparative behavior of the three specimens can be more clearly seen in Fig. 3.16, where the crack length vs

Fig. 3.16. Crack length records for Linde 0091 weld specimens, $R = 0.7$.

cycles records are summarized. It is clear from this figure that specimen D-2WLD behaved differently than the other two. Very little growth occurred in the first 20,000 cycles on this specimen, and even after this length of time, the growth rate was very clearly less than that of the other two.

Another comparison can be made with these specimens, that of water chemistry. Specimen D-4WLD was tested with no boron in the water and with lithium hydroxide added to produce a room temperature pH = 9. This is the environment used for nearly all of the tests reported here, including D-2WLD. For specimen D-5WLD boron was added to the water in the amount of 1000 ppm, and then lithium hydroxide was added for a room temperature pH = 6. Clearly, the behavior of specimens D-4WLD and D-5WLD was nearly identical, indicating that the boron had no effect on growth rates. This conclusion is clouded by the observed difference in the behavior of D-2WLD, however, and further comparisons will be made to confirm the conclusion concerning chemistry effects.

We can conclude from results now in hand that the most damaging wave form is the sinusoidal, and because it is also the most realistic, more attention should be paid to the sinusoidal results. Ramp-reset loadings produce inconsistent results in this environment; in some cases, crack growth is slower than with a sine wave of equivalent frequency, and in other cases, results are equal to sinusoidal results.

From the completed tests of this matrix, we find that the strongest effect on the level of crack-growth rate at high R ratio is the sulfur content. The higher sulfur content results in higher rates. It is judged that the observed differences in behavior are not due to differences in weld and HAZ structure because previous tests have shown nearly identical results in weld and HAZ behavior when the sulfur levels of both were equal.

3.2 CYCLIC FREQUENCY, RAMP AND HOLD TIME, AND WAVE FORM

The study of ramp and hold times extended over more than 2 years, so it is important to summarize the results of the entire program even though many of the individual specimen results have been previously reported. This study was completed in May 1979 before the effect of sulfur on crack growth was discovered, so these tests were done on materials not prone to maximum environmental enhancement.

The study was conducted on specimens taken from two heats of SA 508 class 2 forging steel, machined into 5.08-cm-thick (2-in.) specimens, both WOL and CT type. Tests were conducted by both Westinghouse and NRL, using a test matrix designed to allow comparison between laboratories and to optimize the use of available facilities. Ramp times included were 1 s and 1 and 5 min, and hold times were 1 s and 1, 3, and 12 min. Tests were conducted at both 93°C (200°F) and 288°C (550°F). Results are available for all of the ramp times listed. The test results indicate that ramp time is more important than hold time in affecting fatigue crack-growth rates, so the summary to follow is presented in figures that present different hold time tests with common ramp time.

A summary of results obtained with a 1-s ramp time is found in Fig. 3.17, which shows that at such a fast ramp rate the environment did not have time to enhance the crack-growth rate significantly. Several different hold times were tried with no effect on the data, which show growth rates only slightly higher than results obtained in air. Data of Cullen et al.⁴ from NRL are included and show good agreement with this data even though the R ratio was slightly lower. Two test temperatures were included in this data summary and show that essentially the same data were produced at 288°C (550°F) and 93°C (200°F).

Increasing the ramp time to 1 min results in considerably enhanced crack-growth rates, and results of the tests conducted at 93°C (200°F)

Fig. 3.17. Summary of results for 1-s ramp tests.

are summarized in Fig. 3.18. Growth rates increase very quickly at the beginning of the test and then begin to increase at a slower rate, causing the logarithmic portrayal of growth rate vs range of applied stress-intensity factor to bend over somewhat, in this case trending back toward the air growth rates at $\Delta K = 77$ to $99 \text{ MPa}\cdot\sqrt{\text{m}}$ (70 to 90 $\text{ksi}\cdot\sqrt{\text{in.}}$). The data presented in Fig. 3.18 show a relatively tight scatter band and again reveal excellent agreement with that of Cullen et al.⁴ There is no perceptible effect of hold time on the data and also no effect of specimen type because the "F" specimens were WOL type and the specimens of Ref. 4 were CT specimens.

Results of 1-min ramp time tests conducted in higher-temperature water environment, 288°C (550°F), show considerably more scatter than

Fig. 3.18. Summary of results for 1-min ramp tests at 93°C .

those at the lower temperature (Fig. 3.19). Again, the growth rate increases steeply at the beginning of the test, and the environmental enhancement is clearly evident. It has been found that the crack-growth rate is difficult to sustain in the region where it is increasing so rapidly and has a strong tendency to revert back to the air behavior. This tendency is further enhanced when test interruptions occur (e.g., shutdowns caused by hardware problems or electrical outages).

In this series there were several specimens that showed large decreases in growth rate following such interruptions, and, therefore, the data were ignored after the interruption. Examples are F-10 and F-9, in which the retarded data have been discarded. This behavior has been

Fig. 3.19. Summary of effect of hold time for 1-min ramp tests conducted at 288°C.

verified with striation measurements and is discussed in some depth in Ref. 5 and in Sect. 7 on fractography. In several cases, these arrests in crack growth have occurred early in a test and have been so severe that the specimens were abandoned.

The imposition of hold times also tends to encourage this behavior, as can be seen in Fig. 3.19. The ramp time of 1 min corresponds to a sinusoidal frequency of 0.5 cycles/min, and the data fall well short of the upper band of the sinusoidal data. Thus, it may be concluded that 1-min ramp data with various hold times are somewhat less severe than sinusoidal data for an equivalent loading rate.

Behavior of pressure vessel steels in the light-water reactor (LWR) environment at ramp times >1 min was not studied extensively because the tests require long periods to perform. The available data for a 5-min ramp time and the equivalent 0.1-cycle/min data are summarized in Fig. 3.20. Again, the same general behavior is obtained as with the

Fig. 3.20. Summary of fatigue crack-growth-rate results for 5-min ramp and equivalent sinusoidal tests.

1-min ramp, but two specimens tested at 288°C (550°F) show data that exceed the upper scatter band of the sinusoidal data obtained at 0.5 and 1.0 cycles/min. The previously believed saturation in the growth rates at sinusoidal frequencies of between 0.5 and 1.0 cycles/min is somewhat refuted by these data, but a strong conclusion is not possible because of the lack of sufficient data. Results from a test conducted at 93°C (200°F) fall near the middle of the data.⁴

The growth rates of specimens R-6 and R-11, which exceeded the upper band of the sinusoidal data, were subjected to considerable fractographic study to verify the macroscopically observed rates. The results of these studies did, in fact, verify the macroscopic rates in general. Further discussion of this work is included in Sect. 7.

Besides the intended assessment of the effects of ramp and hold time and temperature on fatigue crack growth, a number of other things were learned in the study. One of the most interesting, and most difficult to explain, is the tendency of the accelerated growth to revert back to the unaccelerated air behavior when interruptions occur in the test, as has been previously discussed. Another very important point is that the accelerated growth is never attained if the specimen is begun at too high a value of applied ΔK or K_{max} . This effect is believed to be related to the mechanisms of crack-growth-rate acceleration in the water environment. It is likely there is a definite relationship between this starting effect and the effect of interruptions in causing the growth rates to revert to a nonaccelerated behavior. This topic was studied in some detail in a matrix of tests discussed in Sect. 3.5.

The process of fatigue crack growth in a water environment is complicated indeed and is believed to be accelerated by the effects of hydrogen produced at the crack tip water-steel reaction. If the mechanical loading rate on the cracked geometry exceeds the rate that hydrogen is produced and diffuses to the region ahead of the crack, the growth will not be accelerated by the water environment. The rate at which hydrogen is produced is affected by many things, only some of which are well understood. On the basis of present evidence, it appears that the electrochemical cell set up at the crack tip that supplies hydrogen to the crack-tip region during a test is easily affected by changes in the specimen loadings. It appears that hydrogen is produced most efficiently during slow, regular loadings, of which the sinusoid is a good example. Irregular loadings, for example, those that include hold times or interrupted tests, cause this cell to operate less efficiently, and, thus, the acceleration is not as pronounced. To result in highly accelerated growth, there must be a very fine balance between the mechanically induced loading rate and the environmentally produced hydrogen. If the mechanical loading, as measured by the stress-intensity factor K , occurs at too fast a rate, the crack will tend to propagate mechanically faster than the environment can influence it, and no enhancement will be observed. This effect has been observed before and is exemplified by comparing 1-s and 1-min ramp test results, but the same mechanism may be the cause of these starting conditions effects. If the specimen is initially loaded at a relatively high ΔK (or K_{max}), the mechanically induced crack-growth rate is faster than the environment can influence it, and no enhancement is observed. In the reactor water

environment, this lack of environmental influence is continued throughout the test; in other steel-environment combinations, however, the environment will gradually catch up to the moving crack, and the enhancement will intensify until a "steady state" behavior of full enhancement is obtained. Further evidence that this effect is strongly dependent on the material and environment combination is provided by a series of tests conducted in hydrogen sulfide gas. In this environment, there is no effect of starting conditions, presumably because the environmental kinetics are so fast that the hydrogen can keep up with the crack no matter how fast it is growing.

3.3 STUDIES OF SULFUR AND CRACK ORIENTATION EFFECTS

The inclusion content of the materials studied clearly appeared to be influencing the amount of environmental acceleration obtained in corrosion fatigue crack-growth-rate tests. In particular, the manganese sulfide inclusions are of interest from this point of view. The level of these inclusions is best measured by the sulfur content listed in the material chemistry.

The effect of inclusions on the fatigue crack growth and toughness of A 533 grade B plate was recently studied by Wilson.^{6,7} He showed that there can be important differences in the inclusion contents of these plates as a function of both steelmaking practice and sulfur content. Lowering the sulfur content has long been known to improve the upper-shelf fracture toughness of these steels, but the effect on fatigue crack growth had not been studied. Wilson found that high inclusion content had a measureable effect on crack-growth rates in air, particularly important in crack propagation through the plate thickness. He found also that when the content was greatest, the crack-growth rate displayed an unsteady effect on a logarithmic plot of crack-growth rate vs range of stress-intensity factor, thus adding to the scatter.

The effect of inclusions on environmental fatigue crack growth had not been studied systematically, but it seemed that the possibility existed for a significant impact here. The water environment could react with the manganese sulfide to form polythionic acid or hydrogen sulfide. Both of these compounds would act to lower the pH level of the water in the crack-tip region. Even minute amounts can cause the corrosion potential to drop very quickly, enhancing a hydrogen mechanism, which was thought to be a strong contributing factor to the environmental acceleration of crack growth.⁸ Thus, the effect of these inclusions on environmental enhancement of crack growth could be much more important than their effect on growth in air. Therefore, the matrix of tests was begun in 1980.

If susceptibility is proportional to size and amount of the manganese sulfide inclusions, it is possible that wide variations in susceptibility could exist in plates of high-sulfur content, as reported on the material test certificate, because the sulfides tend to be concentrated in the upper center of the ingot during solidification and the rolling process does not redistribute them significantly. Therefore, there could be areas of both high and low susceptibility in such a plate. In forgings, the likelihood of this happening is less because the sulfur level

is initially lower in general and the manufacturing process generally results in removal of the center of the ingot and more uniform working of the steel. The sulfide content of a weld is likely to be quite uniform also, particularly those made with an automatic process.

The observations made before this time on the effects of sulfur content on environmental enhancement resulted from the study of many crack-growth tests. None of these tests had been designed specifically to study the relationship between material chemistry and environmental sensitivity, so the pattern that emerged was incomplete. Nonetheless, a strong indication that sulfur could be important arose during these early tests.

To address the material chemistry effects more specifically, a separate matrix of tests was designed. Three different heats of the same material, A 533 grade B class 1 plate, were utilized with the only major difference being the sulfur content. The test matrix was devised to study three parameters: sulfur content, crack orientation, and R ratio. All tests were designed to use the same specimen size and exactly the same loadings for a given R ratio. All tests were run in a simulated PWR environment with the same water chemistry (pH = 9 at room temperature). The matrix, which is now complete, is shown in Table 3.3.

Table 3.3. Matrix study of sulfur effects on A 533 grade B class 1 plate^a

Sulfur content	Orientation		
	TL	LT	LS
<i>Test condition R = 0.2</i>			
0.004	TW-1 W7-2C7	TW-4	TW-7
0.016	PN-1	PN-4	PN-7
0.026	IN-1	IN-4	IN-7
<i>Test condition R = 0.7</i>			
0.004	TW-2 W7-2C6	TW-5	TW-8
0.016	PN-2	PN-6	PN-8
0.026	IN-2	IN-5	IN-8

^aAll of the above tests were carried out in simulated PWR environment at 288°C. All specimens were 2TCT, and the applied loadings were identical for all specimens at a given R ratio. Load form was a 1-cycle/min sine wave.

Using Table 3.3 as a guide, a number of comparisons have been made and will be discussed in some detail in the following paragraphs. Comparisons will all center around the degree of environmental enhancement over air behavior, as characterized by the American Society of Mechanical Engineers (ASME) Sect. XI reference curve for air environment.⁹ Specimen orientation will be studied for each of the heats, and then the effect of sulfur content on growth rates will be characterized for each orientation. These comparisons will be presented for low R ratio (0.2) and then repeated for higher R ratio (0.7). In each figure, the degree of environmental enhancement can be judged relative to the current Sect. XI reference crack-growth laws for water environment. These reference curves are statistically determined estimates of the mean crack-growth rate for steels exposed to LWR environment, as will be discussed in detail in Sect. 9. The data base used in their construction was primarily older steels and welds with relatively high inclusion content (e.g., sulfur).

3.3.1 Orientation Effects

The effect of specimen orientation has been investigated for all three heats, and for all three heats, both R ratios have been studied. The specific orientations studied are shown in Fig. 3.1. Figure 3.21 shows there is a significant effect of orientation in the heat with highest sulfur content at $R = 0.2$ with the TL orientation producing the fastest growth rate. This is not unexpected because this orientation involves the crack propagating along the rolling direction (Fig. 3.1). The LT orientation produces slightly slower crack-growth rates, but the slowest growth rates are produced by the through-thickness (LS) orientation. Figure 3.22 shows a similar effect for the medium-sulfur heat with the TL orientation fastest and the LS slowest. There is an initial acceleration in the growth rate of the LS orientation, after which it settles down to a relatively constant growth rate. This initial acceleration may be related to the fact that crack growth begins almost immediately at the start of the test with no incubation period, as has been observed on some other heats.

The effect of orientation is much different for the low-sulfur heat (Fig. 3.23). Here all of the orientations show similar behavior with the level of crack-growth rate about the same as that observed in the other heats for the LS specimens. This behavior is consistent with other measured properties in these steels (e.g., Charpy energy, which shows no orientation effect in low-sulfur steels) primarily because the inclusion content is low and the manganese sulfide inclusions are small; thus, the rolling process does not produce a weak direction.

Observations on the effect of orientation for tests conducted on the same heats at high R ratio (0.7) are similar to those for the low R ratio. For the low-sulfur heat (TW) there is no orientation effect (Fig. 3.24). For the high-sulfur heat (IN), again the TL orientation is producing the highest growth rate (Fig. 3.25). The results of the high R ratio tests include an initial period of very fast growth in nearly every case, after which the growth rate settles down to a more consistent behavior. This period of initially fast growth is not yet understood but

Fig. 3.21. Effect of specimen orientation on crack growth, high-sulfur plate material (LT, TL, and LS refer to ASTM designations of flaw orientation as found in E399).

Fig. 3.22. Effect of specimen orientation on crack growth, medium-sulfur plate.

Fig. 3.23. Effect of specimen orientation on crack-growth rate for low-sulfur plate.

Fig. 3.24. Effect of specimen orientation of fatigue crack growth for low-sulfur plate TW tested at high R ratio.

Fig. 3.25. Effect of specimen orientation on fatigue crack growth for high-sulfur plate PN tested at high R ratio.

is believed to be a transient effect. Therefore, it has been ignored in the comparisons made.

3.3.2 Sulfur Effects

Another way of looking at the results obtained in this study is the comparison of different sulfur levels for a given orientation. The first orientation discussed will be the TL orientation, which produced the highest growth rates. In Fig. 3.26, summarizing results at $R = 0.2$, the specimen with the lowest sulfur content, TW-1, is clearly showing the lowest growth-rate behavior. There is little or no difference in the growth rates obtained in the two heats of higher-sulfur content, indicating that once the sulfur content is greater than a given level, the environmental sensitivity is constant. At high R ratio, again the higher-sulfur steel produced the most environmental enhancement (Fig. 3.27). Results for these specimens are remarkably parallel with growth rates differing by only about a factor of two. Again, note the high growth rate observed at the beginning of the test.

The effect of the sulfur content on crack-growth rates for the LT orientation is the same as for the TL orientation, and results for $R = 0.2$ are shown in Fig. 3.28. Four heats are shown in Fig. 3.28, and again we see that the low-sulfur heat displays the lowest growth rate and the other three heats fall in a fairly tight scatter band. For the TS or through-thickness directions, the behavior was the same regardless of sulfur content, although there was a lot of scatter at the beginning of the test (see Fig. 3.29).

3.3.3 Summary and Discussion

Results obtained in the matrix of tests have been remarkably consistent with the same trends observed at both high and low R ratio. The trends in the degree of environmental enhancement are the same as those observed earlier for fracture toughness and Charpy energy in which the materials and orientations that display the lowest toughness also display the highest degree of environmental susceptibility.

The highest environmental enhancement was found in the TL orientation; the lowest occurred in the LS orientation. Materials with high-sulfur content showed the most enhancement, but the amount of enhancement did not change once the sulfur level exceeded 0.016.

The observed behavior of the LS orientation was about the same regardless of sulfur content and was equivalent to the behavior of all three orientations of the low-sulfur heat TW. This is an important observation because the most interesting crack orientation from the standpoint of reactor vessel integrity would be the through-thickness direction.

To study further the environmental effects observed, a specimen of the low-sulfur heat was tested in a water environment with hydrogen sulfide introduced in the form of an overpressure. The resulting crack growth rate was linear (Fig. 3.30) and was much more accelerated than a specimen of the same orientation tested under the identical loadings in

Fig. 3.26. Effect of sulfur content on fatigue crack growth in PWR environment at $R = 0.2$, TL orientation.

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Fig. 3.27. Effect of sulfur content on fatigue crack growth in PWR environment at $R = 0.7$, TL orientation.

ORNL-DWG 87-4370 ETD

Fig. 3.28. Effect of sulfur content on fatigue crack growth in PWR environment at $R = 0.2$, LT orientation.

Fig. 3.29. Effect of sulfur content on fatigue crack growth in PWR environment at $R = 0.7$, LS orientation.

Fig. 3.30. Effect of environment on crack-growth rate, $R = 0.2$.

the PWR environment. The behavior of the specimen tested in hydrogen sulfide and water was similar to that of earlier specimens of high-sulfur material tested in pure hydrogen sulfide. This similarity indicates that crack-growth rates for any of these steels, regardless of sulfur content, will be nearly the same in an aggressive environment and gives further basis to the theory that the sulfur content of the steel is, in fact, producing an interaction with the PWR environment to affect the growth rates.

The effects of orientation and sulfur content on the environmental enhancement of fatigue crack growth demonstrated here help to explain some part of the rather large scatter found in earlier attempts to compile available data. It seems very likely that even when all of the variables that affect corrosion fatigue crack growth are known, a considerable amount of scatter will remain because the corrosion influence is not a constant factor in a water environment. Crack-growth behavior is much more consistent in either an inert environment or a very aggressive one, like hydrogen sulfide.

3.4 HIGH-SULFUR PLATE MATERIAL BEHAVIOR

This section will present the results of a study conducted to characterize the fatigue crack-growth properties of a commercial heat of pressure-vessel steel with a sulfur level near the high end of the range of contents typical of contemporary practice. The effects of various steelmaking practices on the size and distribution of the manganese sulfide inclusions were investigated, and a detailed picture of the distribution in the plate under study was determined.¹⁰

A matrix of specimen fatigue crack-growth tests was conducted to study the effects of R ratio and testing frequency on the level of crack growth in a demineralized water environment typical of an operating reactor system. To characterize a range of water environments, the water temperature and pH level were also varied. The environment was maintained by a closed flowing system with a continuous overpressure of hydrogen, which maintained the oxygen level at very low levels (typically, <0.1 ppm).

The environment was also monitored with periodic chemistry and thermal conductivity checks and with an electrochemical potential monitoring probe, using a silver-silver chloride electrode (discussed further in Sect. 5).

Study of the microstructure of the tested specimens has also been made, and these observations will be presented and used to develop a complete picture of the effect of the high-sulfur level on crack-growth behavior, both macroscopically and microscopically, in Sect. 7.

A matrix of tests was constructed to provide a detailed characterization of heat CQ2 (Table 3.4). The matrix was designed to study the effects of R ratio (as before), along with temperature and test frequency. The matrix also allowed the study of the effect of boron level in the environment (Table 3.4). With the exception of the boron addition cases, the environment was precisely the same as that used in the experiments reported in Sect. 3.3.

Table 3.4. Test matrix for study of high-sulfur steel

R = 0.2		R = 0.7	
Frequency (cpm)	Specimen	Frequency (cpm)	Specimen
<i>Test temperature 288°C (550°F)</i>			
1	CQ2-5	1	CQ2-6
1	CQ2-8	10	CQ2-11
1 (borated)	CQ2-9		
<i>Test temperature 204°C (400°F)</i>			
1 (borated)	CQ2-12	1	CQ2-7
1	CQ2-15	10	CQ2-19
10	CQ2-14		

The 2TCT specimens were removed from the plate (Fig. 3.31). Four specimens were taken through the thickness, all with the TL orientation. In addition to the specimens, a thin slice was removed, covering the entire thickness of the plate. A full-thickness sulfur print was then made of the plate, and shown in Fig. 3.32, which is actually a reduced scale version of the full-size sulfur print that was ~9 in. thick by 22 in. long. The sulfur print clearly shows the segregation of the larger manganese sulfide particles toward the plate center. This is confirmed by other sulfur prints of the region immediately below the fracture surface of the corrosion fatigue specimens (Fig. 3.31).

To characterize further the sulfur content of plate CQ2, a series of sulfur measurements were then made at 11 positions through the thickness (Fig. 3.32). Measurements were taken at the two different cross sections shown, and results are also shown in Fig. 3.32. The sulfur content was measured from turnings carefully removed by a carbide tripped drill, using a Leco combustion apparatus. As expected, the sulfur content was found to be lower toward the outer surfaces. The highest levels were observed on either side of the center, near the heavier streaks observed on the sulfur print. Measurements at the ends of the sample show that differences exist even 21 in. apart in the plate. The highest levels measured were 0.015 wt %, which is lower than either the ladle or the check analyses reported on the test certificate. The differences could have resulted from different sampling locations or could also have been caused by different measurement techniques because the plate was produced in 1972. The sulfur content from the most recent measurements is likely to be most reliable. Nonetheless, the content is high enough to put the material in the category for which significant environmental enhancement would be expected.

Fig. 3.31. Cutting sketch for heat CQ2 with examples of sulfur prints for different locations in plate thickness.

To be consistent with previous chemical analyses of specimens, however, the check analysis results will be used in reporting the crack-growth-rate results. Thus, the value of 0.025 will be retained for CQ2 in the figures here.

3.4.1 Environmental Fatigue Behavior of Heat CQ2

Crack-growth-rate results were obtained for a number of different conditions at both 288°C (550°F) and 204°C (400°F). Results obtained at 288°C, 1 cycle/min, and $R = 0.2$ are presented in Fig. 3.33, and it is seen that the growth rates are quite accelerated but still within the

SULFUR MEASUREMENTS

LADLE: 0.019		CHECK: 0.025 wt %	
LOCATION	wt %	LOCATION	wt %
1	0.012	12	0.011
2	0.012	13	0.012
3	0.011	14	0.011
4	0.013	15	0.011
5	0.014	16	0.012
6	0.011	17	0.009
7	0.015	18	0.012
8	0.012	19	0.012
9	0.012	20	0.012
10	0.013	21	0.012
11	0.012	22	0.012

Fig. 3.32. Full-thickness sulfur print of heat CQ2, showing concentration of manganese sulfide inclusions near plate center. Locations of actual sulfur measurements are shown.

Fig. 3.33. Comparison of crack-growth-rate results for high-sulfur steel tested at $R = 0.2$. Note similar results for borated and unborated environments.

reference lines of the present ASME Code, Sect. XI curves for this material and environment.⁹

The test environment for these three specimens was the same with the exception that specimen CQ2-9 was tested in an environment containing 1000-ppm boron; the others had none. This changed the room temperature pH from nine to five, although the pH at the test temperature differs by only about one.

The results shown in Fig. 3.33 are interesting in a number of ways. First, the duplicate tests of CQ2-8 and CQ2-5 provide a measure of the scatter to be expected for this material in a water environment. The scatter results mostly from material variability, as discussed earlier. The scatter is roughly a factor of 2, which is typical for these materials tested in a high-temperature water environment. The larger scatter might have been anticipated if the specimens had not come from very similar locations in the plate (Fig. 3.31). The boron addition for CQ2-9 produced data within the scatter band of the other two specimens, indicating a negligible impact.

The results obtained at 204°C and $R = 0.2$ are presented in Fig. 3.34. Two specimens were tested at 1 cycle/min in borated vs unborated environments, and the crack-growth behavior was essentially the same, as found for the tests conducted at 288°C and discussed above. The crack-growth rates observed at 204°C were much lower than those obtained for the 288°C tests.

The third test shown in Fig. 3.34 showed that higher growth rates are achieved at 10 cycles/min as opposed to 1 cycle/min. This is opposite to the effect of frequency at the higher temperature, where 1-cycle/min data are generally the most accelerated.¹¹ This behavior has also been observed by Van der Sluys in tests on the same class of materials.¹²

The effect of testing frequency at a higher value of R ratio ($R = 0.7$) was also investigated at both 288 and 204°C (Figs. 3.35 and 3.36). At 288°C there was no strong frequency effect although the data points obtained at 1 cycle/min were slightly above those obtained at 10 cycles/min. The results at 204°C showed a clear difference with frequency with the 1-cycle/min results being at a higher rate (Fig. 3.36). Interpretation of these high R ratio test results is complicated by the large scatter in the data, which comes from the tendency for growth rates to decrease occasionally as the tests progress. This behavior has also been observed in many other materials tested earlier under these conditions and is discussed in Ref. 8.

As part of a specimen exchange arranged with Materials Engineering Associates, several specimens of heat CQ2 were tested there, and results were reported recently.¹³ As seen in Fig. 3.35, the crack-growth rates obtained were essentially the same as those reported here for $R = 0.2$ and a 1-cycle/min sinusoidal loading at 288°C. The growth rate did not appear to be affected by the flow rate of the test because the tests were done with water flow rates varying from 50 to 1000 L/h. (The systems used in this program have flow rate of ~50 L/h.)

Fig. 3.34. Comparison of crack-growth-rate results for high-sulfur steel tested at 204°C and $R = 0.2$. Note similar results for borated and unborated environments and enhanced growth at 10 cycles/min.

Fig. 3.35. Effect of frequency on fatigue crack growth for high-sulfur heat CQ2 at $R = 0.7$, 288°C , in simulated PWR environment.

Fig. 3.36. Effect of frequency on fatigue crack growth for high-sulfur heat CQ2 at $R = 0.7$, 204°C , in simulated PWR environment.

3.4.2 Discussion

The material characterized in this work can be viewed as typical of the steel plates fabricated for nuclear service in the early 1970s. The enhanced fatigue crack-growth rates observed in the water environment are thought to be strongly influenced by the presence of the manganese sulfide inclusions in the steel, as discussed earlier.

The scanning electron microscope (SEM) has been used to examine the fractures from the fatigue specimens. Before examination all specimens

were cleaned of corrosion product using an electrolytic technique involving scrubbing with cathodic hydrogen. The oxide removal is facilitated further by conducting the electrolytic cleaning in an ultrasonic cleaning apparatus.

Many of the specimens were examined in this manner, and some of the observations support the conclusions reached earlier on the role of sulfur in crack-growth enhancement. The fracture surfaces of these specimens tested at $R = 0.2$ were relatively smooth, showing ductile fatigue striations with some side cracks, as shown, for example, in Fig. 3.37. The side cracks appear to be emanating from manganese sulfide inclusions, but they affect the fractography only locally. This supports the theory that these inclusions are absorbing hydrogen ahead of the advancing crack and that the hydrogen is then making crack growth easier in these

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Fig. 3.37. SEM fractographs of specimen CQ2-12 (a) 10 mm from root of notch (160 $\times$) and (b) 20 mm from root of notch (800 $\times$).

inclusions than in the remainder of the material.⁸ No clear evidence has yet been found that these inclusions fracture ahead of the advancing crack, as has been suggested by Klemetti et al.,¹⁴ but the other findings here are consistent with their observations.

Figure 3.37(b) shows an example of the above observations. When the growing crack encountered the rather large manganese sulfide inclusion shown here, only a local discontinuity resulted. Examination of other specimens showed similar results for the specimens tested at low R ratio, but the high R ratio specimens did not have visible striations, probably because the lower stress ranges employed produced more-shallow striations, which were obliterated by oxide formation. The fracture surface appearance was unchanged for the various cyclic frequencies examined and for the two different test temperatures. The 204°C test specimens did have a much more adherent oxide film, which reduced the resolution of detail on the fracture faces.

3.5 LOW-SULFUR MATERIAL BEHAVIOR

Now that so many results have confirmed the contribution of the material sulfur content to the level of environmental enhancement, it is of interest to consider materials that are known to have very low levels of sulfur. The aim of this study is to determine if there is a sulfur level below which there is no environmental enhancement.

Tests have now been conducted on four different low-sulfur steel plates, and the materials, specimens, and test conditions are shown in Table 3.5. The sulfur analysis of these four plates ranges from 0.004 to 0.006 wt % (Table 3.1). The steels were made by a variety of practices, including a recently manufactured example of fine-grain practice from Japan Steel Works.

Table 3.5. Cyclic crack-growth tests completed on low-sulfur A 533 grade B class 1 plate

Plate designation	Sulfur (check)	Specimen No. (orientation)	
		R ratio = 0.2	R ratio = 0.7
TW	0.004	TW-1 (TL)	TW-2 (TL)
		TW-4 (LT)	TW-5 (LT)
		TW-7 (LS)	TW-8 (LS)
W72C	0.004	W72C-9 (TL)	W72C-6 (TL)
		W72C-8 (TL)	
B	0.006		B-58 (LS)
L4	0.004	L4-7	
		L4-8	

Tests completed on this low-sulfur heat (L4) from Japan Steel Works at $R = 0.2$ show there is a measurable environmental effect. This conclusion was confirmed with a second test (Fig. 3.38). The amount of scatter between the two tests is very small. These results are compared with the results of other low-sulfur steels at $R = 0.2$ (Fig. 3.38). Clearly, the level of environmental acceleration is not unlike that of the other materials tested.

Two heats of low-sulfur steel (sulfur content 0.004 wt %) have been rather extensively characterized and have resulted in remarkably different crack-growth-rate behavior.

Figure 3.39 presents a summary of the results for these two materials at low R ratio and includes two specimens tested by Cullen et al.¹⁵ The results for heat "W7" show predominantly low fatigue crack-growth rates. There is one anomaly, specimen W72C9, which showed accelerated growth, but this appears to be an exception rather than the rule. Extensive studies have been made with no success to find an explanation for this behavior. The specimens were taken from the same position in the plate, and the test environment chemistry was found to be essentially identical for all tests of the "W7" specimens.

There was no indication in test records that in any way would suggest that different results should be obtained. Given the collection of results on this heat, one must conclude that the behavior of heat "W7" is not accelerated.

The other heat of low-sulfur steel, "TW," has been characterized rather extensively in earlier testing in this program,¹⁶ and the results obtained at low R ratio are consistently much more accelerated than those of heat W7 (Fig. 3.39). The test records for heats TW1 and W72C9 were compared in detail, and no differences are obvious. Even the water chemistry results from both before and after testing compare very well for all of the specimens shown in Fig. 3.39. Further study of these two heats was done fractographically and is reported in Sect. 7.

To show the amount of environmental acceleration to be expected in low-sulfur steels at high R ratio, Fig. 3.40 was prepared. The conclusion from this figure is clearly that the environmental acceleration of low-sulfur steels at $R = 0.7$ can be important. The ASME reference crack-growth-rate curve for high R is still applicable, but growth rates are not markedly below it.

3.6 EFFECTS OF TEST STARTING CONDITIONS

A number of previously reported test results have shown that the fatigue crack-growth-rate results in water environments can be affected significantly by the starting conditions of the test. If the specimen is initially loaded at a relatively high value of ΔK (or K_{max}), the mechanically induced crack growth occurs faster than the environment can influence it, and no environmental enhancement is observed. This lack of environmental influence can be continued throughout the test with no enhanced condition ever being attained. These observations were made on high-sulfur steels and are not expected to be applicable to low-sulfur steels.

Fig. 3.38. Comparison of crack-growth results for various low-sulfur plate materials tested at $R = 0.2$ and 288°C in simulated PWR environment.

Fig. 3.39. Summary of crack-growth results for low-sulfur heat W7 compared with results of low-sulfur specimen TW-1 at $R = 0.2$. Note: all loadings were sinusoidal except W72C2, which was ramp/reset.

Fig. 3.40. Comparison of fatigue crack-growth rates for low-sulfur plate materials tested at $R = 0.7$ in PWR environment, showing growth rates near upper bound of Sect. XI.

A series of tests were designed to study these effects and to quantify them as much as possible. A single heat of A 533 grade B class 1 steel was used for most of the tests, and two values of R ratio were used. Tests were done under three types of conditions. The first was the standard test method in which tests were begun at as low a value of ΔK as was feasible to obtain growth and in a reasonable amount of test time; the entire test was conducted with a 1-cycle/min sine wave. The second test type was done in the same manner but starting at a much higher value of applied ΔK . The third test method involved a combination of rapid cycling and slow cycling. The rapid cycling (60 cycles/min) was done at the beginning of the test to attempt to condition the crack tip and obtain accelerated growth without the long incubation times typical of the tests of type 1. All of these tests were carried out on 2-in.-thick (5.08-cm) specimens in simulated PWR environment at 288°C (550°F) and 7.7 to 14 MPa (1100 to 2000 psi). There were two major reasons for performing these tests: to define further the mechanisms affecting the corrosion fatigue process and to investigate the possibility of shortening the testing time required. This time can often stretch from months to years using the conventional methods of test type 1. The tests appear to have been successful in both objectives. Results are summarized in Figs. 3.41 and 3.42.

Results of one test done in the conventional way at 1 cycle/min are shown in each figure. For the low R ratio tests (Fig. 3.41), this is specimen 04A-103; for the high R ratio (Fig. 3.42) the specimen is No. F-1. The upper band for all of the conventional testing done to date is shown by a solid curved line in each figure.

The second test type, which was started at a relatively high value of ΔK , is represented by one set of test results in each figure as well. In Fig. 3.41 specimen 04AE13 was started at a value of $\Delta K = 13.1 \text{ MPa}\cdot\sqrt{\text{m}}$ ($12 \text{ ksi}\cdot\sqrt{\text{in.}}$) higher than the conventional test and showed behavior very similar to the growth rate obtained in air. For the high R ratio tests, specimen 04AE15 was begun at a value of $\Delta K \sim 16.5 \text{ MPa}\cdot\sqrt{\text{m}}$ ($15 \text{ ksi}\cdot\sqrt{\text{in.}}$) higher than the conventional specimen shown. In this case, the results showed much less accelerated crack growth but somewhat faster than the expected growth in an air environment.

The most interesting results are those from the type three tests (Figs. 3.41 and 3.42). In each case, the test was begun at a cyclic frequency equal to 1 Hz, and after the crack was allowed to propagate a predetermined distance, the frequency was dropped to 1 cycle/min. The test times required for these specimens were a factor of 4 to 6 less than required for the conventional (type 1) tests. The reason for this is twofold. First, the majority of time (or cycles) spent during every crack-growth-rate test is at low ΔK where growth is very slow. Second, in constant load cycling, using the CT or WOL specimen design, the ΔK changes very slowly for a given crack extension when the total crack length is short ($a/w < 0.5$). By consuming the short crack lengths in a 1-Hz test and effectively starting the 17-mHz part of the test at an $a/w = 0.6$ rather than 0.3, the total crack propagation distance is cut in half, and the only penalty is a small loss in attainable ΔK toward the end of the test. By halving the total crack extension and avoiding the crack lengths at which ΔK is low and changing slowly, moving instead to a crack length at which ΔK is equally low but changing somewhat more

Fig. 3.41. Summary of starting conditions effects on corrosion fatigue crack growth at R = 0.2.

Fig. 3.42. Summary of starting conditions effects on corrosion fatigue crack growth at R = 0.7.

rapidly, a significant reduction in test time is achieved. Beginning a test at a high value of ΔK does speed up the test but does not produce the desired enhanced behavior, as shown by the results of the type 2 tests. As seen in Figs. 3.41 and 3.42, at both values of R ratio studied, the type 3 tests were successful in both speeding up test time and reproducing the data obtained from conventional tests.

When the frequency was decreased, each of the specimens tested at high cyclic frequency at the start of the test accelerated immediately to the water-enhanced growth-rate behavior obtained from conventional tests. This is shown most clearly in Fig. 3.41 in which it can be seen that growth accelerates nearly vertically from the air behavior for each of the specimens 04AE11, 04AE12, and 04AE14. Note that the conditioning of specimen 04AE11 by fast cycling before cycling at 1 cycle/min results in considerably different behavior than simply starting the test at $35.2 \text{ MPa}\cdot\sqrt{\text{m}}$ ($32 \text{ ksi}\cdot\sqrt{\text{in.}}$). Also note, in Fig. 3.41 that one specimen was cycled at 1 cycle/min (after conditioning) beginning at a ΔK value lower than had been possible on a conventional test. In this case, the crack in specimen 04AE14 grew at a rate slightly higher than rates obtained thus far from conventional testing.

The same behavioral trends were observed in a series of tests done at high R ratio (Fig. 3.42), but this series was much smaller. Only two tests were completed and were done under very similar conditions to check the effect of material type on this behavior. Specimen R-17 (A 508 class 2 forging) and specimen 04AE16 (A 533 grade B class 1 plate) show similar behavior.

These results indicate that in a water environment there is an incubation time associated with crack growth after which "steady-state" enhanced crack growth occurs. It is apparent that a relatively tight band of data results once this steady-state growth is attained. Therefore, the scatter often observed in water-assisted crack growth tests can be at least partly attributed to non-steady-state growth as a result of starting effects.

Note that the starting effects discussed above may occur only on materials highly susceptible to the water environment (or high in sulfur content). There are several examples in which this effect has not occurred.

3.7 SUMMARY OF MATERIALS STUDIED

A rather large number of materials have been characterized in this program since its inception, and this section is an attempt to collect all of the materials and specimens that have been tested. (The material chemistries are given in Table 3.1.) This collection has been attempted in Tables 3.6–3.9, where all of the specimens, materials, and test conditions are listed. The results of all of these tests are discussed elsewhere in this report, mostly in the earlier subsections of this chapter. One exception is the characterization of the A 106 B pipe and weld material for which a brief discussion has been added here.

3.7.1 Summary Tables

The first table presented, Table 3.6, contains the data for the tests discussed in the majority of this report, sinusoidally loaded specimens in simulated PWR environment, at two R ratios. The following table, Table 3.7, contains the specimen numbers and test conditions for the ramp and hold-time tests discussed in Sect. 3.2.

Table 3.6. Summary of corrosion fatigue crack-growth tests
for PWR environment 288°C, sine wave

Material	Sulfur (check)	R ratio = 0.2		R ratio = 0.7	
		Frequency (cpm)	Specimen	Frequency (cpm)	Specimen
A 533 B plate (CQ2)	0.025	1	CQ2-5 (TL)	1	CQ2-6 (TL)
		1	CQ2-8 (TL)	10	CQ2-11 (TL)
		1	CQ2-9 (TL)		
A 533 B plate (IN)	0.026	1	IN-1 (TL)	1	IN-2 (TL)
		1	IN-4 (LT)	1	IN-5 (LT)
		1	IN-7 (LS)	1	IN-8 (LS)
A 533 B plate (PN)	0.016	1	PN-1 (TL)	1	PN-2 (TL)
		1	PN-4 (LT)	1	PN-5 (LT)
		1	PN-7 (LS)	1	PN-8 (LS)
A 533 B plate (TW)	0.004	1	TW-1 (TL)	1	TW-2 (TL)
		1	TW-4 (LT)	1	TW-5 (LT)
		1	TW-7 (LS)	1	TW-8 (LS)
A 533 B plate (W7)	0.004	1	W72C9 (TL)	1	W72C6 (TL)
		1	W72C7 (TL)		
A 533 B plate (HSST plate 02)	0.019	1	2D3Q-6 (TL)	1	2D1Q2 (TL)
		1	2D3C-3 (TL)	1	02GB-2 (TL)
		0.5	2D1Q-2 (TL)	0.5	02GB-5 (TL)
		1	2D3Q-5 (TL)		
		0.1	02GB8 (TL)		
A 533 B plate (HSST plate 04) (04A)	0.018	1	02EF5 (TL)		
		1	04A-103 (TL)	1	04A-102 (TL)
		1	04AE5 (TL)	1	04AE15 (TL)
		1	04AE6 (TL)		
		1	04AE7 (TL)		
Miscellaneous A 533 B plate (1HT)	0.013	1	04AE13 (TL)		
			1HT20		
EPRI round-robin				1	1021T
B76	0.012			1	B76-226-24
B	0.006			1	B58
A-9404		1	A9404-2-1		
Linde 80 weld HAZ (in A508 C2 forging) (C)	0.008	1	C-3 HAZ	1	C-4 HAZ
Linde 0091 weld HAZ (in A 533 B plate) (D)	0.020	1	D-1 HAZ	1	D-4 HAZ
		1	D-3 HAZ	1	D-2 HAZ
				1	D-6 HAZ
Linde 0091 weld HAZ (in A 533 B plate) (DD)	0.017	1	DD-1 HAZ	1	DD-2 HAZ
Linde 124 weld HAZ (in A 533 B plate) (KH)	0.007	1	KH-1 HAZ	1	KH-2 HAZ
		10	KH-4 HAZ		
Linde 124 weld HAZ (in A 533 B plate) (CQ)	0.019	1	CQ-1 HAZ	1	CQ-2 HAZ
				5	CQ-3 HAZ

Table 3.6 (continued)

Material	Sulfur (check)	R ratio = 0.2		R ratio = 0.7	
		Frequency (cpm)	Specimen	Frequency (cpm)	Specimen
A 508 C2 forging (F)	0.012	0.1 1 1	F-5 F-23 (4T) F-24	1	F-1
Linde 80 weld (C)	0.014	1 1	C-3 WLD C-6 WLD	1 1	C-4 WLD C-7 WLD
Linde 124 weld (CG)	0.012	1 1	CG-2 CG-3	1 1	CG-1 CG-6
Linde 0091 weld (D)	0.008	1 1	D-1 WLD D-3 WLD	1 1	D-2 WLD D-5 WLD
A 508 C2 forging (R)	0.025	1	R-17		
Linde 124 HAZ (G)	0.012	1	G1 HAZ		
A 106 B pipe	0.019	1	A 106 B-L04-31		
A 106 B pipe weld		1	A 106 B-LO-W4		

Table 3.7. Summary of ramp
and hold-time tests

Temperature (°C)	Specimen	Ramp time (min)	Hold time (min)
<i>A 508 class 2 forging (F), sulfur (check) 0.012</i>			
288	F-2	0.0166	1,3,12
93	F-3	0.0166	1,3,12
288	F-10	1	0
288	F-9	1	1
93	F-8	1	1
93	F-11	1	3
<i>A 508 class 2 forging (R), sulfur (check) 0.025</i>			
288	R-8	1	0
288	R-11	5	0
288	R-6	5	1
288	R-9	1	12

Table 3.8. Environmental fatigue crack-growth tests
at high frequency — 1969–1971, PWR
environment, sine wave loading

Specimen	Cyclic frequency (cpm)	R ratio	pH	Material
2T-4	600	0.1	5.1	A 533 B (plate 02)
2T-5	60	0.1	5.1	A 533 B (plate 02)
2C-6	600	0.1	5.1	A 533 B (plate 02)
2C-8	60	0.1	5.1	A 533 B (plate 02)
2C-10	600	0.1	5.1	A 533 B (plate 02)
2C-11	600	0.1	5.1	A 533 B (plate 02)
WA-1	600	0.1	5.1	A 533 B (sub-arc weld)
ZD-4	600	0.1	5.1	A 533 B (HAZ)
ZA-3	60	0.1	5.1	A 533 B (HAZ)
WB-1	600	0.1	5.1	A 533 B (electroslag weld)
WD-2	600	0.1	5.1	A 533 B (manual arc weld)
2C-9	60	0.1	7.0	A 533 B (plate 02)
2C-14	600	0.1	7.0	A 533 B (plate 02)
2C-15	600	0.1	7.0	A 533 B (plate 02)
WA-2	60	0.1	7.0	A 533 B (sub-arc weld)
WB-2	60	0.1	7.0	A 533 B (electroslag weld)
WB-3	600	0.1	7.0	A 533 B (HAZ)
ZD-3	600	0.1	7.0	A 533 B (HAZ)

Tables 3.8 and 3.9 contain specimens and conditions for tests done in the early phases of this program. Table 3.8 lists the early specimens tested in simulated PWR environment at high cyclic frequency, which are discussed in some detail in HSST technical report 16.¹⁷ Table 3.9 summarizes the air tests — which considered the effects of specimen size, frequency, R ratio, and temperature on crack-growth rate — conducted over the length of the program.¹⁸

3.7.2 Piping Material Tests

Testing was also completed on samples cut from SA 106 grade B piping material. The specimens were removed from a welded pipe made of 10-in. schedule 80 material, which has a 1-in.-wall thickness. One specimen was removed from the base metal, and the other was removed from the weld. Both specimens had in-plane dimensions equivalent to a 2TCT specimen but were 1 in. thick. The specimens were taken in the L-C direction (circumferential, as shown in Fig. 3.1) and had been flattened and stress relieved before testing.

The fatigue crack-growth-rate results (Fig. 3.43) show environmental acceleration that is significant for the base metal and much lower for

Table 3.9. Summary of test conditions for
A 533 grade B, class 1 steel fatigue
crack-growth study in air
environment (1968-1970)^a

Specimen identification	Specimen size	Test temperature (°F)	Precrack length (in.)
B-10-1	1T	75	0.10
B-10-2	1T	75	0.06
B-10-3	1T	75	0.05
B-10-4	1T	75	0.10
B-1-T	2T	75	0.45
B-2-T	2T	75	0.48
B-3-T	2T	75	1.07
B-4-T	2T	75	2.90
B-29	2T	0	0.20
B-26	2T	75	0.87
B-27	2T	75	0.31
B-28	2T	75	0.23
B-30	2T	75	0.23
B-31	2T	250	0.13
B-32	2T	400	0.04
B-33	2T	550	0.12
B-11	3T	0	0.18
B-17	3T	0	0.22
B-10	3T	75	0.20
B-12	3T	75	
B-13	3T	75	0.62
B-14	3T	75	0.58
B-15	3T	75	0.30
B-16	3T	75	0.20
B-18	3T	250	0.15
B-21	3T	250	0.63
B-19	3T	400	0.20
B-22	3T	400	0.30
B-20	3T	550	0.25
B-23	3T	550	0.48
B-2	4T	0	0.60
B-8	4T	0	0.40
B-1	4T	75	1.30
B-3	4T	75	1.81
B-4	4T	75	1.80
B-5	4T	75	2.00
B-6	4T	75	0.57
B-7	4T	75	1.03
W1 (weld)	4T	75	0.70
W2	4T	75	0.89
W3	2T	75	0.32
W4	2T	75	0.47
W5	2T	250	0.12
W6	2T	250	0.10
W7	2T	400	0.05
W8	2T	400	0.04
W9	2T	550	0.06
W10	2T	550	0.07
533-HAZ-1	2T	75	0.15
533-HAZ-2	2T	75	0.15
533-HAZ-3	2T	75	0.15
533-HAZ-4	2T	75	0.20

^aTest frequency 600 cpm for all tests.

Fig. 3.43. Fatigue crack-growth-rate results for A 106 class B piping and welds tested in PWR environment at $R = 0.2$.

the weld specimen. This behavior is not unexpected, given that the sulfur level of the pipe was relatively high at 0.019 wt %. The growth rates are within the reference crack-growth-rate curve of ASME Sect. XI, as shown in Fig. 3.43. The lower crack-growth-rate behavior of the weld is not unexpected because the manganese sulfide inclusion size in welds is typically much smaller than in base metals for the same bulk sulfur content.

The crack-growth rates observed in these tests are measurably higher than those observed under similar test conditions in Ref. 19 for SA 106 grade C piping steel. This piping steel had a sulfur content of 0.016 wt %. In all cases, however, the ASME code reference curves were found to bound the data.

4. STATIC-LOAD CRACK-GROWTH TESTING

The behavior of cracks in the pressure-vessel steels and welds of interest in a PWR environment has been under investigation, using bolt-loaded specimens from 1974. The specimens are WOL type, 2.54 cm (1 in.) thick and loaded to a fixed displacement by a bolt of the same material. The specimens were positioned in the bottom of two of the operating corrosion fatigue autoclaves.

4.1 EARLY INVESTIGATIONS

The first step in the investigation was the static loading of a 2T-WOL type specimen in an environmental fatigue chamber for a period slightly in excess of 6 weeks. The specimen was precracked in air, and the material was A 533 grade B class 1. The results of this test showed no static-load cracking and are presented in detail in Ref. 20. A tabulation of the results of the test is given in Table 4.1. These results were inconclusive, however, because of the relatively short exposure time. After this test, a more thorough investigation was begun on three different materials.

A series of bolt-loaded 1T-WOL type specimens was prepared and set on the bottom of two of the environmental fatigue crack-growth chambers. The specimens were precracked in air, and the materials were A 533 grade B class 1 (04A specimens), A 508 class 2, and a weld metal typical of that used on the A 533 grade B class 1 plate with Linde 124 flux

Table 4.1. Static-load test in PWR environment at 288°C (550°F) for specimen No. 2DC-4, 2T WOL

Hold segment	Load [N (lb)]	Crack length ^a [mm (in.)]	K [MPa·√m (ksi·√in.)]	Hold time (min)
1	89,500 (18,200)	53.8 (2.12)	42 (38)	16,502
2	107,000 (20,040)	53.8 (2.12)	48.5 (42)	6,961
3	81,200 (22,400)	53.8 (2.12)	54 (47)	6,872
4	90,600 (21,300)	61.2 (2.41)	55 (50)	8,543
5	99,500 (23,000)	61.2 (2.41)	59.5 (54)	7,940
6	94,000 (20,500)	68.8 (2.41)	62 (56)	12,296
7	102,000 (19,700)	83.3 (3.28)	89 (81)	1,391

^aNote that crack length increases are due to fatiguing to mark crack front position.

(C specimens). These specimens were inserted into the environment [288°C and 14 MPa (550°F and 2000 psi)] on April 2, 1975, and were subjected to changes in environment similar to those that might occur in an operating reactor as the environmental chambers are periodically shut down for maintenance and fatigue crack-growth-rate specimen removal and insertion. These specimens have been carefully monitored and some of them destructively examined and studied in some detail. Many of the specimens remained in the environment for over 11 years.

At the same time that the bolt-loaded specimens were inserted into the chambers, a series of rising load K_{ISCC} screening tests was completed in an atmosphere of gaseous hydrogen sulfide at 0.34 MPa (50 psi) and room temperature. This type of test has been used as a measure of the susceptibility of a given material to static load cracking in hydrogen-containing environments.²¹ A precracked specimen is inserted into a hydrogen sulfide gas environment and is subjected to a monotonic slowly rising load. If the material is susceptible to constant load cracking in a hydrogen-containing environment, the specimen will crack before it reaches the maximum load that a similar specimen would reach in an air environment or vacuum. The stress-intensity factor at which cracking occurs is then a measure of the expected K_{ISCC} in the hydrogen environment. The advantage of this test is that it provides a quick screening of susceptibility to constant load cracking without the long incubation times that are often required in a constant load test.

One 1T-WOL specimen of each of the three materials that had been bolt loaded was precracked in air and loaded at a rate of 890 N/min (200 lb/min) up to 26,690 N (6,000 lb), after which the loading rate was dropped to 89 N/min until the specimen fractured. The results of one of these tests are presented in Fig. 4.1 and show clearly that the specimen failure was not influenced by the hydrogen sulfide, indicating that the submerged arc weld was not susceptible to static-load cracking. Results for the other specimens were similar (Table 4.2). The load at first yielding and limit load calculated in Table 4.2 were obtained by a simple ideally plastic analysis detailed by Dowling.²²

After testing, the fracture surfaces of these specimens were examined both macroscopically and microscopically to determine further whether any cracking had occurred before the maximum load was reached. Photographs of the fracture surfaces of the three specimens are presented in Fig. 4.2. From these pictures, it is relatively easy to see what part of the fracture surface was produced in the hydrogen sulfide environment because the gas darkens the steel surfaces considerably. Note that the darkened areas on all of the specimens are simply the precracked surface. The lines that are partly visible in the precracking area resulted from stepping the loadings down to minimize the plastic zone size of the final crack tip before testing was begun. Specimens 04A-110 and C-15 did not break completely during the test and, thus, were separated by saw cutting, as may be seen in Fig. 4.2.

Microscopic study of the fracture surfaces revealed that there was no indication of cracking as a result of the environmental influence. For example, microphotographs of the boundaries between the precracked area and the cracked area for two of the three specimens are shown in Fig. 4.3. In each case, fatigue striations were found on one side of the

Fig. 4.1. Rising load test record — specimen C-15 in 50-psi hydrogen sulfide.

Table 4.2. Results of rising-load hydrogen sulfide tests

Specimen	Material	Yield strength (ksi)	Crack length a^a (in.)	Predicted loads ^b		Actual maximum load (kip)
				Initial yield	Limit load	
Q4A-110	A 533 class 1 HSST plate 04	63.9	1.06	11.8	23.9	20
F-16	A 508 class 2 forging	66.7	1.20	9.7	19.9	17.1
C-15	Submerged arc weld Linde 124 flux	73.2	1.20	10.7	21.8	21.1

^aIncludes precrack.

$$^b \text{Initial yield } P_y = 0.5 (\sigma_y BW) \frac{\left(1 - \frac{a}{w}\right)^2}{2 + \frac{a}{w}} ;$$

$$\text{limit load } P_L = 1.26 \sigma_y BW \left[2 \left(\frac{a}{w} \right)^2 + 2 \right]^{1/2} - \frac{a}{w} - 1 .$$

2X

O4A-110

2X

F-16

2X

C-15

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Fig. 4.2. Photographs of rising load specimens.

Fig. 4.3. Microphotos of rising load tests.

boundary, and typical overload ductile fracture was found on the other side. There was no cleavage or intergranular propagation found that would be indicative of environmental cracking.

The bolt-loaded specimens inserted into the simulated PWR environment in 1975 are tabulated individually in Table 4.3 in which the loading imposed and the number of accumulated hours are also given. These specimens have been visually examined each time the chambers have been opened to remove fatigue crack-growth specimens, and no evidence of crack extension has yet been seen.* In 1976 three specimens were removed from testing and destructively examined to verify further the conclusion. These specimens were 04A-111, 04A-113, and 04A-115, all plate specimens, but loaded to three different values of stress-intensity factor, as seen in Table 4.3. Photographs of the fracture surfaces of these specimens after separation by fatiguing are shown in Fig. 4.4. Views of the fracture surface in the original crack-tip region of each specimen are shown in Fig. 4.5. Note that there is considerable pitting of the

*With the exception of the A 533 grade B specimens (to be discussed later).

Table 4.3. Final status of bolt-loaded specimens in simulated PWR environment

Material	Specimen	Applied stress-intensity factor (MPa $\cdot\sqrt{m}$)	Total time loaded to date (h)	Time in environment (h)
A 533 grade B class 1 plate (HSST plate 4)	04A-112 ^a	110	74,098	48,638
	04A-114 ^a	99	74,098	48,994
	04A-114 ^a	88	74,098	48,994
A 508 class 2 forging	F-13	110	93,166	60,222
	F-14	99	93,166	61,334
	F-15	88	93,166	61,334
Linde 124 weld	C-13	110	93,166	60,559
	C-14	99	93,166	60,559
Linde 124 weld HAZ	CQ-1-HAZ ^a	110	32,980 (1,816) ^b	14,394 (1,262) ^b
	CQ-2-HAZ	99	46,756 (37,876) ^b	23,186 (19,562) ^b
	CQ-3-HAZ ^a	88	33,408 (3,360) ^b	12,274 (2,616) ^b
Linde 0091 weld HAZ	D-7HAZ ^c	99	46,028 (37,148) ^b	31,816 (28,984) ^b
	D-8HAZ ^a	88	37,148 (7,196) ^b	14,384 (4,160) ^b
	D-9HAZ ^c	77	46,028 (37,148) ^b	31,816 (28,984) ^b
Linde 124 weld HAZ	K-3HAZ	66	22,368 (4,320) ^b	12,059 (4,138) ^b
	K-4HAZ	77	22,368 (4,320) ^b	12,059 (4,138) ^b
	K-5HAZ	88	22,368 (4,320) ^b	12,059 (4,138) ^b
Linde 0091 weld HAZ	DD-3HAZ ^c	66	23,808 (5,760) ^b	13,976 (5,280) ^b
	DD-4HAZ ^a	77	19,080 (5,760) ^b	11,144 (5,280) ^b
	DD-5HAZ ^c	88	23,808 (5,760) ^b	13,976 (5,280) ^b

^aRemoved from testing.

^bHours at first observed crack extension.

^cExtension since last report.

2X

O4A-111

2X

O4A-113

2X

O4A-115

Fig. 4.4. Photographs of fracture surfaces of bolt-loaded specimens removed from environment.

Fig. 4.5. Microphotos of fracture surface of bolt-loaded specimens removed from environment.

fracture surface exposed to the environment but again no evidence of static-load cracking. Striations are found immediately after the precrack where the specimens were fatigued apart. A further check was made for the possibility of constant-load cracking by sectioning of each specimen and nickel plating the fracture surface. The results of this work further indicated the absence of static cracking, and an example of one of these metallographic sections is shown in Fig. 4.6.

Fig. 4.6. Metallographic section of nickel-plated fracture surface - specimen 044A-111.

4.2 LONG-TERM BOLT-LOAD TESTS

In 1978 the study of K_{ISCC} behavior was widened to include HAZ specimens, and this decision proved very enlightening because several of these specimens cracked under static load after relatively short exposures. The materials under study were further widened to include other HAZ specimens as a result of this finding. Two of the A 533 grade B class 1 specimens also cracked after >37,000 h in the environment (47,000 total hours).

A maximum of only 17 bolt-loaded specimens could undergo testing in the environment at once because of space limitations. Specimens from A 533 grade B class 1, A 508 class 2, a Linde 80 weld, a Linde 124 weld, and four HAZs were tested over the years (Table 4.3). The specimens were all prepared in the same manner: precracked in air at low loadings in compliance with ASTM E 399 requirements and then loaded to a predetermined value of stress-intensity factor K , which was obtained by tightening the bolt until the face opening of the specimen corresponds to that K as determined by a compliance relationship.²³ If the crack propagates, it should stop at a length corresponding to the threshold for stress-corrosion cracking K_{ISCC} . This value is easily calculated from the final crack length because the face opening of the specimen remains constant even though the load drops as the crack propagates. Thus, the applied K

will decrease as the crack propagates, stopping at $K = K_{ISCC}$. This process is complicated considerably by the fact that incubation times that may be very long in some cases are involved. Even when cracking occurs, as it has in some specimens, it is irregular and occurs sporadically. The method used to expose the specimens to the environment is deficient because the crack length can only be monitored periodically when the autoclave is opened for specimen insertion or repair of seals. Nonetheless, considerable information has been obtained on the behavior of these specimens through numerous removals.

One specimen, C-11-HAZ, that had experienced static-crack growth was studied fractographically in 1980. Low-magnification examination of the fracture surface revealed a very rough surface containing fibrous ridges that emanate from numerous origins and fan-out over the surface parallel to the direction of crack propagation (Fig. 4.7). At higher magnifications, while cleavage-like river lines that again run along the crack growth direction are seen, the extensive oxidation of the fracture surface has largely obscured any fine fractographic features that could be used to interpret the mechanism of crack advance. It is not clear whether the pock marks seen in Fig. 4.7(b) are due to oxidation or reveal local microvoid formation. Further study of the static crack-growth specimens will be conducted in the future, hopefully using specimens with less exposure of the fracture surface.

Cracking under static load has been observed in several of the HAZ materials and in a single heat of A 533 grade B class plate. The incubation time for the HAZs has been rather short, <2,000 h, but for the plate material the time to first cracking exceeded 37,000 h in the environment. No cracking has occurred in the welds or forging material as yet (>93,000 h, 60,000 in the environment). Table 4.3 shows the status of all specimens tested.

A total of 25 specimens have undergone testing in the environment, including specimens from A 533 grade B class 1, A 508 class 2, a Linde 80 weld, a Linde 124 weld, and four HAZs. A history of the specimens that have shown cracking is presented in Table 4.4.

As can be seen in Table 4.4, crack extension appeared to occur unevenly at first and then to even out on each side. As the cracks propagate into the specimen, the applied stress-intensity factor K decreases and should reach a common value for a given material, regardless of initial loading. This appears to have occurred to a degree, but the uneven crack propagation makes accurate calculation of applied stress-intensity factor difficult. The stress-intensity factor has been calculated based on the average of the two side measurements from each specimen. The applied K values appear to have stabilized in a range of 44 to 55 $\text{MPa}\cdot\sqrt{\text{m}}$ (40 to 50 $\text{ksi}\cdot\sqrt{\text{in.}}$) (Table 4.4).

4.3 COMPARISONS OF STATIC AND DYNAMIC (FATIGUE) CRACK GROWTH

Enough measurements of crack length have now been taken to allow calculation of growth rates per unit time, and the results of this calculation are also shown in Table 4.3. Note that this calculation is also approximate because of the uneven crack propagation and the often long

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(a)

(b)

(c)

Fig. 4.7. Fracture surface of bolt-loaded specimen C-11 HAZ.

Table 4.4. Summary of bolt-loaded growth rates^a

Total time loaded (h)	Environmental time (h)	Crack length [cm (in.)]	da/dt (in./h)	K _I [MPa·√m (ksi·√in.)]
<i>Specimen 04A-112</i>				
0	0	2.59 (1.020)		108.2 (99.98)
47,357	37,266	2.82 (1.107)	0.23346E-05	103.7 (94.23)
47,933	37,506	3.45 (1.357)	0.10417E-02	88.8 (80.73)
49,898	39,354	4.04 (1.587)	0.12446E-03	78.9 (71.72)
50,450	39,690	4.37 (1.717)	0.38690E-03	74.3 (67.52)
51,150	39,882	4.37 (1.717)		74.3 (67.52)
51,894	40,554	4.37 (1.717)		74.3 (67.52)
52,818	40,746	4.64 (1.827)	0.57292E-03	70.4 (63.96)
53,530	41,374	5.38 (2.117)	0.46178E-03	55.9 (50.77)
54,426	41,814	5.38 (2.117)		55.9 (50.77)
55,070	42,430	5.38 (2.117)		55.9 (50.77)
55,670	42,430	5.38 (2.177)		55.9 (50.77)
<i>Specimen 04A-114</i>				
0	0	2.59 (1.020)		99.2 (90.20)
46,789	37,314	4.50 (1.767)	0.20019E-04	65.4 (59.48)
47,261	37,482	4.75 (1.867)	0.59524E-03	62.1 (56.45)
47,597	37,722	5.00 (1.972)	0.43750E-03	58.0 (52.69)
47,933	37,962	5.16 (2.032)	0.25000E-03	55.1 (50.12)
98,989	39,810	5.38 (2.117)	0.45996E-04	50.4 (45.81)
50,450	40,146	5.38 (2.117)		50.4 (45.81)
51,150	40,338	5.38 (2.117)		50.4 (45.81)
51,894	41,010	5.38 (2.117)		50.4 (45.81)
52,818	41,202	5.57 (2.192)	0.39063E-03	45.2 (41.28)
54,426	42,170	5.57 (2.192)		45.2 (41.28)
55,070	42,558	5.57 (2.192)		45.2 (41.28)
55,670	42,972	5.57 (2.192)		45.2 (41.28)
<i>Specimen 04A-116</i>				
0	0	2.59 (1.020)		88.1 (80.14)
46,789	37,314	3.20 (1.262)	0.64855E-05	75.2 (68.39)
47,261	37,482	3.58 (1.412)	0.89286E-03	69.0 (62.75)
47,597	37,722	3.66 (1.442)	0.12500E-03	67.9 (61.75)
47,933	37,962	3.86 (1.517)	0.31250E-03	65.4 (59.44)
49,898	39,810	4.34 (1.707)	0.10281E-03	59.8 (54.35)
50,450	40,146	4.39 (1.727)	0.59524E-04	59.2 (53.84)
51,150	40,338	4.89 (1.927)	0.10417E-02	53.2 (48.32)
51,894	41,010	5.25 (2.067)	0.20833E-03	47.3 (43.03)
52,818	41,202	5.95 (2.342)	0.14323E-02	N.A.
55,070	42,554	5.95 (2.342)	0.0	
55,670	42,794	5.95 (2.342)	0.0	

Table 4.4 (continued)

Total time loaded (h)	Environmental time (h)	Crack length [cm (in.)]	da/dt (in./h)	K_I [MPa·√m (ksi·√in.)]
<i>Specimen C-9 HAZ</i>				
0	0	2.59 (1.020)		112.8 (102.50)
1,396	1,152	3.78 (1.487)	0.40538E-03	85.0 (77.20)
1,564	1,308	3.86 (1.522)	0.22436E-03	83.5 (75.87)
2,212	1,764	4.62 (1.817)	0.64693E-03	72.5 (65.92)
2,884	2,100	4.62 (1.817)		71.0 (65.92)
3,700	2,772	4.72 (1.857)	0.59524E-04	66.6 (64.51)
7,444	5,316	4.98 (1.957)	0.39308E-04	66.6 (60.55)
7,948	5,672	4.98 (1.957)		66.6 (60.55)
8,284	5,960	4.98 (1.957)		66.6 (60.55)
8,765	6,128	4.98 (1.957)		66.6 (60.55)
9,092	6,368	4.98 (1.957)		66.6 (60.55)
10,428	6,608	5.21 (2.047)	0.37500E-03	61.8 (56.16)
12,848	8,892	5.21 (2.047)		61.8 (56.16)
13,292	9,756	5.21 (2.047)		61.8 (56.16)
<i>Specimen C-10 HAZ</i>				
0	0	2.59 (1.020)		101.3 (92.09)
1,396	1,152	3.0 (1.182)	0.14063E-03	90.9 (82.65)
1,564	1,308	3.36 (1.322)	0.89744E-03	83.5 (75.87)
2,212	1,764	2.20 (2.007)	0.15022E-02	57.5 (52.31)
2,884	2,100	2.20 (2.007)		57.5 (52.31)
3,700	2,772	2.20 (2.007)		57.5 (52.31)
7,444	5,316	2.20 (2.007)		57.5 (52.31)
7,948	5,672	2.20 (2.007)		57.5 (52.31)
8,284	5,960	2.20 (2.007)		57.5 (52.31)
8,765	6,128	2.20 (2.007)		57.5 (52.31)
9,092	6,368	2.20 (2.007)		57.5 (52.31)
12,848	8,892	2.20 (2.007)		57.5 (52.31)
13,292	9,756	2.20 (2.007)		57.5 (52.31)
<i>Specimen C-11 HAZ</i>				
0	0	2.59 (1.020)		78.7 (71.59)
1,396	1,152	3.30 (1.297)	0.24045E-03	65.8 (59.85)
1,564	1,308	3.30 (1.297)		65.8 (59.85)
2,212	1,764	4.04 (1.592)	0.64693E-03	56.36 (51.24)
2,884	2,100	5.64 (2.217)	0.18601E-02	
3,700	2,772	5.64 (2.217)		N.A.
7,444	5,316	5.64 (2.217)		

Table 4.4 (continued)

Total time loaded (h)	Environmental time (h)	Crack length [cm (in.)]	da/dt (in./h)	K_I [MPa·√m (ksi·√in.)]
<i>Specimen C-9 HAZ</i>				
0	0	2.59 (1.020)		78.8 (71.59)
1,724	1,520	3.40 (1.342)	0.21184E-03	64.1 (58.30)
1,888	1,624	3.82 (1.542)	0.19231E-02	57.7 (52.47)
2,704	3,160	5.55 (2.187)	0.41992E-03	N.A.
9,478	8,206	5.55 (2.187)	0.0	
<i>Specimen C-12-HAZ</i>				
0	0	2.59 (1.020)		101.3 (92.09)
1,704	1,520	3.14 (1.237)	0.14276E-03	87.8 (79.84)
1,888	1,624	3.41 (1.342)	0.10096E-02	82.5 (75.00)
2,704	3,160	5.48 (2.152)	0.52734E-03	49.2 (44.70)
9,478	8,206	5.46 (2.152)	0.0	49.2 (44.70)
<i>Specimen CQ-1HAZ</i>				
0	0	2.59 (1.020)		109.5 (99.58)
1,816	1,262	2.84 (1.122)	0.80824E-04	101.9 (92.62)
3,784	3,110	3.38 (1.332)	0.11364E-03	89.4 (81.26)
4,336	3,446	3.53 (1.392)	0.17857E-03	86.5 (78.63)
4,936	3,638	4.25 (1.672)	0.14583E-02	73.9 (67.17)
5,680	4,310	4.80 (1.892)	0.32738E-03	58.9 (53.52)
9,120	5,954	4.80 (1.892)	0.0	58.9 (53.52)
9,360	6,194	4.80 (1.892)	0.0	58.9 (53.52)
<i>Specimen CQ-3HAZ</i>				
0	0	2.59 (1.020)		78.8 (71.59)
3,380	2,616	2.84 (1.117)	0.37080E-04	73.7 (67.03)
4,108	3,288	4.04 (1.592)	0.70685E-03	56.4 (51.24)
4,732	3,480	4.58 (1.802)	0.10938E-02	51.0 (46.40)
5,644	4,008	4.87 (1.917)	0.21780E-03	47.8 (43.46)
6,540	4,448	4.99 (1.967)	0.11364E-03	46.2 (41.98)
7,188	4,832	4.99 (1.967)	0.0	46.2 (41.98)
8,788	5,072	4.99 (1.967)	0.0	46.2 (41.98)

^aThe calculation of K is not considered applicable for crack lengths greater than ligament, so it has not been reported here.

periods of time between measurements. Nevertheless, the estimated crack propagation rates per unit time are fairly consistent between specimens of the same material.

A comparable time-based measurement of crack-growth rate can be obtained from fatigue crack-growth-rate data through the relationship

$$\frac{da}{dt} = \frac{da}{dn} \frac{dn}{dt},$$

where the quantity da/dn is the cyclic crack-growth rate and dn/dt is simply the cyclic frequency of the test. This calculation has been carried out for the materials that have shown static-load cracking (Table 4.5). The fatigue crack-growth-rate results for these materials all show a discernable plateau in growth rate, and results for these materials have been reported previously in Sect. 3.1.

Table 4.5. Fatigue crack-growth-rate testing
to complement static-load tests

Test material	Frequency (cpm)	Specimen	Frequency (cpm)	Specimen
Linde 124 weld HAZ material "C"	1	C-24-HAZ-1	1	C-23-HAZ-1
Linde 124 weld HAZ material "CQ"	1	CQ-1 HAZ	1 5	CQ-2 HAZ CQ-3 HAZ
Linde 80 weld HAZ material "C"	1	C-3 HAZ	1	C-4 HAZ
A 533 grade B plate plate "04"	1 1	04A-103 04AE 13	1 1	04A-102 04AE 15
Linde 0091 weld material "D"	1 1	D-1 HAZ D-3 HAZ	1 1 1	D-2 HAZ D-4 HAZ D-6 HAZ
Linde 124 weld material "KH"	1 10	KH-1 HAZ KH-4 HAZ	1 10	KH-5 HAZ KH-2 HAZ
Linde 0091 weld material "DD"	1	DD-1 HAZ	1	DD-2 HAZ
Linde 0091 weld material "G"	1	G-1 HAZ		

Comparisons of the growth rates under static and fatigue loadings are shown in Table 4.6. The "average" growth rate reported here was obtained by averaging the calculated results from each time step (Table 4.4). The growth rates obtained in the two types of tests do not

Table 4.6. Comparison of plateau growth rates in bolt-loaded and fatigue crack-growth-rate tests

Average bolt-load plateau growth rate (in./min)	Average plateau crack-growth rate (in./min)
<i>A 533 grade B plate^a</i>	
7.18×10^{-6} (04A-112)	
4.81×10^{-6} (04A-114)	$1.5-2 \times 10^{-4}$
7.85×10^{-6} (04A-116)	
<i>Submerged arc weld HAZ Linde 80 flux with forging</i>	
1.41×10^{-5} (C-10 HAZ)	6×10^{-5} (C-3-HAZ)
1.52×10^{-5} (C-11 HAZ)	6.7×10^{-5} (C-4 HAZ)
4.86×10^{-6} (C-9 HAZ)	
1.42×10^{-5} (C-12 HAZ)	
<i>Submerged arc weld HAZ Linde 124 flux with A 533 grade B plate</i>	
7.18×10^{-6} (CQ-1HAZ)	Testing not complete
7.21×10^{-6} (CQ-3HAZ)	
<i>Submerged arc weld HAZ Linde 124 flux</i>	
	5×10^{-5} (C-24 HAZ-1)
9.33×10^{-6} (C-12 HAZ-1)	4×10^{-5} (C-23 HAZ-1)

^aSpecimen numbers given in parentheses.

show good agreement, indicating that the crack growth obtained in the fatigue tests may not be related to that obtained in the static tests.

Note that the fatigue crack-growth-rate results for the materials that have shown static cracking do show a discernable plateau in growth rate although many of the other materials tested do not show this behavior. This can be seen in the results that have been collected and reviewed in Sect. 3. It has been suggested that the appearance of a plateau in fatigue crack-growth rate indicates that the material may have a susceptibility to constant load cracking, and this appears to be consistent with the results obtained here. Certainly, a conclusion of this generality cannot be made without considerably more verification than has been obtained in this program and, thus, should be viewed more as a speculation at present.

4.4 ACCELERATED TESTS ON WEDGE-LOADED SPECIMENS

A series of tests were conducted on the HAZ of submerged arc weldment "G," using 1-in.-thick (2.54-cm) compact specimens loaded to a fixed displacement with wedges after precracking. The specimens were loaded to different values of applied stress-intensity factor (Table 4.7).

Table 4.7. Results of accelerated tests in 750°F steam

Specimen No.	Initial stress intensity ($\text{ksi}\cdot\sqrt{\text{in.}}$)	Test duration (h)	Average crack growth (in.)
G1	61.5	600	0.053
G2	65.3	600	0.005
G3	65.5	600	0.0
G4	82.0	600	0.0
G5	81.5	600	0.070
G6	81.6	400	0.113

The test environment was steam at a pressure of 3000 psi with a temperature of 750°F. The steam environment contained a small partial pressure of hydrogen of 11 psia. A substantial body of work has shown that test results in this environment for alloy 600 can be smoothly extrapolated to lower-temperature test data in primary water environment.²⁴ The high temperature of 750°F provides a thermal acceleration effect, and the high pressure of 3000 psi creates a liquid-like environment even though the temperature is above the critical point for water. The thermal acceleration effect for alloy 600 can be expressed in terms of an activation energy, which is ~40 kcal. The magnitude of the thermal acceleration effect, if any, for nuclear pressure steels is unknown. The objective of the present experiments was simply to determine if accelerated cracking will occur in the steam test environment.

The tests were conducted for a period of 600 h, with one of the six specimens removed after 400 h for destructive examination. Table 4.7 lists the test specimens, the initial stress-intensity levels, and the crack-growth values as measured on the specimen fracture surfaces. Substantial crack growth was observed in four of the six test specimens. Figure 4.8 illustrates the posttest fracture appearance of specimen G-5HAZ. The dark oxidized area of crack growth during the steam test is clearly evident. Oxidation attack and heavy oxide layers made most SEM fractographic observations inconclusive, even after repeated attempts at surface cleaning. However, the last stages of crack growth before test termination were occasionally well delineated. Figure 4.9 indicates that crack growth proceeded intergranularly.

Fig. 4.8. Photograph of fracture surface of specimen G-5HAZ showing crack extension.

Fig. 4.9. Fractograph of specimen G-5HAZ showing intergranular cracking morphology.

The presence of static-load cracking in the high-temperature, high-pressure steam environment under controlled oxygen conditions adds important information to the previous studies of static-load cracking. Because the bolt-load specimens are periodically exposed to high-oxygen atmosphere, it has been theorized that cracking only occurs during these exposures and is, therefore, unlikely to occur in service. These tests, although accelerated, definitely indicate that such cracking can occur in controlled environments.

Observation of crack growth in the steam tests does provide an approach to accelerated testing with excellent prospects for a direct extrapolation of results to the temperatures of engineering concern. For example, the estimated crack-growth rate at 750°F is on the order of 10^{-4} in./h. With an assumed activation energy of 40 kcal, this rate would become $\sim 3 \times 10^{-7}$ in./h at 550°F, and over 1 year of testing would be required to observe several mils of crack growth. High-temperature tests in primary water environments between 695 and 600°F could be combined with 750°F steam tests to establish an activation energy for nuclear pressure-vessel steels in a reasonable length of time. Establishment of a direct connection between 750°F steam test results and performance in primary water environments would verify a rapid method to economically evaluate a wide variety of nuclear-pressure steels, weldments, and welding processes. Cracking observed in HAZ specimens in 2,000 to 3,000 h, but not observed in welds and forgings and only after over 37,000 h in a

plate material, suggests that the different materials may have different activation energies. This accelerated test technique could provide an excellent method for determining the cracking susceptibility of these more resistant materials.

4.5 SLOW-STRAIN-RATE TESTING

To supplement the bolt-load tests and provide a possible method of screening for susceptibility to constant-load cracking, a series of slow-strain-rate tests were conducted.

The test conditions for the slow-strain-rate testing were those specified for the round-robin tests described at the Sendai meeting.²⁵

Strain rate	10^{-6} s^{-1}
Temperature	288°C
Oxygen	20 ppb
RT conductivity	0.1 $\mu\text{S}/\text{cm}$ deionized water (increased to 0.2 $\mu\text{S}/\text{cm}$ before degassing)
RT pH	6.4
No lithium hydroxide or boric acid additions	

The specimen gauge length was 4 by 0.375 in. in diameter and the specimens were machined with the HAZ running along the axis of the specimen (Fig. 4.10). Specimen orientation was with the specimen long axis through the thickness of the plate or in the plane of the plate (Fig. 4.11). Testing was conducted in the same chambers as the crack-growth-rate measurements, using preoxidized Zircaloy grip inserts to electrically isolate the specimens from the rest of the apparatus. The strain rate was maintained using a signal generator. The water in the makeup tank was deaerated using a nitrogen purge, and a 35-psig overpressure was maintained for the duration of the tests. Oxygen levels were monitored using the Chem-Mets colorimetric technique with the 0- to 100-ppb test kit and samples taken from the low-pressure side before and after the test chamber.

The data obtained are given in Table 4.8. Relatively low ductility failures were obtained in all specimens, including the air test at room temperature. Fractographic examination of the failed specimens revealed only ductile dimpled rupture on all of the fractures (Fig. 4.12). Comparison of the air and environmental tensile data and the fractographic evidence show that no environmentally assisted cracking occurred during these tests.

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Fig. 4.10. Etched cross section of slow-strain-rate specimen 04-1-1 showing base plate, weld, and HAZ.

Fig. 4.11. Schematic showing orientations and locations of slow-strain-rate specimens.

Table 4.8. Results of slow-strain-rate tests^a

Specimen type	Specimen identity	Diameter (in.)	Test temperature (°F)	Ultimate strength ^b		Elongation in 2 in.		Reduction of area	
				(lb)	(ksi)	(in.)	(%)	(diam)	(%)
B	04.1.3	0.3755	75	9900	89.40	N/A	~2	0.3558	10.2
A	24.1.4	0.3755	550	6500	58.70	N/A	~5	0.3553	10.5
B	04.1.2	0.2740	550	900	81.92	N/A	~7	0.3518	11.5
A	23.1.2	0.3745	550	4300	39.04	N/A	~1	0.3644	5.3
B	04.1.1	0.3750	550	6250	59.56	N/A	~1	0.3662	4.6

^aAll specimens at 550°F were tested in deionized water at 1110 psig with 35 psig of nitrogen overpressure.

^b0.2 offset yield strength not available in either pounds or kips per square inch.

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WELD

HAZ

Fig. 4.12. Fractographs of slow-strain-rate test specimen 04-1-1.

5. ENVIRONMENTAL CHARACTERIZATION

5.1 ENVIRONMENTAL CONDITIONS IN CRACK-GROWTH TESTS

It was recognized at an early stage of this program that maintenance of a consistent environment that accurately modeled a PWR environment was essential to reproducible results.

The crack-growth-rate tests were conducted in environmental chambers that are actually small pressure vessels heated to the desired temperature with electrical heaters embedded in the chamber wall. The PWR chemistry is maintained by a flowing system with carefully monitored oxygen and thermal conductivity. The PWR water chemistry is essentially demineralized water with carefully limited chlorides and fluorides to which is added lithium hydroxide and boron, which is used to control the nuclear reaction. A hydrogen overpressure is also added to the system in both the operating reactor and the test autoclave. The resulting test environment remains essentially demineralized water, however, because the boron dissociates at 288°C (550°F). Oxygen is controlled to be <0.1 ppm.

The makeup water chemistry for the tests is provided in Table 5.1. Deoxygenation before test startup was accomplished by bubbling helium through the makeup water until a reading of 0 to 5 ppb was obtained by the Indigo Carmen method of sampling. With closed test loops such as those employed in this program, the oxygen level in the system would trend toward zero even without these steps because of oxidation of the

Table 5.1. Makeup water chemistry — PWR

Electrical conductivity, μ mhos/cm at 25°C	<20
pH	5 to 9
Oxygen, ppm	<0.10
Chloride, ppm	<0.15
Fluoride, ppm	<0.15
Total solids, ppm	<0.5
Carbon dioxide, ppm	<2.0
Particulates, μ m	Filtered to <25
Silica, ppm	<0.2
Li ⁷ OH, M Li	1×10^{-4} (~0.68 ppm Li ⁷)
Boric acid	Variable (0 for pH = 9) (~1500 ppm B for pH = 5)
Hydrogen, cm ³ /kg H ₂ O	25 to 35

steels in the pressure boundary. The helium purge and hydrogen overpressure were used as added insurance.

5.1.1 Early Test Conditions

The first few tests were completed under uncontrolled oxygen conditions with demineralized water and no hydrogen overpressure. Tests completed later in the program confirmed that these early data were representative of data obtained under controlled oxygen conditions, even though the initial oxygen concentration in the water was very high. In the closed test system, the oxygen is quickly scavenged by the large metal surface area. Closely monitored tests showed that the oxygen level dropped from an atmospheric equilibrium (9 ppm) to below the specified limits in 2 to 3 weeks. Although this is a long calendar time, it occurs at the beginning of the test when very little crack growth is occurring.

An example is provided by specimen 04A-102, tested in early 1975 to characterize the oxygen level vs time in the system when no deoxygenation was performed. The oxygen level began at 9 ppm and within 12 d had dropped to 3 ppm. After a total of 2 weeks, oxygen was measured at <1 ppm and shortly thereafter dropped below the specified minimum of 100 ppb. During the initial 2-week period, 10,000 cycles were applied to the specimen, and the crack was estimated to have grown <0.010 in. This is a small part of the life of the specimen, which had a total crack growth of 1.8 in. over 180,000 cycles.

All tests performed from 1975 onward in the program employed careful deoxygenation before testing and had hydrogen overpressure. The specified PWR water chemistry (Table 5.1) was maintained by periodic measurements, as well as deliberate care.

5.1.2 Range of Water Chemistry Used for Tests Since January 1975

In this study, there was no specific attempt to study the effect of the water chemistry on environmental crack growth, but, in fact, three slightly different controlled environments have been used. The basic PWR environment is described in Table 5.1, in which it can be seen that wide variations in the boron concentration are possible because boron is used to moderate the nuclear reaction and the amount required for this task decreases as core life increases. Within this chemistry specification, two environments have been chosen:

1. 1000-ppm boron with lithium hydroxide added to produce a room temperature pH = 6 and
2. no boron with lithium hydroxide added to produce a room temperature pH = 9.

Most of the tests were conducted with the second of these environments after it was found in the material-effects matrix that there was no impact (Sect. 3.1). Table 3.2 presents the details of which environment was used for each specimen for the material-effects matrix. A third

condition was also used, that of simple demineralized water containing no boron or lithium, which is pH = 6 at room temperature.

The pH level at temperature is nearly the same regardless of the three variations at room temperature. For this reason, the effect of these three variations was not expected to be significant, and no clear effect is evident from the results obtained. To ensure that this was the case for even the most environmentally sensitive materials, a second check was made as part of the high-sulfur material test matrix described in Sect. 3.4.

Even with these measures employed in the test program, there remained a need for a simple, yet reproducible characterizing parameter for such a system so that results from different laboratories could be compared easily and reliably.

5.2 ELECTROCHEMICAL POTENTIAL MEASUREMENT

The monitoring or control of electrochemical potential has for many years been an essential feature of corrosion-fatigue and stress-corrosion experiments on cracking of metals in aqueous environments. Although techniques to achieve these aims are well developed for testing at relatively low temperature, say $<100^{\circ}\text{C}$, reliable methods for hot, pressurized conditions typical of LWRs or conventional high-pressure boiler systems were not readily available.^{26,27} With the advent of extensive research programs on corrosion cracking phenomena in structural materials for LWRs, however, there was considerable renewed interest in developing reliable reference electrodes and other electrochemical test methods suitable for such environmental conditions (i.e., water-based environments at temperatures up to 290°C and pressures up to 17 MPa).

In common with many environmental cracking studies, the electrochemical potential of the specimen was quickly identified as a key variable in experimental studies related to pressure vessel steels in boiling water reactor (BWR) and PWR primary coolant environments.

Added impetus to meet this requirement was provided by a publication by Indig²⁸ showing that extremely small concentrations of oxygen, which were difficult to measure directly, could drastically influence the corrosion potential of ferritic steels in high-temperature water (Fig. 5.1). Further, these changes in corrosion potential were known to affect stress-corrosion susceptibility of ferritic pressure-vessel steels in slow strain rate tests.^{29,30}

It was soon found in surveys of the ICCGR participating laboratories that safe and reliable high-temperature reference electrodes were not widely available. Consequently, the group collated the available information and experience and shared the results of developments to various designs. Now, after a few years, principally between 1982 and 1984, reliable hardware and techniques for various types of simulated LWR environments are available. This short time period to move from a basically unreliable technology to effective and reliable designs demonstrates how effective cooperation and technology transfer between the ICCGR members have been. This survey and the resulting conclusions concerning the most reliable systems were reported by Scott and Bamford in 1985.³¹

Fig. 5.1. Corrosion potential data for SA 333-6 carbon steel in water as function of oxygen concentration and temperature.²⁸

Comparison of the various electrochemical potential measurement systems revealed that the two most reliable systems were the silver-silver chloride (Ag/AgCl) system developed by Andresen³⁰ and the hydrogen reference electrode. Both of these systems were developed for installation on the Westinghouse autoclaves.

Two different types of reference electrodes were incorporated into the corrosion fatigue test loops: an external Ag/AgCl reference electrode and two palladium/hydride (Pd/H) reference electrodes. One Pd/H reference electrode was placed in the corrosion-fatigue chamber. Originally, this electrode was used to measure the potential of the CT specimen, which is electrically coupled to the test chamber. In later tests, the potential of an electrically isolated A 508 test specimen in the fatigue chamber was measured with respect to the Pd/H reference electrode. Other electrodes were located in a separate electrode chamber in the autoclave effluent line. The experimental arrangement is shown in Fig. 5.2. The electrode chamber contained an Ag/AgCl reference electrode, a Pd/H reference electrode, an A 508 steel sample, a thermocouple, and a platinum counter electrode.

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Fig. 5.2. Photographs of electrochemical potential monitoring equipment.

Three experiments will be discussed here to provide an example of the output obtained. The first experiment was a confirmation test of the electrodes in a high-temperature environment containing deionized water with a 20-psig hydrogen overpressure on the room temperature makeup tank. Additions of LiOH were made to the test solution to adjust the room temperature pH to 9.0. In the first experiment, the free corrosion potential of the A 508 coupons was measured vs the Pd/H and Ag/AgCl reference electrodes in the test chamber. During the second experiment, a corrosion-fatigue crack-growth test was conducted on test specimen CQ2-8 in a solution with the same water chemistry, which is typical of the other tests conducted in this program in nonborated water. The third test to be reported here was on a low-sulphur steel, specimen L4-8, in the same environment.

The potential measurements conducted for the first test are shown in Fig. 5.3. The potential readings listed are uncorrected. The equilibrium potential of the 0.01 M KCl Ag/AgCl reference electrode is 25 mV with respect to the standard hydrogen potential at temperature. The Pd/H reference electrodes behave at temperature as reversible hydrogen electrodes that are in equilibrium with the test solution of that high-temperature pH. Calculations of the high-temperature pH of this test solution indicate that the pH is 6.33 at 288°C and 6.25 at 275°C. The reference electrodes appeared to operate satisfactorily during the initial 4 weeks of the test. However, careful examination of Fig. 5.3 shows that the two Pd/H reference electrodes did not behave in an identical manner during the exposure. It was noted that the Pd/H reference electrode in the monitoring cell in the effluent line would become

Fig. 5.3. Summary of potential readings from experiment No. 1.

depleted in H_2 more rapidly than the Pd/H reference electrode in the chamber. When the hydrogen pressure inside the reference electrode was recharged, the measured potential quickly dropped ~ 100 mV. The reference electrode in the fatigue chamber measured the corrosion potential of the static CT specimen in the fatigue chamber. Note that this test specimen in the stand was galvanically coupled to the grips and the test chamber. The area of the specimen was small with respect to the total area exposed to the high-temperature solution. Therefore, the specimen potential response was determined more by the test chamber and the grips than was the A 508 test specimen.

The most consistent readings during this interval were made using the Ag/AgCl reference electrode in the effluent line measuring cell. These potential readings were stable and appeared to predict changes in the solution chemistry during the exposure. In Fig. 5.3 note the decay of the corrosion potential during the first 4 d of the experiment. This decrease was attributed to two separate factors, the slow increase in temperature from 300 to 550°F and the consumption of dissolved oxygen in the system. The system had reached a stable test temperature of 550°F by day five, and the corrosion potential maintained a stable value of approximately -480 mV for the next 14 d. On day 19 the test solution reservoir was refilled with fresh makeup solution, and the potential of the A 508 increased by $+120$ mV. This increase in potential was attributed to a higher level of dissolved oxygen in the makeup water. This potential slowly approached the previous steady-state value of -480 as the oxygen inventory of the system was reduced.

There are two significant findings for this set of potential readings.

1. The external Ag/AgCl reference electrode was judged the most reliable of the three reference electrodes in the chamber. The readings were reliable and consistent with operation of the test system.
2. The potential of the CT specimen in the test chamber may be compromised by the galvanic coupling with the test chamber. Possible ground loops between the CT specimen and the loading part of the fatigue system could also influence the readings. In future experiments, an electrically isolated coupon of A 508 will be included so that the uncoupled potential of the CT specimen can be recorded.

Electrochemical potential measurements were made during the corrosion-fatigue experiment for specimen CQ2-8. The data are shown in Fig. 5.4. The trends were comparable to those reported during the previous experiment. There appeared to be a slight difference between the potential measured by the Pd/H reference electrode and the electrically isolated steel coupon in the fatigue chamber and the Pd/H and steel samples in the monitor cell in the effluent line. As the experiment continued, the two potential readings appeared to converge. The potential readings given by the Ag/AgCl electrodes were stable. Note that this experiment was interrupted at 5860 cycles for ~ 10 weeks because of equipment malfunctions. During this interval, the system was cooled to room temperature until restarted. During this 10-week interval, the Ag/AgCl reference electrode remained in the effluent cell. When the test was restarted, the

Fig. 5.4. Summary of potential readings observed during testing of CQ2-8.

corrosion-potential measurements were ~80 mV more active than before. When the experiment was over, the electrode was calibrated against a commercial standard. The posttest calibration indicated that the equilibrium potential of the Ag/AgCl reference electrode was +200 mV, which can be correlated with an internal chloride concentration of 0.001 M KCl, showing that the electrolyte had become somewhat diluted. Presumably, a large portion of this shift in potential occurred during the system pressure changes from high to ambient and the 10-week exposure to the dilute test environment during shutdown.

Data taken for the potential of the steel vs the Ag/AgCl electrode during testing of specimen L4-8 are shown in Fig. 5.5. The potential started at approximately -300 mV and drifted to more-negative potentials during the course of the experiment. This Ag/AgCl electrode was a slightly different version than was used previously. The internal chamber was filled with a more concentrated solution to maintain a more stable potential. A large proportion of the internal volume was occupied with an asbestos wick to decrease the solution inventory. To test the long-term reliability of the electrode, the electrode was continuously exposed to the test solution for 1 month. The electrode was calibrated vs a commercial Ag/AgCl reference electrode at room temperature at the beginning and at the end of the exposure. Its starting equilibrium potential was +240 mV vs SHE. Figure 5.6 indicates that this potential corresponds to a starting chloride concentration of 0.8 M. The calibration of the electrode at the end of the experiment was 282 mV vs SHE, indicating that the chloride concentration had dropped to ~0.015 M. This

Fig. 5.5. Potential readings, A 508 steel vs Ag/AgCl references electrode, corrosion fatigue test L4-8.

40-mV shift, although not ideal, can be tolerated. A much lower potential shift would be anticipated if the electrode were periodically valved into the solution stream.

The potential shift of the reference electrode is much less than the 300-mV negative shift in corrosion potential of the pressure-vessel steel. These data suggest that the corrosion potential of the steel in the corrosion-fatigue test environment is decreasing with time as the dissolved oxygen in the test solution decreases and the protective magnetite film develops. The potential values reported at the end of the exposure period are consistent with those reported by other investigators.

The other electrode measured is the Pd, an improved version of the electrode that had been previously used on this program. The key development for this electrode was inclusion of an internal support system to help support the high external pressure. With this internal support, it is no longer necessary to use a high hydrogen pressure to balance the pressure across the membrane. The potential measurements for this electrode are shown in Fig. 5.7. The electrode was generally operated as a

Fig. 5.6. Equilibrium potential of Ag/AgCl reference electrode (25°C).

hydrogen electrode. As a hydrogen electrode, its potential varied with pH and the dissolved hydrogen partial pressure. The pH of the test solution was calculated to be 6.7 at 280°C in a solution containing 500 ppm boron as boric acid and 1.0 ppm lithium as LiOH. For a hydrogen partial pressure of 1 atm (unit activity), the reversible potential of this electrode should be -736 mV vs SHE. A small correction should be made for the actual dissolved hydrogen partial pressure in the test environment, which is determined by Henry's law for hydrogen in water.

An equilibrium partial pressure of 50 psia was placed on the room temperature makeup tank. The equilibrium-dissolved hydrogen gas concentration for this partial pressure at room temperature is ~60 mL (STP)/kg

Fig. 5.7. Potential readings, A 508 steel vs a palladium hydride electrode — corrosion fatigue test L4-8.

of water. This dissolved hydrogen concentration would be expected to remain essentially constant as the test environment is pumped around the system. At any location in the test loop, the hydrogen partial pressure should be defined on the basis of the Henry's law constant at that temperature. At 280°C this partial pressure should be 11 psia. Attempts were made to measure the dissolved hydrogen concentration by using the Pd electrode in reverse. The inside of the electrode was evacuated, and an equilibrium condition was established. Hydrogen gas diffuses through the Pd tube until the hydrogen partial pressure on the inside of the sensor equals the hydrogen partial pressure on the outside of the sensor exposed to the test environment. These measurements suggest that the hydrogen partial pressure in the test environment is 4.5 psia, only 40% of the anticipated value. The reason for the difference between the expected hydrogen partial pressure and the measured value remains unknown. There were small solution leaks at some of the sliding seals in these experiments. Perhaps hydrogen is being depleted from these regions.

As was expected, the potential measured between the steel sample and the reference electrode changed with changes in the hydrogen partial pressure of the Pd tube. When the inside of the Pd tube was evacuated with a mechanical roughing pump, the potential of the Pd increased with respect to the steel. Likewise, when the hydrogen partial pressure inside the Pd electrode was increased to >1 atm, the potential of the Pd

electrode decreased, and the potential difference between the Pd electrode and the steel sample approached zero.

This series of potential readings demonstrated the usefulness and feasibility of measuring corrosion potential during actual corrosion-fatigue experiments. The importance of pretest and posttest calibration of the reference electrodes vs a commercial standard was also emphasized. Additionally, the Ag/AgCl reference electrode should be removed from the test solution if down times in excess of a few days are anticipated. Properly maintained, the Ag/AgCl electrode appears to be a more sensitive monitor of the environ than the others.

In summary, the potential measurements appear to have proceeded well. Readings obtained were reasonably consistent with those reported by other laboratories testing in simulated PWR primary water. The equilibrium potential of the Ag/AgCl electrodes was still decreasing with time as the chloride inventory decreased inside the electrode. This decrease should be much lower when the electrodes are valved into the system periodically. Interesting data are being obtained with the palladium electrode relating the corrosion potential to the local hydrogen partial pressure.

6. MECHANISMS OF ENVIRONMENTAL CRACK GROWTH

It was recognized after several years of testing in this program that it would be impossible to examine all of the variables that had been found to affect the level of environmental enhancement. Some of the needed tests would simply take too long to accomplish, and, furthermore, each new test series seemed to turn up new and unstudied effects. Therefore, a study of the mechanisms that are active in high-temperature water environment crack growth was begun in 1974. This section will review the early work conducted in this program⁸ and provide links with later work that was an outgrowth of the ICCGR group, a cooperative group discussed in Sect. 8.

6.1 EARLY WORK

In attempting to understand the mechanisms of corrosion-fatigue crack growth in the PWR environment, a comprehensive review of the literature was undertaken, and extensive microscopic examination of the crack surfaces of tested specimens was conducted. Also, an auxiliary series of crack-growth-rate tests was conducted on these steels in an environment of 60 psi of hydrogen sulfide gas. These tests were conducted at room temperature and 93°C (200°F), and results were obtained at $R = 0.2$ and 0.7 . The crack-growth-rate behavior was very similar to that obtained in the PWR environment at the same R ratio at low and intermediate values of ΔK , as shown, for example, in Fig. 6.1. At the higher values of ΔK , the crack-growth rate in the PWR environment increases at a slower rate with ΔK , and, thus, the data tend to curve over, but the growth rate in hydrogen sulfide continues to increase steadily in this region (Fig. 6.1).

Crack-growth-rate data available on pressure-vessel steels in PWR environment indicated that there are two distinctly different regimes of behavior for the most accelerated crack-growth-rate data. At low values of ΔK , the crack-growth rate increases rapidly with ΔK ; however, at higher values of ΔK , the rate of increase declines, and the data curve over. These two regimes will be discussed separately; then the results of microscopic examination of the fracture surfaces will be presented.

6.1.1 Low ΔK Behavior

The present data indicate that there is an apparent threshold ΔK below which there is little or no crack growth. This apparent threshold is dependent on the R ratio of the test, as shown schematically in Fig. 6.2. From the two R ratios tested (0.2 and 0.7), the data can be correlated to a function of $K_{\max}$, the maximum applied stress-intensity factor. This agrees with the model for thresholds ΔK developed by Richie, using a critical stress criterion for crack growth.³² On the basis of this model, the load ratio R only influences threshold ΔK (ΔK_{th}) in an environment where the influence of hydrogen in reducing the fracture stress is related to the stress-assisted accumulation of hydrogen in

Fig. 6.1. Schematic of frequency and R ratio effects — pressure vessel steels in LWR environment.

Fig. 6.2. Fatigue crack-growth-rate results — pressure vessel steels in PWR and hydrogen sulfide environments.

the region of hydrostatic tension near the crack tip. For a given yield strength, temperature, and environment, the amount of hydrogen accumulated is influenced only by $K_{\max}$. The model further predicts that ΔK_{th} should be influenced by the environment. The greater the unstressed equilibrium concentration of hydrogen that the environment introduces into the steel, the lower ΔK_{th} .

Based on this model, it would be expected that the hydrogen-sulfide (H&S) environment would produce a lower ΔK_{th} than the high-temperature PWR water environment. The present data for the H&S environment have a finite slope, and results are not available at very low growth rates where ΔK_{th} would be defined; thus, a precise value has not been obtained. It is clear from the data now available (e.g., Fig. 6.1) that the data extend beyond the apparent ΔK_{th} observed for the PWR environment and are, thus, consistent with the model.

The model predicts that threshold ΔK should be independent of frequency or loading rate because it is based on equilibrium segregation behavior. The data now available are consistent with this model, as are other data obtained for similar material/environment combinations in the literature.³³⁻³⁵

Other models for environmental threshold ΔK based on the idea of a critical crack-opening displacement (COD) to expose fresh metal to the environment do not explain this dependence on environment. Although the lack of any static stress-corrosion cracking at that time implied the necessity for cracking a protective film to allow hydrogen to enter the material, the crack-growth enhancement, as well as the threshold behavior, appears to be directly related to the amount of hydrogen present. This conclusion may still be valid, even though static-load cracking has been observed in some cases. The lack of a definite relationship between the static and dynamic cracking was discussed in Sect. 4. The apparent threshold ΔK occurs at a value well above the threshold ΔK measured for these steels in an air environment of $\sim 2 \text{ ksi} \cdot \sqrt{\text{in.}}$.³⁶

Although no deliberate attempt was made in this study to determine ΔK_{th} for the PWR environment, specimens have been cycled with no observable growth for up to a month at lower values of ΔK than the apparent threshold, which would indicate that the effect is real.

Apparent thresholds for fatigue crack growth in the environment greater than the value in air and corrosion-fatigue crack-growth rates higher than those in air have been frequently observed in previous work^{33,34,37-39} on fatigue crack growth of steels below K_{ISCC} , although no attempt has been made to study this effect in detail by using the precautions necessary for ΔK_{th} determination. One of the major reasons for this lack of information is that the environmental enhancement only occurs at low frequencies, and, thus, the time required for such tests would be prohibitive. Several investigations indicate that the apparent threshold measured coincides with the initially applied ΔK .^{35,39} This agrees with behavior observed in the present study when higher initial values of ΔK were used. This effect is illustrated and discussed in Sect. 3.6, which shows that specimens started at different levels of ΔK show a different crack-growth behavior. The effect of higher initial loading is to retard the environmental effect for some time, and this time can often last the full period of the test. The cause of this behavior appears to be related to the kinetics of the water-steel system

but is not yet well understood. Such behavior is not obtained in systems with slower kinetics (steel-air) or systems with faster kinetics (steel-hydrogen sulfide).⁴⁰

6.1.2 Behavior at High ΔK

Above the threshold ΔK the crack-growth rate increases with ΔK much faster in the PWR environment than in air (Fig. 6.2). As in static stress-corrosion cracking, the synergistic effect of increasing K and increasing the available hydrogen combine to give this effect.

The steep rise does not continue, but at a particulate value of ΔK , the growth rate slows down, becoming nearly independent of ΔK . The point at which this occurs is dependent on the frequency, increasing with decreasing frequencies (Fig. 6.1). This effect was also observed by Vosikivsky.³³ It is also dependent on the environment — in the hydrogen sulfide atmosphere no plateau region is developed (Fig. 6.2).

Although plateau regions in fatigue crack growth are frequently seen in aluminum, magnesium, and titanium alloys,⁴¹ their incidence in low-alloy steels is only occasionally reported^{33,34,37} with results more often being parallel to air or to a slight bending over.^{35,38,42,43}

This plateau region analogous to stress-corrosion cracking behavior can be related to an environmentally controlled phenomenon, the requirement that the crack must wait for the hydrogen to build up to the concentration required for propagation. Testing at a slower frequency allows more hydrogen to accumulate per cycle and, thus, allows a faster crack-growth rate. By adding a kinetically faster environment to produce hydrogen — the hydrogen sulfide — no plateau was seen in the growth rate because hydrogen can be supplied at a fast enough rate to sustain growth.

One aspect of this plateau region is the tendency of the cracks to branch; in the absence of a plateau region, however, crack branching is self-limited because of decreasing K at the tip when a branch forms and the high K dependence of the crack growth. With the plateau region the strong K dependence no longer exists, and when the applied ΔK level is sufficiently high ($\sim 3 \Delta K_{th}$), then macrobranched cracks have sufficient driving force to allow propagation.

The macrobranching is also strongly dependent on the material being tested. Macrobranches have not been observed in forgings, for example, and seem to be limited to plate material, which displays some effect of the rolling process. Earlier specimens from HSST plate 02 tested at $R = 0.2$ with the crack plane parallel to the rolling direction showed extensive macrobranching at a value of $\Delta K = 2.3 \Delta K_{th}$.⁴⁰ Subsequent tests on the same plate with a crack plane perpendicular to the rolling direction with $R = 0.67$ showed much less propensity for macrobranching with only one specimen (02GB4) showing an extensive branch, although microbranching was observed on all specimens, both plate and forging. One explanation for the lack of extensive branching is that at the higher R ratio the test must be concluded before the applied ΔK reached the required level of $\sim 3 \Delta K_{th}$.

It is not clear why specimen 02GB4 branched, but it appears to have originated at a beach mark produced at 10 Hz to correlate compliance and crack length before returning to the 0.1-cycle/min testing rate. Because

no data were taken during the beach marking process, a check on the plateau requirement could not be made. More limited crack branching events were also apparent in specimen 02GB-5, originating during the loading at 1 Hz used to mark the specimen and continuing once the specimen was returned to the test frequency of 1 cycle/min. A similar behavior occurred in specimen 04A-102 from HSST plate 04 (Fig. 6.3).

Fig. 6.3. Branching originating at beachmark specimen 04A-102, HSST plate 04 in PWR environment - 1 cycle/min, $R = 0.7$.

Again, the branch was initiated at the 1-Hz beach mark, and although this mark was only 0.025 cm (0.01 in.) long, the branch continued for 0.508 cm (0.2 in.) into the specimen after the test conditions were returned to 1 cycle/min. The crack-growth behavior for this specimen was well developed and shows a gradual development of the plateau region (Fig. 6.4); however, some arrests were seen.

The hydrogen-sulfide specimens, which did not show any plateau region, also were completely free of macrobranches. This is consistent with the proposed model.

Fig. 6.4. Fatigue crack growth — PWR environment specimen 04A-102, A 533 grade B class 1, 1.10 cycles/min, $R = 0.7$.

6.1.3 Fractography

The mechanisms of fatigue crack growth in the present experiments are dependent on both the material and environment. Examination of the fracture surfaces indicates that different mechanisms of crack growth are occurring in the base metal and the weld specimens (Fig. 6.5).

In the base metal, both A 533 grade B class 1 and A 508 class 2, all of the crack propagation in the PWR environment occurs by ductile striation formation. The striation spacings measured agree completely

ORNL PHOTO 5564-87

2000X

(a) SPECIMEN F-1 PWR ENVIRONMENT, 1 cpm, $R = 0.7$

2000X

(b) SPECIMEN C-1 PWR ENVIRONMENT, 1 cpm, $R = 0.7$

Fig. 6.5. Fracture surface comparison of plate and weldment showing different appearances (striations vs dimples).

with the macroscopic growth throughout the range of applied stress-intensity factor where measurements could be made. This agreement held for all values of R ratio and cyclic frequency, confirming the absence of any other environmentally induced cracking mechanism.

Another mechanism of crack growth, void coalescence, occurs in the weld specimens. This mechanism occurs in both air and PWR water environment, but the crack growth in the PWR environment is considerably

enhanced, as has been observed previously in both base metal^{44,45} and welds.⁴⁶ In the present experiments the transition in fractography from striations to void coalescence is related to the much higher density of large carbide particles in the weld metal (Fig. 6.6). Voids form at particle interfaces as a result of plastic strain at the crack tip, and the strain required decreases with increasing particle density.

The influence of hydrogen on the nucleation and growth of microvoids in the plastic zone has been recently reviewed⁴⁷ and could include increased nucleation through a decrease in the particle matrix cohesive energy or growth because of a buildup of hydrogen pressure in the void.

ORNL PHOTO 5565-87

Fig. 6.6. Extraction replica of carbides in (a) base metal specimen 02GB-3, (b) weld metal specimen C-2 showing larger spherical particles in weld metal.

The enhancement of growth rates in the water environment implies that hydrogen is able to penetrate the plastic zone to the extent of about the COD during the cyclic test, which is 7.6×10^{-4} cm (~ 1 μ in.) for an applied stress-intensity factor of $27.5 \text{ MPa}\cdot\sqrt{\text{m}}$. With a diffusion coefficient of $\sim 10^{-7} \text{ cm}^2/\text{s}$, only ~ 6 s would be required for hydrogen transport by diffusion, so clearly this does not impose a kinematically limiting effect at the test frequencies imposed (1 cycle/min imposes a 30-s rise time). In view of this, it is not surprising that although the microvoid coalescence represents a much different mechanism, it results in a similar crack-growth rate.

The mechanism whereby the environment is able to assist the crack growth occurring by the plastic flow process in ductile striation formation is not clear because of the minor role played by hydrogen on the plastic flow properties of low-alloy steels.⁴⁸ The observed effect is, however, consistent with the proposal of Beacham in which hydrogen is envisioned to aid whatever process the crack tip wants to take, implying hydrogen lowers the stress required for dislocation motion.⁴⁹ This idea is recently supported by experiments of Bernstein, which could be interpreted to imply that hydrogen reduces the frictional stress and, thus, makes dislocation motion easier.⁵⁰

Another mechanism that has been advanced that could explain the influence of hydrogen is the reduction in surface energy associated with the creation of a slip step on the surface.⁵¹ This model also considers the effect of hydrogen in lowering the surface energy associated with cleavage crack propagation. The mechanism of crack advance by ductile plastic deformation is shown to be in delicate balance with the cleavage crack propagation (illustrated schematically in Fig. 6.7).

In a PWR water environment at 288°C (550°F), the specific surface energy associated with slip γ_G is lower than that required for cleavage γ_{cl} , resulting in ductile striations. As the temperature of the test is lowered, the crack-growth mechanism begins to switch to a cleavage mode. A specimen tested at 93°C (200°F) in the water environment showed a mixture of ductile striations and cleavage,⁵² indicating that γ_{cl} and γ_G were nearly equal. Other tests conducted in room-temperature demineralized water have produced almost entirely the cleavage mode of crack growth,⁵³ which agrees well with the model because γ_{cl} is lower than γ_G .

Hydrogen sulfide gas lowers the value of γ_{cl} to a much greater extent than γ_G , resulting in a cleavage mode of propagation at both room temperature and 93°C (200°F). Although the diagram of Fig. 6.7 is schematic and drawn arbitrarily to be consistent with observations, it does predict, assuming the environment is able to induce sufficient hydrogen, that the mode of propagation will revert to ductile striations at higher temperatures in the hydrogen sulfide.

One of the main results coming out of the fractographic studies is the correlation of high-sulfur content with the appearance of brittle regions connecting adjacent areas of striated growth that had deviated in elevation from the macroscopic fracture plane. Previous studies had noted the appearance of this component in plate material and in HAZ fracture surfaces.

In the test matrix on sulfur level and orientation (Sect. 2.3), going from 0.004 to 0.026 sulfur clearly promotes the brittle component

Fig. 6.7. Influence of environment on crack-tip behavior.

independent of orientation. Attempting to correlate the presence of this brittle component with crack-growth enhancement is, however, not straightforward because the low-sulfur series of specimens shows definite enhancement in crack-growth rate compared with baseline air data. One of the clearest examples of the effect of sulfur level is in the TL orientation, specimens IN-1, PN-1, and TW-1 (Fig. 3.26). Although TW-1 does show enhanced growth at low ΔK , at higher ΔK values the effect diminishes. Apparently, in this series of tests, the sulfur is making an additional contribution to the crack-growth rate so that the higher-sulfur specimens are able to sustain a higher growth rate even though the low-sulfur specimen also is enhanced relative to air data.

Additional indications that sulfur can contribute to the growth are provided by the tests conducted in water doped with H_2S (Fig. 6.8). In this case, the growth rate of the low-sulfur material is accelerated greater than the high-sulfur specimens, and the bend-over in growth rate normally experienced by the low-sulfur specimen is avoided. In the $H_2O + H_2S$ tests the environment has become sufficiently aggressive that the primary mechanism of crack propagation has been shifted from the ductile striations typical of high-temperature water to cleavage propagation seen previously in gaseous H_2S testing. Apparently, in the testing of the high-sulfur specimens, insufficient modification of the crack-tip water

Fig. 6.8. Effect of environment of fatigue crack growth, R = 0.2.

chemistry has occurred to allow this propagation mechanism to operate in precisely the same manner.

The mechanism responsible for the development of the brittle areas and the crack-growth enhancement of the high-sulfur specimens is not clear. Based on the sulfur prints and fractography, it appears unlikely that the effect results from a simple mechanical effect associated with inclusions, as discussed by Wilson.⁷ Only in specimen IN2 (see Sect. 7.3) was there any evidence of inclusion segregation on the fracture surface, and even in this specimen there appears to be no strong influence on local crack behavior as a result of the inclusions present.

A more likely possibility is the local alternation in crack-tip chemistry caused by a reaction of the sulfide with the water. Apparently, this creates a condition at the crack tip that promotes local crack-front deviations from the main fracture plane and produces brittle striations. It would be interesting to conduct a series of tests to determine if a low level of sulfur in the water would produce this fractography and, if so, what level would be required.

This local enhancement of crack-tip chemistry model is also consistent with the results obtained on specimen TW5. Apparently, at the start of the test, the environment was sufficiently aggressive to cause brittle areas to form and very high crack-growth rates (Fig. 3.29). As the test proceeded, whatever the origin of the aggressive species in the crack-tip, water was dissipated, and the fractography and growth rate returned to that more typical of the low-sulfur material.

The crack-tip environment is, therefore, the key in understanding the mechanisms of crack-growth enhancement. One of the best methods of characterizing this environment has been found to be the free corrosion potential of the system. One of the first characterizations of the potential of a LWR system was produced by Indig.²⁸ He found that the oxygen level was clearly reflected by the measured electrochemical potential, as discussed further in Sect. 5. One of the key areas of emphasis in this program has been to determine the most-reliable methods of measuring the electrochemical potential and to implement them. Results of some of the tests with different measurement systems are discussed in Sect. 5. This work has been coordinated with a large number of other laboratories through the ICCGR international group, which has advanced the state of the art considerably in this area.³¹

Another important variable influencing the crack-tip behavior is the flow rate. Faster flow sweeps the water out of the crack-tip region, making it more difficult for species (such as sulfur) to concentrate there. An example of this effect (Fig. 6.9) involves testing the same material in a high-flow-rate rig [United Kingdom Atomic Energy Authority (UKAEA)] and a low-flow-rate rig (Westinghouse). The crack growth was only slightly accelerated in the high-flow-rate rig.

Another way of explaining environmental influences has been suggested by Scott³¹ and others and involves the crack-tip strain rate. This strain rate affects the ability of the environment to influence the cracking and does explain the frequency and wave-form effects that have been observed. More work is under way in this area, and the approach shows promise of providing a consistent characterization of environmental effects.

Fig. 6.9. Crack-growth-rate data obtained on pressure vessel steels in water environment using two different autoclave systems.

6.2 ENVIRONMENTAL CHARACTERIZATION AND MODEL DEVELOPMENT

A great deal of progress has been made in relating the character of the water environment to the level of crack-growth enhancement. It is evident that the water in the immediate vicinity of the crack tip is of the most importance in determining the rate of crack advance, and the corrosion potential and pH are the key characteristics. A number of studies⁵⁴⁻⁵⁶ have confirmed that under controlled oxygen conditions the potential and pH conditions at the crack tip are essentially the same as

those in the exposed crack mouth. These values are measurable in operating power systems and offer some hope of quantification of these effects.

Another important finding, which complicates the environmental quantification, is that the flow of various species in and out of the crack-tip vicinity can have an important impact on the crack-growth rate. This flow is directly related to the bulk flow, which is also a measurable quantity in an operating system, but the flow rate varies considerably, from nearly static to over 0.305 m/s (1 ft/s). Attempts to characterize the effects of flow on crack-growth enhancement are only in their early stages.

The development of global models to characterize the enhancement of crack growth from water environments is complicated because there appears to be a strong interaction between the water itself and the material in which the crack is progressing. This interaction results because manganese sulfide inclusions in the steel dissolve readily and produce in the crack-tip region sulfur-rich anions that clearly affect crack-growth rates. Of course, the amount of manganese sulfide exposed to the environment is dependent on the size and distribution of the inclusions in the steel, as mentioned earlier.

Two different global models have been developed to quantify these effects, one based on slip dissolution (or film rupture) and the other based on a hydrogen embrittlement mechanism. The slip dissolution model is based on the hypothesis that the crack propagation rate is governed by the frequency of oxide rupture at the crack tip and its relation with the passivation rate of the newly exposed material and has been described in detail.⁵⁷ The hydrogen embrittlement model is based on the theory that hydrogen is trapped in the material ahead of the crack, making its advance easier.⁵⁸ Because manganese sulfide inclusions are known for trapping hydrogen, this theory relates more strongly to the material effects on crack-growth rate. It has also been suggested recently that crack advance may occur by a combination of the two models.⁵⁹ An extensive amount of experimental work has been done to develop both of these theories, and although neither is considered mature at this time, the slip dissolution model is more fully quantified. An attempt has been made to develop a model for crack growth based on the available data, but its application to an actual cracking incident⁶⁰ indicated that a more refined approach was needed.

6.3 CONCLUSIONS

Thus, it may be concluded that significant progress has been made in quantifying environmental enhancement of crack growth in the years since publication of the present code model. In addition to more-advanced models of cracking, a multitude of additional data has been obtained. These new developments have called attention to the complexity of the actual process and offer the hope of significant improvements in the prediction of environmental cracking. In a recent review of the available theories,⁶¹ Ford found that the crack-growth rates predicted by the slip dissolution model could exceed those of the present code reference

curves,⁹ but these growth rates cannot be sustained and soon fall back under the code predictions. Ford also suggested that a more realistic crack-growth prediction should be based on a time-dependent analysis, as opposed to the cyclic basis of the current code curves. Such an approach clearly has merit but will require more work before revision of the ASME Code reference curves is contemplated.

7. FRACTOGRAPHY OF TESTED SPECIMENS

Fracture surface examination has been carried out on a considerable number of the tested specimens discussed in this report. A detailed list is included as Table 7.1. Many of the specimens so studied have been discussed in earlier reports, but only occasionally have comparisons been made. This section will attempt to provide a more comprehensive comparison of the materials tested and examined to date, including plates, forgings, submerged arc welds using three different flux types, and HAZs in both plate and forging materials.

After a typical crack-growth-rate test is completed, the fracture surface of the specimen is covered with a relatively thin black oxide. To see the fracture surface clearly, this oxide was removed with a special vibratory cleaning in Endox solution. Specimens were examined under the SEM.

The appearance of the fracture surfaces of tested specimens has been found to be generally a stronger reflection of the material microstructure than the crack-tip environment or the degree of environmental enhancement. Several different fracture surface appearances, including void coalescence and striation formation, were observed. The characteristic common to the static-loaded specimens that have cracked and to the more accelerated fatigue crack-growth specimens is a brittle-appearing comment, which results from the tendency of the main crack to propagate on different levels, eventually linking up. The density of these regions can be correlated to the macroscopic roughness of the fracture surface, and this generally increases with sulphur content. This local branching is often associated with the formation of plateaus of growth that are essentially independent of the applied stress-intensity factor and suggests a more aggressive crack-tip environment.

Another feature that has been frequently encountered in specimens tested is the occurrence of bands of light and dark shading on the fracture surface at various crack lengths (e.g., Fig. 7.1). Fractography of these banded regions indicates that these areas correlate to where the mechanism of crack propagation changes. In the weld specimen C-2, it represents a shift from predominantly dimpled rupture to striated growth. In base metal specimen TW-5 the banded region of fracture surface correlated to a shift from highly nonplanar to a more planar crack growth with little noticeable change in the macroscopic growth rate (Fig. 7.2).

7.1 MATERIAL-EFFECTS MATRIX

7.1.1 Base Metal Specimens

Two base metals were included in this study, A 533 grade B class 1 plate and A 508 class 12 forging. Although only limited samples of the

Table 7.1. Summary of corrosion fatigue crack-growth tests for PWR environment 288°C, sine wave

Material	Sulfur (check)	R ratio = 0.2		R ratio = 0.7	
		Frequency (cpm)	Specimen	Frequency (cpm)	Specimen
A 533 B plate (CQ2)	0.025	1	CQ2-5 (TL)	1	CQ2-6 (TL)
		1	CQ2-8 (TL)	10	CQ2-11 (TL)
		1	CQ2-9 (TL)		
A 533 B plate (IN)	0.026	1	IN-1 (TL)	1	IN-2 (TL)
		1	IN-4 (LT)	1	IN-5 (LT)
		1	IN-7 (LS)	1	IN-8 (LS)
A 533 B plate (PN)	0.016	1	PN-1 (TL)	1	PN-2 (TL)
		1	PN-4 (LT)	1	PN-5 (LT)
		1	PN-7 (LS)	1	PN-8 (LS)
A 533 B plate (TW)	0.004	1	TW-1 (TL)	1	TW-2 (TL)
		1	TW-4 (LT)	1	TW-5 (LT)
		1	TW-7 (LS)	1	TW-8 (LS)
A 533 B plate (W7)	0.004	1	W72C9 (TL)	1	W72C6 (TL)
		1	W72C7 (TL)		
A 533 B plate (HSST plate 02)	0.019	1	2D3Q-6 (TL)	1	2D1Q2 (TL)
		1	2D3C-3 (TL)	1	02GB-2 (TL)
		0.5	2D1Q-2 (TL)	0.5	02GB-5 (TL)
		1	2D3Q-5 (TL)		
		0.1	02GB8 (TL)		
A 533 B plate (HSST plate 04) (04A)	0.018	1	02EF5 (TL)		
		1	04A-103 (TL)	1	04A-102 (TL)
		1	04AE5 (TL)	1	04AE15 (TL)
		1	04AE6 (TL)		
		1	04AE7 (TL)		
Miscellaneous A 533 B plate (IHT)	0.013	1	04AE13 (TL)		
		1	IHT20		
EPRI round-robin				1	1021T
B76	0.012			1	B76-226-24
B	0.006			1	B58
A-9404		1	A9404-2-1		
Linde 80 weld HAZ (in A508 C2 forging) (C)	0.008	1	C-3 HAZ	1	C-4 HAZ
Linde 0091 weld HAZ (in A 533 B plate) (D)	0.020	1	D-1 HAZ	1	D-4 HAZ
		1	D-3 HAZ	1	D-2 HAZ
Linde 0091 weld HAZ (in A 533 B plate) (DD)	0.017	1	DD-1 HAZ	1	D-6 HAZ
					DD-2 HAZ
Linde 124 weld HAZ (in A 533 B plate) (KH)	0.007	1	KH-1 HAZ	1	KH-2 HAZ
		10	KH-4 HAZ		
Linde 124 weld HAZ (in A 533 B plate) (CQ)	0.019	1	CQ-1 HAZ	1	CQ-2 HAZ
				5	CQ-3 HAZ
A 508 C2 forging (F)	0.012	0.1	F-5	1	F-1
		1	F-23 (4T)		
		1	F-24		
Linde 80 weld (C)	0.014	1	C-3 WLD	1	C-4 WLD
		1	C-6 WLD	1	C-7 WLD
Linde 124 weld (CG)	0.012	1	CG-2	1	CG-1
		1	CG-3	1	CG-6
Linde 0091 weld (D)	0.008	1	D-1 WLD	1	D-2 WLD
		1	D-3 WLD	1	D-5 WLD
A 508 C2 forging (R)	0.025	1	R-17		
Linde 124 HAZ (G)	0.012	1	G1 HAZ		
A 106 B pipe	0.019	1	A 106 B-L04-31		
A 106 B pipe weld		1	A 106 B-L0-W4

Fig. 7.1. Photographs of fracture surfaces, fatigue crack-growth specimen R-6 and static-load specimen CQ-3HAZ.

ORNL PHOTO 5567-87

Fig. 7.2. Specimen TW-5, showing light and dark bands. Note (a) brittle-appearing texture in white band and (b) striated appearance in dark band, at crack length of 2.54 cm (~1 in.).

forging material were examined (F-9* and F-1), the results indicate that the crack growth at both $R = 0.2$ and $R = 0.7$ proceeds well-developed ductile striations (e.g., Fig. 7.3).

Many more samples of the plate material were studied, and although the predominant propagation mechanism is also ductile fatigue striations, in this material, they are less well defined and are more difficult to discern because of the background roughness produced by oxidation (Fig. 7.4). On these specimens, ductile striations can be clearly seen typically only on about one-half of the fracture surface. Another feature not seen in the forging material is the occasional occurrence of regions exhibiting a cleavage-like appearance at low magnification (Fig. 7.5, region A). These regions have riverlike markings that emanate from a single source and fan out over the region. At higher magnification these regions also have periodic markings (Fig. 7.5, region A); however, they are not as well developed as adjacent regions of ductile striations (Fig. 7.6, region B).

The fracture surface appearance of the plate material at the higher R ratio (0.7) shows a similar pattern. At low magnification it has the general pattern of fatigue [Fig. 7.7(a)] with some areas containing the more brittle-appearing component [Fig. 7.7(b), region A]. At higher magnification the brittle-appearing areas seem to be devoid of any periodic markings (Fig. 7.8). However, the extensive oxidation that occurred has nearly obliterated the striations in the ductile areas (Fig. 7.9), so it is not possible to draw firm conclusions about the fine structure of these brittle areas.

7.1.2 Weld Metal — Linde 124 Flux

The fracture surfaces of all specimens of this weld exhibit a predominance of dimpled rupture, whether tested in air (Fig. 7.10) or in the PWR environment. The fracture surface tested in air is the cleanest and clearly shows the role being played by the large carbide particles in the weld metal, serving as nucleation sites for dimpled rupture. A minor amount of ductile striation formation can also be seen in Fig. 7.10.

The fracture surface features of the specimens tested in water are considerably obscured by oxidation, but at both $R = 0.2$ and $R = 0.7$ the surfaces are very similar to the air specimen with a preponderance of dimpled rupture. This behavior is shown, for example, in Fig. 7.11, where it can be seen that the amount of dimples is enhanced at the higher R ratio.

7.1.3 Weld Metal — Linde 80 Flux

The Linde 80 flux has been primarily used in welding forgings; the Linde 124 flux discussed above, however, is primarily used for welding plates, and the welds tested in this program are typical in this respect.

*F-9 was tested in PWR environment with 1-min ramp, 1-min hold, and $R = 0.2$ and, thus, is not part of the main matrix tests. See Sect. 7.2 for discussion of this specimen.

ORNL PHOTO 5568-87

F-1

1000X

F-1

1000X

Fig. 7.3. Specimen F-1, tested at 1 cpm, $R = 0.7$, showing well-defined ductile striations.

ORNL PHOTO 5569-87

Fig. 7.4. Typical fracture surface morphology of A 533 class 1 plate material tested at 1 cycle/min, $R = 0.2$.

ORNL PHOTO 5570-87

Fig. 7.5. Specimen 1N4 tested at 1 cycle/min, $R = 0.2$. Note occurrence of brittle structure at Point A.

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Fig. 7.6. Higher magnification of Fig. 7.5 showing ductile striations at B and markings in brittle area A.

Fig. 7.7. Specimen 02GB2, tested at 1 cycle/min, $R = 0.7$.
(a) Typical fatigue surface at low magnification, (b) area showing brittle component at A.

ORNL PHOTO 5573-87

Fig. 7.8. Specimen 02GB2 — higher magnification of brittle type area.

ORNL PHOTO 5574-87

Fig. 7.9. Specimen 02GB2 — higher magnification of area showing ductile striations.

ORNL PHOTO 5575-87

2000X

Fig. 7.10. Specimen C-3, portion of sample tested in air after water testing.

ORNL PHOTO 5576-87

500X

2000X

Fig. 7.11. Specimen C-1 tested at $R = 0.7$, 1 cycle/min.

In contrast to the Linde 124 weld metal, the Linde 80 weld metal specimens are characterized more by ductile striation formation than dimpled rupture, a fact that may be related to the relative ease of ductile striation formation in the forging as compared with plates.

At the low R ratio and low-to-intermediate values of applied K, the fracture surface has a rough texture with occasional ductile striations discernible on about one-half of the surface (Fig. 7.12). The density of carbide-nucleated dimples is very low. At higher applied K, the incidence of secondary cracking and dimples increases, along with a decrease in density of well-defined ductile striations (Fig. 7.13).

At a higher R ratio, again the density of carbide-assisted dimples is very low with the dominant fractographic feature being ductile striations (Fig. 7.14). In specimen C-2WLD, tested with a 1-min ramp/reset, and interesting transition in fractography was noted. This specimen had a series of light and dark bands clearly visible on the macrophotograph of the fracture surface (Fig. 7.15). The bands were found to correspond to a transition from entirely dimpled growth to ductile fatigue striations in adjacent areas of the specimen (Fig. 7.16). Attempts to correlate these bands with some change in the operating test conditions were unsuccessful although a similar type of behavior has been observed in isolated other specimens, for example, a forging specimen tested at $R = 0.2$ and discussed elsewhere.⁶²

7.1.4 Weld Metal — Linde 0091 Flux

The Linde 0091 flux has been generally used to weld A 533 grade B plates together. The fracture appearance of these specimens appears to be intermediate between the predominantly dimpled structure of the Linde 124 weld and the ductile striations seen in the Linde 80 weld. The fracture surface at the low R ratios appears to be a mixture of carbide-nucleated dimples interspaced throughout with poorly defined ductile striations (e.g., Fig. 7.17). At a higher R ratio (0.7) the structure has less well-defined striations and is more dominated by dimples (Fig. 7.18).

7.1.5 Heat-Affected Zone Material

HAZ specimens were examined from both a forging (C-3 HAZ) and a plate (D1-HAZ). In both specimens, the results were similar (Figs. 7.19 and 7.20). At low magnification the familiar rumpled markings that characterize fatigue (region A in both figures) were the predominant feature, but there were occasional regions exhibiting a more brittle appearance. Noted previously only in the plate material, these areas have river markings that lie nearly perpendicular to the direction of main crack propagation. At higher magnification these brittle-like areas are generally devoid of any distinctive pattern that could indicate the mechanism responsible for their appearance [Fig. 7.20(b)]. The area surrounding these regions contains ductile fatigue striations [Fig. 7.20(c)].

D6-HAZ, a HAZ sample, was made using Linde 0091 flux and A 533 grade B class 1 plate, which contained relatively high sulphur (0.020). This

Fig. 7.12. Specimen C-3 WLD tested at 1 cycle/min, $R = 0.2$, $\Delta K = 48.8 \text{ MPa}\cdot\sqrt{\text{m}}$ ($44 \text{ ksi}\cdot\sqrt{\text{in.}}$).

ORNL PHOTO 5578-87

Fig. 7.13. Specimen C-3 WLD tested at 1 cycle/min, $R = 0.2$, $K = 60.5 \text{ MPa} \cdot \sqrt{\text{m}}$ ($55 \text{ ksi} \cdot \sqrt{\text{in.}}$).

ORNL PHOTO 5579-87

Fig. 7.14. Specimen C-4 WLD tested at 1 cycle/min, $R = 0.7$.

ORNL PHOTO 5580-87

Fig. 7.15. Specimen C-2 WLD showing banded structure on fracture surface (1-min ramp/reset, $R = 0.7$). Note: The V-shaped water mark does not correspond to the crack front, which is marked by a series of wide light bands and narrow dark bands. Crack-growth-rate results were not reported after the crack became uneven, near the end of the test.

Fig. 7.16. Specimen C-2 WLD showing (a) transition from dimpled to striated growth at arrows, (b) high magnification of dimples, and (c) striated growth.

ORNL PHOTO 5581-87

ORNL PHOTO 5582-87

Fig. 7.17. Specimen D-3 WLD tested at $R = -0.2$, 1 cycle/min.

ORNL PHOTO 5583-87

Fig. 7.18. Specimen D-2 WLD tested at $R = 0.7$, 1 cycle/min.

ORNL PHOTO 5584-87

Fig. 7.19. Specimen D1-HAZ showing brittle-appearing areas, region B, contained in a more typical fatigue surface, region A.

Fig. 7.20. Specimen C-3 HAZ. (a) Low magnification indicating typical fatigue features, region A, and a more brittle component, region B; (b) higher magnification of region B; (c) higher magnification of region A.

specimen was tested at $R = 0.7$ and 0.017 Hz. In the macrophotography of this specimen, a rough-appearing region was present in the center of the specimen (Fig. 7.21). Higher-magnification examination of this region indicated an appearance similar to that primarily seen in the static K_{ISCC} testing. The region contains numerous "clam-shell"-like areas formed by fracture initiating at numerous origins and then fanning out over the surface (seen more clearly in Fig. 7.22). In regions removed from this very rough area, there are still regions containing a brittle-appearing component (Fig. 7.23). These brittle areas are now much smaller and elongated parallel to the direction of primary macroscopic crack growth. The fibrous ridges contained within these areas [Fig. 7.23(b)], which indicate the growth direction, are now oriented perpendicular to the main direction of crack propagation. This would indicate that these areas may form when adjacent regions of the primary crack become split and there is a need for propagation perpendicular to the main growth direction to connect them up. The other component of the fracture surface in this area that surrounds the brittle areas is ductile fatigue striations [Fig. 7.23(c)].

The fractography of this specimen is similar to that of D1-HAZ discussed above, tested at $R = 0.2$, with the exception that at the lower R ratio the rough region (Fig. 7.22) was not present. Similar results were also obtained for the HAZ made with the Linde 80 flux and A 508 class 1 steel at low R ratios (e.g., specimen C-3 HAZ¹). Note, however, that in a similar specimen (C-1 HAZ) tested with a 1-min ramp that did not experience accelerated growth (Sect. 3.1), the fractography does not contain any of the brittle-appearing regions (Fig. 7.24). This will be considered further in Sect. 6.1.6.

7.1.6 Discussion

The fractographic results further reflect the conclusion reached in Sect. 3.1, "Material-Effects Matrix," that there are significant differences in the crack-growth behavior from material to material. Unlike the rather clearly defined trend in macroscopic crack-growth results that appears related to sulfur content, however, the fracture morphology appears to be more related to the basic microstructure of the material.

Studies to date have found no evidence in the fractography to suggest that this sulfur effect is directly related to mechanical mechanisms, such as accelerated growth caused by tearing at sulfide inclusions. In no case was evidence for such effects seen although sulfides had been previously noted by Wilson⁷ to influence the fatigue crack growth of A 533 grade B class 1 steel in air.

From the limited number of samples the forging material examined, it does appear that the ferrite present in the microstructure of this material helps to promote ductile fatigue striations (Fig. 7.25). The fractography of specimen F-1 is one of the best examples of well-defined ductile fatigue striations observed. Contrasted to this, the striations in the plate material are much less pronounced, reflecting the greater difficulty of local plastic flow in the entirely bainitic microstructure.

The fractography of the sample having the lowest sulfur and the least enhancement in crack-growth behavior at $R = 0.2$, D-1WLD, gives

ORNL PHOTO 5586-87

1.5X

Fig. 7.21. Macrograph of specimen D6-HAZ.

ORNL PHOTO 5587-87

8.5X

Fig. 7.22. Enlargement of rough fracture surface at crack length of 0.6 in. from machined notch in specimen D6-HAZ.

GROWTH DIRECTION

(b) REGION A

(c) REGION B

Fig. 7.23. Specimen D6-HAZ (a) 1 in. from machined notch, (b) enlargement of region A, and (c) enlargement of region B.

Fig. 7.24. Specimen C-1 HAZ containing no evidence of brittle-like areas seen in D6-HAZ, Fig. 5.3.

ORNL PHOTO 5590-87

Fig. 7.25. Comparative microstructure of plate and forging material.

little indication of the reason for this unusual behavior. The fracture surface (Fig. 7.16) is intermediate between the heavily dimpled structure of the Linde 124 weld and the striation-dominated appearance of the Linde 80 welds. The large amount of dimpling and microcracks present in the Linde 124 weld metal is probably related to the density and size of the large, globular carbide particles present. The microstructure of the Linde 124 weld contains a much higher density of these particles than is present in the plate material (Fig. 7.26).

Note that at higher R ratio (0.7) where the Linde 0091 weld does show enhanced crack growth (Fig. 7.10), the fractography reverts from the mixed striated-dimpled structure to one of almost entirely dimpled rupture, as seen in specimen C-1 (Fig. 7.11) although the size of the dimples is considerably smaller (Fig. 7.17).

One of the most puzzling aspects of the work is the variation in fractography of weld specimens from dimples to striations with the apparent shift from one mode to another within adjacent regions of the same sample (Figs. 7.15 and 7.16). It is clear from the macrophotograph that the transition is more probably related to the local variation in test conditions than to a microstructural feature of the sample because the changes track the position of the crack front.

7.2 RAMP AND HOLD-TIME STUDIES

The ramp and hold-time studies were conducted primarily on forging materials, as detailed in Sect. 3.2. Because observations of the fracture appearance of the A 508 materials were made in the previous section, only the differing features that may have resulted from this loading form will be discussed here. Also here, a series of striation measurements will be highlighted. These were conducted to ensure the measured macroscopic growth rates were accurate because of the wide variations in observed crack growth.

For specimen R-6 the striation measurements follow the specimen behavior even through a sizeable arrest event (Fig. 7.27). The striation measurements for specimen R-11 are also in agreement with the macroscopic data at the beginning and at the end of the test (Fig. 7.28) but do not reflect the apparent arrest that occurred near the middle of the test. This "arrest" may have been caused by the fact that the crack front was very uneven during the middle of the test (Fig. 7.29). Certainly, the striation measurements on both specimens confirm that the growth rates were indeed higher than the upper band of the sinusoidal data although the high growth rate quickly subsided.

Specimen R-11 displayed a character in its fracture surface that had not been observed before, an alternating light and dark banded appearance, which was visible to the naked eye. This is shown in Figs. 7.29 and 7.30 in which magnifications of both the light and dark bands are provided. The reason for this banded behavior is currently unknown, but striations are clearly visible in both regions and give comparable growth-rate measurements. The only real difference between the two types of regions is that the dark regions appear to have a rougher texture with more crack branching (Fig. 7.30).

ORNL PHOTO 5565-87

Fig. 7.26. Extraction replica of carbides in (a) base metal specimen 02GB-3, and (b) weld metal specimen C-2 showing larger spherical particles in weld metal.

ORNL-DWG 79-6117A ETD

Fig. 7.27. Fatigue crack growth and striation spacing measurements for specimen R-6.

Fig. 7.28. Fatigue crack-growth rate and striation spacing measurements for specimen R-11.

Fig. 7.29. Fracture surface of specimen R-11.

Fig. 7.30. Closeup of banded regions of specimen R-11.

A number of other specimens were studied fractographically to confirm the macroscopic growth rates, and the resulting measurements are shown in Table 7.2. The striation measurements matched the macroscopic growth in all cases for tests conducted at 288°C (550°F) as shown further for specimen F-9 in Fig. 7.31. Tests conducted at 93°C (200°F) produced striations that did not match the macroscopic growth, however, as shown in Fig. 7.32 for specimen F-8. This is particularly interesting because the macroscopic growth rates of the specimens tested at the two temperatures are comparable. This indicates the presence of a corrosion component that adds to the mechanical growth in the lower-temperature specimen.

The fracture surfaces of specimens tested at the two temperatures are considerably different, however, and this at least partially explains the inconsistencies between macroscopic and striation-measured growth rates. The fracture surfaces of specimens F-9 and F-8 are shown in Figs. 7.33 and 7.34. The fracture surface of the high-temperature specimen shows entirely striated growth; that of the low-temperature specimen displays a mixture of cleavage and ductile-striated growth. This is shown very clearly in the bottom picture of Fig. 7.34.

7.3 SULFUR AND CRACK ORIENTATION EFFECTS

The macrophotographs of the high-sulfur (IN1) and the low-sulfur (TW1) specimens in the TL orientation are given in Fig. 7.35. It can be seen from this figure that IN1 has a slightly rougher appearance than TW1, particularly for longer crack lengths, even though they were tested under identical loadings. This rougher appearance in IN1 results from the brittle-appearing areas that are elongated parallel to the macroscopic crack-growth direction (Fig. 7.36). Within these brittle areas, the pattern of ridges indicates that the growth occurs perpendicular to the macroscopic growth direction. These brittle areas apparently form to link up adjacent regions of the main crack that have split and begun to propagate at different elevations. This is consistent with both the direction of propagation within the brittle region — perpendicular to the main growth direction — and the rough appearance of the fracture surface. In specimen TW1, which has a much smoother appearing fracture surface, all of the crack propagation occurs by ductile fatigue striations (Fig. 7.37).

In the LT orientation a similar trend in sulfur content and fracture surface appearance is found. Both specimens of the intermediate-sulfur level (PN4) and the high-sulfur level IN4 have a rough appearance in the macrophotographs (Fig. 7.38) with the higher-sulfur specimen having a somewhat rougher appearance. This rough appearance was again correlated with the appearance of brittle areas in the fracture surfaces (Figs. 7.39 and 7.40).

In the LS orientation specimens of all three sulfur levels were examined, again confirming the trend of rougher-appearing fracture surfaces as the sulfur level is increased (Fig. 7.41). Specimen TW7 contained only ductile fatigue striations; specimen IN7, however, contained a mixture of brittle areas in addition to ductile striations, as observed

Table 7.2. Crack-growth-rate determinations
from striation spacing measurements of
ramp and hold-time specimens

Applied ΔK (ksi $\sqrt{\text{in.}}$)	Striation spacing (in./cycle)	Macroscopic data (in./cycle)
<i>Specimen F-2</i>		
51.1	3.0×10^{-5}	6.7×10^{-5}
51.3	4.0×10^{-5}	
51.9	5.0×10^{-5}	
<i>Specimen F-9</i>		
25.8	9.0×10^{-5}	9.7×10^{-5}
25.8	1.0×10^{-4}	
30.0	1.1×10^{-4}	
<i>Specimen R-6</i>		
35	2.7×10^{-4}	1.44×10^{-4}
35	3.8×10^{-4}	
40	4.4×10^{-4}	4.0×10^{-4}
40.6	6.0×10^{-4}	
43.4	6.0×10^{-4}	6.95×10^{-5}
57.7	6.7×10^{-5}	
71.0	9.0×10^{-5}	
<i>Specimen R-11</i>		
26	1.1×10^{-4}	7.7×10^{-5}
26	1.6×10^{-4}	
30	4.3×10^{-4}	1.1×10^{-4}
35	4.0×10^{-4}	
44	2.9×10^{-4}	5.9×10^{-4}
44	3.4×10^{-4}	
50	4.8×10^{-4}	1.23×10^{-4}
50	7.3×10^{-4}	
77	5.6×10^{-4}	1.24×10^{-3}
77	6.0×10^{-4}	
<i>Specimen F-8 (93°C)</i>		
47	3.0×10^{-5}	1.62×10^{-4}
47	4.0×10^{-5}	
47	5.0×10^{-5}

Fig. 7.31. Striation spacing measurements for specimen F-9.

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Fig. 7.32. Striation spacing measurements for specimen F-8.

ORNL PHOTO 5592-87

500X

2000X

Fig. 7.33. Fracture surfaces of specimen F-9: 1-min ramp, hold time, $R = 0.2$, temperature = 288°C .

ORNL PHOTO 5593-87

500X

2000X

Fig. 7.34. Fracture surfaces of specimen F-8: 1-min ramp, 1-min hold, $R = 0.2$, temperature = 93°C .

Fig. 7.35. Macrophotographs of specimens IN1 and TW1.

Fig. 7.36. (a) Specimen IN1 showing brittle-appearing areas elongated parallel to growth direction, and (b) higher magnification of nonbrittle areas in (a).

ORNL PHOTO 5596-87

Fig. 7.37. Fracture surface of specimen TW1 containing ductile fatigue striations.

ORNL PHOTO 5597-87

(a) PN4

1.5X

(b) IN4

1.5X

Fig. 7.38. Macrophotographs of specimens (a) PN4 and (b) IN4.

Fig. 7.39. (a) Specimen IN4 tested at 1 cycle/min, $R = 0.2$; note occurrence of brittle structure at Point A; (b) higher magnification of Fig. 5.11(a), showing ductile striations at B and marking in brittle area A.

ORNL PHOTO 5598-87

Fig. 7.40. Typical regions of fracture surface of specimen PN4 containing brittle areas elongated parallel to growth direction.

Fig. 7.41. Macrophotographs of specimens (a) TW7, (b) PN7, and (c) IN7.

for the other orientations. In specimen IN7 occasional indication of interaction with inclusions oriented perpendicular to the growth direction was seen (Fig. 7.42). This result is consistent with the expected elongation of inclusions parallel to the surface of the plate (Fig. 3.1).

At the higher R ratio ($R = 0.7$) the high-sulfur specimen IN2 still has a rougher appearance (Fig. 7.43), and the lower-sulfur specimen TW2 is somewhat rougher than for the lower R ratio (Fig. 7.43). At higher magnification the fracture surface of TW2 has the appearance seen in Fig. 7.44. Although no brittle areas elongated parallel to the growth direction can be seen, the striated surface has a much more brittle appearance than at the low R ratio (Fig. 7.45). In specimen TW2 there are distinct tear ridges that tend to fan out over the surface and separate adjacent areas of striated growth.

Specimen IN2 contains a mixture of brittle and striated areas, similar to the low R ratio specimens (Fig. 7.45). In this specimen, some indications of massive inclusions were found (Fig. 7.46), elongated parallel to the growth direction, which is consistent with that expected from Fig. 3.21 for the TL orientation.

To determine the distribution of sulfur inclusions in each specimen, a sulfur print was made of each fracture by polishing a section parallel to the fracture surface $\sim 1/4$ in. below that surface. These sulfur prints correspond exactly to the region indicated by the macrophotograph of each specimen.

In addition to providing a distribution of sulfides in the plane of crack propagation, these sulfur prints show the change in sulfur levels through the plate. The specimens were taken in the orientations shown in Fig. 3.1, and all three heats (TW, PN, and IN) were numbered consistently. For each of the TL and LT orientations, three specimens were taken through the thickness. For each heat, specimen one was at the top, two in the middle, and three at the bottom, and similarly for the LT orientation, specimen four was at the top, five in the middle, and six at the bottom. Some material at the very top and bottom of the plate was discarded, so each of the specimens on the top and bottom rows had its center at approximately the quarter-thickness location of the plate. All of the W7 series of specimens were also taken from approximately the quarter-thickness location.

For the high-sulfur specimens (Fig. 7.47) the only specimen that had other than a uniform distribution was IN2, which was taken from the center of the plate. It is apparent that the ingot segregation is still present in the center of the plate while the surfaces are much more uniform. This explains the presence of the large stringer seen in the fracture surface of IN2 (Fig. 7.46).

Even in the low-sulfur specimens, there is evidence of sulfur stringers in the center of the plate in the TL orientation, specimen TW2 (Fig. 7.48). Other specimens examined, W7-2C7 and PN4, have a very uniform sulfur print (Fig. 7.49).

One low-sulfur specimen, TW10, tested in a mixture of water and H_2S was also examined fractographically. Unlike a specimen of the same orientation, TW7, tested in PWR water (Fig. 7.47), which contained ductile striations, TW10 was entirely brittle-type crack propagation (Fig. 7.50). Unlike the other specimens in this section, the brittle areas are made up of small regions, and the tear ridges within each

ORNL PHOTO 5601-87

Fig. 7.42. Inclusions oriented perpendicular to growth direction in specimen IN7 (at crack depth = 0.7 in.).

ORNL PHOTO 5602-87

Fig. 7.43. Macrographs of specimen (a) TW2 and (b) IN2.

Fig. 7.44. Typical fracture surface appearances of specimens TW2 (1 cycle/min, $R = 0.7$).

Fig. 7.45. Specimen IN2 containing (a) brittle appearing areas surrounded by (b) region containing ductile striations.

ORNL PHOTO 5605-87

Fig. 7.46. Appearance of inclusion stringer elongated parallel to growth direction in specimen IN2.

ORNL PHOTO 5696-87

(a) IN4

(b) IN1

(c) IN2

GROWTH
DIRECTION

Fig. 7.47. Sulfur prints of section cut parallel to fracture surface, $\sim 1/4$ in. below surface; specimens IN1, IN2, and IN4.

Fig. 7.48. Sulfur prints of specimens (a) TW1, (b) TW5, and (c) TW2.

ORNL PHOTO 5608-87

(a) PN4

(b) W7-2C7

Fig. 7.49. Sulfur prints of specimens (a) PN4 and (b) W7-2C7. Both specimens were taken with midplane at quarter-thickness location in plate.

ORNL PHOTO 5609-87

Fig. 7.50. Specimen TW10 tested in $H_2O + H_2S$ at $93^\circ C$ [crack length 2.79 cm (1.1 in.)].

region run, in general, parallel to the main direction of crack propagation. This type of crack growth results in a very smooth macroscopic surface appearance. Similar results have been reported previously for room-temperature H_2O tests and gaseous H_2S tests.⁸

7.4 HIGH-SULFUR MATERIAL FRACTOGRAPHY

The specimens CQ2-9 and CQ2-12 were tested in borated water at 288°C (550°F) and 204°C (400°F), respectively (Table 3.4). Both fractures were relatively smooth and exhibited ductile fatigue striations with side cracks on most areas of the fracture (Figs. 7.51). Where the growing crack apparently intersected an inclusion [Fig. 7.51(b)], there was no effect on the fractography other than the local discontinuity. The different test temperatures had no effect on the general fractography of the material.

Specimens CQ2-6 and CQ2-11 were tested in water at 288°C (550°F) at 1 and 10 cpm, respectively. Specimen CQ2-7 was tested at 1 cpm in the same environment but at 204°C (400°F). All three specimens exhibited a flat fracture macroscopically. Ductile striations were readily observed on specimen CQ2-6. Inclusions had no effect on the fractography. Specimens CQ2-11 and CQ2-7 had a rougher texture, and no fatigue striations have yet been observed.

Specimens CQ2-7 and CQ2-12, both tested at 204°C (400°F), had a more adherent oxide that has proven more difficult to remove. This has reduced the resolution of detail on the fracture faces. A series of sulfur prints was made on these specimens, including a full-thickness sulfur print of the plate, and this is discussed in Sect. 2.4 and shown in Figs. 3.31 and 3.32.

7.5 LOW-SULFUR MATERIAL FRACTOGRAPHY

Specimens TWI and W72C9 both exhibited higher crack-growth rates than specimen W72C7, especially at ΔK levels approaching 20 MPa $\sqrt{m}$ (Fig. 3.39). No significant differences have yet been found among the fractographic details observed on these three specimens. All three exhibit striated fatigue crack growth with side cracks and tear ridges separating the striated regions (Fig. 7.52).

7.6 STATIC LOAD AND COMPLEMENTING FATIGUE SPECIMEN FRACTOGRAPHY

The specimens discussed in this section were from the HAZ of two weldments. The fatigue specimens (Table 4.5) exhibited similar fractographic details to those displayed by the base metal specimens described previously. Specimens CQ2-HAZ and DD-1HAZ, both tested at 1 cpm at $R = 0.2$, exhibited flat fractures with ductile striations and side cracks (Fig. 7.53). No striations have been resolved on specimen DD-2HAZ,

ORNL PHOTO 5787-84

Fig. 7.51. SEM fractographs of specimen CQ2-12. (a) 10 mm from root of notch (1600 $\times$), (b) 20 mm from root of notch (800 $\times$).

ORNL PHOTO 5788-84

Fig. 7.52. SEM fractographs of specimen W72-C9. (a) 400 $\times$, (b) 1600 $\times$. Similar fractography was observed on W72-C7 and TW1.

ORNL PHOTO 5789-84

Fig. 7.53. SEM fractographs of (a) CQ-2-HAZ at 20 mm and (b) DD-1 HAZ at 10 mm from root of notch. 800×

tested at 1 cpm at $R = 0.7$. Side branching [(Fig. 7.54(a))], along with some featureless areas [Fig. 7.54(b)], is seen. Some features are similar to dimples [Fig. 7.54(c)] but also could be corrosion pits, possibly formed after the test. The fracture on this specimen started as a flat smooth surface and then moved out of plane and became rough.

The bolt-loaded specimens were significantly different. Macroscopically, the cracks were rough with cracks "fanning out" on different planes and linking up. It has proven difficult to remove the oxide easily from specimen CQ1-HAZ, and any fine detail is not visible below the magnetite crystals. The cracking is relatively featureless at the start [Fig. 7.55(a)], but river marks similar to those seen on cleavage cracks are visible at greater depth [Fig. 7.55(b) and (c)]. An attempt to remove the remaining oxide without causing significant degradation of the underlying fracture was made.

Specimen DD-4HAZ, tested under the same conditions as CQ2-HAZ, exhibited significantly different fractography. Figure 7.56(a) shows the end of the fatigue precrack and the branching at the start of the crack whose fracture morphology is not obvious at this point. However, within 5 mm the cracking is obviously intergranular [Fig. 7.56(b)]. The duplex grain size of this specimen is readily apparent in Fig. 7.57. In Fig. 7.57(b) some small regions of transgranular cracking can be seen. The transgranularity becomes more evident at longer crack lengths; Figs. 7.58 and 7.59(a) can be compared with the river pattern, cleavage-like appearance seen on specimen CQ2-HAZ. However, a significant amount of intergranular cracking is still present [Fig. 7.59(b)].

7.7 INCLUSION MEASUREMENTS

It has been theorized for some time now that the level of environmental acceleration in pressure-vessel steels and associated welds is a function of the sulfur content of the steel. Recently, it has become evident that the level of environmental acceleration is not directly related to the bulk sulfur content but may be more closely correlated with both the size and distribution of the manganese sulfide inclusions. To provide a quantitative measure of the inclusions present in these steels, a detailed study of the inclusions was made.

The inclusion content of selected specimens was determined using ASTM E45-76, Method D. The determination was made on surfaces ~ 0.38 cm (0.15 in.) from, and parallel to, the fracture surfaces of the specimens. The first three steels listed in Table 7.3 are all low sulfur (0.004%). Heat L4 is a modern, fine-grained steel; W7 is a relatively modern steel; and TW is an older steel. The three steels are clean with low inclusion contents. Interestingly, the more modern steels exhibit slightly higher inclusion ratings on the counts listed in Table 7.3, but the older steels have areas where larger inclusions are observed (e.g., Fig. 7.60). The finely dispersed, more homogeneous nature of the inclusions is typical of the newer steels. Sulfur prints from the same surfaces also demonstrate the same finer dispersion in the modern steels (Fig. 7.61). In these figures, it can readily be seen that the sulfides

ORNL PHOTO 5790-84

Fig. 7.54. SEM fractographs of specimen DD-2HAZ at 15 mm from root of notch. (a) 400 $\times$, (b) 40 $\times$, and (c) 400 $\times$.

ORNL PHOTO 5791-84

Fig. 7.55. SEM fractographs of bolt-loaded specimen CQ-1-HAZ at (a) 20 mm (160 $\times$), (b) 25 mm (80 $\times$), and (c) 25 mm (80 $\times$) from root of notch.

ORNL PHOTO 5792-84

Fig. 7.56. SEM fractographs of specimen DD-4HAZ. (a) End of pre-crack and start of SCC at 5 mm below root of notch 160, (b) SCC at 10 mm below root of notch. 80×

ORNL PHOTO 5793-84

Fig. 7.57. SEM fractographs of specimen DD-4HAZ at 10 mm below root of notch. 80×

ORNL PHOTO 5794-84

Fig. 7.58. SEM fractographs of specimen DD-4HAZ at 19 mm below root of notch. 160×

ORNL PHOTO 5795-84

Fig. 7.59. SEM fractographs of specimen DD-4HAZ at 20 mm below root of notch. 160×

Table 7.3 Inclusion ratings of plates
and forgings

Material	Inclusion rating No.	Type			
		A	B	C	D
L4 (Plate, S = 0.004)	0.5 thin	13	8	3	19
	heavy				2
	1.0 thin	6	11		
	heavy				
	1.5 thin				1
	heavy			1	
	2.0 thin				
W7 (Plate, S = 0.004)	heavy				
	1.0 thin	3	9	1	10
	heavy	4	2	1	1
	1.5 thin	1	6		13
	heavy		2	5	1
	2.0 thin		2	1	11
	heavy		2	1	
TW (Plate, S = 0.004)	2.5 thin		2		
	heavy		2		
	0.5 thin		2		
	heavy			1	
	1.0 thin	2	2		3
	heavy				
	1.5 thin		2	1	3
CG (Plate, S = 0.019)	heavy			1	
	2.0 thin		1		1
	heavy				
	0.5 thin	11	3	13	8
	heavy	8	5	17	35
	1.0 thin	1	5	5	22
	heavy		13	13	15
D (Plate, S = 0.020)	1.5 thin		3		3
	heavy		3	3	3
	2.0 thin				
	heavy				
	2.5 thin				
	heavy				
	0.5 thin	2	2	1	4
R (Forging, S = 0.025)	heavy		4	2	
	1.0 thin	1	3		
	heavy		3	23	
	1.5 thin		5		
	heavy		3	68	
	2.0 thin				
	heavy				
	2.5 thin				
	heavy				

ORNL PHOTO 5648-86

(a) TW5

100X

(b) W72C9

100X

Fig. 7.60. Photomicrographs of polished sections of A 533 grade B class 1 plate, showing different sizes of manganese sulfide inclusions.

ORNL PHOTO 5649-86

(a) TW

(b) W72C

(c) L4

Fig. 7.61. Sulfur prints for A 533 class 1 plate material, S = 0.004 wt %, different steelmaking practices.

in the older (TW) steel are coarser. If the sulfur content and distribution in the steels significantly affect the environmental response, then variations in crack-growth rate could be expected as the crack grows through the regions of sulfide segregation.

The high-sulfur (0.025) forging (R) exhibits a higher inclusion count, and although the sulfur content is six times that of the previously described plate steels, the sulfur print for this material displays a dispersed, relatively homogeneous sulfide dispersion (Fig. 7.62). The high-sulfur plates [0.019 (CG) and 0.020 (D)] exhibit higher inclusion ratings, particularly of the classical sulfide type, and there are more of the heavier-type inclusions. Although the weldments have sulfur contents of 0.014 (C) and 0.008 (D), the inclusion sizes were below the scale, and the sulfur prints did not reveal any sulfides.

These findings are, in general, consistent with fatigue crack-growth-rate data obtained for these steels. There is clearly a different relationship between bulk sulfur content and environmental acceleration in the different product forms. Very high sulfur levels produce enhanced

ORNL PHOTO 5650-86

Fig. 7.62. Sulfur prints for high-sulfur forging (R) and high-sulfur weld (CG). Note that weld specimen also shows print for adjacent plate ($S = 0.019$).

crack growth in plates, forgings, and welds (Figs. 3.12, 3.21, and 3.44). For plates, as discussed earlier and shown in Fig. 3.21, even a very low sulfur plate results in enhanced growth. On the other hand, Fig. 3.12 shows that a forging with relatively high sulfur content (heat F, S = 0.012) can show very little environmental enhancement. Figure 3.44 shows a similar result for a relatively high sulfur weld (weld D, S = 0.008). Therefore, the threshold for environmental enhancement is different in plates, forgings, and welds.

8. INTERLABORATORY COMPARISONS AND THE ICCGR GROUP

An important element of this program has long been to exchange information with other laboratories doing the same or similar work. This task was made much more convenient and disciplined with the founding of the ICCGR review group. This section of the report will review some of the early exchanges, discuss the work of ICCGR, and review some of its accomplishments since its inception.

8.1 EARLY SPECIMEN EXCHANGES

The first specimen exchange accomplished within this program was with NRL as part of cooperative work on the ramp and hold-time test matrix described in Sect. 3.2. Specimens from heats "F" and "R" were given and were then renumbered "FW," "R1," and "R2" when tested. Some of these results are presented in Sect. 3, Figs. 3.17 and 3.18, and show very good agreement between the two labs.

To provide further comparison between test results obtained at Westinghouse and those obtained at NRL, a series of specimens was exchanged in 1981. Results are now available on heat W7, of which three specimens were sent to Westinghouse from NRL. This heat was a low-sulfur plate (0.004 wt %), and the specimens were oriented in the TL direction.

This plate has been tested, and the results have been reported in Sect. 3.5 for specimens at both $R = 0.2$ and 0.7 . For both R ratios, the agreement is very good with the specimens showing very little environmental enhancement. One exception was specimen W72C9, which showed much more environmental enhancement than the others. No explanation has been found for this; thus, because of the other specimens, it is probably invalid. The frequency in all cases was 1 cycle/min, 0.017 Hz. This, as well as a number of earlier comparisons, has confirmed the consistency of results produced in directly comparable tests.

8.2 THE ICCGR GROUP

ICCGR is an interdisciplinary group set up to share information on corrosion fatigue. The group was originally formed in 1977 at the initiative of the Nuclear Regulatory Commission and the Electric Power Research Institute (EPRI), the two major U.S. sponsors of research in corrosion fatigue of pressure-vessel steels in LWR environments. Both realized that a systematic study of the variables affecting corrosion fatigue and their possible synergistic interactions would be prohibitively expensive and time consuming if undertaken by a single organization.

This initiative received enthusiastic support in western Europe and Japan and resulted in the formation of the ICCGR Group, which was formally chartered in November 1978. A review of the group's function and achievements in its first 4 years of existence has been given by Slama and Jones,⁶³ and a followup was given in 1984 by Jones.⁶⁴

The general objective of the ICCGR Group is to coordinate the research and development efforts on environment-assisted cyclic crack growth in pressure boundary materials in light-water service environments undertaken in western Europe, North America, and Japan. Specific objectives are to

1. provide a forum for the timely interchange of information, ideas, and test data among research workers;
2. plan and perform cooperative research programs;
3. develop consensus test procedures and data-reduction methods and encourage the transfer of such procedures and methods to the appropriate "standards" activities;
4. develop well-qualified crack-growth-rate data bases and improved understanding of the rate-determining factors; and
5. make recommendations on the application of research results and laboratory test data to engineering practice and encourage the transfer of this information by organizing symposia, interacting with code committees and regulatory bodies, etc.

The focus of the activities of the Group to date has been on the pressure-vessel steels A 533 grade B class 1, A 508-C12, and A 508-C13, but work on the other nuclear pressure boundary materials is also reported and discussed during the semiannual meetings, which alternate between the United States and western Europe. Currently, over 40 organizations from 12 countries participate in the activities of the ICCGR Group. Members include experts in mechanics, corrosion chemistry, and metallurgy. As required, additional experts are invited to the meetings.

Task groups have been formed to concentrate on specific topics. Four such task groups have existed up to the present: Test Methods and Results, Mechanisms, Data Collection and Evaluation, and Applications.

Participation in ICCGR meetings was one of the key elements of this work because it is the best forum presently available for exchanging information on current research in the areas of corrosion and corrosion fatigue in LWR materials. Therefore, the status of the groups activities was reviewed periodically in the progress reports.

ICCGR has been very helpful to this program in three key areas. First, a number of round-robins have been performed by the group, and these will be reviewed briefly. Second, the task groups have been a valuable forum for the exchange of information and for guiding the program. Third, ICCGR has made specimen exchanges among laboratories much more convenient.

8.3 ROUND-ROBIN TESTS

A number of round-robin tests have been carried out for both corrosion fatigue and constant extension rate tests and have been reviewed in detail recently.⁶⁵ The two most interesting were the low R ratio and high R ratio round-robin tests.

8.3.1 Test Results and Analyses

Detailed accounts of the outcome of the initial group of tests have been presented elsewhere.^{64,65} In summary, the first test results showed unacceptable scatter, and an extensive testing and analysis program was undertaken to isolate the important interlaboratory variables. Tests at a frequency of 1 Hz showed that differences in the methods used to measure load and crack length were not important contributors to scatter and confirmed that data reduction was not a significant problem. By May 1981 a number of other possible sources of interlaboratory scatter had been identified, including

1. waveshape (some participants used a triangular waveform instead of the sinusoidal waveform specified),
2. control mode (some participants used K control instead of load control),
3. dissolved oxygen content of the water (which ranged from near zero to well above 100 ppb),
4. the measures used to ensure water purity (which were widely different in different test systems),
5. hydrodynamic conditions (test conditions ranged from no flow to highly turbulent flow), and
6. galvanic effects (a few participants electrically isolated the specimen from the test system, but in most cases, the specimen was electrically connected to a variety of other materials).

Detailed analysis of the test results⁶⁴ indicated that acceptable interlaboratory agreement could be obtained if the compilation was limited to those tests in which the aspects of the test conditions listed above were similar (Fig. 8.1). Additional tests performed since 1981 have confirmed this conclusion: at least 20 ICCGR Group members have now reported low-R test results falling within the scatter band of Fig. 8.1, which is felt to be acceptable.

Since 1981 further work has also been conducted on several of the potential sources of interlaboratory scatter identified previously.

1. Tests in two laboratories have confirmed the existence of an effect of waveform. Under the low-R round-robin test conditions, a sinusoidal loading waveshape leads to more-rapid crack growth than a triangular waveshape.
2. Duplicate tests in two laboratories suggest that there is no important effect of control mode. Constant ΔK and constant load range tests both give results falling within the Fig. 8.1 scatter band.
3. Tests in several laboratories have shown that the dissolved oxygen content of the water has a large effect on the corrosion potential of low-alloy steel and can significantly affect the behavior in cyclic crack-growth tests. Impurities in the water, particularly sulfur species, have also been shown to affect the crack-growth rate.

Fig. 8.1. Compilation of low-R round-robin test results from tests conducted under similar conditions. More than 20 ICCGR members have now reported low-R round-robin test results that fall within the scatter band shown here.

4. Conflicting results have been obtained on the importance of fluid flow. Turbulent conditions usually appear to lead to lower crack-growth rates than stagnant or laminar conditions, but the details of the hydrodynamic situation seem to be important. This variable is felt to require additional study.
5. Galvanic coupling to the load train and autoclave can affect the crack-growth rate, particularly if there is a large difference in corrosion potential between the specimen and any of the system components electrically connected to it. (This effect also depends on the conductivity of the water.)

As well as additional testing, the Group undertook an extensive fractographic study of the specimens from the low-R test program. The results are reported in detail in Ref. 66. One of the interesting observations made in that study has been illustrated in Sect. 7: the brittle-appearing fracture surface features that are often seen in association with sulfide inclusions become more prominent as the environmental enhancement of the crack-growth rate increases. This observation is felt by some Group members to suggest that an embrittlement process, perhaps hydrogen embrittlement, is predominantly responsible for the environmental enhancement of the cyclic crack-growth rate observed in low-frequency tests. However, attempts by high-temperature tests in hydrogen gas to demonstrate a direct hydrogen effect have not yet been successful.

The low-R round-robin test program is now complete, but tests under the conditions specified for this round-robin are still conducted occasionally by ICCGR Group members as a qualification test when they are commissioning new test rigs.

8.3.2 High-R Test Program

Based on the satisfactory outcome of the low-R testing round-robin, the Group formulated a plan for a collaborative program of cyclic crack-growth tests at an R ratio of 0.7. This program, which is currently approaching completion, is being coordinated by EPRI, which provided the test material, a heat of A 533 grade B class 1 steel with an intermediate sulfur content of 0.013%. Table 3.1 provides details about this material "T." Fifty-millimeter-thick (2-in.) specimens were machined from a slab cut from the center of the plate. These specimens were distributed to the 12 participating laboratories conducting tests at 288°C in 1 of 3 environments: BWR water, PWR water, and dry air. The water chemistries and the test procedures were tightly specified (Table 8.1) in an attempt to minimize interlaboratory scatter, and measurements on corrosion coupons are being used to provide an integrated indication of the chemical/electrochemical conditions in the autoclaves during the tests. The intent of these high-R tests is to provide a body of well-pedigreed data under low-frequency, low ΔK , high R conditions to supplement the rather sparse information currently available in this particularly important regime.

The result of a high-R test in air at 288°C at a frequency of 0.1 Hz was predictable. The data fell along a line parallel to the ASME XI subsurface flaw reference curve at about a factor-of-two-faster crack-growth rate. This result is similar to data reported by a number of recent studies of crack growth in air at low levels of ΔK and high values of R. (A compilation of data for tests in air can be found in Refs. 67 and 68.)

Figures 8.2 and 8.3 present the results of the high-R tests performed to date in PWR and BWR environments. Considerable scatter is apparent, in part because not all participants followed the test specification precisely, particularly with respect to the value of ΔK at which the tests were started. Other differences in the test conditions probably are also important: note that the lowest rates reported in the

Table 8.1. Specification for ICCGR high-R tests

<i>Mechanical</i>	
Specimen	2TCT (initial a/w = 0.5)
Starting ΔK , MPa $\cdot\sqrt{m}$	11
R ratio	0.7
Frequency, Hz	0.0167
Wave form	Sine
Test type	Constant load range
System pressure, MPa	8.3
System temperature, °C	288
<i>Water chemistry — BWR tests</i>	
Conductivity, $\mu\text{mho/cm}$	<1
pH (room temperature)	7.0 $\pm$ 0.5
Dissolved O ₂ , ppb	200 $\pm$ 50
Cover gas	None (use N ₂ if necessary)
Chloride, ppm	<0.1
Fluoride, ppm	<0.1
<i>Water chemistry — PWR tests</i>	
Conductivity, $\mu\text{mho/cm}$	<20
pH (room temperature)	6.5 + 0.2 (control with LiOH addition)
Dissolved O ₂ , ppb	<5
Boron (as boric acid), ppm	1000 (~6000 boric acid)
Hydrogen, cm ³ /kg H ₂ O	15 to 50 st.
Chloride, ppm	<0.1 ppm
Fluoride, ppm	<0.1 ppm
<i>Startup procedure</i>	
(Report actual procedure on extra page attached to data sheets)	
Demineralize water	
Insert specimen in chamber	
Replace air in autoclave with N ₂ (BWR) or H ₂ (PWR)	
Fill chamber with water	
Remove dissolved oxygen if necessary by bubbling N ₂ (BWR) or H ₂ (PWR)	
Add <1 ppm hydrazine if required	
Heat autoclave to test temperature as quickly as possible (record time taken on data sheet)	
Start cycling as soon as test temperature and desired O ₂ level are achieved	
<i>General</i>	
No deliberate test interruptions before 3 months (record all inadvertant interruptions)	
ΔK can be increased 10% after 3 months if 2 mm of crack growth has not occurred	
No hydrazine additions once test has started (record all water chemistry transients)	
No beach marks	
Isolation of specimen from autoclave — not required but OK if desired	
Corrosion potential measurements — not required but encouraged if feasible	
Corrosion coupons. Place two coupons per specimen in the autoclave close to the specimen, one electrically connected to it, the other isolated from both the specimen and the autoclave. Report the surface preparation and preexposure and postexposure weights on the form provided with the coupon. Then return the coupons and the forms to Paul McConnell at Fracture Control Corporation, noting which coupon is which.

Fig. 8.2. High-R test program results for simulated PWR primary water at 288°C.

PWR environment were from the only test that was performed under highly turbulent flow conditions and that the highest rates were from a specimen tested in an environment that did not meet the PWR water chemistry specification (Table 8.1). A limited amount of metallography suggests that material variability in the center-plate slab used for these tests could also be the source of some of the interlaboratory scatter. For example, the crack front of the specimen giving the slowest rates in the BWR environment (Fig. 8.3) showed severe tunnelling, which is believed to be attributable to compositional and microstructural variations through the specimen thickness. Although average crack-growth-rate data are plotted in Fig. 8.3, the rates near the center of the specimen were at least a

Fig. 8.3. High-R test program results for simulated BWR water at 288°C.

factor of two faster, which would give better agreement with the results reported by the other laboratories.

Although further analysis is required, some tentative observations and conclusions can be drawn from the information available to date.

1. The maximum crack-growth rates in both environments are similar and are comparable with the ASME XI surface flaw reference curve for R > 0.65.

2. The specimens tested in BWR water mostly show "hook-shaped" curves with a very rapid increase of crack-growth rate with increasing ΔK in the early part of the test. Examination of the crack length vs cycles

data from some of the tests indicates that the rapid increase of crack-growth rate was often preceded by an incubation period of several days during which the crack grew very slowly. This behavior suggests that the threshold ΔK for the onset of environmentally assisted cracking for this test material in BWR water under the high-R program test conditions is $<11 \text{ MPa}\cdot\sqrt{\text{m}}$.

3. The specimens tested in PWR water mostly exhibit da/dN vs ΔK curves that are not much different from the behavior in air between 11 and $20 \text{ MPa}\cdot\sqrt{\text{m}}$. At $\sim 20 \text{ MPa}\cdot\sqrt{\text{m}}$, several specimens show a rapid, ten-fold increase of crack-growth rate up to values comparable with the ASME surface flaw curve. This behavior suggests that the threshold ΔK for the onset of environmentally assisted cracking for this test material in PWR water under the high-R program test conditions is $\sim 20 \text{ MPa}\cdot\sqrt{\text{m}}$. (One test result supports a lower ΔK threshold, but that specimen was tested in a different environment from that specified.)

4. The behavior observed is qualitatively consistent with what would be expected if the establishment at the crack tip of a critical concentration of sulfur anion species (from solution or from sulfide dissolution) is necessary to trigger the environmentally assisted cracking process. The oxygen concentration cell present in the BWR case would be expected to facilitate the concentration of sulfur anions at the crack tip, thereby allowing the onset of environmental enhancement to occur under less-severe mechanical conditions than those required in PWR water. However, once the required crack-tip environment is established, it would not be unexpected for the rate of crack growth to be independent of the bulk environment, as observed. The high-flow PWR result is also consistent with this idea because high-flow conditions would tend to increase diffusion rates and reduce the crack-tip sulfur anion concentration compared with stagnant or low-flow conditions. This raises the possibility that some of the interlaboratory scatter may be attributable to differences in the sulfur contents of the test specimens or test environments. Additional work to investigate this possibility is under consideration.

8.4 ICCGR TASK GROUPS

As the sophistication and level of knowledge in the area of environmental crack growth have increased, the topics emphasized in the various task groups have changed. In this section, a brief description of each of the four task groups, along with some of their key accomplishments, is provided.

8.4.1 Test Methods Task Group

In the early days of the ICCGR Group, the Test Methods Task Group provided a valuable opportunity for members to exchange ideas on test-rig design and on methods of measuring load and crack length. Subsequently, the Task Group was heavily involved in the work that was undertaken to understand the results of the low-R testing round-robin and in the development of the test specification for the high-R collaborative test program.

One of the outcomes of the low-R testing round-robin was a recognition of the need to define more precisely the environmental conditions under which cyclic crack-growth tests were performed. Accordingly, the Test Methods Task Group has been pursuing methods of reliably measuring the corrosion potential of the specimens during the tests. A number of reference electrode designs have been built and installed in test rigs to gain experience. Particularly good results have been obtained using an external pressure-balanced silver-silver chloride electrode designed by Andresen.⁶⁹ Hydrogen electrodes also appear to be useful for use in PWR-type environments, particularly in long-term tests if hydrogen is used as the cover gas. At this stage, almost all of the test laboratories of the ICCGR Group have some experience of corrosion potential measurements, and most of them plan to be in a position to routinely monitor test specimen corrosion potential within the next year. A detailed description of the implementation of potential measurement on this program is provided in Sect. 4.

As confidence has increased in the comparability of test results from different laboratories, the Test Methods Task Group became an increasingly important forum for comparison of recent test results. These comparisons generally led to ideas for follow-up tests, which, in turn, helped to aid in developing improved understanding of the behavior under study. Also, as the test methods became standardized, the need for this group decreased, and the comparisons of new results were passed on to the Data Collection task group. This task group was disbanded in 1985.

8.4.2 Mechanisms Task Group

The activities of the Mechanisms Task Group have covered a very broad range and have included fractographic studies, stress-corrosion cracking investigations, cyclic crack-growth work in hydrogen environments, crack-tip environment characterization work, repassivation rate studies, measurements of temperature effects, and the development of mechanistic models. The central theme of the work has been the development of a sufficient mechanistic understanding of the crack-growth behavior of pressure-vessel steels to permit reliable extrapolation of laboratory test data to the service situation that, as noted earlier, often lies outside the variable space that can be covered by practical specimen tests. Consideration is being given to two types of crack-tip advancement mechanisms — anodic dissolution and hydrogen-assisted cracking — but it is understood that they may not be distinguishable because both depend on the same crack-tip rate-limiting processes (transport in solution and oxide film rupture and repassivation).

The major cooperative effort of the Mechanisms Task Group has been an extensive fractographic examination of specimens from the low-R testing round-robin. This work was coordinated by the Technical Research Center of Finland, and the results obtained are reported in full elsewhere.⁶⁹ In summary, it was found that different fractographic features became apparent as the environmental enhancement of the crack-growth rate became more pronounced. In particular, an increasing areal density of brittle-appearing regions was seen as the environmental enhancement increased. The brittle-appearing regions were typically fan-shaped and

emanated from manganese sulfide inclusions. Detailed examination in the crack-tip region suggested that fracture around the inclusions was occurring ahead of the crack front. After exposure to the solution, the inclusions appeared to dissolve slowly.

The detection of cracking ahead of the crack front represents persuasive evidence in favor of the hydrogen-assisted cracking model of crack advance. Additional support comes from the results of cyclic crack-growth tests in hydrogen gas recently conducted at Central Electricity Research Laboratories (CERL) in the United Kingdom. These tests exhibited similar crack-growth rates and similar fractography to specimens tested in simulated reactor water. However, further development of the hydrogen model awaits an improved understanding of the factors that control the amount of crack advance per cycle (which, on the basis of a classical hydrogen embrittlement type of model, is a great deal smaller than would be expected).

As a result of its application to other cracking problems,⁵⁷ the anodic dissolution model has reached a more advanced stage of development than the hydrogen model and currently can rationalize many of the features seen in the cyclic crack-growth tests. For example, the effect of the sulfur content of the steel can be understood through its effect on crack-tip chemistry, which, in turn, controls the repassivation rate. Figure 8.4 shows predictions of the crack-growth rate as a function of

Fig. 8.4. Theoretical crack propagation rate/check-tip strain-rate relationships for low-alloy steels in water at 288°C. The curves for high-sulfur content steels are made on the assumption that sulfate anions predominate in the crack enclave; those for low-sulfur steels are made on the assumption that molybdate anions predominate in the crack enclave.⁷⁰

crack-tip strain rate for low- and high-sulfur steels, based on measurements of repassivation behavior in simulated crack-tip environments.⁷⁰ Although an unequivocal definition of the crack-tip strain rate presents a quantitative problem, it is clear that the predictions of Fig. 8.4 are qualitatively correct. It is also fairly straightforward to use qualitative arguments based on crack-tip chemistry changes to explain why environmental acceleration is suppressed at high flow rates, why an addition of sulfate to the bulk environment accelerates cracking, and why transients in crack-growth rate follow decreases in frequency. If these arguments can be made more quantitative, they could provide the basis for the future development of a reliable, physically based crack-growth prediction procedure.

8.4.3 Data Collection and Evaluation Task Group

The Data Collection and Evaluation Task Group was formed at the inception of the ICCGR, and data-reduction methods have remained one of its concerns up to the present time. However, as its name implies, the scope of this Task Group, which represents the major technology transfer activity of the ICCGR Group, includes data collection, compilation, and analysis, as well as data reduction.

Fairly early in the activities of the ICCGR Group, it became apparent that the volume of data being produced by the members was too great for any one organization to collect and compile on a voluntary basis. Accordingly, EPRI placed a contract with Battelle Columbus Laboratory to develop the computerized data base now called EDEAC (EPRI Database on Environmentally Assisted Cracking). EDEAC, which uses existing interactive data base software supplemented by some special plotting capabilities, currently contains most of the cyclic crack-growth-rate data generated by the ICCGR Group members before 1981, which amounts to 1417 data sets. A quality assurance and updating effort is currently in progress.

Because of concerns about data-reduction methods, raw crack length vs cycles data, as well as reduced da/dN vs ΔK data, are included in EDEAC. Work is now in progress to develop a consensus data-reduction scheme for application to these crack length vs cycles data. The method currently under development contains a number of features identified as necessary in extensive previous Task Group discussions, including

1. linear or bilinear fits of known confidence (in $\log da/dN$ vs $\log \Delta K$ space) of any preselected group of crack length vs cycles data sets,
2. applicability to a variety of crack-growth-rate measurement techniques that weigh reduced data based on the amount of crack growth represented, and
3. ability to separate "scatter" because of real variations in material behavior from that caused by measurement error or local material inhomogeneities.

8.4.4 Applications Task Group

This group was formed in 1985, and its emphasis is on providing a forum for development of crack-growth-rate reference laws that are

improvements over the present laws. Both air and water environments are being considered, and with the help of the data base tended by the Data Collection task group, several candidate laws have begun to take form.

8.5 SPECIMEN EXCHANGES WITHIN ICCGR

Since the formation of ICCGR, specimen exchange has become much more straightforward, and many have been carried out. A summary of the cooperative testing conducted by ICCGR members was given recently by Cullen,⁷¹ so only the exchanges directly involved with this program will be reviewed here.

Exchange of specimens with Materials Engineering Associates (MEA), to whom the test program at NRL was transferred, continued up through the completion of this program. The low-sulfur heat "W7" sent to Westinghouse was followed by a high-sulfur heat "CQ2" sent to MEA. The test results have shown extremely good agreement, as demonstrated in Refs. 72 and 73, which were companion papers in a recent International Atomic Energy Agency (IAEA) Specialists meeting. This high-sulfur steel "CQ2" was also furnished to Centro Informazione Studi Esperienze (CISE) with similar confirmation although results are not yet published.

Specimens were exchanged with Framatome with a number of useful results. The first comparison led to the conclusion, later confirmed by the low-R round-robin tests, that the triangular wave form produced less environmental enhancement than a sinusoidal loading form. Follow-up exchanges of various materials led to the first evidence of the sulfur chemistry effect on environmental enhancement.¹⁶

Another exchange was made with the UKAEA, Harwell, and the results have already been reviewed in Sect. 6 (Fig. 6.9). The comparison of the results is rather poor in this case, and this led to an extensive effort to identify the reasons for interlaboratory variability within ICCGR. The largest identifiable difference between the two laboratories was the environmental flow rate. The Harwell flow rate was at least an order of magnitude faster than that in the Westinghouse system, and it is theorized that the crack-tip region was washed free of the species that would act to accelerate growth. This was the only specimen exchange that did not produce comparable results.

8.6 CONCLUSIONS

ICCGR has always displayed an openness and willingness to cooperate on both test techniques and recent findings. This atmosphere has led to very quick advances in the state of the art, sometimes in a year or less. A good example is the development of reliable environmental monitoring of the electrochemical potential (Sect. 5). A second example is the discovery of the effect of sulfur material chemistry on environmental crack-growth enhancement,¹⁶ which has now been extensively verified by the group.

During the years since its inception, the available research work on environmentally enhanced crack growth has expanded by more than an order of magnitude. The major sources of interlaboratory scatter have been identified and eliminated, and the important real variables have been identified. A large body of data has been generated and compiled on the EDEAC data base, and great progress has been made in understanding the mechanisms of the environmental cracking process.

The group has organized under the auspices of the IAEA two international specialists meetings, open forums held in 1981 and 1985 in Germany and Japan, respectively.^{74,75} These symposia have included over 100 technical papers from at least 12 different countries and have each represented a good summary of the state of the art.

The main challenge remaining for the group is implementation of the research findings into revisions of the reference crack-growth-rate laws for pressure-vessel steels. This work is discussed further in Sect. 9.

9. REFERENCE CRACK-GROWTH-RATE LAWS

9.1 INTRODUCTION

The objective of the research carried out in this program has been to give improved understanding of the crack-growth-rate properties of LWR materials exposed to primary coolant environments. This understanding can then be used to improve for the industry the reference fatigue crack-growth laws contained in Sect. XI of the ASME Code.

This section begins with a summary of the background and technical basis of the reference fatigue crack-growth laws presently contained in Sect. XI. This discussion will provide a starting point for the rest of the section, in which recent findings will be related to these existing curves and conclusions will be reached about their applicability. Reference laws for both air and water environments will be discussed.

Two key variables have emerged from recent work: the effect of material chemistry and microstructure and the characteristics of the water environment in which crack growth occurs. The material effect is presently believed to be best measured by the sulfur content of the steel, in which high sulfur levels have been found to increase environmental acceleration of crack growth; several examples of this effect will be shown. Work on the mechanisms of environmental acceleration that confirms this effort and provides an explanation for it has been done (Sect. 6).

The environmental effect has been found to be best characterized by the free corrosion potential that exists in the system, and a great deal of development work has been done to find reliable measurement techniques (Sect. 5). This has been coordinated through ICCGR.

One environmental factor shown to be important is the flow rate of the water, and observations on this subject will be made. This factor is closely related to the water chemistry effects, and mechanistic explanations for this will be reviewed.

Finally, the impact of these new observations on potential revision of the reference curves will be discussed, and conclusions will be reached on the likelihood of such changes.

9.2 BACKGROUND AND TECHNICAL BASIS OF THE PRESENT ASME SECT. XI REFERENCE CRACK-GROWTH CURVES (WATER ENVIRONMENTS)

The present ASME Code reference fatigue crack-growth curves for pressure-vessel steels in water environments were adopted with the Winter Addendum to the 1980 Code. The technical basis for these curves was provided in Ref. 76 and will be reviewed briefly here to provide a basis on which to judge recent developments.

The data available from 1978 and 1979, along with the first reference curves that were developed, are summarized in Fig. 9.1. These curves are statistically developed 95% global confidence limits on the mean of the data, which correspond to predictions that the mean behavior of 95% of the future data will fall below the reference curves. The

Fig. 9.1. Summary of data used for 95% global confidence limit on mean data - 1979.

original Sect. XI reference curve for water environments is also shown in Fig. 9.1 for reference.

The general fatigue crack-growth behavior of ferritic steels in a water environment is shown schematically in Fig. 9.2. The growth rate increases rapidly with the increasing range of stress-intensity factor ΔK . At a point that depends on frequency and R ratio, the slope changes, and the curve then continues until it reaches the curve corresponding to air behavior. The growth-rate curve that follows is strongly dependent on R ratio and loading rate or cyclic frequency. The R ratio is accounted for directly in the reference curves (Fig. 9.3). The effect of welds vs base metal on the crack-growth-rate data is shown, for example, in Fig. 9.4, where it is seen that weld crack-growth data are slightly lower and contain frequent retardations. The reference curves account for frequency or loading rate effects by using data obtained at

Fig. 9.2. Schematic of frequency and R ratio effects — pressure vessel steels in LWR environment.

Fig. 9.3. Comparison of revised curves with previous reference curve.

Fig. 9.4. Comparison of crack growth of base metal and submerged arc weldments in pressurized-water environment.

the frequency for which the highest growth rates were found, 1 cycle/min (e.g., Fig. 9.5). Available data at the time indicated no measurable difference between PWR and BWR environments, so all the data were treated together. Although irradiation effects were unknown in a water environment, they were assumed to be negligible.

The resulting curves were checked for adequacy by using them to predict the individual specimen results on which they were based and to predict the behavior of two specimens loaded with a range of conditions designed to simulate an operational condition. Loadings and frequencies applied to the two specimens are shown in Table 9.1. The predictions using the reference curves are shown in Table 9.1 and plotted in Figs. 9.6 and 9.7. Results show that the ASME revised reference curves

Fig. 9.5. Effects of test frequency on crack growth for series of medium- and high-sulfur steels.

are conservative predictors of the specimen behavior in both cases. In these figures, curve 1 is the old Sect. XI law, curve 2 is a graphically determined revision (not adopted), and curve 3 is the revised ASME law.

The revised curves are double-sloped curves on a logarithmic plot (Fig. 9.8). This was a subjective decision based on the overall appearance of the data, along with some insight into the mechanisms involved in the crack-growth-rate process. The data were grouped into two families according to R ratio, and two curves resulted. A continuous correction for R ratio was not employed because an adequate parameter for this approach could not be found.⁷⁶

Note that the selection of a point-of-slope change is an important one, so this was done statistically. A series of break points was tried, and for each one, a straight line was fitted to the data on either side

Table 9.1. Multiple-load experiments used for verification of the ASME reference crack-growth-rate law

Cycle type	Relative number of occurrences	R ratio	Initial ΔK	Final ΔK	Total growth (%)	
					Sect. XI (old)	Sect. XI (present)
<i>Specimen R-16</i>						
A 1/2-min ramp	1	0.17	33.5	65.4	75	42
B 1 cpm	6	0.68	15.4	30.1	25	58
<i>Specimen R-15</i>						
A 1/2-min ramp	1	0.18	34.6	70.2	91	61
B 1 cpm	10	0.92	3.6	7.4	0.3	0.3
C 1 cpm	5	0.75	12.1	24.6	8.7	36.7

^aCycle details given in the schematic below.

of it on the ΔK axis by the least-squares method. For each fit, the residual variance was obtained, and then the residual variance of the combined curve was determined by a standard relationship. The best cut point was that which produced the smallest residual variance.

The equations for the revised reference curves are given in the following:

R ratio < 0.25

$$(\Delta K < 19 \text{ ksi}\cdot\sqrt{\text{in.}}) \frac{da}{dn} = 1.02 \times 10^{-6} \Delta K^{5.95} ,$$

$$(\Delta K > 19 \text{ ksi}\cdot\sqrt{\text{in.}}) \frac{da}{dn} = 1.01 \times 10^{-1} \Delta K^{1.95} ,$$

Fig. 9.6. Results of verification test R-15.

Fig. 9.7. Results of verification test R-16.

Fig. 9.8. Present Sect. XI reference fatigue crack-growth-rate laws for ferritic pressure vessel steels.

R ratio $\geq$ 0.65

$$(\Delta K < 12 \text{ ksi}\cdot\sqrt{\text{in.}}) \frac{da}{dn} = 2.52 \times 10^{-1} \Delta K^{5.95} ,$$

$$(\Delta K < 12 \text{ ksi}\cdot\sqrt{\text{in.}}) \frac{da}{dn} = 2.52 \times 10^{-1} \Delta K^{1.95} ,$$

where crack-growth rate is in microinches per cycle (10^{-6} in./cycle). To obtain $\text{MPa}\cdot\sqrt{\text{m}}$ from $\text{ksi}\cdot\sqrt{\text{in.}}$, multiply by 1.1, and to obtain millimeters from microinches, multiply by 2.54×10^{-5} .

For R ratios between 0.25 and 0.65, interpolation is recommended, and a method has been developed to accomplish this (Fig. 9.8):

$0.25 < \text{R ratio} < 0.65$

$$(\Delta K < X \text{ ksi}\cdot\sqrt{\text{in.}}) \frac{da}{dn} = 1.02 \times 10^{-6} Q_1 \Delta K^{5.95} ,$$

$$(\Delta K < X \text{ ksi}\cdot\sqrt{\text{in.}}) \frac{da}{dn} = 1.02 \times 10^{-6} Q_2 \Delta K^{1.95} ,$$

where

$$Q_1 = 26.9 R - 5.725 ,$$

$$Q_2 = 3.75 R + 0.06 ,$$

and X is determined from the intersection of the two curves for da/dn .

Considering Fig. 9.3, it is seen that this revised law is considerably more conservative than the old Sect. XI law for lower values of ΔK and for high R ratio. At higher values of ΔK , the revised law is less severe than the old Sect. XI law, and also the slopes for the two R ratio values are identical, as might be expected theoretically. As a result of the statistical method determining the confidence limits, the curves are higher than the true mean of the data by a factor of 1.5 to 2.0. A single factor is not possible to calculate because of the two slopes and two R ratios with different amounts of data for each.

The reference curves in their present form were based on ~25 specimen results, including base metal and welds. Since that time (circa 1978), the available data base has increased by nearly an order of magnitude, and many carefully controlled studies, including mechanistic ones, have been carried out. This work has resulted in many new findings and revealed the essence of environmental enhancement effects in these materials. These environmental effects are now known to result from three sources: mechanical, metallurgical, and environmental variables (Table 9.2). The present reference curves were primarily based on mechanical variables but have held up remarkably well in light of the new findings in these other areas. The next two sections will consider the new findings in some detail.

Table 9.2. Variables affecting environmental fatigue crack growth

<i>Mechanical</i>	
Stress ratio or load ratio R	
Frequency and waveform	
<i>Environment</i>	
Temperature	
Level of oxygen and other dissolved gases	
Electrochemical potential	
Velocity of water	
<i>Metallurgical</i>	
Level and distribution of alloying elements and impurities	
Microstructure	
Mechanical properties (strength and toughness)	

9.3 METALLURGICAL VARIABLES

It has been found that the level of sulfur in steel can have a remarkable effect on its sensitivity to environmental enhancement of crack growth. This effect was demonstrated in a carefully constructed matrix of tests carried out in this program (Sect. 3.3).

Results obtained in the matrix of tests have been remarkably consistent, and the same trend has been observed at both high and low R ratio. The trends in the degree of environmental enhancement are the same as those observed earlier for fracture toughness and Charpy energy, where the materials and orientations that display the lowest toughness also display the highest degree of environmental susceptibility (e.g., Fig. 3.26).

The highest environmental enhancement was found for the TL orientation; the lowest occurred in the LS orientation. (See Fig. 3.1 for identification of orientations, per ASTM.) Materials with high sulfur content showed the most enhancement, as well as the strongest orientation effects (e.g., Fig. 3.21). The observed behavior of the LS orientation was about the same regardless of sulfur content and was equivalent to the behavior of all three orientations of the low-sulfur heat TW (Fig. 3.23). This is an important observation because the LS crack orientation is of most interest from the standpoint of vessel integrity.

Corrosion-fatigue crack growth in submerged arc welds has been found to follow the same pattern as base metal, which has similar material chemistry, except that retardation occurs with some regularity (Fig. 9.4). These retardations are more pronounced at higher R ratios, so the net effect is that crack growth in welds will be somewhat slower.

Although various fluxes have been used for the submerged arc welds, differences in crack-growth-rate behavior appear to be more dependent on

the sulfur content of the weld than on the flux used. This can be seen in Figs. 9.9 and 9.10, in which welds made with two different fluxes with nearly equal levels of sulfur (0.012 to 0.014 wt %) are compared with a lower-sulfur weld (0.008 wt %) made with a third flux.

These findings are, in general, consistent with fatigue crack-growth-rate data obtained for these steels. There is clearly a different relationship between bulk sulfur content and environmental acceleration in the different product forms. Very high sulfur levels produce enhanced crack growth in plates, forgings, and welds (Figs. 9.9—9.12). For plates, as discussed earlier and shown in Fig. 9.9, even a very low sulfur plate results in enhanced growth. On the other hand, Fig. 9.11 shows that a forging with relatively high sulfur content (heat F, S = 0.012) can show very little environmental enhancement. Figure 9.12 shows a similar result for a relatively high sulfur weld (weld D, S = 0.008). Therefore, the threshold for environmental enhancement is different in plates, forgings, and welds.

9.4 POSSIBILITIES FOR REVISION OF REFERENCE LAWS (WATER ENVIRONMENTS)

The discussion in Sect. 9.3 shows that there can be large differences in the degree of environmental enhancement produced by steels that are nominally the same but have different sulfur contents and large differences in enhancement for different product forms with the same nominal sulphur content. As may be seen in the previous figures, even the highest crack-growth rates are still covered by the present reference fatigue crack-growth curves. Lower growth rates are clearly in evidence for lower-sulfur steels, but this effect seems to be much less in evidence at high R ratios (e.g., Fig. 9.13).

The mechanisms of environmental enhancement of crack growth are now being aggressively investigated by many laboratories. A number of important variables have been identified (Sects. 5 and 6). The key variables include water chemistry, water flow rate, and the crack-tip strain rate.

Recent work by Scott⁷⁷ and Ford⁵⁷ has led to development of a relationship between the crack-tip strain rate and the crack-growth rate in environments. The crack-tip strain rate is directly proportional to the macroscopic growth rate in air, but environmental effects were not considered until this recent work, which has been discussed in some detail in Sect. 6.

Inspired by this work, Gilman⁷⁸ attempted to develop a time-based crack-growth model for water environments, using the approach first suggested by Shoji et al.⁷⁹ Because of the large amount of scatter in the data, upper and lower bounds were developed (Fig. 9.14). These bounds correspond roughly to high- and low-sulfur data, but a direct correspondence could not be established because of the more complex effects in evidence, as discussed above.

This candidate crack-growth approach was recently applied to an example of cracking that occurred in BWR power plants some years ago,⁶⁰ and the results are presented in Figs. 9.15 and 9.16. Here, note that

Fig. 9.9. Comparison of Linde 80 and Linde 0091 sample weld crack-growth behavior.

Fig. 9.10. Comparison of Linde 124 and Linde 0091 sample weld crack-growth behavior.

Fig. 9.11. Effect of sulfur content on fatigue crack growth in PWR environment at $R = 0.2$, A 508 class 2 forgings.

Fig. 9.12. Effect of sulfur content on fatigue crack growth in PWR environment at $R = 0.2$, submerged arc weldments.

Fig. 9.13. Comparison of fatigue crack-growth rates for low-sulfur plate materials tested at $R = 0.7$ in PWR environment.

Fig. 9.14. Development of a time-based fatigue crack-growth-rate model.⁷⁸

Fig. 9.15. Comparison of fatigue crack-growth predictions with field cracking data for BWR old nozzle/sparger design, assumed initial flaw 0.25 in. deep.

Fig. 9.16. Comparison of fatigue crack-growth predictions with field cracking for BWR old nozzle/sparger design, assumed initial flaw 0.10 in. deep.

the upper-bound curve significantly overpredicted the growth that occurred, and the lower-bound curve significantly underpredicted the growth. This result is somewhat encouraging because refinement of the model may produce a more accurate prediction; it is also noteworthy that the present Sect. XI law provided a rather accurate prediction in this case.

9.5 REFERENCE LAWS FOR AIR ENVIRONMENTS

The reference crack-growth-rate law in the ASME code for an air environment has been present since 1972 when the flaw evaluation appendix of the code (Appendix A) was first drafted. At that time, there were available only a small amount of environmental fatigue crack-growth data, and these data had been obtained at very high cyclic frequencies (60 and 600 cycles/min). Because these results showed little or no environmental influence, they were combined with the available air data, and a hand-drawn upper bound to these data became the reference curve. The majority of these data were obtained in this program in its early years, 1969-1971. The specimens and test conditions have been summarized in Sect. 3.6.

A review of the air curve relative to tests conducted since that time on a very wide spectrum of loadings on the pressure-vessel steels provides some insight into its overall accuracy. The original formulation, which remains today, was in terms of the driving force ΔK by itself. There were no known effects of R ratio at the time, and the curve, along with its supporting data, is shown in Fig. 9.17. Work conducted by Dowling²³ at very high loadings confirmed the validity of the present curve in the high ΔK range (Fig. 9.18).

In the low- ΔK region, there has been a recent review by Eason et al., which showed that the slope of the present curve is too steep here.⁸⁰ His review of available data is summarized in Fig. 9.19. It is clear from these results that the Sect. XI air curve should be revised in the low- ΔK region.

The impact of such a revision on calculated crack-growth rates for most applications will not be strongly felt, however, because the growth rates in this region are extremely low.

9.6 SUMMARY AND CONCLUSIONS

A great deal of study of the issue of environmentally assisted crack growth has been completed since the revision of the ASME reference crack-growth curves in 1980. This work has resulted in a data base almost an order of magnitude larger than that available at that time. These new data have generally confirmed the adequacy of the present curves and have provided much more insight into the mechanisms that are operating. It is possible that a modification of the present curves may be possible for newer, low-sulfur steels in a tightly controlled environment, but more-complete understanding of the operable mechanisms will be necessary before such a step is warranted.

Fig. 9.17. Section XI air curve with supporting data.

Fig. 9.18. Comparison of fatigue crack-growth curve for air environment with recent data.²³

Fig. 9.19. Comparison of fatigue crack-growth results for low-alloy steels tested in air environment with Sect. XI reference curve for air environments.

10. SUMMARY AND ACCOMPLISHMENTS

While the assessment of single severe loadings on a structure is a very important element of an integrity analysis, the net risk to the structure is relatively low because the high-probability events are not severe and, generally, the severe events are low in probability. Fatigue loadings present perhaps the highest net risk to the propagation of an existing flaw because they are virtually certain to occur.

This fact was recognized in the early stages of the Heavy-Section Steel Technology Program, and an experimental program was developed to characterize the fatigue crack growth of pressure-vessel steels in LWR environments in the late 1960s. It was also recognized in the development of flaw evaluation criteria in Sect. XI of the ASME Code, in which fatigue crack growth is one of the considerations.

The tests conducted under this program provided the majority of the data upon which the first Sect. XI reference crack-growth-rate curve was based in 1972. This curve was formulated based on air and high-frequency water data.

This program provided the first verification of Kondo's discovery in 1971 that low-frequency loadings in water environments result in significant acceleration in crack-growth rates above those for air environments.¹ The data produced by this program provided the basis for the early incorporation of water-environment effects on crack growth in the ASME Code in the 1974 edition of Sect. XI Appendix A.⁹ This was the first incorporation of environmental effects on crack growth in a code or standard in the world.

The program continued with emphasis in the mid-1970s being directed at the effects of environment on plates, forgings, welds, and HAZs to determine if all behaved similarly. The findings were that they all behaved similarly, but more-complete studies revealed that the crack-growth rate did not increase in a linear fashion on a logarithmic plot, as with inert environments, but tended to flatten out at a growth rate that was dependent on the frequency of loading. A significant effect of R ratio was also discovered and verified by other investigators. Again, the data produced in this program formed the majority of the data base used to develop a revision to the ASME Code reference crack-growth curves for water environment, which was accomplished in the winter addendum to the 1980 Code.⁷⁶

Most recent work has resulted in the discovery that the degree of environmental enhancement for pressure-vessel steels can be different for different heats of the same steel. It was found that the sulfur content of the steel was the key factor: high-sulfur levels resulted in significant enhancement but low-sulfur steels showed only slight enhancement over growth rates observed in air.¹⁶ This was a significant finding because it shows that the newer, low-sulfur steels used in recent construction will be less susceptible to environmental fatigue.

Another recent discovery of this task was that cracks in pressure-vessel steels can grow under constant load in water environments under certain conditions.⁸¹ The relationship between this mode of subcritical crack growth and fatigue is currently under investigation, and efforts are also being aimed at better defining the conditions under which this growth can occur.

The program has continued in recent years with emphasis being placed on further understanding the mechanisms involved in the process of environmental enhancement of fatigue crack growth. This understanding is essential to enable accurate predictions of crack growth during service because it is impossible to model all of the types of loadings that can occur. The overall goal to develop accurate characterization of subcritical crack growth for use in the assessment of defects found in operating plants has remained.

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13. ABSTRACT (200 words or less)

This report presents a collection of the information developed in the crack-growth studies at Westinghouse Electric Corporation from 1969 to 1986 under a subcontract with the Heavy-Section Steel Technology Program. Although progress reports were issued regularly over the entire period, the program results have not previously been integrated in a single document.

A large number of tests were performed to characterize the effects of pressurized-water environment on the enhancement of fatigue crack growth. Specimens were taken from plates, forgings, welds and heat-affected zones. This work provided a data base for development of ASME Code reference crack-growth curves, as well as insights into some of the mechanisms of environmental enhancement. Static-load crack-growth-rate tests were also conducted, with some specimens loaded in excess of 11 years.

Extensive fractography was performed and the observations were correlated with the microstructure and the level of environmental enhancement. All the data obtained in this program were compared with results obtained at other laboratories.

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