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ABSTRACT

Designers of high-temperature power generation plants, both nuclear and fossil-fired, have made extensive use of 2 1/4 Cr-1 Mo steel for piping, shell, and tubing and as a tubesheet material in steam generators. While this material has been researched as much if not more than any other boiler material, an understanding of the time-dependent fatigue behavior for long-term service applications is still incomplete. Progress is reported in obtaining formulations that can be used for low and high cycle continuous cycle fatigue, time-dependent fatigue, and crack growth behavior. Material variables such as melting practice, heat-to-heat, and heat treatment variations are discussed. The importance of environment is shown by comparing the results of elevated-temperature strain and load-controlled tests of different waveforms conducted in both air and helium. Interim progress is reported in characterizing the crack growth behavior of HAZs adjacent to dissimilar weld joints of this material.

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INTRODUCTION

Materials specialists in the United Stated selected 2 1/4 Cr-1 Mo steel for use as the primary steam generator material in the Clinch River Breeder Reactor Plant (CRBRP) after consideration of the material's good service record in boiler tubes and main steam piping systems in fossil-fired steam generating plants. Further, 2 1/4 Cr-1 Mo steel had been qualified to meet most of the high-temperature ASME Boiler and Pressure Vessel Code requirements, and it was also used as the steam generator material in EBR-II, which has operated virtually trouble-free since 1965. Peak temperatures in the CRBRP steam generators were set at about 494°C.[1]

As tubing, this material [ASME Specification SA-213, minimum yield, and ultimate tensile strength of 207 and 413 MPa (30 and 60 ksi), respectively] is also under consideration as a High Temperature Gas-Cooled Reactor (HTGR) steam generator material with peak temperatures as high as 528°C.[2]

Because components of these systems would be subject to creep-fatigue interaction due to thermal transient-induced strain cycling and hold periods, possible Departure from Nucleate Boiling (DNB) in evaporator tubing, as well as flow-induced vibrations, it was recognized that design against fatigue failure would be an important consideration. Further, in the case of the CRBRP steam generators, the need for both longitudinal as well as girth welds in the shell necessitated crack growth or flaw propagation data in both base as well as weldment material to confirm design adequacy with respect to flaw propagation. Accordingly, it is the objective of this paper to review progress in obtaining low and high cycle fatigue, creep-fatigue, and crack propagation data on generally annealed 2 1/4 Cr-1 Mo steel for steam generator usage.

Continuous Low and High Cycle Fatigue Behavior

Continuous cycle fatigue curves are not presently in ASME Code Case 1592 for 2 1/4 Cr-1 Mo steel; however, much of the data required has been generated and analyzed.[3] An example of the data base from strain-controlled fully-reversed uniaxial fatigue tests covering temperatures of 428 to 593°C is shown in Fig. 1. Data are compared from specimens that were in the annealed and isothermally-annealed condition. The isothermal anneal is used for tubing and is accomplished in a continuous-annealing furnace that contains varying temperature zones. The tubing is moved from a zone where austenitization occurs to a lower temperature zone, usually between 677 and 732°C. The data comparisons given in Fig. 1 indicate that there is no difference in the low cycle fatigue life of this material for either heat treatment. Current efforts are directed at obtaining high cycle fatigue data in the range of 10^6 to 10^8 cycles and beyond.

In generating the strain-controlled continuous cycle data base a number of material variables had to be considered, such as melting practice, heat-to-heat as well as heat treatment variations, loss of carbon due to mass transfer, and other environmental factors. Melting and heat treatment practice imposed on the steam generator material was considered to be an important element in the design of the U.S. CRBRP. Accordingly, Vacuum-Arc Remelted (VAR) and Electro-Slag Remelted (ESR) 2 1/4 Cr-1 Mo steel were specified for the CRBR tubesheets and tubes, respectively, with air-melted material selected for the shell, nozzles, etc. It was felt that remelting, which reduces both the inclusion content and certain impurity levels as well as increasing weldability and workability, was worth the extra effort. Further, a post-weld

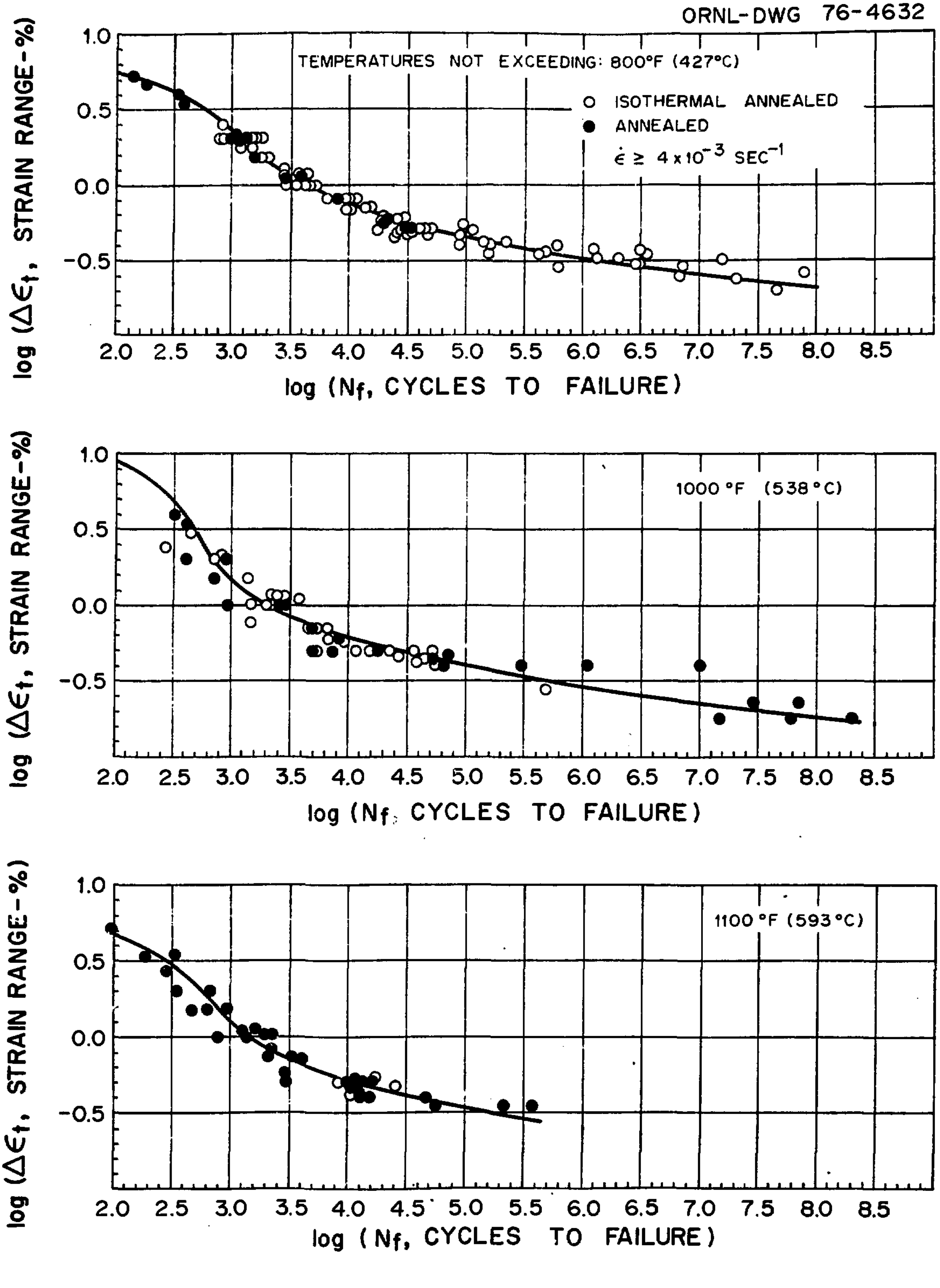

Fig. 1. Strain-Controlled Fully-Revresed Fatigue Data Generated at a Strain Rate of $4 \times 10^{-3}s^{-1}$ for Air-Melted and Annealed 2 1/4 Cr-1 Mo Steel.

heat treatment (PWHT) consisting of heating to 727 ± 13°C and holding for up to 40 h was specified in certain instances. These variations required fatigue property evaluations of prototypic material to determine the influence of these factors on subsequent behavior. In Fig. 2 a comparison is made between a best fit curve obtained from data (omitted for clarity) generated from air-melted and annealed material and data from ESR and VAR material with the PWHT tested at 538°C. Apparently these variables have little or no influence on continuous cycle fatigue life as is shown for this temperature and range of cycle life.

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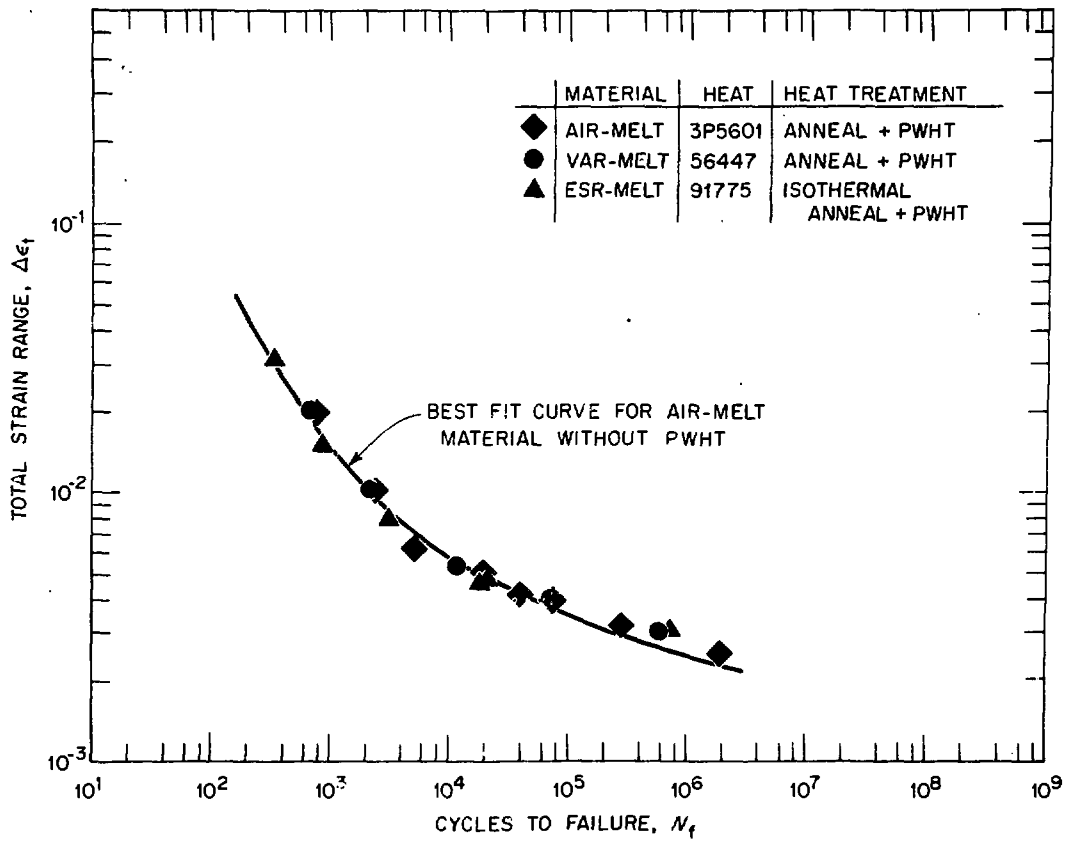

Fig. 2. Effect of Melting Practice and Postweld Heat Treatment (PWHT) on the Fatigue Life of 2 1/4 Cr-1 Mo Steel at 538°C.

Generally large heat-to-heat variations have not been noted in the continuous cycle strain-controlled fatigue life of this material except in the dynamic strain aging temperature range of 316 to 427°C.[4] These variations occurred primarily in the high cycle region as shown by a comparison of the response of two air-melted heats (20017 and 3P5601) given in Fig. 3. Note that heat 3P5601 showed a superior high cycle fatigue life in comparison to heat 20017 (data omitted for clarity) for fatigue lifetimes in excess of about 10^4 cycles. Since the carbon content was nearly the same for the two heats, the differences in fatigue life were attributed to subtle differences in chemistry

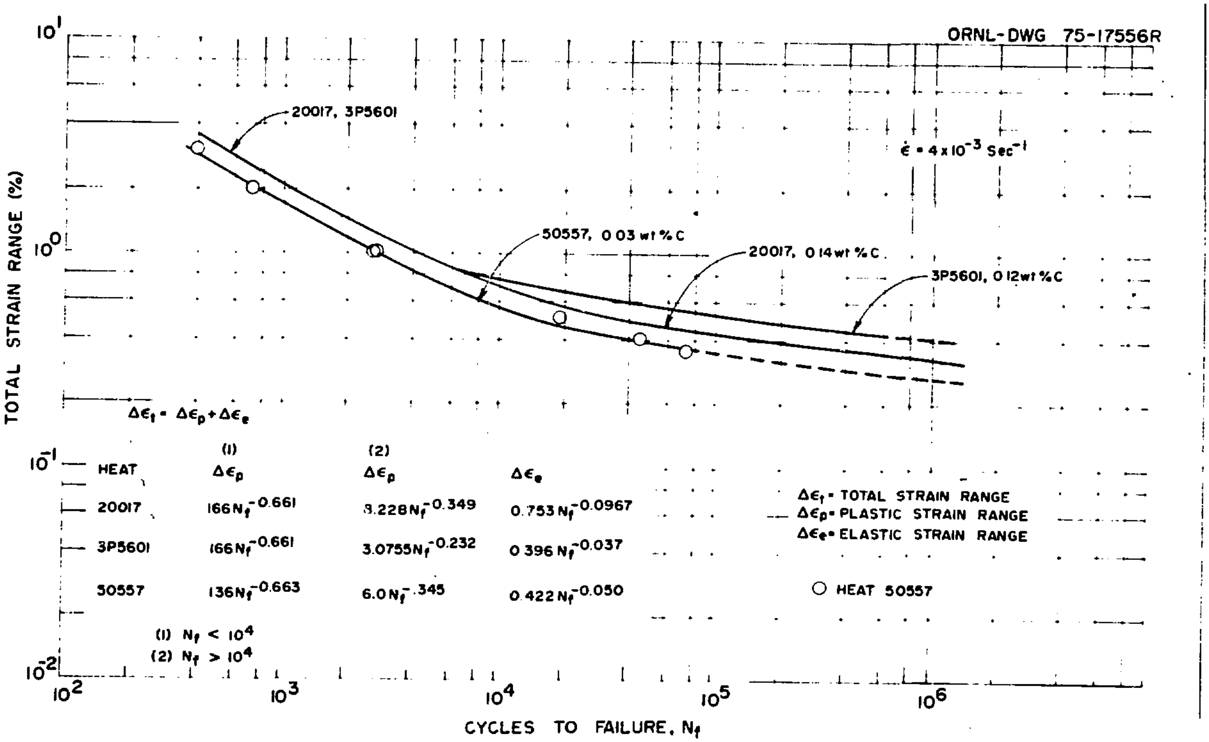

Fig. 3. Comparison of the Strain-Controlled Fatigue Behavior of Three Heats of Isothermally Annealed 2 1/4 Cr-1 Mo Tested at 427°C at a Single Strain Rate. Carbon contents were as indicated.

and/or response to heat treatment. Also shown for comparison are fatigue data generated on a low carbon heat (heat 50557), which might be indicative of material in the partially decarburized condition subsequent to high temperature service where carbon reduction occurred. Data used to construct the curves in Fig. 3 were fit by an equation of the form

$$\Delta\varepsilon_t = A\Delta\varepsilon_p{}^a + BN\Delta\varepsilon_e{}^b ,$$

where $\Delta\varepsilon_t$, $\Delta\varepsilon_p$, and $\Delta\varepsilon_e$ are the total, plastic, and elastic strain ranges respectively, and the coefficients and exponents are given in Fig. 3.

Results of continuous cycling fatigue tests generated in air, high vacuum, and low oxygen containing sodium have generally shown that over the temperature range of 538 to 593°C that low oxygen-containing environments increase the fatigue life by factors as high as 2 to 4 in comparison to air data.[5] Currently, continuous cycle as well as strain cycle with a hold period (creep-fatigue) tests are being conducted at ORNL in one of several possible HTGR environments. The gaseous environment is helium with typical controlled impurity levels given in Table 1.

Table 1. Compositional Limits of Test Gases[a]

Impurity	Level Noted (ppm)
H_2	150–184
O_2	10^{-22} atm[b]
N_2	∿1
CH_4	12–16
H_2O	∿5

[a]Test chambers operate at 184 kPa (12 psig, 26.7 psia). Multiply ppm × 1.82 to obtain μatm; by 0.184 to obtain Pa.

[b]Gas passed through furnace at 500°C to react O_2 with H_2. Estimate based on equilibrium.

Representative data from tests conducted in the above environment are tabulated for comparison purposes with data generated in air in Table 2.

Table 2. Comparison of Results of Continuous Cycling and Creep-Fatigue Tests[a] Conducted in Air and a Simulated HTGR Helium Environment at 538°C

Total Strain Range (%)	Environment	Hold Mode[b]	Hold Period (h)	Values at $N_f/2$, Plastic Strain Range $\Delta\varepsilon_p$ (%)	Cycles to Failure, N_f	Time to Failure, t_f (h)
2.0	Air	—	0	1.703	881	2.4
2.0	Impure He	—	0	1.644	1,059	3.1
2.0	Air	T	0.5	1.789	640	322.2
2.0	Impure He	T	0.5	1.716	546	273.3
2.0	Air	C	0.5	1.772	714	359
2.0	Impure He	C	0.5	1.745	715	358
2.0	Air	T+C	0.5	1.872	533	535
0.5	Air	—	0	0.276	16,036	11.1
0.5	Impure He	—	0	0.208	21,645	15.0
0.5	Air	C	0.1	0.320	4,496	453.0
0.5	Impure He	C	0.1	0.301	11,188	1126.3
0.5	Air	T	0.1	0.284	15,450	1555.6
0.5	Impure He	T	0.1	0.284	5,860	589.9
0.5	Air	T+C	0.1	0.314	1,954	392
0.5	Impure He	T+C	0.1	0.314	2,953	592.4
0.31	Air	—	0	0.144	63,906	31

[a]All tests were conducted in completely reversed strain control at a strain rate, $\dot{\varepsilon} = 4 \times 10^{-3}$/s.

[b]T implies a hold period at peak tensile strain amplitude; C, compression

Results of environmental continuous cycle zero hold time tests conducted at 538°C are compared in Fig. 4. Data from continuous cycle tests generated in air were mostly deleted with a "best fit" of the air data line shown for clarity covering the range of about 10^3 to 10^6 cycles to failure. Comparing the cyclic lives of specimens tested in impure helium with the air data (Table 2 and Fig. 4), it is seen that

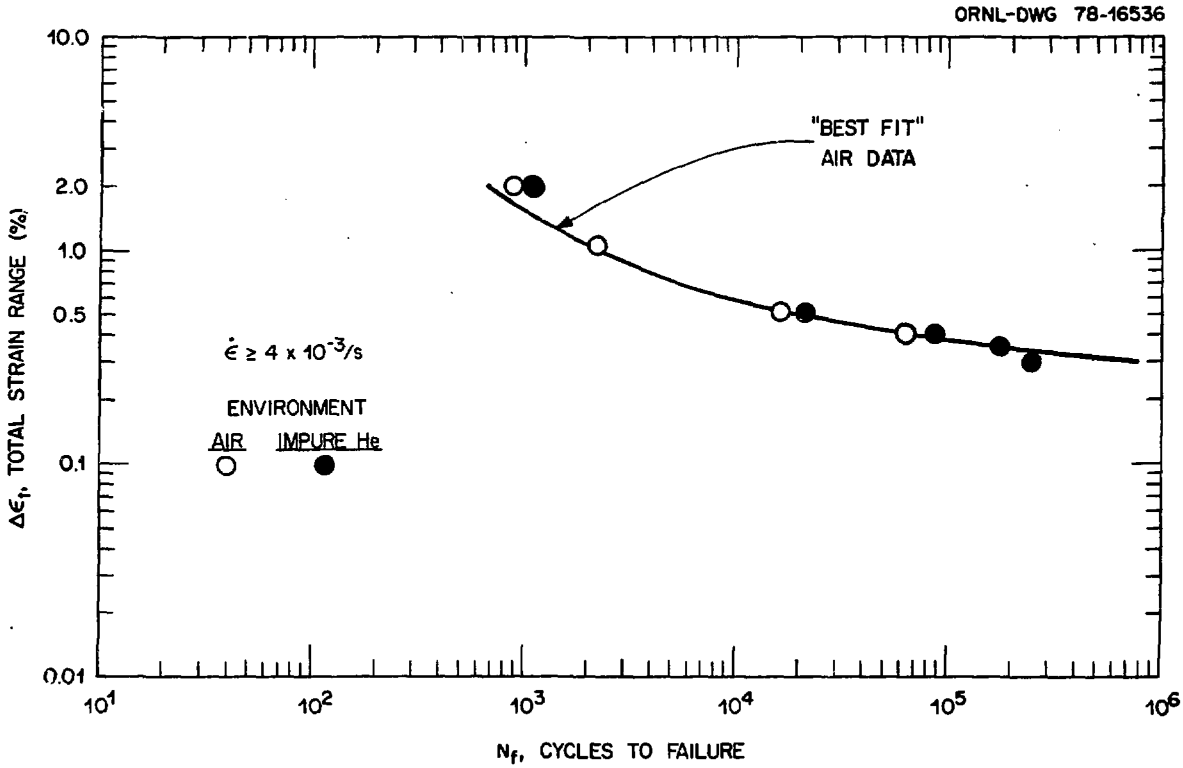

Fig. 4. Comparison of Continuous Cycling Strain-Controlled Fatigue Data from Tests Conducted on 2 1/4 Cr-1 Mo at 538°C.

the cyclic lives of specimens tested in impure helium exceed or are equivalent to those tested in air. The slight increase in fatigue life in the helium environment is attributed to the low oxygen rather than to decarburization, which can occur after prolonged exposure in this environment at higher temperatures, as shown in Fig. 5.

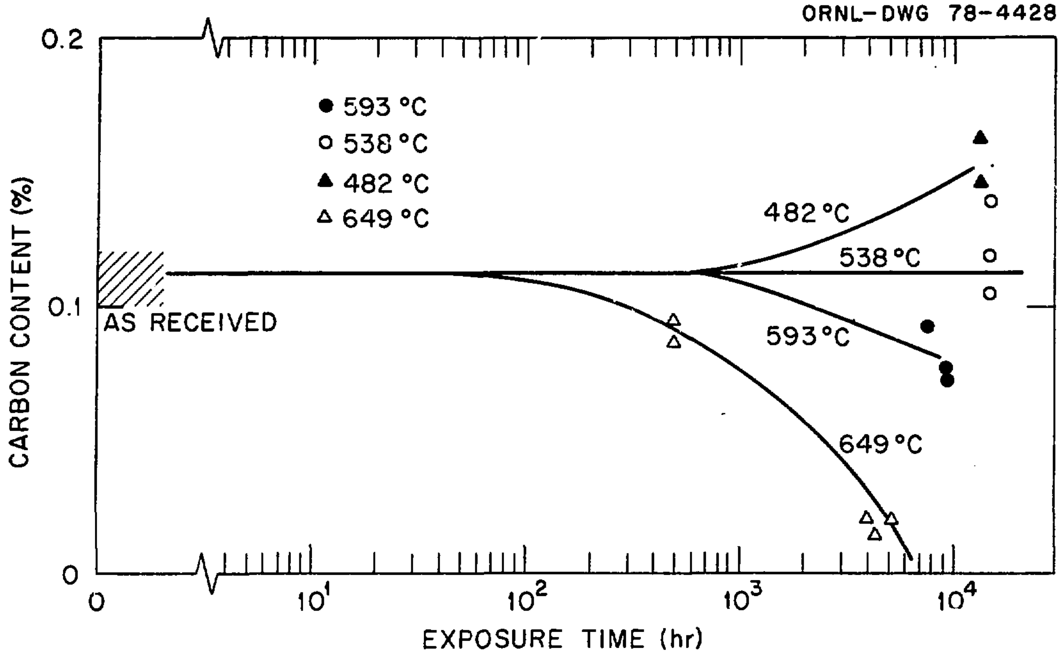

Fig. 5. Carbon Contents of 2 1/4 Cr-1 Mo Steel Exposed to HTGR Helium. (Specimens were sheet 1.6 × 13 mm cross-sections and carbon contents are average values over the entire cross-section.)

Time-Dependent Fatigue Behavior

Section III, Division 1, of the ASME Boiler and Pressure Vessel Code provides design rules for avoidance of structural failure by creep or time-dependent fatigue. Therefore, it is important that a satisfactory data base be generated to permit development of time-dependent fatigue rules for this material. Essentially, strain-controlled cyclic hold periods occur during operation at temperatures within the creep range such that relaxation or creep damage occurs. Therefore, a large number of uniaxial tests have been run on this material to duplicate this mode of failure.[5]

Representative cyclic lives of strain-controlled tests conducted in both impure helium (Table 1) and air are compared in Fig. 6 for several strain-time waveforms repeated each cycle as shown. All tests were conducted at a single strain range of 0.5% and at temperatures of either 482 or 538°C.

Comparing the cyclic lives of the specimens tested in air, it is apparent that (1) hold periods, i.e. 0.1 h, reduce the cyclic lives below that of the continuous cycle or no hold time test results, (2) compressive holds are more damaging than tensile holds, and (3) combined tension and compression holds of equal duration are more damaging than either tensile or compressive holds. Comparing the representative results of 2% strain range tests run at 538°C and conducted in air given in Table 2, it is apparent that these same trends are also true at the higher strain ranges with the exception that there was an indication that tensile holds were equally or slightly more damaging than compressive holds. Similarly, tests

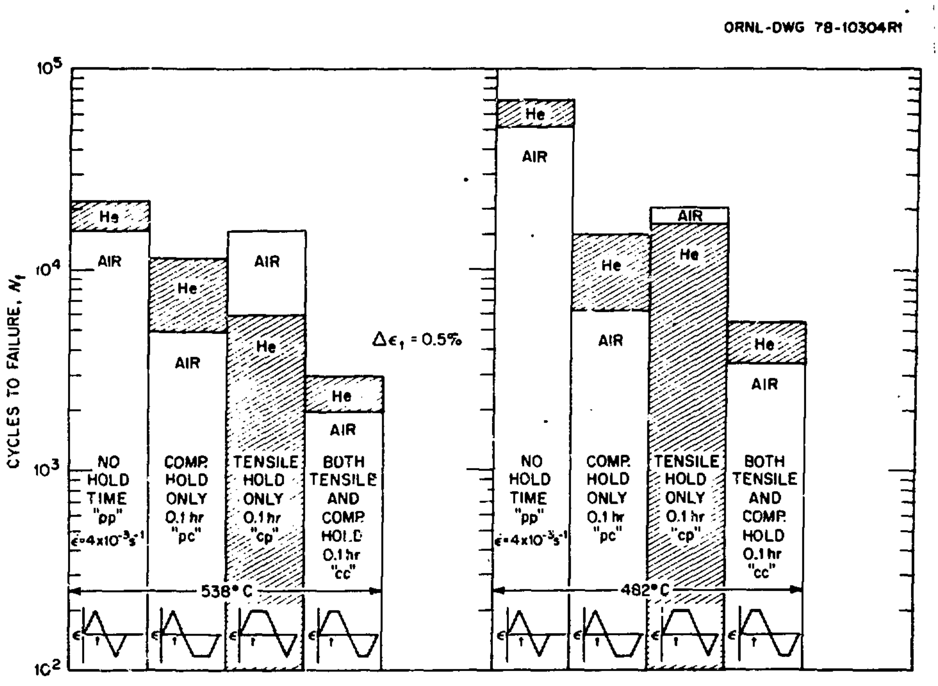

Fig. 6. Tests Conducted in He with Controlled Impurities Show Improved Cyclic Lifetimes for Several Waveforms in Comparison to Air. Tensile hold times only, however, appear to be an exception.

conducted in impure helium showed reductions when hold periods were introduced into each cycle as shown in Fig. 6. However, in contrast to the air data, the limited data suggested that tensile holds were slightly more damaging than compressive holds at 538°C as shown in Fig. 6 and Table 2. Accordingly, there is limited evidence that environment plays a role in determining the mode of the hold period that is most damaging.

There is an indication that mean stress development as a consequence of hold periods may also be important in explaining why one mode of hold is more damaging than another. Lord et al.[7] characterized the low-cycle fatigue and hold time behavior of cast Rene 80 and found that compressive hold times were more damaging than tensile hold times. This behavior was qualitatively expalined in terms of the ability of the material to sustain a mean stress at cycle lives beyond the transition fatigue life; that is, the elastic strain range was greater than the plastic strain range. These workers postulated that in this material compressive tensile mean stresses could be maintained that enhanced crack growth rates. For 2 1/4 Cr-1 Mo steel tested at 482°C, the cyclic transition life was around 10,000 cycles, such that specimens tested at a strain range of 0.5% or less clearly had elastic strain ranges in excess of the plastic strain ranges. In the case of this material tested at 538°C, the transition point occurs at about 10,000 cycles as well as at a total strain range of about 0.39%. If the mean stress postulate was entirely correct, one would expect to see hysteresis loops shifts as shown in Fig. 7. An examination of large data base[5,6] generated for this material indeed revealed that in nearly all cases when a tensile only hold period was imposed a small mean compressive stress developed.

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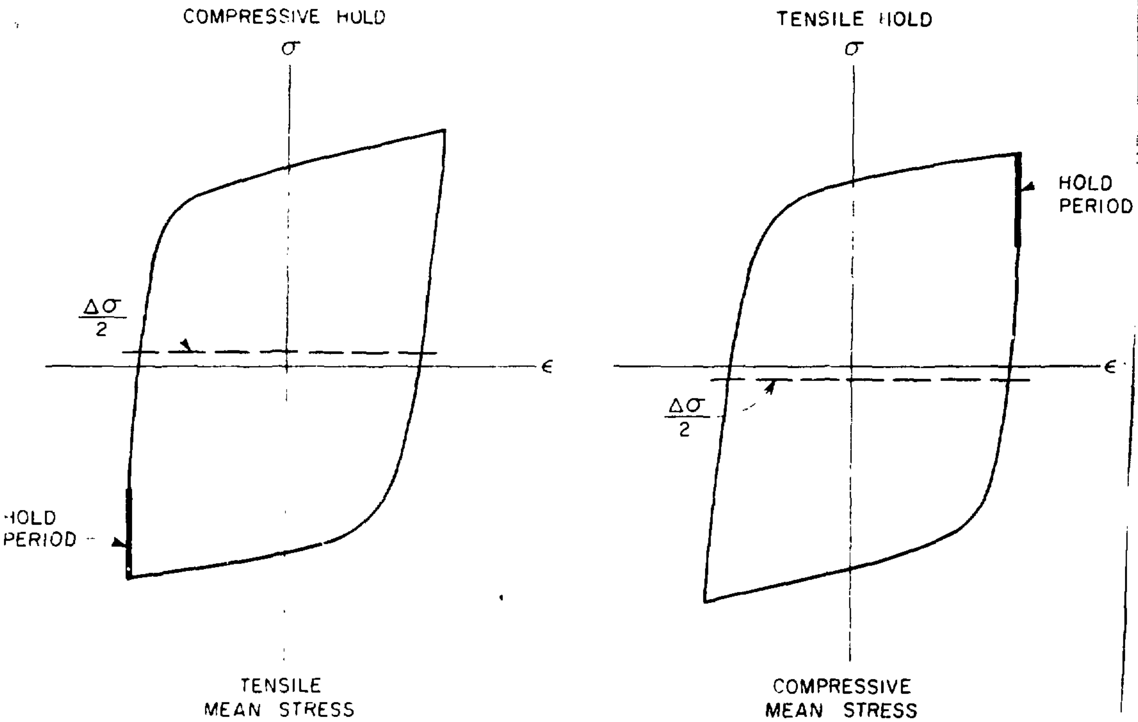

Fig. 7. Representative Hysteresis Loops Occurring During Elevated Temperature Strain-Controlled Fatigue Tests.

However, when a compressive hold only cycle was imposed it did not necessarily result in a mean tensile stress. That is, no consistancy was apparent.

Another possibility as to why compressive hold periods are more deleterious to cycle life than tensile hold periods for this material at low strain ragnes may be due to differences in surface deformation characteristics resulting from the different strain-time waveforms. Thus, compressive holds with resultant compressive relaxation may aggrevate surface nucleation of cracks. Since fatigue life generally becomes increasingly nucleation-rate dependent as the strain range is decreased, this would explain the increased sensitivity of fatigue life of this material to compressive hold periods at low strain ranges, i.e. below about 1%. Additional long-term tests in both air and helium as well as metallographic observations of fatigued specimens are presently underway, which hopefully will provide additional information concerning why compressive holds are more damaging than tensile holds at low strain ranges.

Various empirical methods, such as linear damage summation and strain range partitioning,[6] are being employed in an effort to extrapolate trends seen in the present data base to service times as long as 30 years. An example of the use of strain range partitioning for predicting the influence of compressive hold times of increasing duration on the hold time reduction factor (N_f^0/N_h) is shown in Fig. 8.

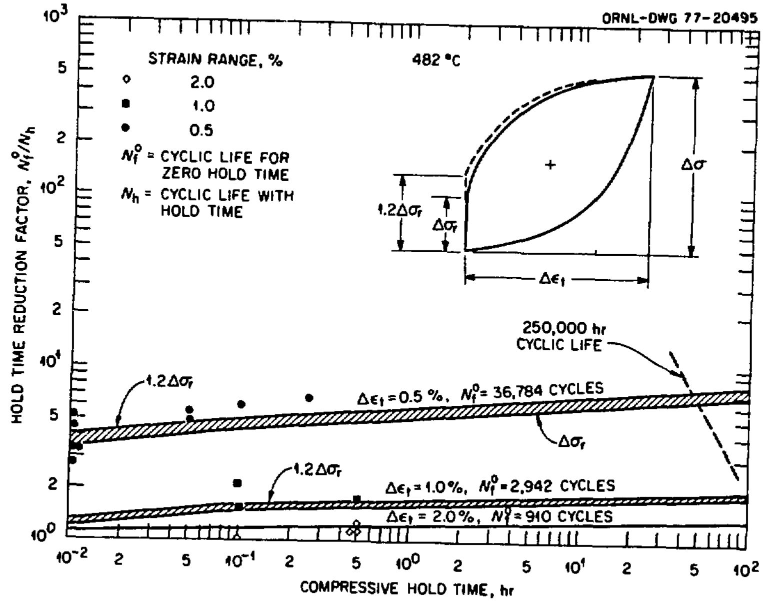

Fig. 8. Comparison of Measured and Predicted Hold-Time Reduction Factors for 2 1/4 Cr-1 Mo Steel Subjected to Compressive Hold Times at 482°C

The hold time reduction factor is simply the quotient of the cyclic life (N_f^0) for a test run at a given strain range and temperature without a hold time and the cyclic life (N_h) of a test run at the same strain range, temperature, and hold time indicated. The bands 1.2 $\Delta\sigma_r$ and $\Delta\sigma_r$ reflect different assumptions with respect to extrapolation as as shown by a schematic of the hysteresis loop in Fig. 8. The method shows promise, and additional generation of data in both air and impure helium is presently underway for further verification purposes.

Fatigue Crack Propagation

An extensive amount of work is currently underway to characterize the fatigue crack propagation behavior of primarily 2 1/4 Cr-1 Mo steel in the annealed condition.[8] Heat-to-heat variations are being investigated in commercially available air, electro-slag, and vacuum-arc remelted material. Further, the influence of various environments appropriate to Fast Breeder and High Temperature Gas-Cooled Reactors are also being determined.

The influence of temperature and frequency on essentially stage II crack growth behavior is shown in Figs. 9 and 10 for air-melted 2 1/4 Cr-1 Mo steel tested in air. The tests were performed with wedge-opening load test specimens. As the temperature increases into the creep range, the crack growth rate characteristically shows an increase (Fig. 9). Also, at temperatures within the creep range a decrease in frequency results in an increase in crack growth rate, as shown in Fig. 10.

The influence of temperature, frequency, stress intensity range (ΔK), and load or R ratio on stage II crack propagation rates in an air environment has been modeled by Booker.[8] The resultant equation is as follows:

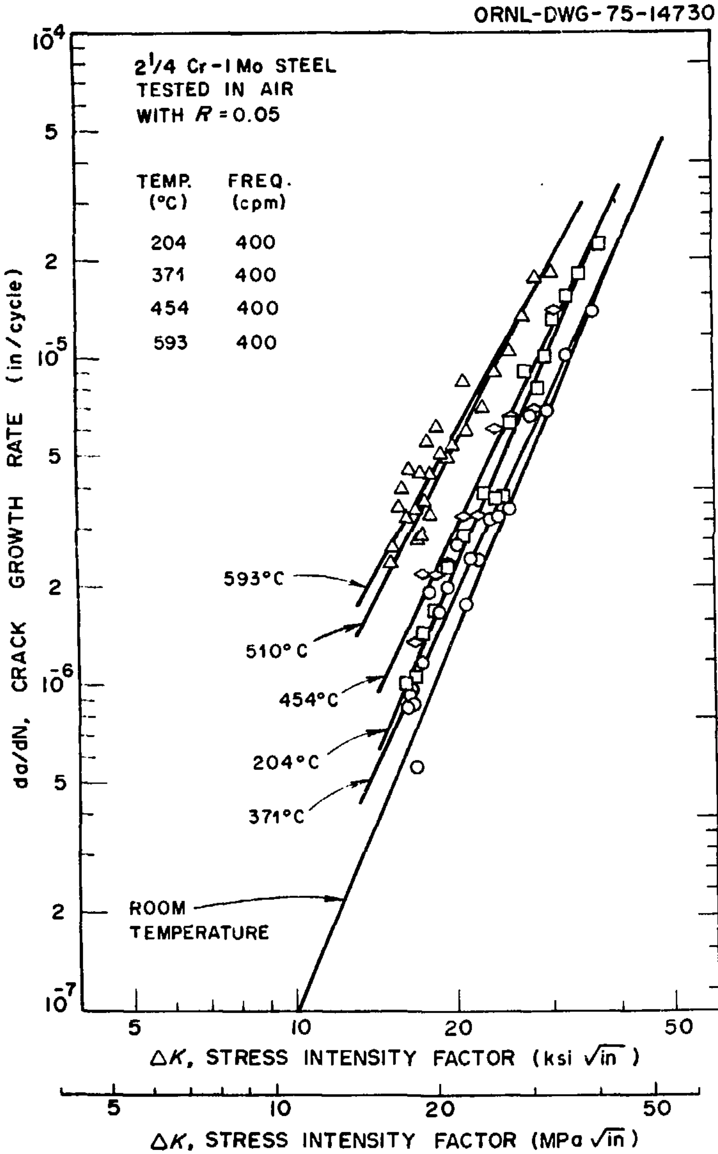

Fig. 9. Effects of Temperature on Crack Growth Rate.

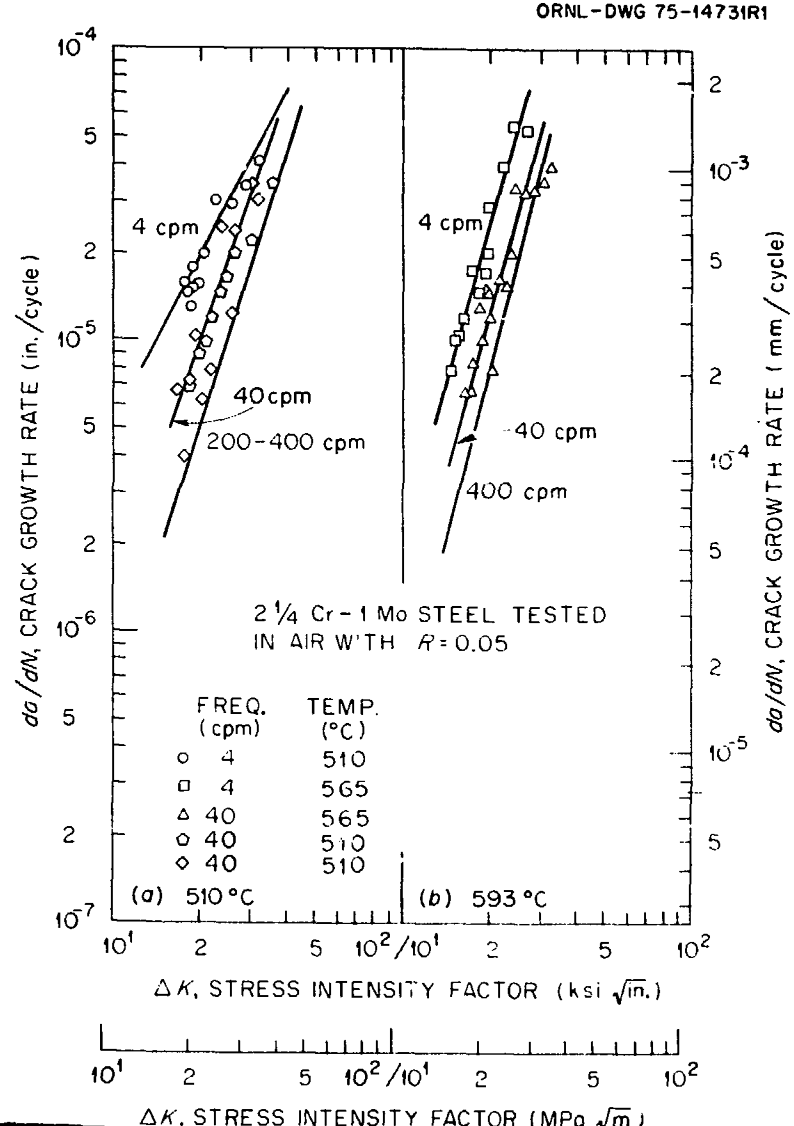

Fig. 10. Effects of Frequency at 510°C and 593°C.

$$\frac{da}{dN} = A_0(DK)^{3.5} , \qquad (1)$$

where

$\frac{da}{dN}$ = crack growth rate (mm/cycle),

$$DK(\text{MPa}\ \sqrt{\text{m}}) = \Delta K\ \sqrt{1 - R} , \qquad (2)$$

$$R = \frac{K_{\text{min}}}{K_{\text{max}}} = \frac{\text{minimum stress intensity}}{\text{maximum stress intensity}} , \qquad (3)$$

$$\Delta K = K_{\text{max}} - K_{\text{min}} . \qquad (4)$$

The exponent 3.5 in Eq. (1) was found to be essentially independent of temperature, but the influence of temperature and frequency was accounted for in coefficient A_0. The expression for A_0 was determined by generalized regression analysis of the available data between 371 and 593°C and is given as follows:

$$\log A_0 = -10.106 + 1.359 \times 10^{-5}T^2 - 1.432 \times 10^{-8}T^3$$
$$- 1.7 \times 10^{-9}T^3 \log \nu , \qquad (5)$$

where

ν = frequency, Hz, and

T = temperature, °C.

At room temperature no frequency effects were seen, A_0 being given by a constant value of 7×10^{-10}. Between room tempeature and 371°C, insufficient data were available to characterize A_0 as a function of ν and T. Current data support the validity of Eq. (1) over the following ranges of frequencies and DK values:

ν = 0.067 to 6.7 Hz from 371 to 482°C,

ν = 0.0067 to 5.0 Hz from 482 to 593°C,

DK = 10 to 70 MPa $\sqrt{m}$ at room temperature, and

DK = 10 to 50 MPa $\sqrt{m}$ from 371 to 593°C.

Fig. 11 schematically shows the limited range of ΔK over which Eq. (1) appears to be valid at room temperature.

Limited data available shows that environment is a particularly important consideration for crack growth behavior at temperatures

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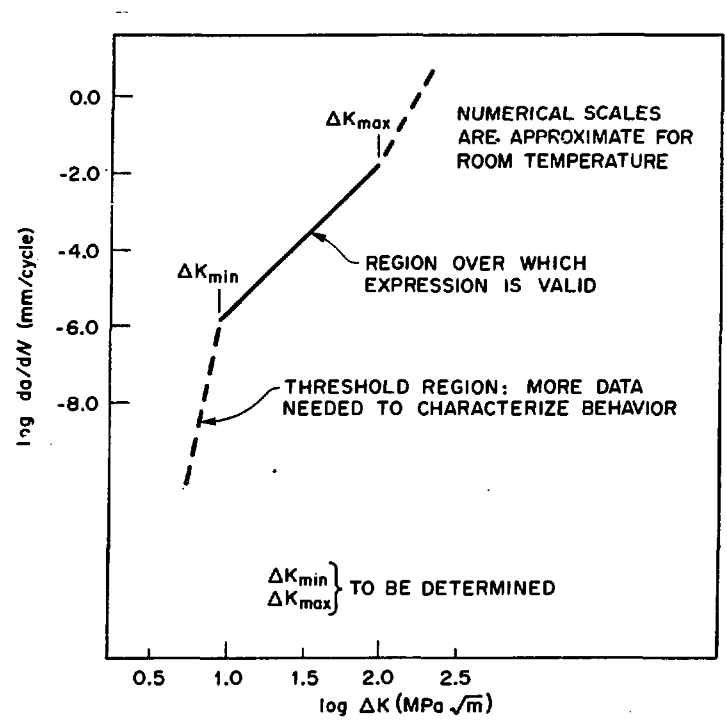

Fig. 11. The Current Expression is Valid Over a Limited Range of ΔK. More data are needed before low ΔK behavior is modeled.

within the creep range as shown in Fig. 12. Comparisons are given between crack growth rates in air, impure helium (Table 1), and vacuum (1×10^{-4} Pa or 1×10^{-6} torr) in Fig. 12. Low oxygen environments are clearly beneficial in retarding cyclic crack growth rates at a given ΔK.

Fatigue crack propagation studies are also being conducted on the 2 1/4 Cr-1 Mo steel or HAZ side of dissimilar metal joints. The joint involves 2 1/4 Cr-1 Mo steel and ERNiCr-3 weld metal, a nickel base alloy with approximately 20% Cr, several percent Fe, Mn, and Nb, and lesser amounts of several other elements. Weldments involving these materials are expected to be used in the United States in both HTGR and LMFBR systems.

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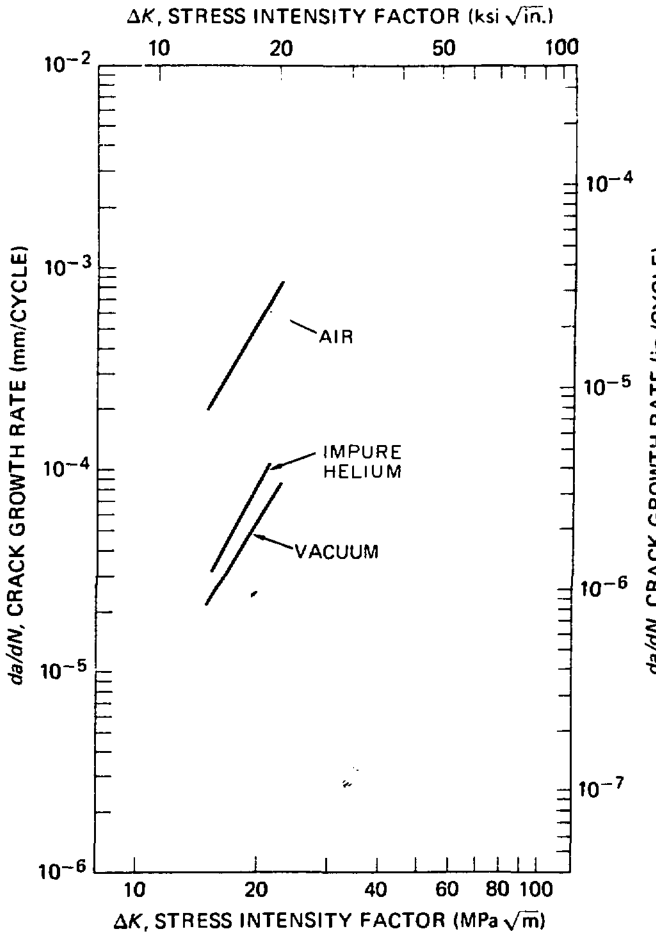

Fig. 12. Comparison of the Fatigue Crack Growth Rate of 2 1/4 Cr-1 Mo Steel in Air, Simulated HTGR Primary Coolant Helium, and Vacuum at 593°C and 0.67 Hz with R = 0.05.

Historically, some weldments involving ferritic and austenitic metal have failed after prolonged service at elevated temperatures. An example of such a weldment taken from a coal-fired utility boiler after a service life of 17 years is shown in Fig. 13.[9] The weld metal in this case was Inconel 132, a nickel-base alloy with approximately 16% Cr, 9% Fe, and lesser amounts of several other elements. Fig. 13 shows that crack propagation did occur in the HAZ of the 2 1/4 Cr-1 Mo steel in an area adjacent to and in the fusion line.

Comparisons are shown in Fig. 14 of crack propagation rates between 2 1/4 Cr-1 Mo Steel, ERNiCr-3 weld metal,[10] and HAZ material. ERNiCr-3 weld metal, AWS A5.14 commonly known as Inconel 82, has a nominal composition of 67% Ni, 20% Cr, 3% Mn, 2.5% Nb, and lesser amounts of several other elements. It is a prime candidate for use in CRBRP as well as HTGRs for dissimilar metal joints. The region that is of particular concern is the HAZ adjacent to the fusion line on the 2 1/4 Cr-1 Mo side. Accordingly, composite specimens were designed which would emphasize crack growth behavior in this region. The composite HAZ specimens were prepared from welded plates such that the fusion line was perpendicular to the large surface of the specimen as shown in Fig. 14. Welding was performed by the automatic gas tungsten-arc process with hot-wire filler additions. Prior to machining of the specimens the weldments were given a 1-h stress relief anneal at 732°C. Current comparisons between crack propagation behavior in the HAZ material in comparison to annealed 2 1/4 Cr-1 Mo or all weld metal (ERNiCr-3) material showed no unusual behavior. Fig. 15 shows the tip of a crack that propagated in the HAZ of one of the composite specimens with no side branching or intergranular cracking. Composite material is now undergoing thermal aging for prolonged periods of time at

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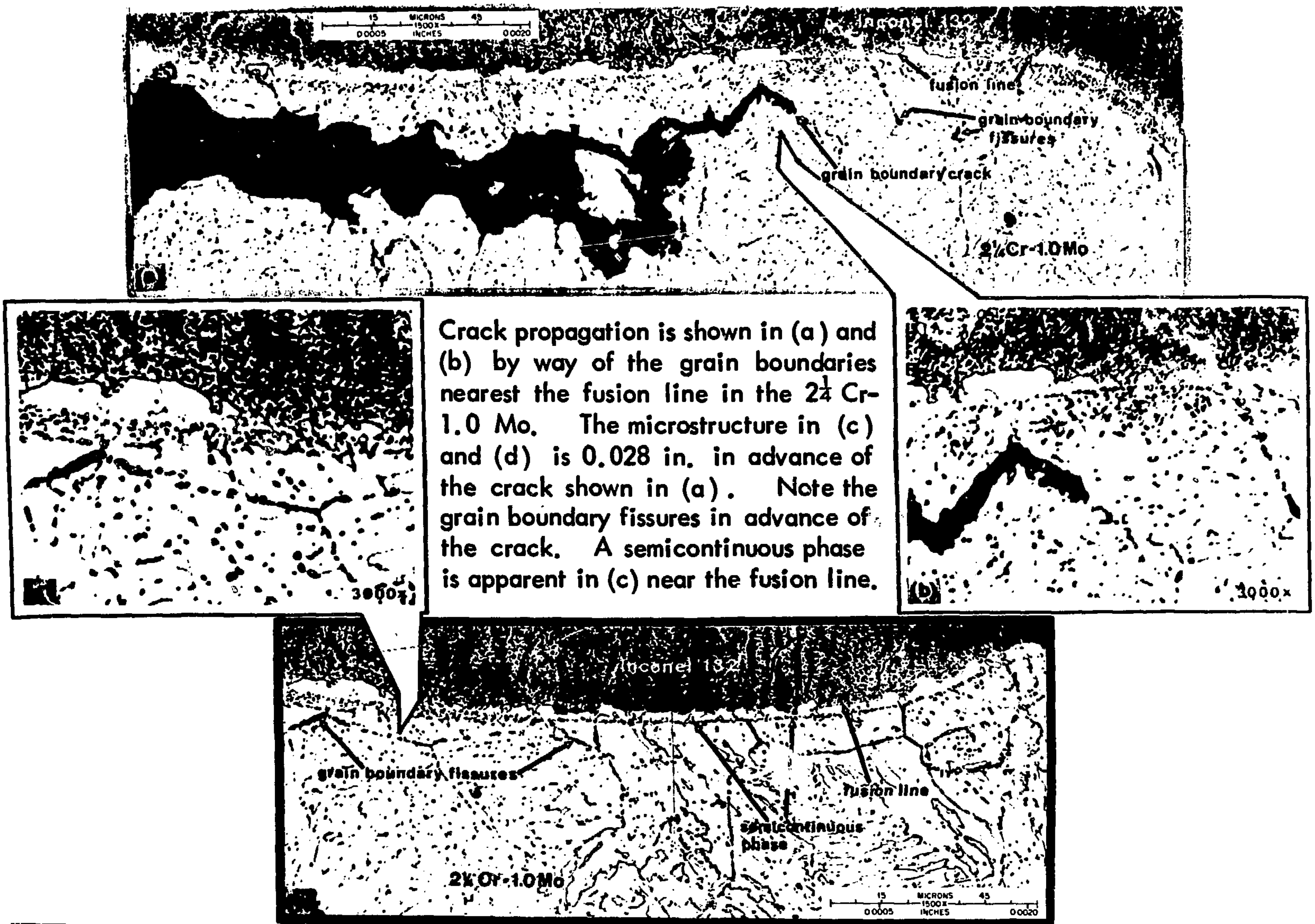

Fig. 13. Failed Transition Joint: Crack Propagation is Concentrated in Semicontinuous Phase Near Fusion Line.

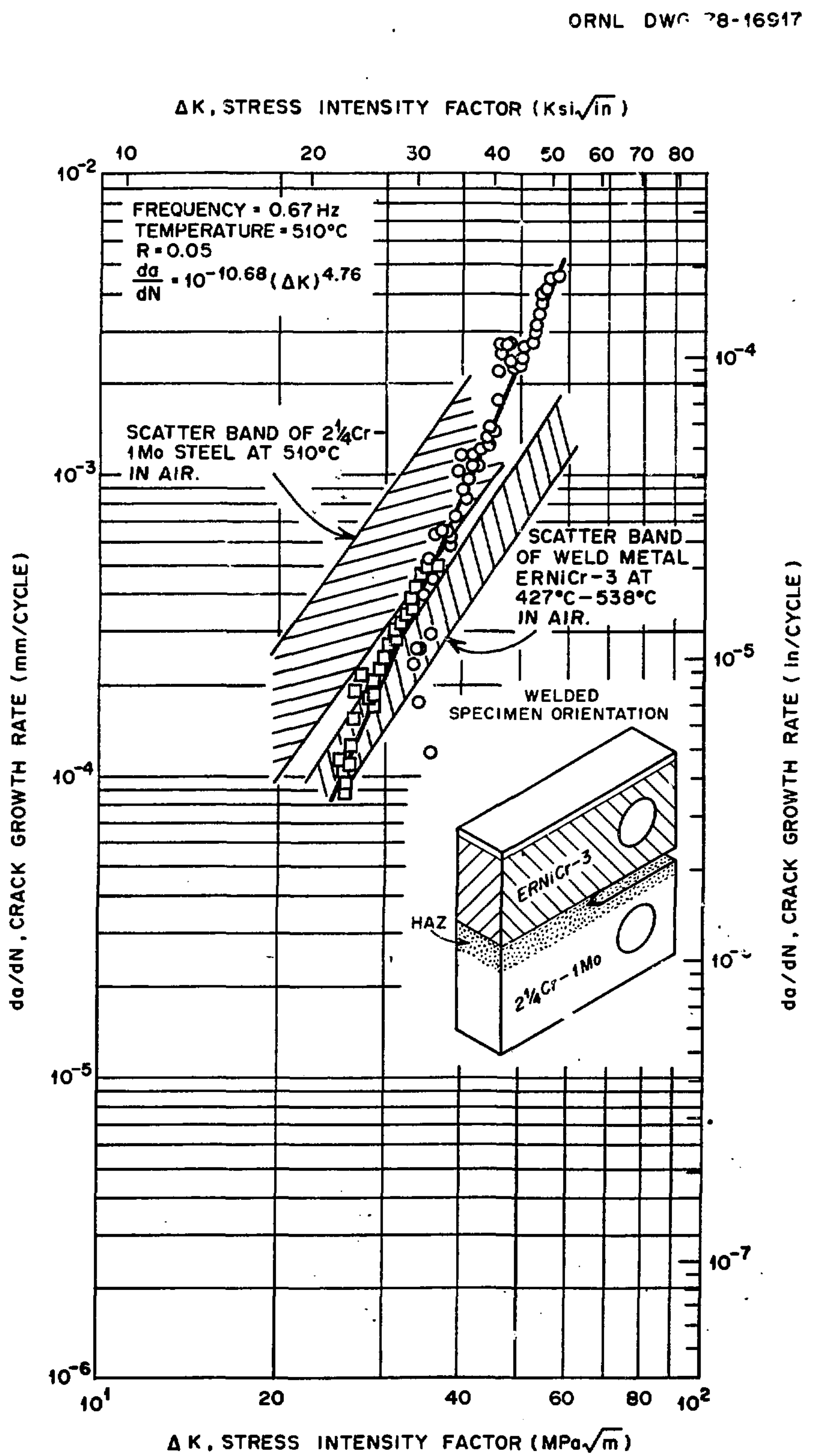

Fig. 14. Comparison of the Elevated-Temperature Crack Propagation Rates of Several Materials.

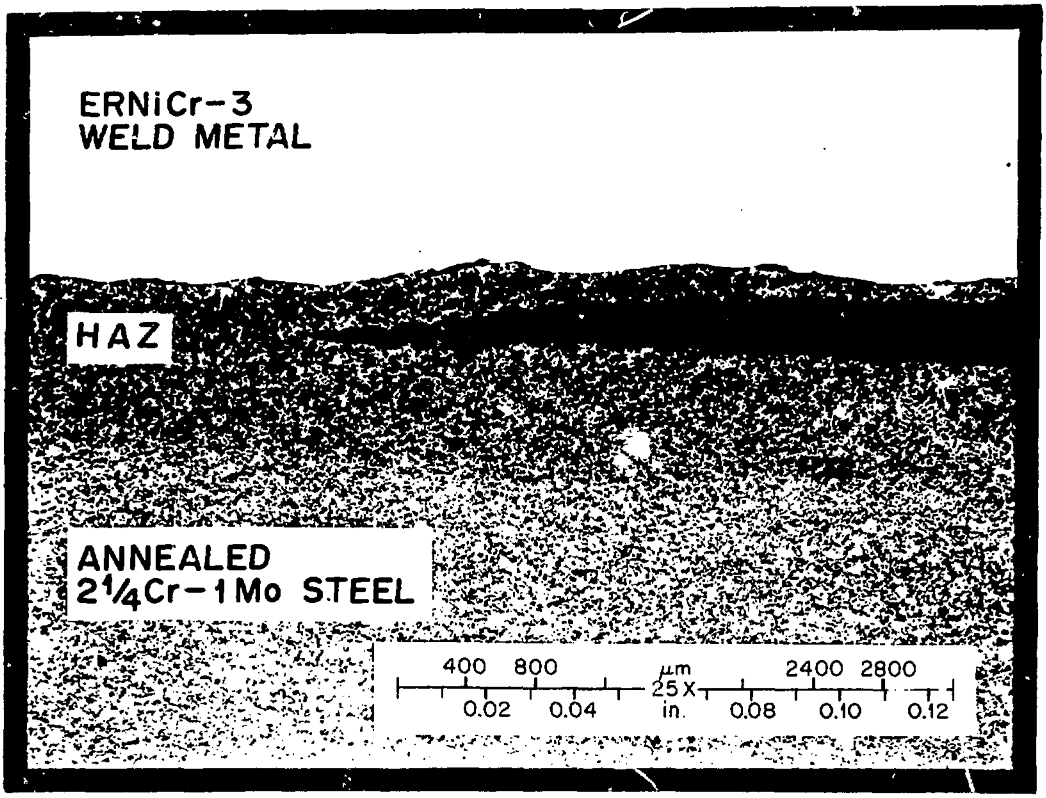

Fig. 15. Photomicrograph of the Surface of a Composite Crack Propagation Specimen of ERNiCr-3 and 2 1/4 Cr-1 Mo Steel. Crack is propagating in the HAZ.

elevated temperatures in preparation for additional crack growth studies. It is expected that these tests will determine if any degradation occurs as a result of this pre-exposure treatment. Results will be presented in subsequent reports.

CONCLUSIONS

1. Low cycle rain-controlled fatigue data comparisons indicated that variations in melt practice and annealing heat treatment did not significantly alter cyclic fatigue life. However, heat-to-heat variations were noted in the high cycle fatigue life for specimens tested at 427°C.
2. Comparisons of continuous cycle fatigue data generated in air and impure helium indicated that the fatigue lives of specimens tested in impure helium were equal to or exceeded that of specimens tested in air. However, when strain holds were introduced into each cycle of specimens tested at temperatures within the creep range, there was an indication that environment could be important in determining the most damaging strain-time waveform.
3. An expression was given for stage II cyclic crack growth behavior over the temperature range of room temperature to 593°C.
4. Comparisons in crack growth rate data were given for annealed 2 1/4 Cr-1 Mo steel, ERNiCr-3 weld metal, and HAZ material in 2 1/4 Cr-1 Mo steel adjacent to the weldment fusion line. HAZ material generally showed crack growth rates at 538°C intermediate between those of annealed 2 1/4 Cr-1 Mo steel and the weld metal.

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