ORNL/TM-8392

ornl

OAK
RIDGE
NATIONAL
LABORATORY

UNION
CARBIDE

Response of Bainitic 2 1/4 Cr–1 Mo Steel to Creep-Fatigue Loadings at 482°C

R. W. Swindeman
J. P. Strizak

Fossil
Energy
Program

OPERATED BY
UNION CARBIDE CORPORATION
FOR THE UNITED STATES
DEPARTMENT OF ENERGY

Printed in the United States of America. Available from
National Technical Information Service
U.S. Department of Commerce
5285 Port Royal Road, Springfield, Virginia 22161
NTIS price codes—Printed Copy: A03 Microfiche A01

ORNL/TM-8392

METALS AND CERAMICS DIVISION

RESPONSE OF BAINITIC 2 1/4 Cr-1 Mo STEEL TO CREEP-FATIGUE LOADINGS AT 482°C

R. W. Swindeman and J. P. Strizak

Date Published - November 1982

Notice: This document contains information of a preliminary nature. It is subject to revision or correction and therefore does not represent a final report.

Prepared by the
OAK RIDGE NATIONAL LABORATORY
operated by
UNION CARBIDE CORPORATION
for the
U.S. DEPARTMENT OF ENERGY
Under Contract No. W-7405-eng-26

CONTENTS

RESPONSE OF BAINITIC 2 1/4 Cr-1 Mo STEEL TO CREEP-FATIGUE LOADINGS AT 482°C*

R. W. Swindeman and J. P. Strizak

ABSTRACT

Creep-rupture, fatigue, and creep-fatigue tests were performed on 150-mm plate of bainitic 2 1/4 Cr-1 Mo steel at 482°C. The purpose was to produce exploratory data to assess the sensitivity of the material to time-dependent damage processes. Two material conditions were examined — quenched and tempered (QT) for 6 h at 662°C and simulated postweld heat treated (PWHT) for 28 h at 690°C.

Continuous-cycling fatigue data for the PWHT material revealed better fatigue resistance than reported in the literature for this material. This difference may be an effect of specimen geometry. Under creep-fatigue conditions, with relaxation hold times in the range 0.01 to 0.5 h, the fatigue strength was substantially reduced. The PWHT material exhibited slightly poorer fatigue properties under tensile relaxation holds than under compression holds at the 1% strain range level. This behavior pattern is the opposite of the trend reported for annealed 2 1/4 Cr-1 Mo steel and suggests that the mechanism of damage accumulation in this class of material may depend on strain range and microstructure as well as environmental conditions. However, the failure mode for bainitic material was similar to that reported for ferritic (annealed) material.

INTRODUCTION

This paper presents data bearing on the sensitivity of bainitic (normalized and tempered) 2 1/4 Cr-1 Mo steel to creep-fatigue loadings. The purpose was to aid in the determination of whether creep-fatigue data are needed in the design and analysis of dissolver vessels for coal liquefaction plants. Such vessels are expected to operate at temperatures to 480°C and undergo thermally or mechanically induced strain cycles associated with normal startup and shutdown procedures as well as anticipated upset conditions.

*Research sponsored by the Solvent Refined Coal Projects Office, U.S. Department of Energy/Oak Ridge Operations, under contract W-7405-eng-26 with the Union Carbide Corporation.

In spite of the extensive use of bainitic 2 1/4 Cr-1 Mo steel in the petrochemical industry,[1] very little information is available in the open literature concerning the creep-fatigue properties. Continuous-cycling data have been reported by the National Research Institute for Metals (NRIM) in Japan.[2] Here, data obtained on tubing are provided for temperatures in the range 25 to 600°C and at several strain rates. A significant strain rate effect was observed at high temperatures; it implied the existence of a creep-fatigue or time-dependent fatigue effect. In the United States, the alloy was included in a Metals Properties Council program[3] on creep-fatigue interspersion testing. All tests were at 538°C, but several strain ranges and creep stresses were examined. In support of the program, control data consisting of continuous-cycling fatigue and constant-stress creep-rupture were obtained by Jaske and Mindlin.[4] The continuous-cycling fatigue data reported by NRIM and Jaske and Mindlin indicated considerably less fatigue resistance than observed for ferritic (annealed) material by Ellis et al.[5] Further, the interspersion test data on the same heat of material indicated that a deleterious creep-fatigue interaction was more pronounced in bainitic material than in ferritic material. These differences suggested that the creep-fatigue damage interaction diagram developed for ferritic material may not apply for bainitic material. It was partly to explore this hypothesis that our experimental work was undertaken.

MATERIAL

Specimens were machined from a 150-mm plate of 2 1/4 Cr-1 Mo steel provided to us by the Lukens Steel Company. The heat (A6660) was melted in an electric furnace, vacuum degassed, and calcium treated. Part of the heat was produced as 300-mm plate and characterized by Swift[6] in terms of composition, microstructure, tensile properties, and toughness. The ladle analysis was as follows: C, 0.13%; Mn, 0.33%; P, 0.012%; Si, 0.05%; Cr, 2.34%; Mo, 0.99%; and Al, 0.004%.

The material was received in what we will call the quenched and tempered condition (QT). The specific history was as follows: 6 h at 927°C, accelerated cooled, tempered for 6 h at 662°C. This temperature was below the minimum tempering temperature (690°C) required to meet SA-387, Grade 22, Class 2. Metallographic examination revealed a fully bainitic microstructure with a prior austenitic grain size near 80 µm. The tensile properties of this material are reported in Table 1. The strengths fell near the upper limits expected for SA-387, Grade 22, Class 2, as determined by Booker et al.[7]

Specimens were machined from the quarter and three-quarter thickness locations and half were subjected to a simulated postweld heat treatment (PWHT) of 28 h at 690°C. This PWHT did not change the microstructure observed optically at magnifications up to 1000× but decreased the strength properties to levels around average for Class 2 material (see Table 1).

Table 1. Tensile properties of 2 1/4 Cr-1 Mo steel 150-mm plate

Condition[a]	Temperature (°C)	Strain rate (s^{-1})	0.2% Yield (MPa)	Ultimate strength (MPa)	Uniform strain (%)	Elongation (%)	Reduction of area (%)
QT	25	6.7 E-5	545	659	6	24	71
QT	482	6.7 E-5	448	507	4	27	72
PWHT	25	6.7 E-5	441	558	8	26	59
PWHT	427	6.7 E-5	376	459	6	25	61
PWHT	482	6.7 E-6	354	393	2.5	27	60
PWHT	482	6.7 E-5	365	419	3.2	26	62
PWHT	482	6.7 E-4	352				
PWHT	538	6.7 E-5	328	392	1.2	30	64

[a]QT = 6 h at 662°C, PWHT = 6 h at 662°C + 28 h at 690°C.

Threaded-end bar specimens were used for tensile, relaxation, and creep-rupture tests. The reduced section was 32 mm and its diameter 6.3 mm. Tensile tests conformed to ASTM recommended practice E 21, and creep-rupture tests conformed to ASTM E 138. Buttonhead specimens, described in ASTM recommended practice E 606, were used for fatigue testing. In most instances, we tested the hourglass-type specimen. Here the reduced section had a radius of 25 mm and a minimum diameter of 5.0 mm. We also tested a few uniform-gage specimens under fatigue loading. Here the reduced section was 19 mm and its diameter 6.3 mm.

RESULTS

Since the emphasis of the work was on the creep-fatigue behavior, the tensile, creep, and relaxation tests were intended to produce what we judged to be the minimum of characterization data. Table 1 and Fig. 1 provide the tensile data, as mentioned earlier, and Table 2 and Fig. 2 provide creep-rupture data. Similar to the tensile test results, the creep-rupture strength of the QT material was greater than the strength of PWHT material. This correlation between short-time tensile and creep rupture has been reported by Smith[8] and Booker.[9]

Typical relaxation data are provided in Fig. 3. These curves agree fairly well with the relaxation curves produced by other cyclic tests and were used in the analysis of the creep damage component in creep-fatigue tests.

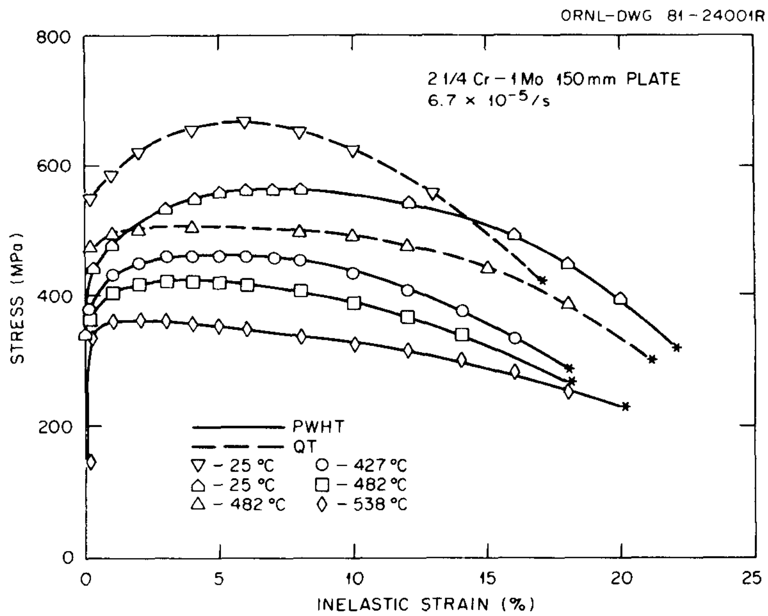


Fig. 1. Tensile curves for 2 1/4 Cr-1 Mo steel quenched and tempered and postweld heat-treated.

Table 2. Creep-rupture data for 2 1/4 Cr-1 Mo steel at 482°C

Condition[a]	Stress (MPa)	Loading strain (%)	Secondary creep rate (%/h)	Rupture life (h)	Elongation (%)
QT	380	0.4	4.5 E-3	638	21
QT	345	0.27	2.3 E-3	1627	19
PWHT	380	0.43	6.6 E-2	33.7	8
PWHT	345	0.46	1.1 E-1	24.4	21
PWHT	276	0.27	4.8 E-3	638.7	24
PWHT	207	0.27	2.7 E-4	>3000	

[a]QT = 6 h at 662°C, PWHT = 6 h at 662°C + 28 h at 690°C.

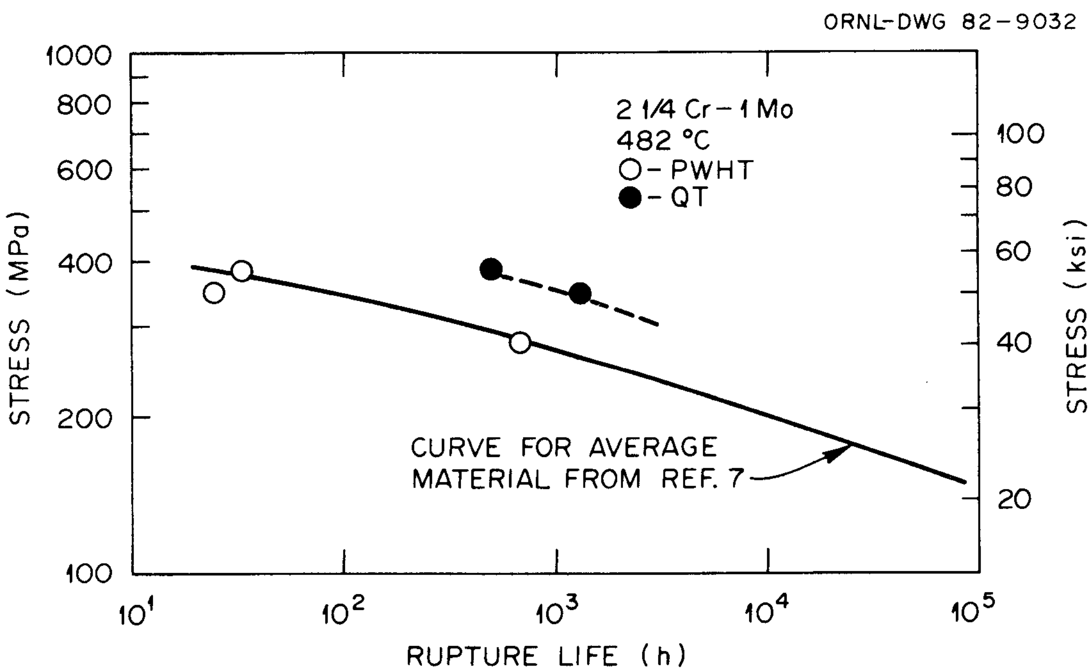


Fig. 2. Creep-rupture curves for 2 1/4 Cr-1 Mo steel.

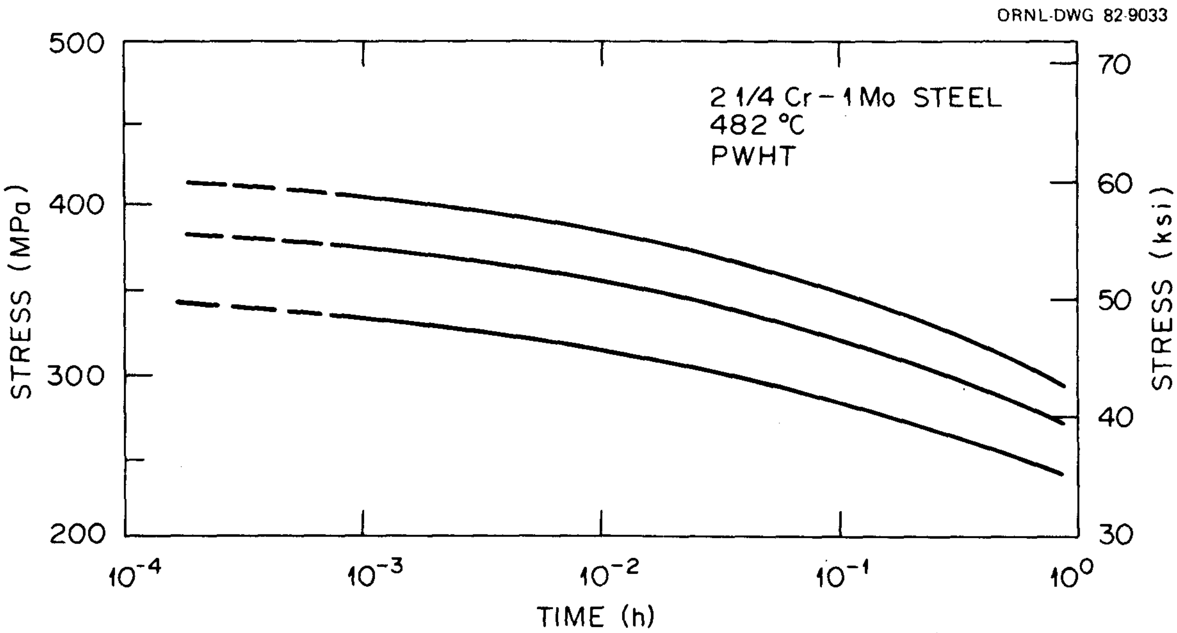


Fig. 3. Typical relaxation curves obtained from cyclic tests at 482°C on postweld heat-treated material.

Results from continuous-cycling fatigue tests on PWHT material are provided in Table 3. These cover strain ranges ($\Delta\varepsilon$) from 0.35 to 2% and cyclic lives from 600 to 500,000. The reported stress ranges ($\Delta\sigma$) represent values at half life. In all cases, we observed cyclic softening; hence the stress range data in Table 3 do not represent the maximum observed in each test. Information on the stress range and the cycle dependence of softening is contained in Fig. 4, where $\Delta\sigma$ is plotted

Table 3. Continuous-cycling fatigue data for 2 1/4 Cr-1 Mo steel 150-mm plate at 482°C

Strain range (%)	Tensile stress (MPa)	Stress range (MPa)	Cycles to failure
2.0	355	724	645
1.0	314	646	2,465
0.6	265	552	24,554
0.45	243	504	56,366
0.35	234	476	500,145

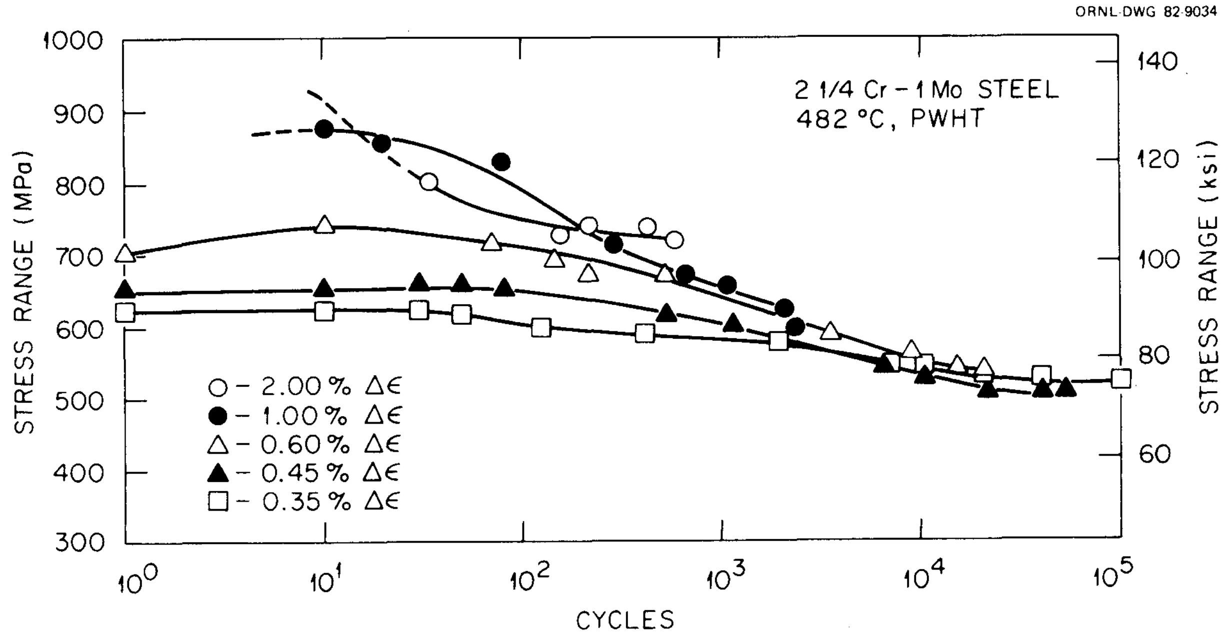


Fig. 4. Stress range vs log cycles for continuous-cycling fatigue tests on postweld heat-treated at 482°C.

against cycles (N) for five continuous-cycling tests. The high-strain-range tests exhibited relatively more softening than did the low-strain-range tests. In fact, the Δs vs N curves tended to converge at high cycles.

The log Δe vs log cyclic life (N_f) data revealed some interesting trends when compared with literature data. These are shown in Fig. 5. Our continuous-cycling data, obtained on hourglass samples, are plotted as filled circles. These data fell close to a trend (solid curve in Fig. 5) observed by Ellis et al.[5] in continuous-cycling tests of ferritic (annealed) material at 482°C. The curve represents data for hourglass specimens tested at fairly high strain rates (1 to 4 × 10^{-3}/s).

The most extensive work on bainitic material, on the other hand, is represented by the dashed curve in Fig. 5. This curve represents the trend for data obtained by NRIM at 500°C on uniform-gage specimens with a strain rate of 10^{-3}/s. The cyclic life observed by NRIM was about half the life that we observed for bainitic material (hourglass specimens). Open symbols are also plotted in Fig. 5. These symbols represent other test data for uniform gage specimens and include two tests by Ellis et al.[5]

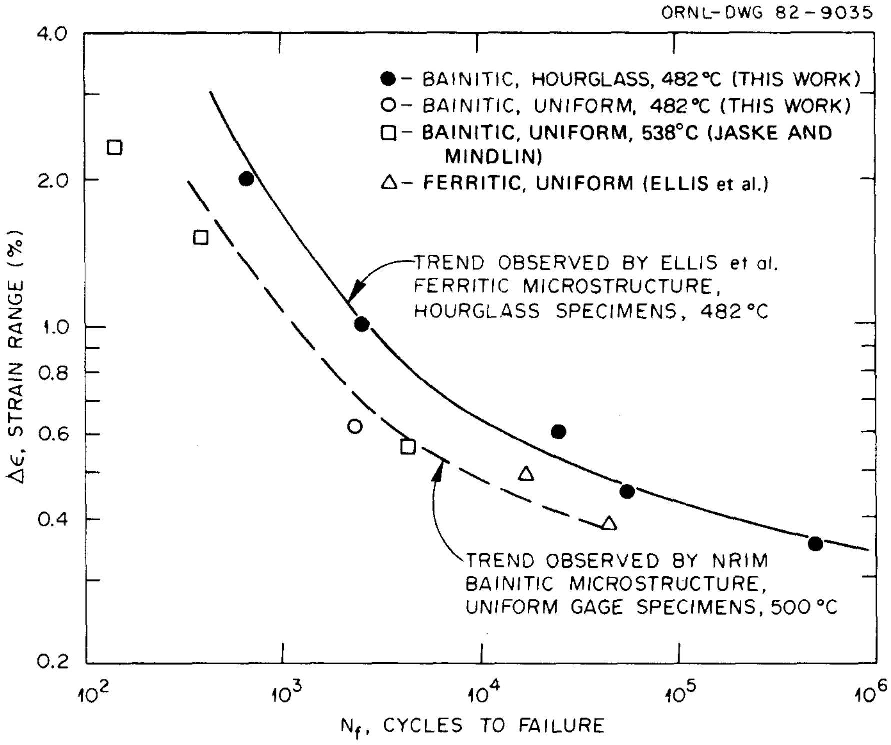


Fig. 5. Continuous cycling-fatigue data from tests on postweld heat-treated material at 482°C.

on ferritic material at 482°C and 4×10^{-3}/s strain rate, three tests by Jaske and Mindlin[4] on bainitic material at 538°C, and one test that we performed on bainitic material at 482°C and a strain rate of 10^{-4}/s. All the data for uniform-gage specimens fall short of the curve for hourglass specimens by a factor of 2 or more in cyclic life. To us, this indicated that the specimen geometry was more important than metallurgical structure in affecting the fatigue life at constant strain range and temperature. We did not perform any continuous cycling tests on the material in the QT condition. However, from the data for QT material reported by Jaske and Mindlin, it appears unlikely that the continuous-cycling behavior in the QT condition would be much different from that in the PWHT condition.

The data for tests involving relaxation hold times on hourglass specimens are provided in Table 4. The hold time was at either the maximum tensile or compressive strain amplitude and ranged from 0.01 to 0.5 h duration. Similar to the behavior under continuous cycling, the hold time tests exhibited softening from the first cycle onward. Data trends are

Table 4. Relaxation-hold fatigue data for 2 1/4 Cr-1 Mo steel

Condition	Strain range (%)	Hold time[a]	Stress range (MPa)	Peak stress (MPa)	Relaxed stress (MPa)	Cycles to failure	Fatigue damage	Creep damage
QT	2.0	0.1 T	885	407	310	365	0.57	0.027
QT	2.0	0.1 C	882	−447	−345	420	0.66	0.093
QT	2.0	0.5 T	855	407	276	344	0.54	0.041
QT	2.0	0.5 C	855	−434	−310	455	0.71	0.21
QT	1.0	0.5 T	745	345	248	1239	0.52	0.12
QT	1.0	0.5 C	738	−359	−276	1367	0.57	0.18
PWHT	2.0	0.1 T	745	359	248	324	0.50	0.117
PWHT	2.0	0.1 C	724	−352	−276	483	0.75	0.142
PWHT	2.0	0.01 C	717	−359	−303	581	0.90	0.212
PWHT	1.0	0.5 T	586	276	193	1089	0.44	0.08
PWHT	1.0	0.5 C	586	−290	−214	1375	0.33	0.06

[a]T, tensile; C, compressive.

summarized in Figs. 6 and 7. We observed that the stress range was the same regardless of whether tensile or compressive hold times were introduced. Further, hold time had little or no influence on the stress range. In several instances the starting stress and relaxed stress were greater (in an absolute sense) for the compressive hold. This had an effect on the calculated creep damage during the hold period, as will be seen later.

The fatigue lives under hold time conditions were about one-half the continuous cycling lives (see Fig. 8). In every case the tensile hold produced a shorter life than the compressive hold. Ellis et al.[5] and Brinkman et al.[10] found a similar trend for ferritic material at the 2% strain range level. However, at the 1% strain range, and especially below 1%, compressive hold times were generally observed to be more damaging.[10–12] At present we lack evidence to suggest that bainitic material will behave differently from ferritic at 0.5% $\Delta\epsilon$. However, we plan to do more testing at the lower strain ranges to explore this possibility.

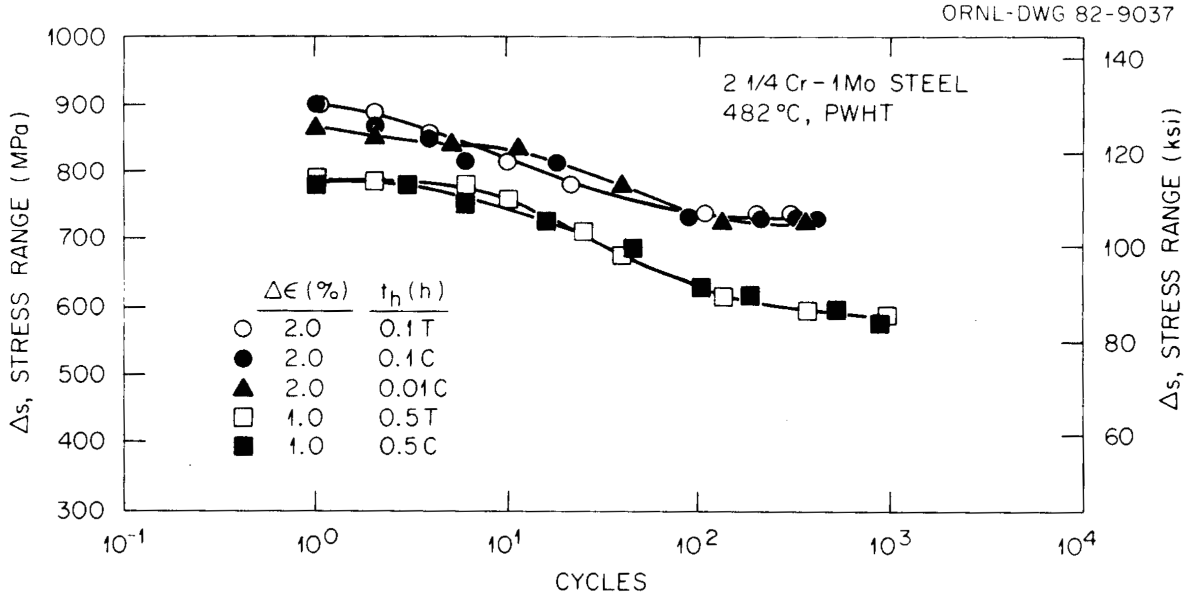


Fig. 6. Stress range vs log cycles for tests on postweld heat-treated material with hold time at the tensile or compressive strain limit.

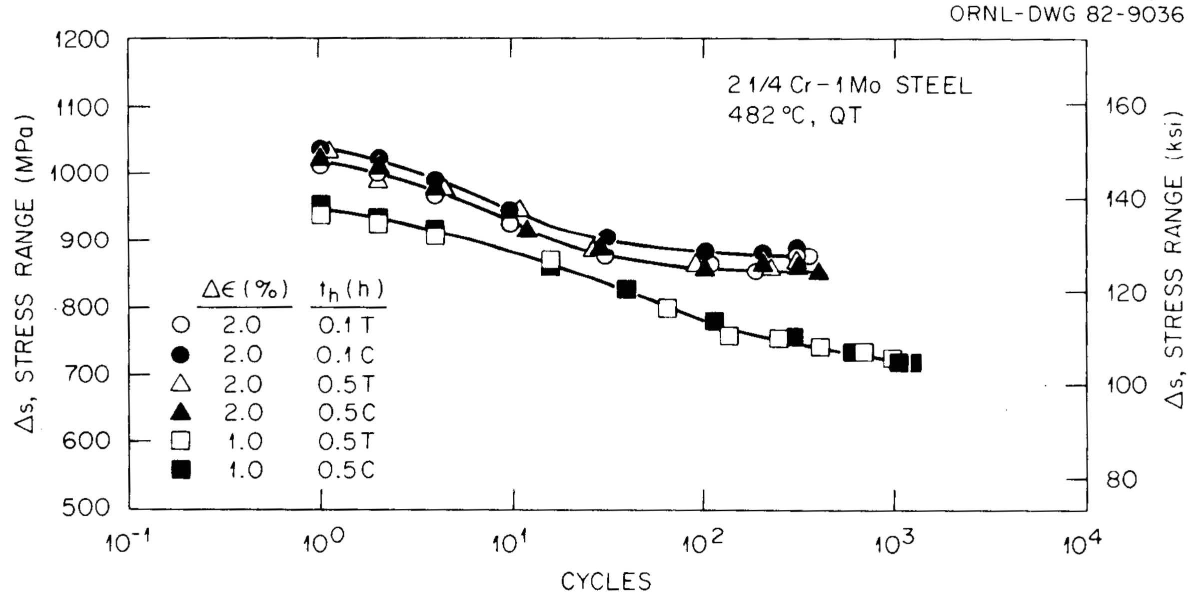


Fig. 7. Stress range vs log cycles for tests on quenched and tempered material with hold time at the tensile or compressive strain limit.

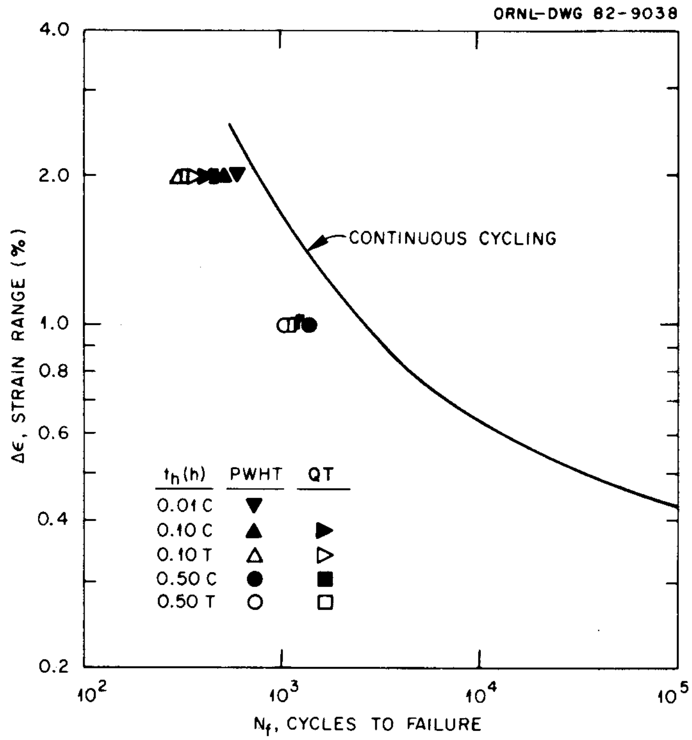


Fig. 8. Comparison of hold time strain fatigue with the curve for continuous cycling at 482°C.

Creep and fatigue damage ratios were calculated for the relaxation hold tests by the method of Campbell.[13] Here, the creep damage was based on the stress versus time behavior during relaxation and the stress versus life curve obtained from constant-stress creep-rupture data. Thus, for a single relaxation period

$$\delta_c = \sum_{0}^{t_h} dt_i/t_{R_i} \quad ,$$

where δ_c is the creep damage in the cycle, t_h is the hold time per cycle, dt_i is the time at a given stress (σ_i) during the relaxation period, and t_{R_i} is the time to failure at that stress. The total creep damage, D_c, was obtained by assuming that δ_c did not change from cycle to cycle, and damage at half-life was representative of the entire test:

$$D_c = N_f \delta_c \quad .$$

The fatigue damage ratio, D_f, was simply the number of cycles to failure under hold time conditions (N_f) divided by the number of cycles to failure under continuous cycling conditions at the same total strain range (N_{f_c}):

$$D_f = N_f/N_{f_c} \ .$$

The D_c and D_f data for 11 tests are provided in Table 4. The "damage interaction" diagram is plotted in Fig. 9 for D_f versus D_c. Curves included in the figure represent (1) no interaction, (2) linear damage interaction, and (3) severe interaction of the type observed by Brinkman et al. for ferritic material. Clearly, the tensile hold time data fell closest to the severe interaction curve, but most of the compressive hold time data fell between this curve and the linear interaction damage curve. In only one case did the damage summation exceed unity. This fact is in contrast to the summations reported by others,[5,10,11] in which many of the tensile hold time tests produced summations exceeding unity.

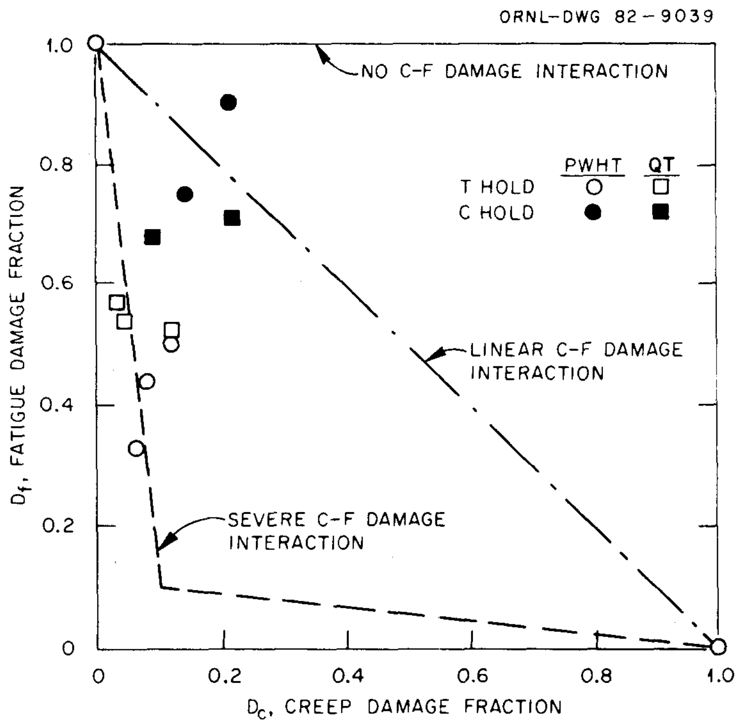


Fig. 9. Creep-fatigue damage interaction data for bainitic 2 1/4 Cr-1 Mo steel at 482°C.

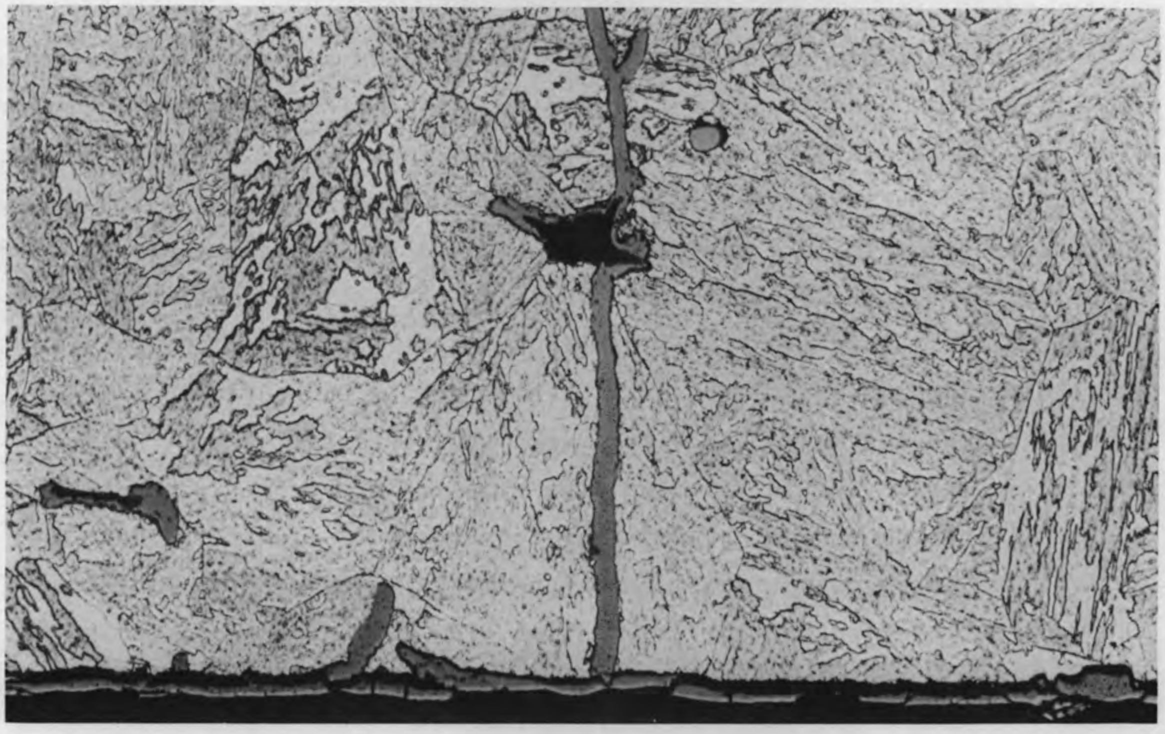
METALLURGICAL INVESTIGATION

The tensile and creep specimens exhibited cup-and-cone failures indicative of a plastic instability failure mode. Metallography at magnifications up to 1000× revealed transgranular failure and no evidence of cavity formation on prior austenite grain boundaries. Examination of the fatigue specimen failure surfaces was hampered by the presence of a heavy oxide. This was particularly the case near the failure initiation sites. Away from the initiation site striations indicative of a transgranular ductile failure mode were present. Metallography on sectioned fatigue specimens revealed many secondary cracks that were filled with oxide. Typical cracks moving in from opposite sides of a specimen are shown in Fig. 10. The transgranular nature of the cracks is clearly evident as well as the heavy oxidation progressing inward. The general pattern of the fatigue cracking and its association with surface oxide cracking was similar to the pattern observed by Brinkman et al.[10]

DISCUSSION

Although few, the exploratory creep-fatigue test data described in this paper raise some interesting questions. Of particular concern is the resolution of the specimen geometry effect. Uniform-gage specimens seem to have shorter fatigue lives than hourglass specimens at the same nominal strain range. We know that the bainitic steels have very little strain hardening capability, as indicated by the trends shown in Figs. 1, 4, 6, and 7. If true, then strain concentration within the uniform gage section is possible and could lead to earlier fatigue crack initiation in the uniform-gage specimen than in the hourglass specimen. A second explanation for geometry effects relates to the dimensional stability problem in uniform-gage samples. Buckling is very difficult to avoid at high strain ranges. This, of course, could shorten life. Further, local stress concentrations occur near the ends of the gage length at low strain ranges and also could shorten life. To avoid failure at the gage length-radius intersection, a very small taper is commonly introduced in the gage section with the minimum diameter at the center of the specimen.

Y-178668



Y-178669

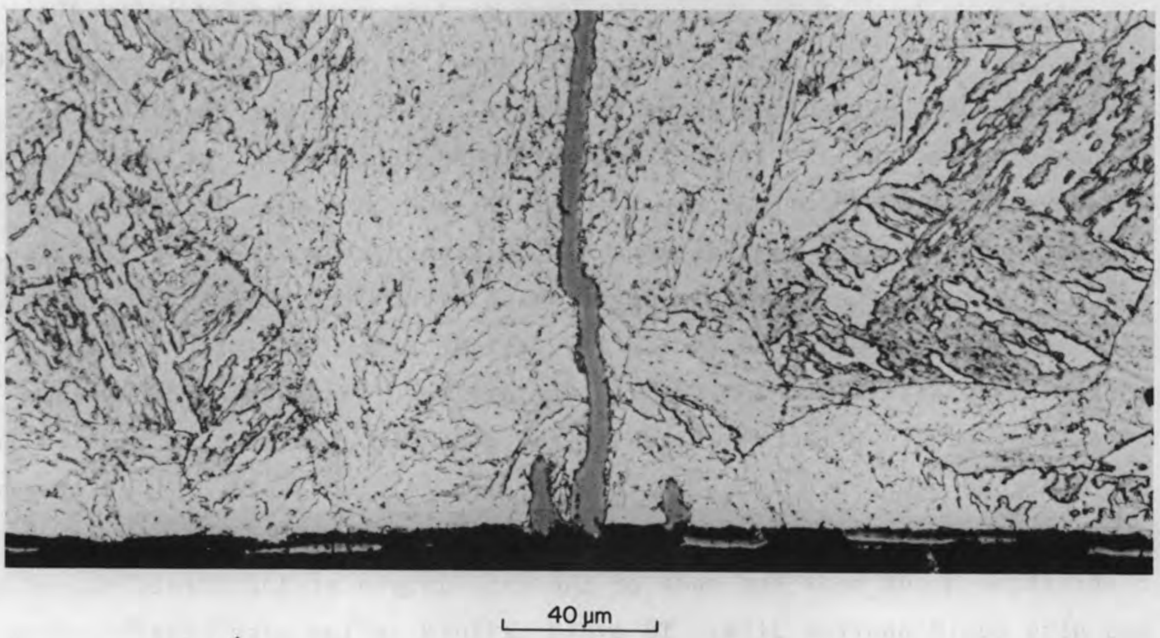


Fig. 10. Oxide-filled secondary fatigue cracks on the opposite sides of a specimen tested at 482°C and 0.65% strain range.

The validity of data produced from hourglass specimens is not without question. First, a multiaxial stress state exists at the minimum section.[14] This has an unknown influence on the stress versus strain response relative to data produced on uniform-gage specimens. Second, the strains are measured diametrally and used to calculate the axial strains. Assumptions must be made regarding the partitioning of the diametral strain into elastic and plastic components. This in turn requires assumptions that have not been adequately verified regarding the values of Poisson's ratio and Young's modulus in the elastic-plastic-creep regime. Third, the volume and surface area under maximum strain are smaller in the hourglass specimen than in the uniform-gage specimen. From a statistical viewpoint failure initiation may occur later in the hourglass specimen. Fourth, the possibility of dimensional instability and rachetting outside the controlled region still exists in hourglass specimens. This possibility further influences the stress versus strain behavior at the control point in a relatively unknown way. Finally, the hourglass specimens may have a distinctly different response from the uniform gage specimen once the fatigue cracks reach sizes and densities that affect the compliance of the specimen. We need to know more about the relative times spent in crack initiation and growth to assess the severity of this problem. The above issues suggest that it is important to fully understand the nature of work hardening and geometrical stability under cyclic conditions when testing specimens that have little or no work-hardening capability. Before leaving the subject, we point out that the work-hardening capability of the ferritic material tested by Ellis et al. was significant, yet differences were observed between uniform and hourglass specimens.

Another issue that bears on the significance of creep-fatigue data for low-alloy steels concerns the effect of oxide cracking on fatigue initiation and growth. The models suggested by Brinkman et al.,[10] Teranushi and McEvily,[12] and Challanger et al.[11] seem to explain most of the observed hold-time behavior patterns in both ferritic and bainitic steels. Here the oxide that forms during the tensile hold period will be relatively stress free, because any stresses associated with volume differences in base metal and oxide will be relieved by relaxation or plasticity in the base metal. During the subsequent excursion to the compressive strain limit the oxide can accommodate the compressive strain without cracking. In the case of the compressive hold time, the oxide must survive an excursion to the tensile strain limit. If $\Delta\varepsilon$ in the excursion exceeds a critical strain for oxide cracking, which is less in tension than compression, then the subsequent oxide crack can act as a strain concentration plane in the base metal and can lead to an eventual fatigue crack. The fraction of life spent in the crack initiation stage becomes greater with decreasing strain range, hence the effect of oxide cracking on fatigue crack initiation becomes more important as the strain is decreased. This is the trend that has been observed by others.[10,12,13] As pointed out by Challenger et al., the tensile hold mode may be more damaging once the fatigue crack has formed. At high strains initiation is quick, hence the tensile hold mode could produce higher crack growth rates and shorter life than the compressive hold mode. The understanding of the process is complicated by the use of hourglass specimens, however. Here the oxide can appear as a diametral strain and influence the test. Long-time tests produced oxide layers as thick as 20 μm. This thickness corresponds to a compressive axial strain near 0.8%. The machine reacts to this increase in diameter by stretching the specimen. The spalling and reforming of the oxide complicates the situation even further.

True creep-fatigue effects could develop in the bainitic material in any of several ways. First, the hold time produces relaxation with the concomitant increase in the inelastic strain range per cycle. Simply on the basis of the Coffin-Manson equation we would expect a small decrease in life relative to the continuous cycling test. Second, once the cracks have formed the hold time in tension could produce accelerated crack growth, as is often observed in fatigue crack growth experiments.[15] Third, with very long hold times, microcavity growth as observed in creep rupture[16] could occur on prior austenite grain boundaries and change the failure mode from transgranular to intergranular. At present we have no evidence that any of these time-dependent phenomena contribute significantly to the creep-fatigue effect observed in the bainitic material for the conditions of testing.

CONCLUSIONS

1. Under continuous-cycling strain-fatigue conditions data for bainitic 2 1/4 Cr-1 Mo steel fall close to the trends observed for ferritic material at comparable temperatures and cyclic strain rates.

2. Under hold time conditions at 1 and 2% strain range the fatigue life of bainitic material is less than that observed for ferritic material, and for all conditions examined the life was less under tensile hold time than under compressive hold time.

3. The fatigue life of uniform-gage specimens was less than that of hourglass specimens by at least a factor of 2.

4. The loss in fatigue life with hold time suggested that time-dependent damage occurred and was as severe as that reported for ferritic (annealed) material.

5. The failure mode in the bainitic material was transgranular with heavy oxide penetration of secondary fatigue cracks.

ACKNOWLEDGMENTS

The authors acknowledge the assistance of L. K. Egner and B. C. Williams in performing the mechanical tests and the work of C. W. Houck in the preparation of metallurgical samples. Helpful reviews were performed by C. R. Brinkman, M. K. Booker, and V. K. Sikka. The authors also acknowledge Sigfred Peterson for editing and Denise Sammons and Shirley Frykman for preparation of the final document.

REFERENCES

1. A. O. Schaefer, ed., *Symposium on 2 1/4 Chrome-1 Molybdenum Steel in Pressure Vessels and Piping*, American Society of Mechanical Engineers, New York, 1970.

2. *Data Sheets on Elevated-Temperature, Low-Cycle Fatigue Properties of SCMV4 (2.25 Cr-1 Mo) Steel Plate for Pressure Vessels*, National Research Institute for Metals, Tokyo, 1978, NRIM Fatigue Data Sheet No. 7.

3. R. M. Curran and B. M. Wundt, "A Study of Low-Cycle Fatigue and Creep Interactions in Steels at Elevated Temperatures," *Current Work on Behavior of Materials at Elevated Temperatures*, A. O. Schaefer, ed., American Society of Mechanical Engineers, New York, 1974, pp. 1–104.

4. C. E. Jaske and H. Mindlin, "Elevated Temperature Low-Cycle Fatigue Behavior of 2 1/4 Cr-1 Mo and 1 Cr-1 Mo-1/4 V Steel," *Symposium on 2 1/4 Chrome-1 Molybdenum Steel in Pressure Vessels and Piping*, A. O. Schaefer, ed., American Society for Mechanical Engineers, New York, 1971, pp. 137–210.

5. J. R. Ellis et al., "Elevated Temperature Fatigue and Creep-Fatigue Properties of Annealed 2 1/4 Cr-1 Mo Steel," *Structural Materials for Service at Elevated Temperatures in Nuclear Power Generation*, MPC-1, Schaefer, A. O., ed., American Society of Mechanical Engineers, New York, 1975, pp. 213–246.

6. R. A. Swift, "The Effects of Composition and Heat Treatment on the Mechanical Properties of 300 mm gage 2 1/4 Cr-1 Mo Steel Plate," in *The Application of 2 1/4 Cr-1 Mo Steel for Thick Wall Pressure Vessels*, A. O. Schaefer, ed., American Society of Mechanical Engineers, New York, in press.

7. M. K. Booker, B.L.P. Booker, and R. W. Swindeman, *Analysis of Elevated-Temperature Tensile and Creep Properties of Normalized and Tempered 2 1/4 Cr-1 Mo Steel*, ORNL/TM-8075, Jan. 1982.

8. G. V. Smith, "Elevated Temperature Strength and Ductility of Quenched and Tempered 2 1/4 Cr-1 Mo Steel," *Current Evaluation of 2 1/4 Chrome1 Molybdenum Steel in Pressure Vessels and Piping*, A. O. Schaefer, ed., American Society of Mechanical Engineers, New York, 1972, pp. 1–32.

9. M. K. Booker, "Analytical Description of the Effects of Melting Practice and Heat Treatment on the Creep Properties of 2 1/4 Cr-1 Mo Steel," *Effects of Melting and Processing Variables on the Mechanical Properties of Steel*, MPC-6, G. V. Smith, ed., American Society of Mechanical Engineers, New York, 1977, pp. 323–343.

10. C. R. Brinkman, J. P. Strizak, M. K. Booker, and C. E. Jaske, "Time-Dependent Strain-Controlled Fatigue Behavior of Annealed 2 1/4 Cr-1 Mo Steel for Use in Nuclear Steam Generator Design," *J. Nucl. Mater.* **62**: 181–204 (1976).

11. K. D. Challenger, A. K. Miller, and C. R. Brinkman, "An Explanation for the Effects of Hold Periods on the Elevated Temperature Fatigue Behavior of 2 1/4 Cr-1 Mo Steel," *J. Eng. Mater. Technol.* **103**, 7–14 (1981).

12. H. Teranishi and A. J. McEvily, "The Effect of Oxidation on Hold Time Fatigue Behavior of 2 1/4 Cr-1 Mo Steel," *Metall. Trans. A* **10A**, 1806–07 (1979).

13. R. D. Campbell, "Creep-Fatigue Calculation Procedures for Code Case 1592," pp. 45–56 in *Advances for Design for Elevated Temperature Environment,* S. Y. Zamrik and R. I. Jetter, eds., American Society of Mechanical Engineers, New York, 1975.

14. C. C. Schultz and H. M. Zien, "Verification of Specimens for Low-Cycle Fatigue and Cyclic Plasticity Testing," ASME paper 77-PVP-3S, presented at the Energy Technology Conference and Exhibit, Houston, Texas, Sept. 18–22, 1977.

15. W. R. Corwin, M. K. Booker, B.L.P. Booker, and C. R. Brinkman, "Characterizing Fatigue Crack Propagation in 2 1/4 Cr-1 Mo Steel for Steam Generator Applications," pp. 1–7 in *Trans. 5th Annual Conference on Structural Mechanics in Reactor Technology,* Vol. F, T. A. Jaeger and B. A. Boley, eds., North Holland, Amsterdam, 1979.

16. B. J. Cane, "Mechanistic Control Regimes for Intergranular Cavity Growth in 2.25 Cr-1 Mo Steel Under Various Stresses and Stress States," *Met. Sci.* **15**, 302–10 (1981).

INTERNAL DISTRIBUTION

1–2. Central Research Library
3. Document Reference Section
4–5. Laboratory Records Department
6. Laboratory Records, ORNL RC
7. ORNL Patent Section
8. M. K. Booker
9. R. A. Bradley
10. C. R. Brinkman
11. G. L. Copeland
12–14. F. R. Cox
15. J. F. Fisher
16. W. L. Greenstreet
17. M. L. Grossbeck
18. D. O. Hobson
19. J. A. Horak
20. R. R. Judkins
21. R. L. Klueh
22. L. E. McNeese
23. R. K. Nanstad
24. J. C. Ogle
25. A. R. Olsen
26. P. L. Rittenhouse
27. G. M. Slaughter
28. C. B. Smith
29–33. J. P. Strizak
34–38. R. W. Swindeman
39. M. H. Yoo
40. Alan Lawley (Consultant)
41. T. B. Massalski (Consultant)
42. R. H. Redwine (Consultant)
43. K. M. Zwilsky (Consultant)

EXTERNAL DISTRIBUTION

44. OAK RIDGE ASSOCIATAED UNIVERSITIES LIBRARY, Oak Ridge, TN 37830
Mert Division

45. DOE, OFFICE OF FOSSIL ENERGY, 20 Massachusetts Avenue, N.W., Washington, DC 20545

S. J. Dapkunas

46–47. DOE, OAK RIDGE OPERATIONS OFFICE, P.O. Box E, Oak Ridge, TN 37830

Office of Assistant Manager for Research and Development

48–74. DOE, TECHNICAL INFORMATION CENTER, P.O. Box 62, Oak Ridge, TN 37830