SOME CORRELATIONS OF HIGH- AND LOW-CYCLE FATIGUE DATA FOR $2\frac{1}{4}$ Cr-1 Mo STEEL IN THE ANNEALED, AND NORMALIZED AND TEMPERED CONDITIONS

PETER SOO AND JOE G.Y. CHOW

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DEPARTMENT OF APPLIED SCIENCE
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ABSTRACT

High-cycle fatigue data for both annealed, and normalized and tempered 2½Cr-1Mo steel were obtained from load-controlled tests at 1000°F and compared to strain-controlled low-cycle results from the literature. In the range of cycles where the two sets of data overlap, good correlations were obtained provided that the high-cycle stress values were converted to equivalent strains using appropriate cycle stress-strain curves. The fatigue strengths for both heat treatments were strongly dependent on the carbon level indicating that any carbon losses during fabrication or service are likely to detrimentally affect the fatigue resistance of reactor components. Specimen geometry is also shown to be an important consideration in the production of conservative fatigue data. Fatigue testing at ultrasonic frequencies yields much lower endurance limits when compared to normal high-cycle push-pull tests at 1000°F and this is possibly connected with localized deformation within the sample.

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1. Introduction

Within the power industry, $2\frac{1}{2}\text{Cr-1Mo}$ steel is used as a general structural material for various water/steam-side component applications. The principle reasons for its usefulness lie in its relatively low cost and its resistance to chloride and caustic stress-corrosion cracking. Specific uses in nuclear reactor technology include its use in the High-Temperature Gas-Cooled Reactor (HTGR) as a steam generator tubing material and for components within the helium circulators. In the Liquid-Metal Fast-Breeder Reactor (LMFBR) it is also a candidate for use as a steam generator tubing material.

A major effort is currently in progress at Brookhaven National Laboratory (BNL) in the area of high-cycle fatigue, where failure typically occurs in 10^5 cycles or greater, in an attempt to evaluate the structural integrity of critical HTGR components which are subjected to low-amplitude, high-frequency vibrations. At this time there are few relevant data available so that this type of safety-related analysis is extremely difficult.

The studies are being conducted on material in two heat-treated conditions. Firstly, the fully-annealed state is being evaluated since this is relevant to many reactor components. In addition, the normalized and tempered condition is being studied since this will provide basic insight into the properties of weld material which undergoes rapid air cooling from the weld temperature, followed by a tempering treatment at approximately 1300°F to minimize the probability of embrittling effects. For this heat treatment, steels with a range of carbon

contents were evaluated in order to determine the effects of possible carbon depletion in components during heat treatment, fabrication or service. An example where this could present concern lies in the $2\frac{1}{4}\text{Cr}-1\text{Mo}/\text{Incoloy-800H}$ weld joint in the HTGR steam generator tubes. Carbon transfer from the ferritic to the austenitic side of similar weld joints is a known phenomenon¹⁻³ and could lead to a decreased resistance to fatigue failure. A major part of this study involves a correlation of the high-cycle data produced in this work with available low-cycle information from the literature. The intent is to determine whether a unique fatigue curve may be obtained for cycles-to-failure, N_f , between 10^2 and 10^8 , or whether there is a discontinuity at the cross-over from high- to low-cycle data which is usually taken to be at 10^5 cycles. Such a curve would be of significant value in analytical studies of component structural integrity.

2. Materials and Test Procedures

The annealed material was supplied by the Metal Properties Council, New York, and the normalized and tempered steels were obtained from Oak Ridge National Laboratory (ORNL). Table 1 gives the chemistries of these materials and their heat treatments. For the normalized and tempered heats the chemistries were taken to be those determined by ORNL since these are the most recent and comprehensive.⁴

Figure 1 shows the two specimen configurations used in this work. The pipe material was received in the fully-annealed condition, viz.

heated to 1700°C for two hours followed by furnace cooling at 50°F/hour to 1100°F, followed by air cooling to room temperature. Samples were machined from locations close to the internal and external surfaces and also from mid-wall locations. All were oriented along the pipe axis and no additional heat treatment was given. For the remaining steels the starting materials were heated to 1700°C and then allowed to air cool to room temperature. This was followed by tempering at 1300°F followed by air cooling. Specimens were then machined from the heat-treated stock. All samples were given a surface-ground finish.

Figure 2 shows the microstructure of a specimen cut from a mid-wall location in the annealed pipe material. The grains are predominantly ferrite and contain a fine dispersion of carbides. Interspersed between the ferrite grains is approximately 10-20% pearlite.

Figure 3 give the microstructure of all of the normalized and tempered materials. For the 0.009% carbon steel virtually all of the structure is pro-eutectoid ferrite. In the case of the 0.030% carbon material, however, significant amounts of a bainitic second phase are seen, amounting to about 10% of the volume. There is a noticeable decrease in the grain size of this steel when compared to the low-carbon material. For the 0.120% carbon rod and the 0.135% carbon commercial plate the structure is nearly all tempered bainite with only a few grains of ferrite readily identifiable. The microstructures of the two materials are essentially identical.

Three MTS Systems Corporation closed-loop, electro-hydraulic fatigue

machines were used in this program. They were all operated in the push-pull mode with a sinusoidal load waveform at a frequency of 40 Hz under load-controlled conditions. Limit switches were set so that any deviation greater than $\pm 2\%$ of the selected load stops the test. The specimens were heated with a resistance-type furnace and all tests to date have been carried out in air. Temperatures were controlled to within $\pm 2^\circ\text{F}$ by a thermocouple attached to the specimen surface. During the initiation of a test the load is steadily increased from zero to the required steady-state value.

A limited number of experiments were conducted using an ultrasonic fatigue machine operating at a frequency of $20,000 \pm 1,000$ Hz in an attempt to extend the N_f range beyond 10^8 . A description of this apparatus is given in reference 5.

Some tests were conducted in a previous BNL program described in report numbers BNL 50363 and BNL 50376.

3. Correlation of High- and Low-Cycle Data for Annealed Material

As discussed in prior reports,^{5,6} high-cycle and low-cycle fatigue test procedures are basically different since the former involves load control at low stress levels, whereas the latter is concerned with strain control at higher stresses. The major differences lie in the way steady-state, stress-strain conditions are achieved, and in the large differences in the cycling rates. In the earlier studies it was shown that high- and low-cycle data for Incoloy-800H and Type 304 stainless steel were correlatable as functions of both the applied

stress, σ , and total strain range, $\Delta\epsilon_t$, if the following criteria are met:

- For the $\sigma - N_f$ data the effective stresses for the low-cycle data are taken to be the values at half the fatigue life, i.e. at $N_f/2$.

- For the $\Delta\epsilon_t - N_f$ curves the effective strains for the high-cycle tests are obtained from cyclic stress-strain curves, which relate the stable stress and stable strain levels during test.

The above guidelines were used here in an attempt to correlate the high-cycle data for annealed 2½Cr-1Mo and available low-cycle data from the literature.^{7,8} Reference 8 contains a compilation of information from several sources including Battelle Memorial Institute (BMI), General Atomic (GA) and Mar Test Laboratory. These are acknowledged in the figures of this report. Most of the low-cycle data considered are for a strain rate of $4 \times 10^{-3} \text{ sec}^{-1}$, since this was most common.

Table 2 shows the BNL high-cycle data and Figure 4 shows the correlation with the low-cycle data as a function of the alternating stress. The stress values for the low-cycle data were those cited by the various experimentalists, and represent the values at $N_f/2$. Comparisons with data other than those from ORNL and BMI were not possible since the stress levels were not given. A good correlation between the ORNL and BNL results was obtained although some scatter was evident in the high-cycle data. However, there was a major discrepancy with the BMI data which gave fatigue lives about two orders of magnitude

lower than those obtained by ORNL and BNL. Possible reasons for this anomalous behavior are:

- Surface finish
- Strain rate
- Material chemistry
- Size effects and geometry.

The BNL samples were given a machine-ground finish whereas the BMI samples were given a final hand finish through 600-grit emery paper. Since these are similar finishes the wide difference in N_f values is probably not due to surface effects. Strain rate effects are also unlikely to be responsible since the respective strain rates for the ORNL and BMI tests were 4×10^{-3} and $1.7 \times 10^{-4} \text{ sec}^{-1}$, i.e. they differ by about an order of magnitude. From the data given in Figure 5, it is clear that in the high-cycle fatigue range strain rate effects are not sufficiently large to account for the 100 times difference in fatigue life between the ORNL and BMI tests. Material chemistry is also unlikely to explain the lower fatigue strengths in the BMI tests since the BNL and BMI results were obtained from the same pipe material.

It appears that the most plausible explanation is connected with specimen geometry. Both the ORNL and BNL samples had an hourglass-shaped gauge length, whereas the BMI samples have a straight-sided one-inch gauge length. Studies have demonstrated that a large effective surface area increases the probability of crack initiation, and this would help explain the anomalous BMI results.⁹ Specimen geometry must, therefore, be considered an important variable in determining fatigue life in reactor structural steels since parallel-sided samples appear

to yield more conservative data when compared to the more commonly used hourglass type.

An important observation in Figure 4 is that there is not a well-defined endurance limit at 1000°F in annealed 2½Cr-1Mo steel. This indicates that fatigue failure may present a problem, even at relatively low stress levels. Such behavior is in contrast to Incoloy-800H and annealed Type 304 stainless steel which show stable fatigue strengths for N_f values in excess of about 10^5 cycles (see Figure 6). The lower fatigue strengths of 2½Cr-1Mo will be magnified if any in-service decarburization occurs since decreased carbon levels will lead to a general weakening of the steel.

The data in Figure 4 also show a dependence of fatigue strength on the location from which the specimens were machined from the pipe. Samples cut from regions adjacent to the outside surface were generally stronger, especially for the very high-cycle range. Samples cut from the mid-wall regions were intermediate in strength, and internal diameter samples were the weakest. This observation is compatible with chemical analyses conducted by Jaske, et al on the same pipe material.⁷ They showed that for two specimens cut from the outer pipe surface the carbon levels were 0.17 and 0.15%, whereas for two samples from the mid-wall region the concentrations were both 0.13%. It is, therefore, apparent that significant changes in high-cycle fatigue strength may result from differences in carbon concentration brought about by processing variables. Care must be taken to account for such deleterious effects in design.

In order to correlate the high- and low-cycle data as a function of strain range, available cyclic stress-strain curves from the literature have been plotted in Figure 7. The cyclic stress-strain curve obtained from the ORNL hourglass sample shows a significantly higher strength than parallel-sided BMI samples. Similar behavior has been previously observed from monotonically-loaded austenitic materials,¹⁰ indicating that this is a general geometrical effect. Hence, care must be taken to ensure that the cyclic stress-strain curve used for stress-to-strain conversion are for the relevant geometry. If, for example, the BNL high-cycle fatigue stresses were converted to equivalent strains using the BMI instead of the ORNL curve in Figure 7, a larger effective strain range would be obtained, and give much less conservative fatigue curves.

The fatigue data in Figure 8 were, accordingly, obtained using the ORNL curve in Figure 7. A good correlation is obtained between the BNL and ORNL hourglass samples, although some scatter exists in the BNL results at the higher strain levels. This effect could be associated with internal heating which was observed in a few specimens during the very early stages of deformation.

4. Correlation of Data for Normalized and Tempered Material

Table 3 and Figure 9 show the high-cycle data obtained for the four heats of normalized and tempered $2\frac{1}{4}\text{Cr}-1\text{Mo}$ steel described in Table 1. As expected, the two high-carbon heats show comparable behavior whereas lower carbon levels give rise to extremely large drops

in fatigue strength. This strongly suggests that significant losses in fatigue strength are likely to occur in components which are decarburized either during fabrication, heat treatment, or service. In this respect, the 2½Cr-1Mo/Incoloy-800H weld joint in the HTGR steam generator is an area susceptible to losses in high- and low-cycle fatigue strength because of the likelihood of in-service carbon transfer from the ferritic to the austenitic side of the joint. Also, since fatigue failure is usually a surface initiated phenomenon, surface decarburization alone may give rise to a noticeable degradation of fatigue strength.

Figure 10 shows a comparison of the high-cycle results and available BMI low-cycle fatigue data. There is a large discrepancy between the two sets of fatigue results. In fact, the BMI fatigue lives are two orders of magnitude less than the high-cycle values. Again, this is best attributed to differences in specimen geometry, as discussed above for the annealed material. Since most fatigue data being generated at this time within the nuclear industry are for hourglass samples, care must be exercised to account for these geometrical effects in component design and analysis. Failure to do so may result in a non-conservative estimate of the fatigue behavior.

5. Ultrasonic Fatigue Tests

Ultrasonic fatigue test procedures were used in an exploratory attempt to extend the range of fatigue data beyond 10^8 to determine whether the fatigue curves obtained from the MTS machines tend towards

an endurance limit. Figure 11 and Table 4 show the BNL high-cycle and ultrasonic fatigue data obtained from the 0.120% carbon material. The endurance limit is about 40% lower in the ultrasonic tests and could be associated with non-homogeneous deformation such as that observed for similarly-tested Incoloy-800H samples.⁵ For the latter material it was shown that deformation tended to be in highly-localized areas which were surrounded by regions which were essentially free of plastic flow. It was postulated that the local stresses within these regions of deformation were sufficient to cause fatigue crack initiation even though the macroscopic stress levels were much lower.⁵ For the Incoloy-800H samples there was a distinct possibility that the cracks were initiated at sub-surface sites whereas for the 2½Cr-1Mo steel the cracks were formed at the surface. At this time, therefore, it seems clear that ultrasonic test procedures are of limited application with respect to nuclear component design and analysis. Consequently, additional testing in this area is being terminated.

6. Discussion and Conclusions

Although the high-cycle test data produced in this work are limited, several significant conclusions may be made regarding the fatigue resistance of 2½Cr-1Mo steel, for both the annealed and the normalized and tempered conditions. These findings are of relevance to the structural integrity of specific HTGR and LMFBR components such as the steam generators and HTGR helium circulators. Among the more

important findings are the following:

(a) For both the annealed and the normalized and tempered states the fatigue strength is strongly dependent on the carbon concentration. Any losses in carbon level during heat treatment, fabrication, or service are likely to result in reduced fatigue resistance. In the HTGR steam generator, for example, the $2\frac{1}{4}\text{Cr}-1\text{Mo}/\text{Incoloy-800H}$ bimetallic weld joint is an area of potential concern since it is well known that high-carbon ferritic materials are prone to lose carbon to austenitic materials if there is a diffusion bond between the two. Because of this, the ferritic side of the bimetallic joint will weaken and show a decreased fatigue resistance. Testing of prototypic thermally-aged welds must be conducted to quantify the extent of carbon transfer and its impact on fatigue properties.

(b) For the annealed condition, low- and high-cycle fatigue data are correlatable as a function of the alternating stress. There does not appear to be a significant effect arising from the large differences in cycling rate for the test data evaluated.

(c) For the annealed material, high- and low-cycle fatigue data are correlatable as a function of strain provided the high-cycle stress levels are converted to equivalent strains by the use of cyclic stress-strain curves.

(d) For the normalized and tempered state, high- and low-cycle data are also correlatable as a function of stress. Strain rate differences are not apparently significant in the 10^5 cycle range.

(e) For both the annealed and the normalized and tempered states

there seems to be a clear difference in the cyclic work hardening rates of hourglass and parallel-sided test samples. The latter geometry results in a pronounced decrease in fatigue life when compared with the usual hourglass sample. At this point it seems that the straight-sided samples could also give a lower endurance limit. Hence, a careful study is needed in order to determine the extent to which hourglass samples may be used to generate conservative design data for reactor component design and analysis.

(f) Samples fatigue tested at an ultrasonic frequency of 20,000 Hz give an extremely low endurance limit when compared to samples tested under more conventional push-pull conditions at 40 Hz. It is postulated that, in the ultrasonic tests, plastic deformation is restricted to discrete zones which may be the sites for crack nucleation. Owing to the specialized nature of this deformation it is thought that ultrasonic fatigue test data are of strictly limited utility in the nuclear industry.

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TABLE 1

Chemistries and Heat Treatments of Starting Materials

Material	Chemical Composition %								Heat Treatment
	C	Mn	Si	Cr	Mo	Ni	S	P	
Pipe	0.145	0.53	0.40	2.2	0.94	0.30	0.019	0.014	Fully Annealed*
Low - C Rod	0.009	0.51	0.58	2.3	0.98	0.05	0.009	0.002	Normal. 1700°F, Temp. 1300°F
Med - C Rod	0.030	0.38	0.27	2.4	1.14	< 0.05	0.025	0.019	Normal. 1700°F, Temp. 1300°F
High - C Rod	0.120	0.44	0.40	2.2	0.91	0.21	0.015	0.010	Normal. 1700°F, Temp. 1300°F
High - C Plate	0.135	0.57	0.37	2.2	0.92	0.16	0.016	0.012	Normal. 1700°F, Temp. 1300°F

*Annealed 1700°F/2h, furnace cooled at 50°F/h to 1100°F, then air cooled to room temperature.

TABLE 2

High-Cycle Fatigue Data for Fully Annealed

2½Cr-1Mo Steel Pipe at 1000°F⁽¹⁾

Specimen No.	Location in Pipe ⁽²⁾	Alternating Stress (ksi)	Cycles to Failure ⁽³⁾
MC1-70	MW	42.0	4.8×10^3
MC1-75	MW	38.0	4.8×10^3
MC1-78	ID	32.0	7.4×10^3
MC1-66	OD	32.0	9.9×10^3
MC1-64	OD	28.0	3.5×10^6
MC1-62	OD	25.0	1.0×10^7
MC1-79	ID	25.0	1.1×10^6
MC1-65	OD	22.0	7.1×10^7
MC1-80	ID	22.0	2.9×10^7
MC1-76	MW	20.0	6.1×10^7
MC1-82	ID	20.0	1.5×10^7
MC1-63	OD	20.0	$>2.0 \times 10^8$

NOTES: (1) The pipe was 10" OD x 1.972" wall. It was heated for 2 hours at 1700°F, furnace cooled at 50°F/h to 1100°F and then air cooled to room temperature. The average carbon content is 0.145%.

(2) Samples were cut along the axial direction of the pipe and were from the ID, OD, or the mid-wall (MW) locations.

(3) Cycling rate is 40 Hz.

TABLE 3

High-Cycle Fatigue Data for Normalized and Tempered
2½Cr-1Mo Steel at 1000°F

<u>Material</u>	<u>Specimen No.</u>	<u>Alternating Stress (ksi)</u>	<u>Cycles to Failure</u>
Experimental Heat 0.009% C	MC1-04	34.0	2.2×10^5
	MC1-03	32.0	2.0×10^5
	MC1-02	30.0	1.2×10^6
	MC1-01	28.0	4.9×10^6
	MC1-07	28.0	2.1×10^6
	MC1-05	26.9	3.0×10^5
	MC1-06	26.9	1.4×10^6
	MC1-08	26.4	4.4×10^7
	MC1-09	24.0	1.5×10^6
	MC1-10	22.0	6.2×10^7
	MC1-11	20.0	7.3×10^7
Experimental Heat 0.030% C	MC1-13	36.0	2.7×10^4
	MC1-12	34.0	4.5×10^6
	MC1-14	32.0	2.7×10^6
	MC1-15	30.0	4.0×10^6
	MC1-16	28.0	3.8×10^7
Experimental Heat 0.120% C	MC1-27	42.0	1.2×10^5
	MC1-28	40.0	1.0×10^6
	MC1-29	39.2	3.3×10^6
	MC1-30	38.4	5.9×10^5
	MC1-31	38.4	8.1×10^7
	MC1-32	38.4	4.3×10^7
	MC1-26	38.0	$> 6.3 \times 10^6$
Commercial Plate Heat 20017 0.135% C	MC1-50	40.0	1.7×10^6
	MC1-51	38.0	1.3×10^7
	MC1-52	36.0	1.2×10^7
	MC1-53	35.0	$> 4.8 \times 10^7$

Notes:

1. All materials normalized from 1700°F and tempered at 1300°F.
2. Fatigue cycling rate is 40 Hz.

TABLE 4

Ultrasonic Fatigue Data for Normalized and
Tempered 2½Cr-1Mo Steel at 1000°F

<u>Specimen</u>	<u>Alternating Stress (ksi)</u>	<u>Cycles to Failure</u>
UC1-31	25.0	3.7×10^7
UC1-28	23.4	4.2×10^8
UC1-35	22.6	3.2×10^8
UC1-27	22.5	$>1.2 \times 10^{10}$
UC1-17	22.2	$>1.2 \times 10^{10}$
UC1-34	21.9	$>1.2 \times 10^{10}$
UC1-33	21.2	$>2.1 \times 10^{10}$
UC1-29	20.7	1.3×10^8
UC1-32	20.7	2.7×10^8
UC1-24	20.0	$>1.3 \times 10^{10}$

Notes:

1. Material contains 0.120% carbon and was normalized from 1700°F and tempered at 1300°F.
2. Fatigue cycling rate is 20,000 Hz.

MTS FATIGUE SPECIMEN

ULTRASONIC FATIGUE SPECIMEN

Figure 1. Standard specimen configurations for high-cycle and ultrasonic fatigue testing.

Figure 2. Microstructure of annealed $2\frac{1}{4}\text{Cr}-1\text{Mo}$ steel pipe.
Mag. 250X.

Figure 3. Microstructure of normalized and tempered $2\frac{1}{4}\text{Cr}-1\text{Mo}$ steels as a function of carbon content: (a) 0.009% C, Mag. 250X, (b) 0.030% C, Mag. 250X, (c) 0.120% C rod, Mag. 400X, (d) 0.135% C plate, Mag. 250X.

Figure 4. High-cycle fatigue of annealed 2 $\frac{1}{4}$ Cr-1Mo steel at 1000°F as a function of stress.

Figure 5. Effect of cycling rate on the high-cycle fatigue of annealed 2 $\frac{1}{4}$ Cr-1Mo steel at 800°F.

Figure 6. Comparison of the 1000°F fatigue strengths of annealed 2 $\frac{1}{4}$ Cr-1Mo steel, type 304 stainless steel and incoloy-800H.

Figure 7. Cyclic stress - strain curves for annealed 2½Cr-1Mo steel at 1000°F.

Figure 8. High-cycle fatigue of annealed 2½Cr-1Mo steel at 1000°F as a function of strain.

Figure 9. Effect of carbon content on the high-cycle fatigue behavior of normalized and tempered 2 $\frac{1}{4}$ Cr-1Mo steel at 1000°F.

Figure 10. Comparison of high and low cycle fatigue data for normalized and tempered 2 $\frac{1}{4}$ Cr-1Mo steel at 1000°F as a function of stress.

Figure 11. Comparison of high-cycle and ultrasonic fatigue tests for normalized and tempered 2 $\frac{1}{4}$ Cr-1Mo steel at 1000°F.