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SOME QUESTIONS REGARDING THE INTERACTION OF CREEP AND FATIGUE

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ABSTRACT

Data are presented from fatigue-crack growth tests conducted on Type 304 S.S. in inert environments at elevated temperatures which show that the thermal-activation noted in similar tests run in air environments is not present in the inert environment. Similar observations from the literature are reviewed, including the observation that the time-dependency noted in tests conducted in elevated temperature air environments is also greatly suppressed in inert environments. These findings suggest that an interaction between the fatigue process and the corrosive air environments is responsible for the thermally-activated time-dependent behavior often attributed to creep-fatigue interaction. Data are also presented which show that the fatigue-crack growth behavior of Type 304 S.S. subjected to significant creep damage prior to fatigue testing does not differ appreciably from the behavior of material not subjected to prior creep damage; again indicating minimal interaction between creep and fatigue. It is suggested that in the temperature range where pressure vessels and piping are generally designed to operate (i.e. below about one-half the absolute melting temperature of the alloy), the interaction between creep and fatigue is far less significant than once supposed, and that the major parameter interacting with the fatigue process is that of high temperature corrosion.

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INTRODUCTION

Fatigue tests on structural metals have been conducted at elevated temperatures for a number of years. These tests have generally been of two types: the "traditional" fatigue test where the specimen life is expressed as a function of the stress (or strain) range, and the more recent fracture mechanics tests where the rate of crack extension is expressed in terms of the stress (or strain) intensity factor. The majority of these elevated temperature tests have been conducted in an air environment. Although the two types of tests are of a quite different nature, they generally have two observations in common: decreasing fatigue lives (or increasing crack growth rates) with 1) increasing test temperature and 2) decreasing cyclic frequency. In other words, the process responsible for the observed changes in fatigue behavior is thermally-activated and time-dependent. Since the phenomenon of creep at elevated temperatures is also thermally-activated and time-dependent, these changes in fatigue behavior have often been attributed to a "creep component" or a "creep-fatigue interaction". The relative importance of a creep component in the fatigue process at elevated temperatures has been questioned by previous investigators. This paper reviews some of the previous findings and contributes further information to support the position that the interaction

between creep and fatigue at temperatures of general structural interest (i.e. below about one-half the absolute melting temperature) can be rather minor compared to the contribution of the air environment itself.

It has been long recognized that, even at room temperature, an air environment could exert an aggressive action (relative to behavior in vacuo) in the fatigue of metals[1]. In recent years investigators have noted substantial improvements in fatigue behavior (i.e. longer fatigue lives and/or lower crack growth rates) when tests conducted at elevated temperatures in vacuo or in inert gasses are compared to identical tests conducted in an air environment[2-6]. In most of these cases the improvements in fatigue behavior were so dramatic that the fatigue behavior in the inert environment at elevated temperature was approximately the same as that observed in tests in ambient air at room temperature; thereby suggesting that the aggressive action of the air environment played a far greater role in influencing the behavior than did creep. In other words, the thermal activation noted in the tests conducted in elevated temperature air should be attributed to corrosion-enhanced fatigue rather than creep. This will be discussed in more detail later.

It has also been noted that when fatigue tests are conducted at elevated temperature in vacuo or in an inert environment, the time-dependency observed in elevated temperature air environments is reduced considerably, if not eliminated altogether[3,4,7]. Hence we see that when fatigue tests are conducted in inert environments at elevated temperatures, much if not all of the thermally-activated and time-dependent behavior often attributed to a "creep-fatigue interaction" is eliminated. This behavior will be discussed in more detail later.

Finally, new results of fatigue-crack propagation tests on material which had received significant creep damage prior to fatigue testing will be presented, and it will be shown the behavior was essentially identical to tests on material which had received no creep damage prior to fatigue testing.

Although none of these observations is in itself completely sufficient evidence of the lack of an appreciable interaction between creep and fatigue, taken together they do raise serious questions regarding the extent of such an interaction and the way in which such assumed interactions are being accounted for in present design codes (e.g. linear or non-linear damage summation equations).

EXPERIMENTAL PROCEDURE

Specimen Material

The material utilized in the present study was from a heat of exceptionally well-characterized solution-annealed Type 304 stainless steel (Allegheny Ludlum heat 55697). This heat of material has been employed extensively in the author's laboratory to investigate the effects of several parameters upon the crack growth behavior of Type 304 stainless steel. Among the parameters investigated using this heat are the effect of temperature[8,9], cyclic frequency[10], loading waveform (or hold-time)[11], neutron irradiation[12], cyclic stress ratio[13,14], thermal aging[15], surrounding environment[6], and crack growth in weldments[16]. In addition, this heat has been employed in studies of the effect of temperature[17], strain rate[18], thermal aging[19], and neutron irradiation[20] upon the tensile properties, as well as employed in studies of flow behavior[21,22], low-cycle fatigue[23,24] and creep and stress rupture behavior[25,26]. The elastic constants of this heat at several temperatures are given in Reference 23.

The chemical composition and room temperature mechanical properties of heat 55697 are listed in Tables I and II, respectively*, and the thermal/mechanical processing history of this heat is detailed in Reference 8.

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*Note: Minor differences may be seen in the listed values for chemical composition and mechanical properties between some of the studied conducted on heat 55697. These differences are probably due to normal experimental scatter, or due to the different product forms (e.g. plate and rod) produced from this heat.

Environmental Testing Appratus

A series of elevated temperature fatigue-crack propagation tests were conducted in two relatively inert environments: vacuum and liquid sodium. These results have been presented elsewhere[6], but because of their importance in the present discussion, the apparatus used will be briefly described here. Reference 6 should be consulted for more details.

Vacuum tests were conducted in a commercially-available (Richard Brew Co.) chamber that allowed vacuum of approximately 6×10^{-6} torr (8×10^{-4} Pa) to be achieved at elevated temperatures. A servo-controlled electro-hydraulic testing machine (MTS Systems Corp.) was utilized to apply the cyclic loads, and load was used as the control parameter. The heating elements in the chamber were of tungsten, and hence acted as an "oxygen getter" at elevated temperatures. Therefore, oxygen content in the chamber environment during the fatigue tests was probably quite low. ASTM "Compact Specimens"[27] were used in the fatigue tests.

The tests conducted in the sodium environment were performed in a special test chamber (see Ref. 6) which allowed liquid sodium to continuously flow around an ASTM Compact Specimen. A servo-controlled electro-hydraulic loading system (MTS Systems Corp.) was utilized to apply the cyclic loads, using load as the control parameter. The test chamber was installed on a relatively large sodium loop which was operated approximately isothermally for these tests. Impurity levels within the sodium (principally oxygen and hydrogen) were controlled with the use of a cold trap. Crack length measurements in both the sodium and vacuum tests were made by use of a fracture face marking technique[6].

Creep/Fatigue Specimens

The second phase of this study involved the fatigue-crack propagation testing of specimens which had received significant creep damage prior to fatigue testing. The specimens employed were of the center-cracked design (see Figure 1).

However, the specimens were initially machined without the center-notch, as the creep phase of the testing was done on unnotched specimens. The reason for utilizing this specimen design for the creep/fatigue testing was that, unlike the ASTM Compact Specimen, it possesses a uniform stress across the entire specimen width. This is important in assessing the extent of creep damage. The specimens were creep tested in a conventional creep testing machine (SATEC Systems Corp.) fitted with an air-circulating furnace. Creep exposure was 3000 hours at 1000°F (583°C) at 23,475 psi (162 MPa). This stress level is 75 percent of the 3000-hour rupture stress at 1000°F (538°C) for this heat of material. (See the Appendix for the stress-rupture behavior and an assessment of the creep damage).

Following this exposure the specimens were removed from the creep tester and measured to determine creep deformation (see Figure 1). Electrical-discharge machining was then used to produce a starter notch in the specimen center. The specimens were then fatigue precracked after which the regular fatigue testing commenced. Specimens were tested in an air-circulating furnace at 1000°F (538°C). A servo-controlled electro-hydraulic testing machine (MTS Systems Corp.) employing load for the control parameter was utilized. One specimen was tested at 40 cpm (0.667 Hz) utilizing a "sawtooth" waveform (see Figure 2), and one specimen was tested at 0.083 cpm (1.39×10^{-3} Hz) utilizing a "square" waveform with a hold-time. From Reference 11 it is known that in an air environment at 1000°F (538"C) the former waveform-frequency combination produces crack extension which is predominately transgranular, while the latter waveform-frequency produces cracking that is predominately intergranular. Thus, the effect of prior creep damage was investigated for both modes of crack extension: transgranular and intergranular.

In addition, several specimens (both ASTM Compact Specimens and center-cracked specimens) which had not received prior creep damage were tested for comparison purposes. (See Table III for a listing of specimens and test parameters).

Some of these specimens had received the same thermal exposure (but without the applied stress) as the specimens that were creep tested. The stress ratio R ($R = K_{min}/K_{max}$) was 0.05 for these tests.

Crack lengths were determined using a travelling microscope. The crack growth rate (da/dN) for a given increment of cracking was calculated by dividing the crack extension (Δa) by the number of cycles in that increment (ΔN). The stress intensity factor (K) was based on the average crack length for that increment using the relationships found in Reference 27 and 28 for the Compact Specimens and center-crack specimens, respectively.

EXPERIMENTAL RESULTS

Environmental Tests

Tests were conducted in a liquid sodium environment and *in vacuo* at two test temperatures, 800°F (427°C) and 1000°F (538°C), and the results are shown in Figures 3 and 4, respectively. The fracture face marking technique mentioned previously required the use of more than one single stress ratio during the testing of each specimen. For this reason the crack growth rates are plotted as a function of the "effective stress intensity factor" since it has been shown[13,14] that this parameter provides an excellent correlation of the data when more than one stress ratio is utlized.

The scatter bands for tests conducted in an air environment at 800°F (427°C) and 1000°F (538°C) are also plotted in Figures 3 and 4, respectively, as well as the scatter bands for tests conducted in an air environment at 75°F (24°C). These results form a basis with which the tests in sodium and vacuum environments may be compared. A number of observations may be made concerning these data: 1) at each of the elevated test temperatures, the crack growth behavior in sodium and *in vacuo* are essentially identical (i.e. a sodium environment is approximately as

inert as a vacuum), 2) at a given value of K_{eff}, the crack growth rate in sodium or in vacuo is considerably lower than that in an air environment at the same temperature and, 3) at both elevated test temperatures the crack growth rates in sodium or in vacuo are approximately equivalent to that observed in a room temperature air environment. These results will be discussed in a later section.

Creep/Fatigue Tests

Fatigue tests were conducted at 1000°F (538°C) on specimens that had received prior creep damage as previously described, and on specimens which had received no prior creep treatment. Waveform-frequency combinations were used which produced either predominately transgranular cracking or predominately intergranular cracking.

It will be noted in Figure 5 that Specimens 228 and 308 (neither of which received prior creep damage) exhibit essentially identical behavior at 1000°F (538°C). Since the former is a center-cracked specimen while the latter is an ASTM Compact Specimen, this excellent correlation between specimen designs is yet another example of the successful use of fracture mechanics techniques in describing cracking behavior at elevated temperatures. However, a much more significant observation concerning the data in Figures 5 and 6 is that for both modes of crack extension, predominately transgranular and predominately intergranular, the specimens receiving the rather severe creep treatment prior to fatigue testing exhibit behavior that is essentially identical to that of the specimens which had received no prior creep damage. This result is particularly dramatic since it is estimated (see the Appendix) that, prior to fatigue testing, Specimen 226 had received 36% of the creep strain to rupture, and Specimen 384 had been subjected to 50% of the creep strain to rupture.

Metallographic examination of Specimens 226 and 384 at the conclusion of fatigue testing revealed the presence of numerous grain boundary cracks. Typical examples of these cracks are shown in Figures 7a and 8a. These photomicrographs were obtained from areas far removed from the crack tip area, and presumably were not influenced by the elevated stresses in the vicinity of the crack tip. Such grain boundary cracks are often associated with creep damage[29]. Figures 7b and 8b illustrate the predominately transgranular and predominately intergranular fatigue cracking associated with Specimens 226 and 384, respectively. Note the grain boundary cracks in the vicinity of the crack tip in Specimen 384 (Figure 8b). Such cracks were also noted in the crack tip vicinity in Specimen 226, but they were less numerous and not as pronounced as in Specimen 384. Although quantitative grain boundary crack densities were not determined it was noted that Specimen 384 contained a somewhat larger number of cracks per unit area. This somewhat greater crack density could be due to the more severe creep damage sustained by this specimen during the creep testing (see the Appendix), or it could be due to additional creep damage sustained during fatigue testing. (Total fatigue test time for Specimen 384, including hold-times, was approximately 2200 hours, while the total fatigue test time for Specimen 226 was only about 34 hours).

It will be noted that the specimens tested at the low frequency with the hold-time (Fig. 6) exhibit somewhat greater scatter in the results than those tested at the higher frequency with the sawtooth waveform. Most of this increased scatter is attributed to the greater difficulty encountered in determining the lengths of the intergranular cracks.

DISCUSSION

As previously mentioned, a number of investigators have found that when elevated temperature tests are conducted in an inert environment, substantial improvements are noted relative to identical tests in an air environment. Hill[2] studied 1% Cr-Mo-V steel at 1022°F (550°C) and found appreciable increases in endurance for specimens tested in vacuo relative to those tested in air. Nachtigal et al[30] studied the fatigue behavior of a cobalt-base alloy and a nickel-base alloy at 1500°F (816°C) and found the fatigue life in vacuo to be greater than in air under similar conditions. White[3], studying 1/2% Mo steel at 932°F (500°C) observed similar results, noting that under conditions of continuous cycling the behavior in an elevated temperature vacuum were comparable to that in room temperature air. White obtained essentially identical fatigue lives for specimens continuously cycled in vacuo at 932°F (500°C) as for those continuously cycled at room temperature. Tests conducted in vacuo at 932°F (500°C) with a 30 minute hold-time exhibited somewhat shorter lives than those continuously cycled under identical conditions, but the lives were still considerably longer than for tests with a 30 minute hold time in air at 932°F (500°C). White observed light oxidation on the specimens tested in vacuo at elevated temperature, and it is therefore possible that environmental attack could account for at least part of the difference in fatigue lives between in vacuo tests with continuous cycling and those with the hold-times. White concluded that, for the conditions he studied, oxidation appeared to have a greater influence on fatigue behavior than creep.

Soloman and Coffin[7] have shown that A286 steel tested in vacuo at 1100°F (593°C) exhibits (at a given frequency for frequencies less than 1 cpm) crack growth rates about 2 orders of magnitude lower than those observed in air at the same temperature. Their tests in vacuo did exhibit a frequency dependence at low frequency, presumably due to the presence of a creep interaction, but it should be pointed out that in the low frequency regime, the aggressive action of the air

environment was 100 times more damaging than was creep. It is often observed in fatigue tests at high temperatures that as the frequency is decreased (especially in conjunction with a tensile hold-time) the mode of crack extension changes from transgranular to intergranular. Woodford and Coffin[31] have shown that, for A286 steel tested at 1100°F (593°C), the principal cause for this transition to intergranular crack propagation is one of environment, rather than creep.

Mahoney and Paton[5] tested Types 316 and 321 stainless steels in air, nitrogen, and argon environments at 1200°F (649°C). They found that the crack growth rates observed in the inert environments (argon and nitrogen) were approximately equal to those observed in a room temperature air environment, although for Type 316 the crack growth rate in air at 1200°F (649°C) was a factor of 22 higher than in room temperature air. Again, the conclusion reached was that the environmental interaction with the crack front was primarily responsible for the increase in crack growth rates with increasing temperature in air.

The results shown in Figures 3 and 4 indicate that in the equally-inert environments of liquid sodium and vacuum at both elevated test temperatures, the fatigue-crack growth rate is approximately equal to that observed in room temperature air, indicating once more that that thermal-activation disappears when the corrosive environment is removed.

When the observations of these previous investigators are combined with the results from the vacuum and sodium environments of the present study, one point becomes quite clear: the thermal-activation and time-dependency so frequently observed in elevated temperature fatigue tests conducted in an air environment becomes greatly suppressed, if not eliminated altogether, when the

aggressive environment is removed. The literature abounds with studies, conducted in air environments, which attribute the thermal-activation and time dependency to "creep-fatigue interaction" or to a "creep component", and fatigue design rules for elevated temperatures[32] include creep-fatigue interaction calculations.

A necessary condition for creep is that it be both thermally-activated and time dependent, and a body of data is emerging which shows greatly suppressed thermal-activation and time-dependency when fatigue tests are conducted at elevated temperatures in inert environments. Another phenomenon which is known to interact with fatigue is both thermally-activated[33] and time-dependent[34]: corrosion-enhanced fatigue. The analogies between the large amount of data that has been generated on corrosion-enhanced fatigue at room temperature and crack propagation at elevated temperatures in air have been discussed in a recent paper by Speidel[35], and Speidel points out again that much of the latter should be considered as corrosion fatigue.

It should be quite apparent that much of the thermally-activated time dependent behavior that has been attributed in the past to a creep-fatigue interaction should more properly be attributed to a corrosion-fatigue interaction. The situation, however, is not at all simple since creep probably does interact with fatigue to some degree, especially above about one-half the absolute melting temperature. The effect of environment upon creep or stress-rupture behavior is not even simple, with some studies indicating a greater creep resistance in vacuo relative to in air[30], others showing lower resistance in vacuo[36] or mixed results depending upon stress level[37,38], while still others show no difference[39].

The discussion on the interaction between creep and fatigue can be approached from a different direction: examining the fatigue-crack propagation behavior in

material that has received prior creep damage. As previously discussed, Figures 5 and 6 reveal little or no effect of prior creep damage for either mode of crack extension (predominately transgranular or predominately intergranular) when compared to the behavior of specimens receiving no creep damage. One might not expect to see a large effect of prior creep damage upon the fatigue-crack growth behavior at a moderate frequency when the mode of cracking is predominately transgranular such as for Specimen 226 (Figure 5). However, if an interaction between prior creep damage and fatigue-crack growth behavior were going to be readily apparent, it should be at a low cyclic frequency coupled with a tensile hold-time such as with Specimen 384. The fact that there is no observable effect, at least for prior creep damage up to 50% of the creep strain to fracture, suggests that creep and fatigue may be generally separate phenomena that interact with one another only to a limited extent. In the limit of massive creep damage, of course, there must be some interaction between a very high density of grain boundary creep cracks, and a predominately intergranular fatigue crack. One might expect that the linking up of the creep cracks with the fatigue crack would tend to increase the macroscopically-observed fatigue-crack growth rate.

CONCLUDING REMARKS

The present study has attempted to examine the interaction between creep and fatigue from two different standpoints: by studying the fatigue-crack growth behavior when the aggressive air environment is removed, and by examining the fatigue-crack growth behavior in material which has been subjected to reasonably severe prior creep damage. The experimental results presented in this study taken by themselves would probably not be totally convincing of the lack of a significant interaction between creep and fatigue. They do, however, add further to the growing body of data which suggests that such interactions have been greatly over-estimated in the past.

This study (and others) has shown that when fatigue tests at elevated temperatures are conducted in an inert environment, the thermal-activation and time-dependency which is a necessary prerequisite for creep is greatly suppressed if not eliminated. In addition, a reasonably severe amount of prior creep damage had no apparent effect upon the rate of fatigue-crack propagation.

Creep and fatigue probably do interact, perhaps even to a significant degree, at temperatures in excess of about one-half the absolute melting point of the alloy. However, most structural components (especially pressure vessels and piping) are generally not designed to operate for any appreciable length of time in this temperature regime. Creep and fatigue probably also interact at temperatures below about one-half the absolute melting point of the material, but the data tend to show that the effect is not large. Creep cracks and fatigue cracks probably do interact with one another, especially when the density of the creep cracks becomes very high. Again, however, structural components such as pressure vessels and piping are generally not designed to sustain such massive creep damage.

The author realizes that the work presented in this paper does not solve the problem of quantifying creep-fatigue interaction--it may even raise more questions. The problem is quite complex and neither simplistic view of all creep-fatigue interaction or no creep-fatigue interaction appears to be correct. However, there is emerging a body of data which suggests that, at the temperatures where pressure vessels and piping are used in practice, the interaction is minimized. The impact of this view upon presently used design rules is quite obvious.

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TABLE I

HEAT 55697 CHEMICAL COMPOSITION (PERCENT BY WEIGHT)

C	Mn	P	S	Si	Cr	Ni	Cu	Mo	Pb	Ca	Sn	Ti	N	Total Rare Earths
0.053	0.87	0.019	0.010	0.49	18.25	9.51	0.21	0.18	<0.001	0.11	0.006	<0.01	0.030	<0.001

TABLE II

HEAT 55697 ROOM TEMPERATURE MECHANICAL PROPERTIES

0.2% Yield Strength	Tensile Strength	Percent Elongation	Percent R.A.	Hardness R_B	Grain Size ASTM
39,600 psi 273.0 MPa	77,050 psi 531.2 MPa	65.0	60.7	73.2	4.5

TABLE III

SUMMARY OF TEST PARAMETERS

Specimen Number	Specimen Design	Thermal Aging Prior To Fatigue Test	Applied Stress During Aging	Cyclic Freq.	Wave-Form	Data Ref.
58	C.S.	None	None	0.083 cpm	Square	Ref. 11
62	C.S.	None	None	0.083 cpm	Square	Ref. 11
67	C.S.	None	None	0.083 cpm	Square	Ref. 11
226	C.C.	3000 hr. @ 1000°F	23,475 psi	40 cpm	Sawtooth	
228	C.C.	3000 hr. @ 1000°F	None	40 cpm	Sawtooth	
308	C.S.	3000 hr. @ 1000°F	None	40 cpm	Sawtooth	Ref. 15
319	C.S.	3000 hr. @ 1200°F	None	0.083 cpm	Square	Ref. 15
384	C.C.	3000 hr. @ 1000°F	23,475 psi	0.083 cpm	Square	

C.S. = ASTM Compact Specimens

C.C. = Center-cracked specimen

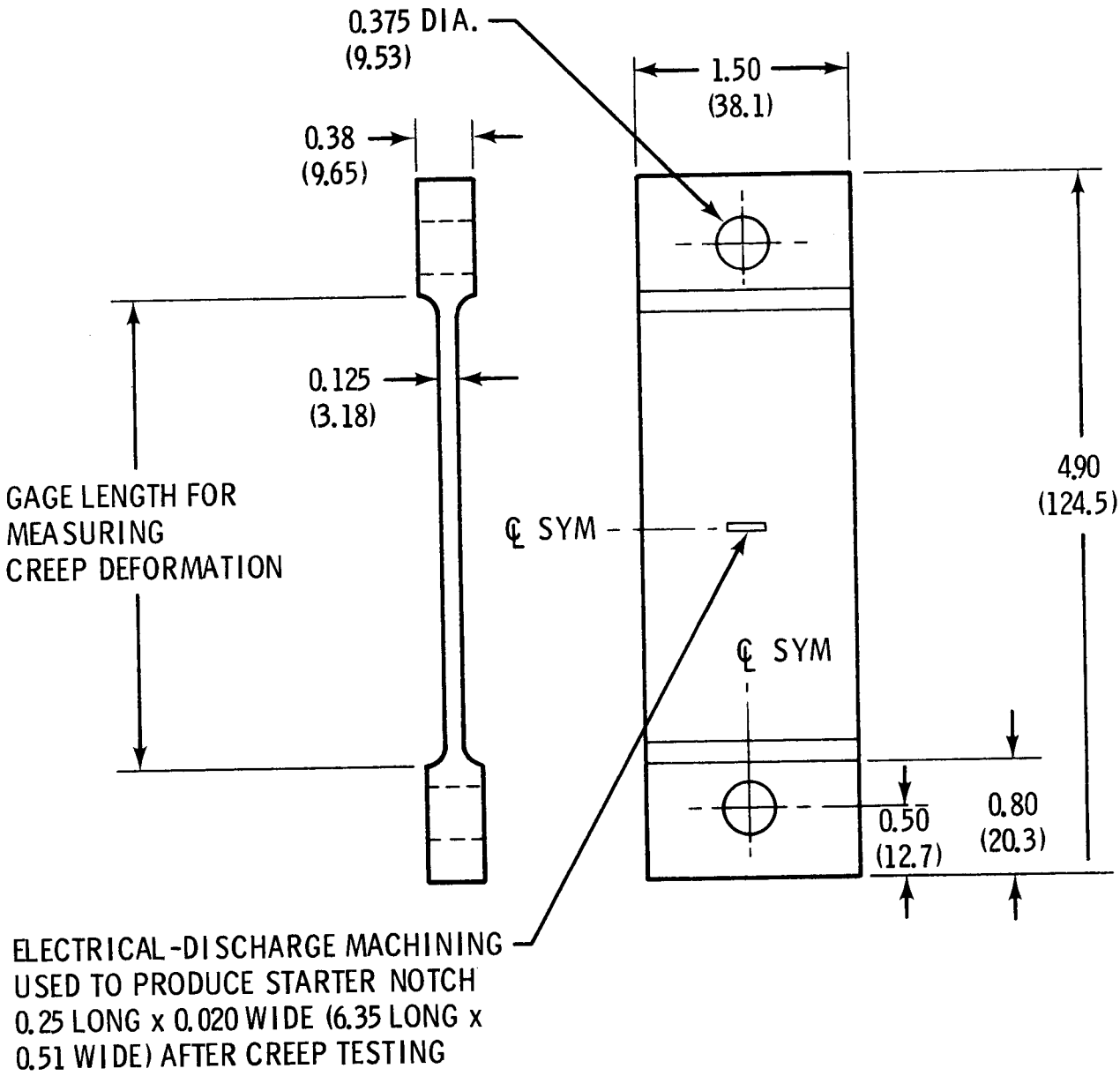


FIGURE 1. Specimen employed for creep-fatigue tests. Dimensions are in inches (millimeters).

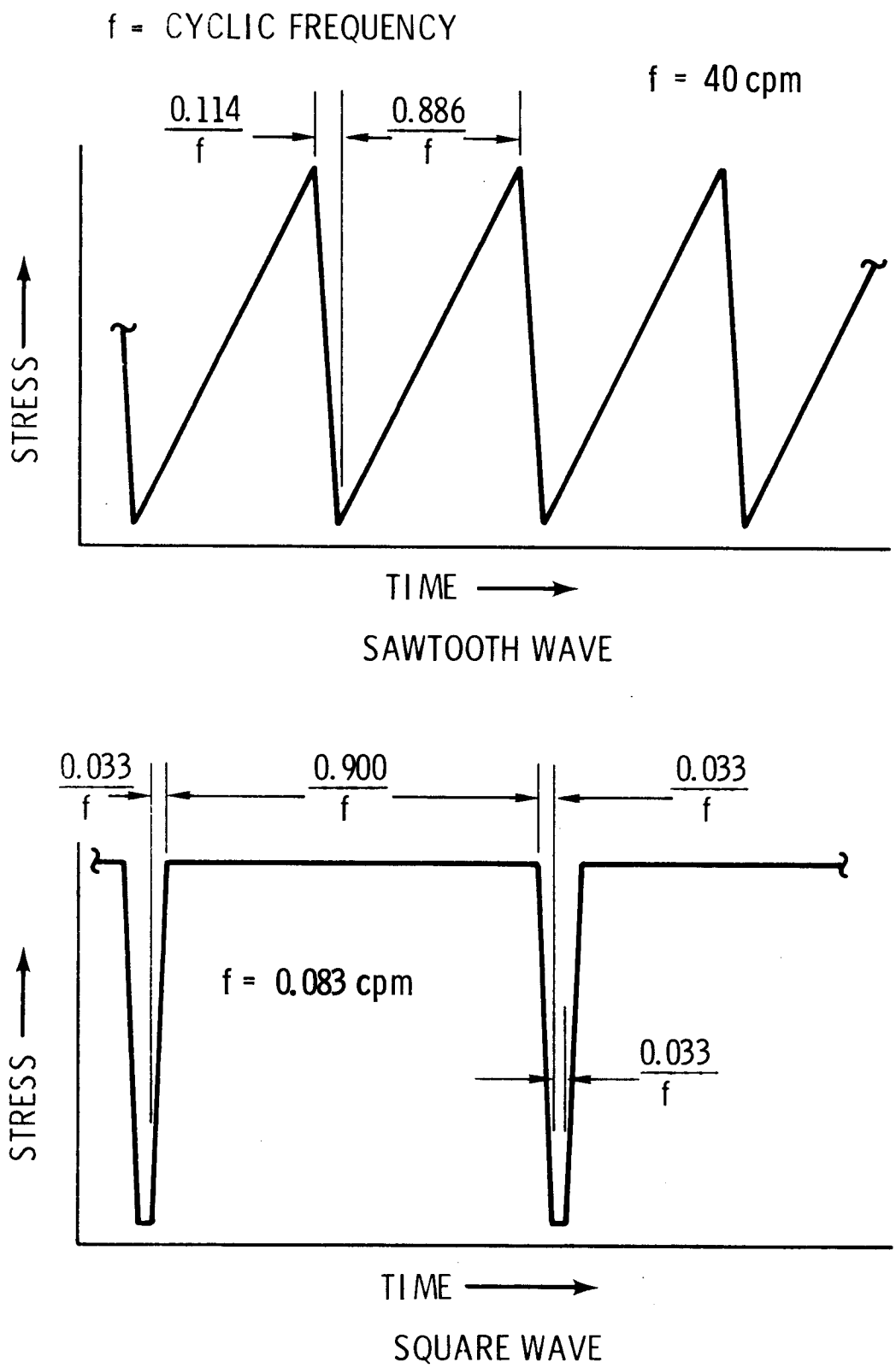


FIGURE 2. Loading waveforms employed in creep-fatigue tests.

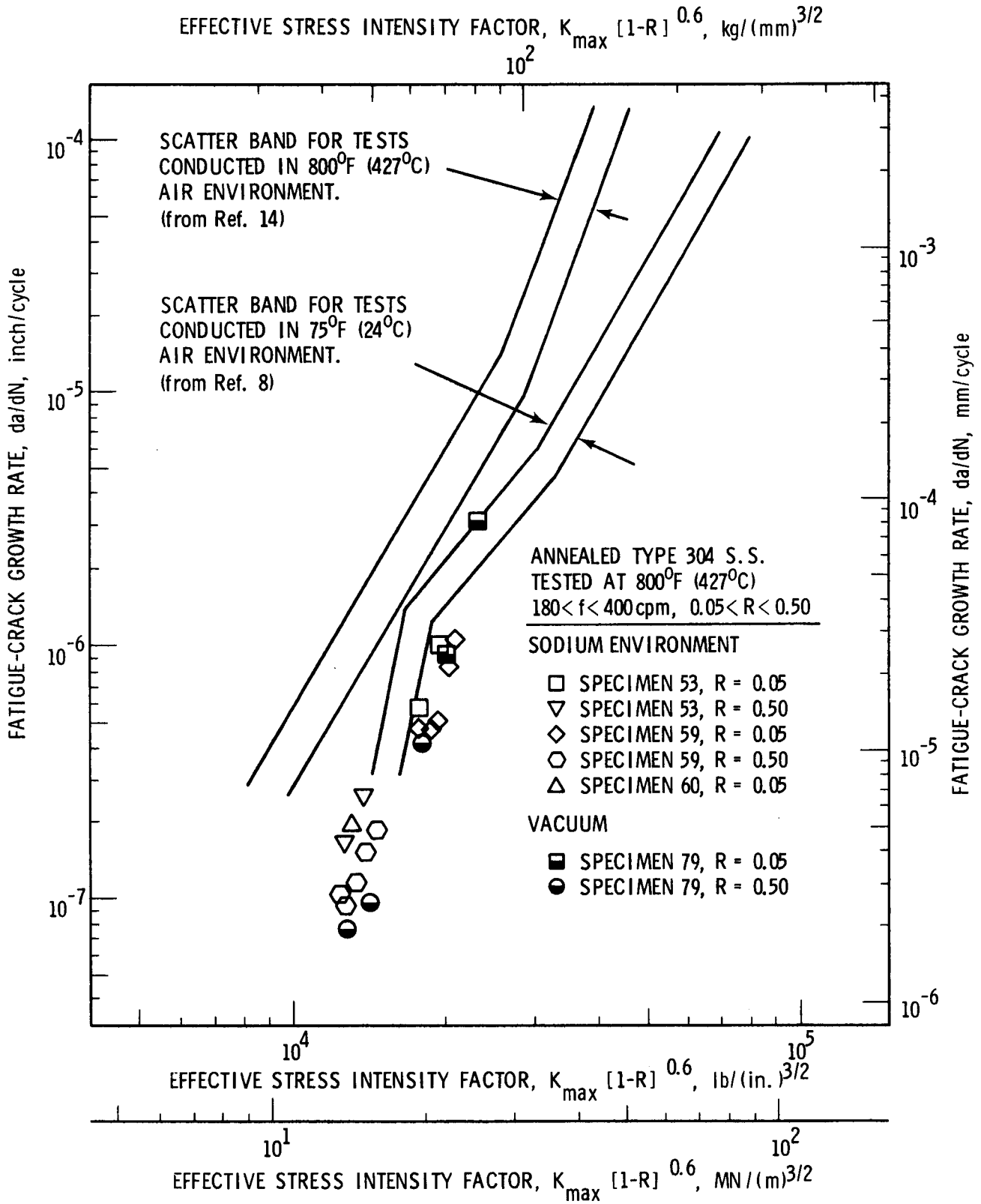


FIGURE 3. Fatigue-crack growth behavior of Type 304 S.S. in sodium, vacuum, and air environments at 800°F (427°C). (from Ref. 6).

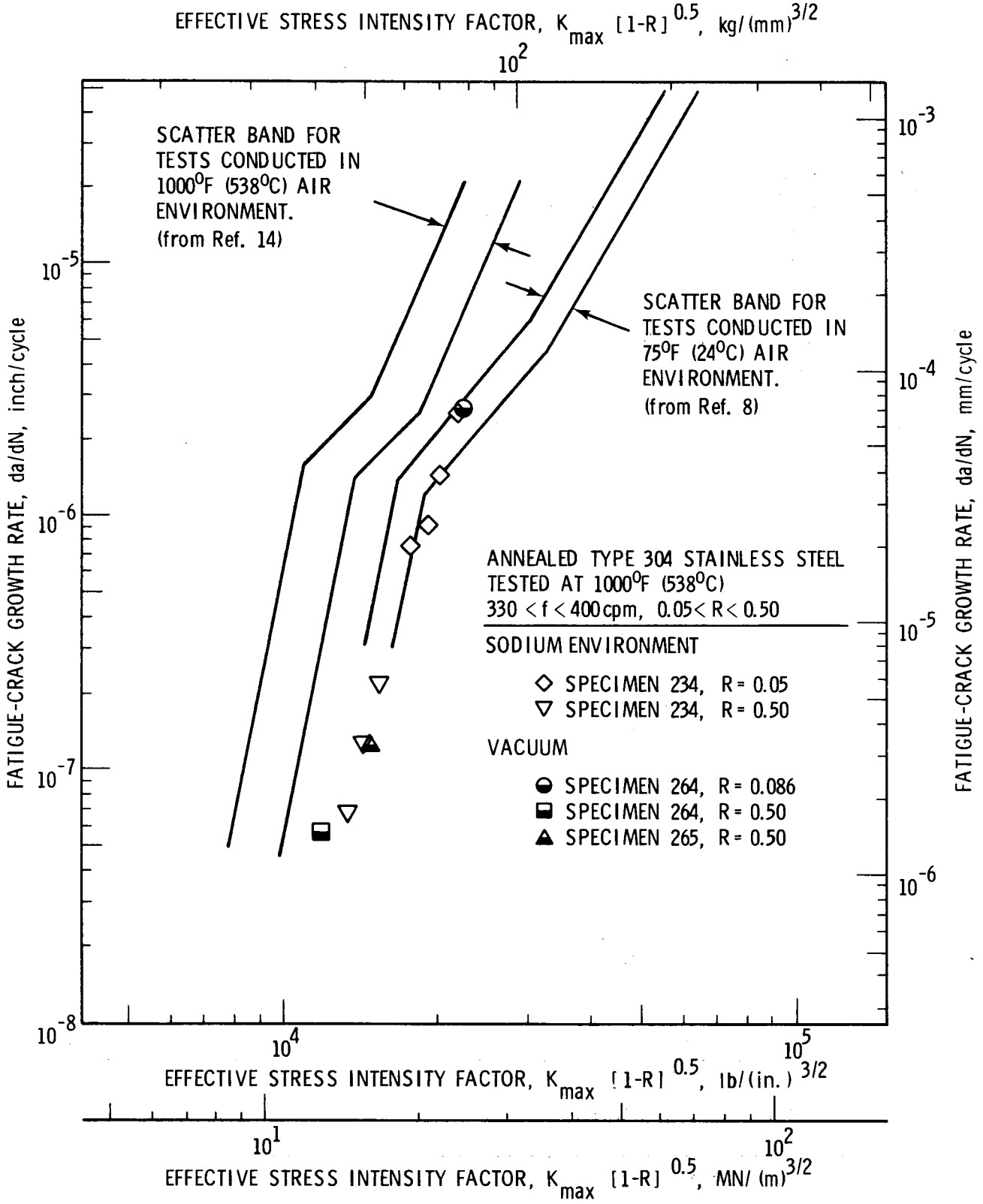


FIGURE 4. Fatigue-crack growth behavior of Type 304 S.S. in sodium, vacuum, and air environments at 1000°F (538°C). (from Ref. 6).

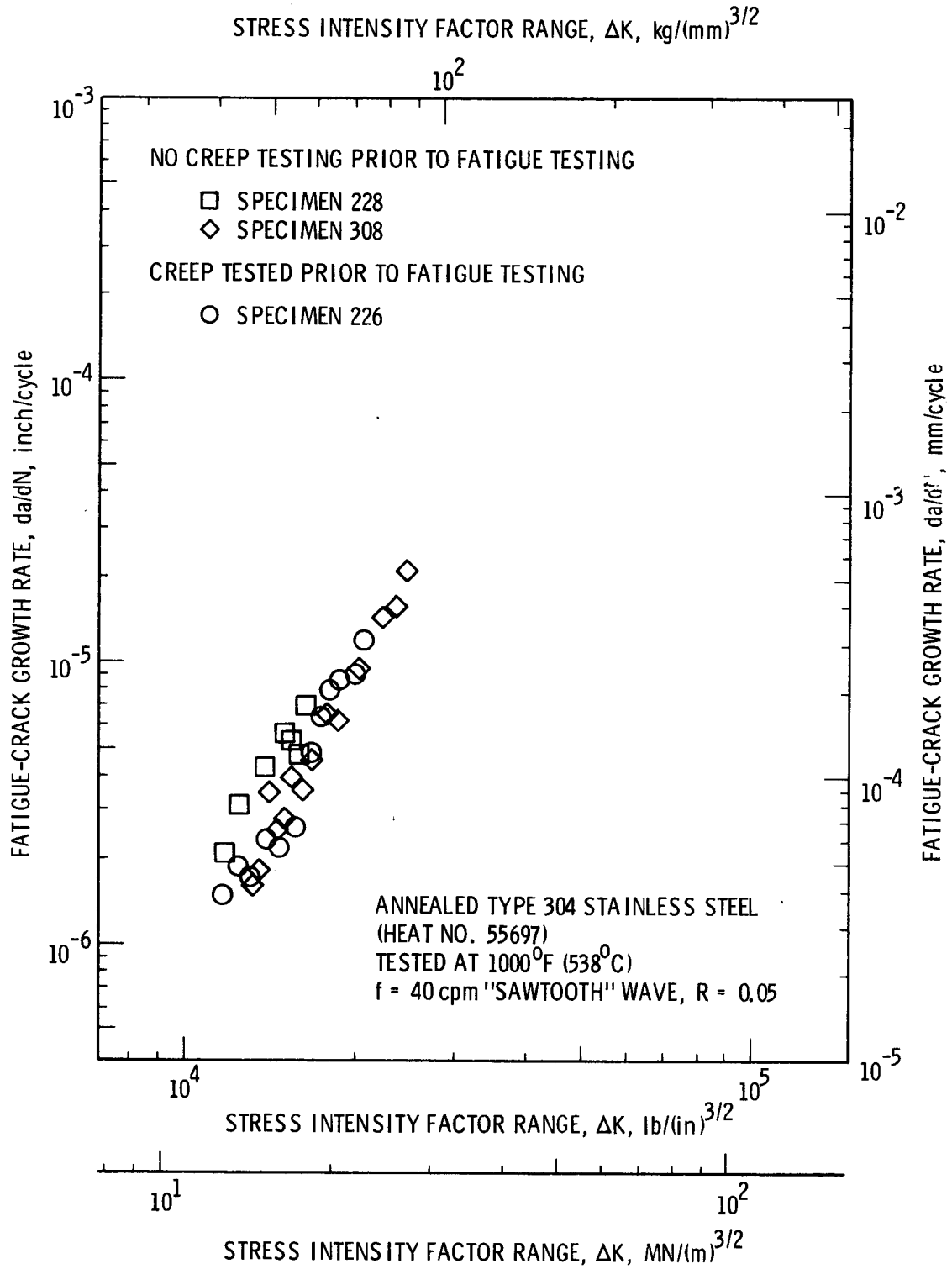


FIGURE 5. Comparison of the fatigue-crack growth behavior of specimens subjected to prior creep damage with those receiving no prior creep damage. (Predominately transgranular cracking).

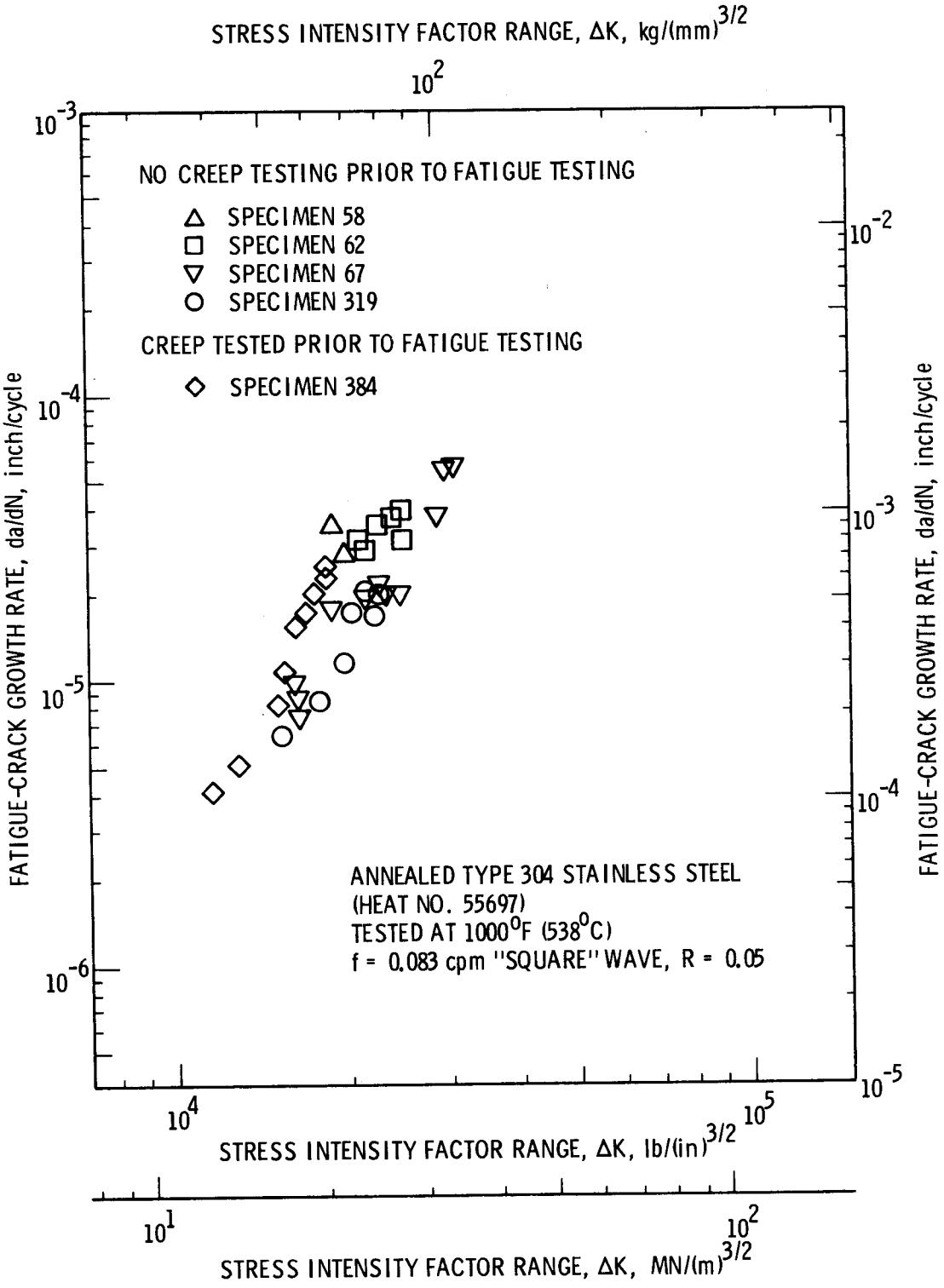


FIGURE 6. Comparison of the fatigue-crack growth behavior of specimens subjected to prior creep damage with those receiving no prior creep damage. (Predominately intergranular cracking).

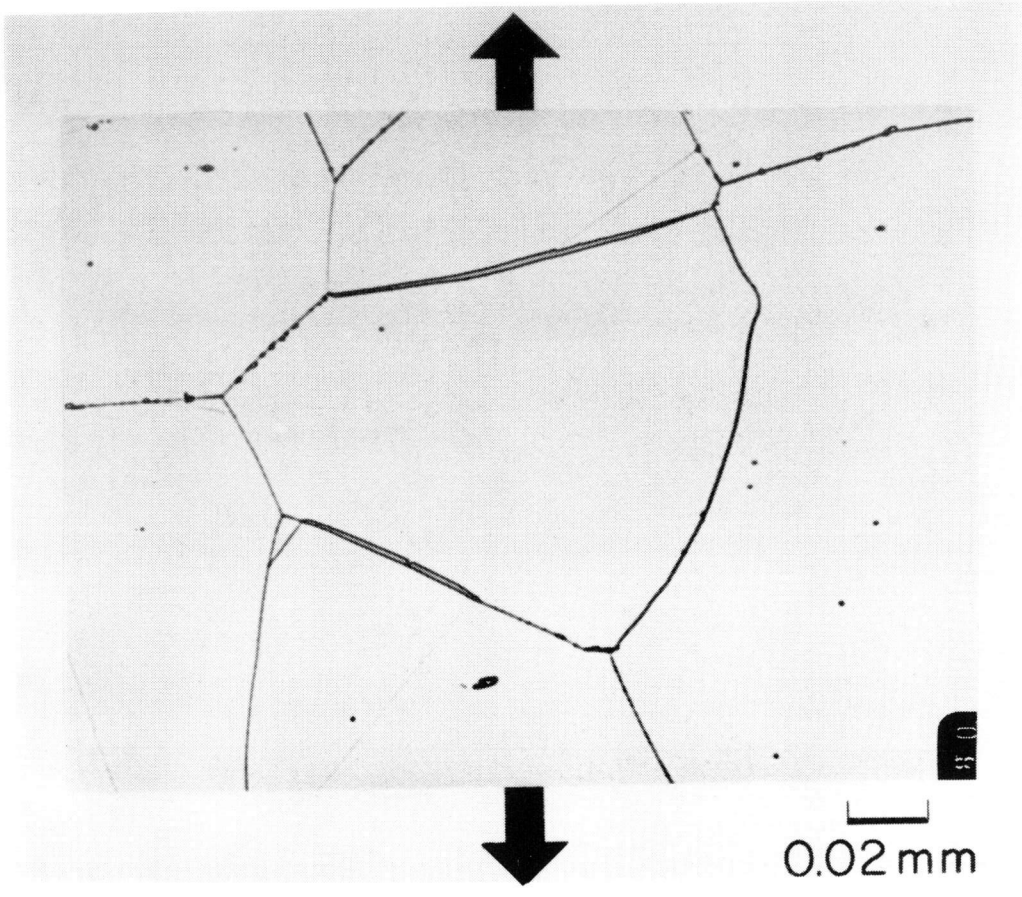


FIGURE 7a. Typical grain boundary cracks from area remote from crack tip. Arrows indicate direction of creep loading. (Specimen 226).

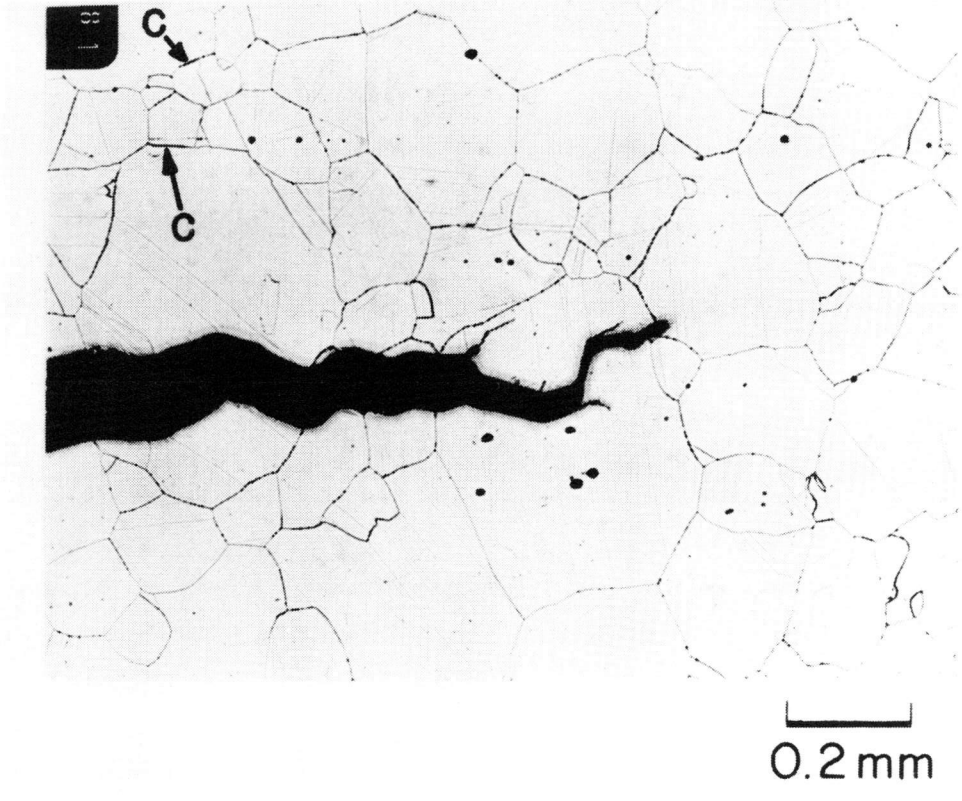


FIGURE 7b. Predominately transgranular fatigue-crack extension in Specimen 226. Grain boundary cracks are identified with "C".

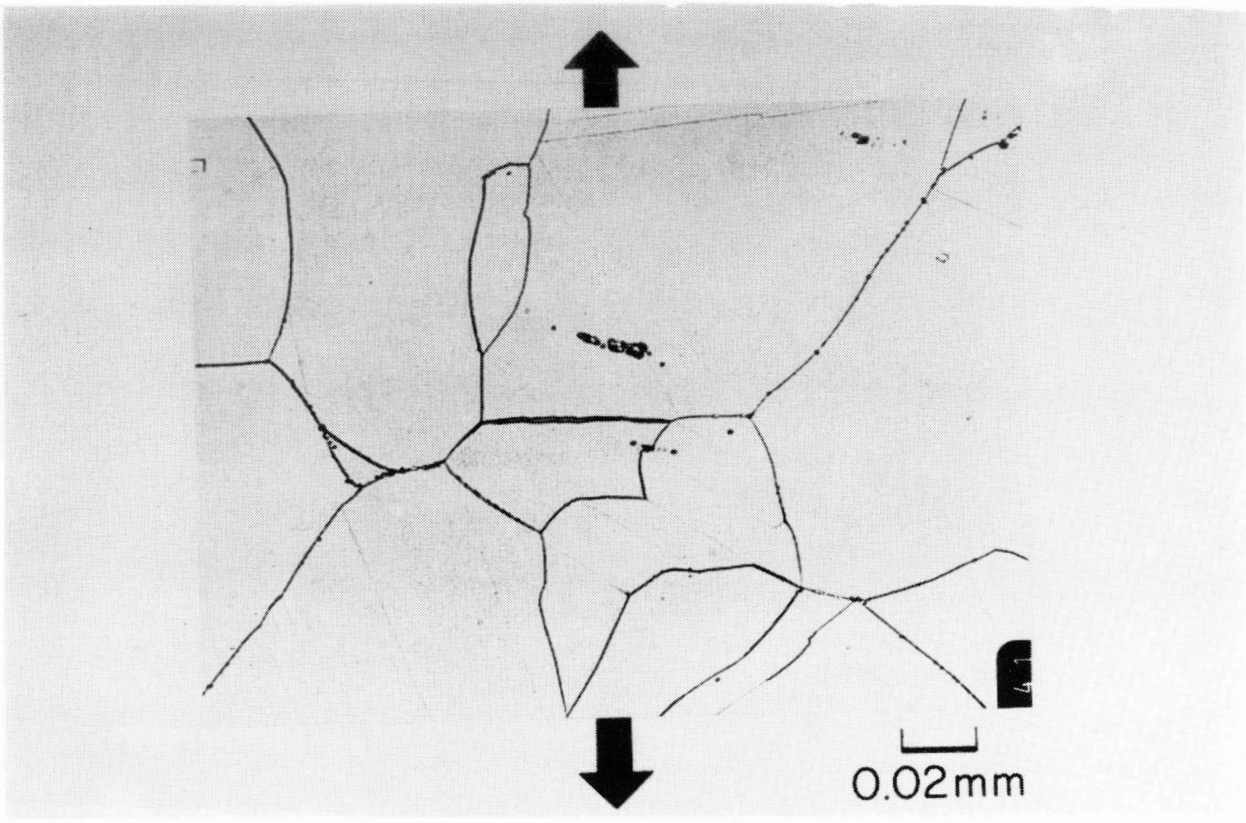


FIGURE 8a. Typical grain boundary crack from area remote from crack tip. Arrows indicate direction of creep loading. (Specimen 384).

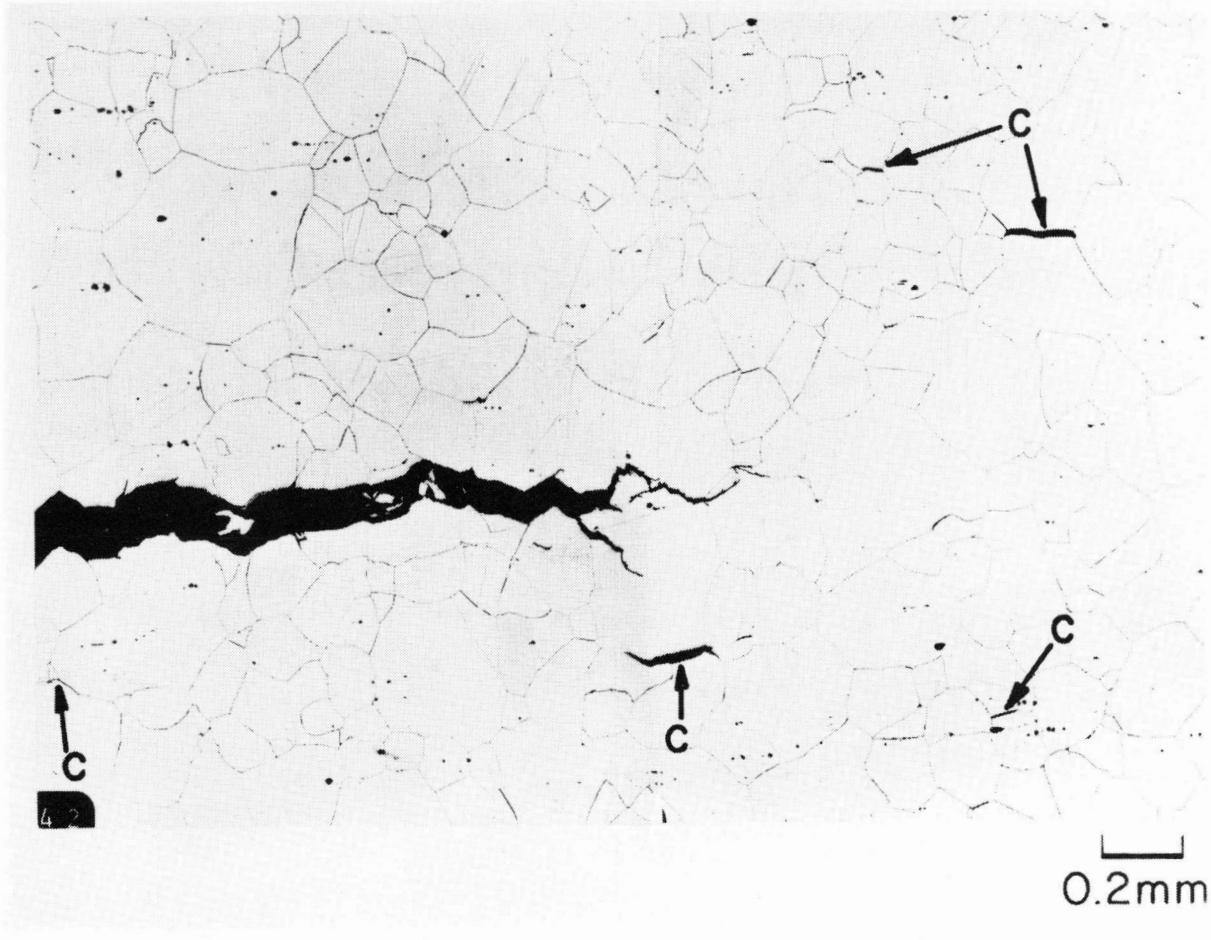


FIGURE 8b. Predominately intergranular fatigue-crack extension in Specimen 384. Grain boundary cracks are identified with "C".

APPENDIX

ESTIMATION OF THE EXTENT OF CREEP DAMAGE

Specimens number 226 and 384 were creep tested 3000 hours at 1000°F (538°C) at a stress level of 23,475 psi (162 MPa). The stress-rupture behavior of this heat of material is shown in Figure A-1. As will be seen, 23,475 psi (162 MPa) represents 75 percent of the stress to rupture for 3000 hours at 1000°F (538°C).

As previously mentioned, measurements were made across the shoulders of the specimens (see Figure 1) before and after creep testing. The total strain measured (in the unloaded state) at the completion of creep testing was 3.2 percent and 3.9 percent for Specimens 226 and 384, respectively. In order to assess the extent of creep damage, it is helpful to examine a schematic strain-time curve for a creep specimen. Such a curve is illustrated in Figure A-2. ε_R is defined as the total strain at rupture, or at some time prior to rupture, ε_L as the time-independent strain imposed upon loading, and the quantity $(\varepsilon_R - \varepsilon_L)$ as the creep strain.

With these definitions in mind, we see from Figure A-3 that the creep strain at fracture in this heat of material at 3000 hours is approximately 5 percent. Figure A-4 is a stress-strain curve for this heat of material which shows that the time-independent strain (the quantity ε_L) imposed upon loading a specimen to 23,475 psi (162 MPa) is approximately 1.4 percent. The calculations of creep damage are as follows:

- Specimen 226 had 3.2% total strain after creep testing.

  Creep strain after testing = $(\varepsilon_R - \varepsilon_L) = 3.2 - 1.4 = 1.8\%$.

  Creep strain = 5.0% at rupture.

  Therefore creep damage $\approx$ 1.8/5.0 = 36% of the creep strain to fracture.

- Specimen 384 had 3.9% total strain after creep testing.

  Creep strain after testing = $(\varepsilon_R - \varepsilon_L) = 3.9 - 1.4 = 2.5\%$

  Creep strain = 5.0% at rupture.

  Therefore creep damage $\approx$ 2.5/5.0 - 50% of the creep strain to fracture.

Hence, we see that prior to fatigue testing, Specimen 226 was subjected to approximately 36 percent of the creep strain to rupture and Specimen 384 was subjected to approximately 50 percent of the creep strain to rupture.

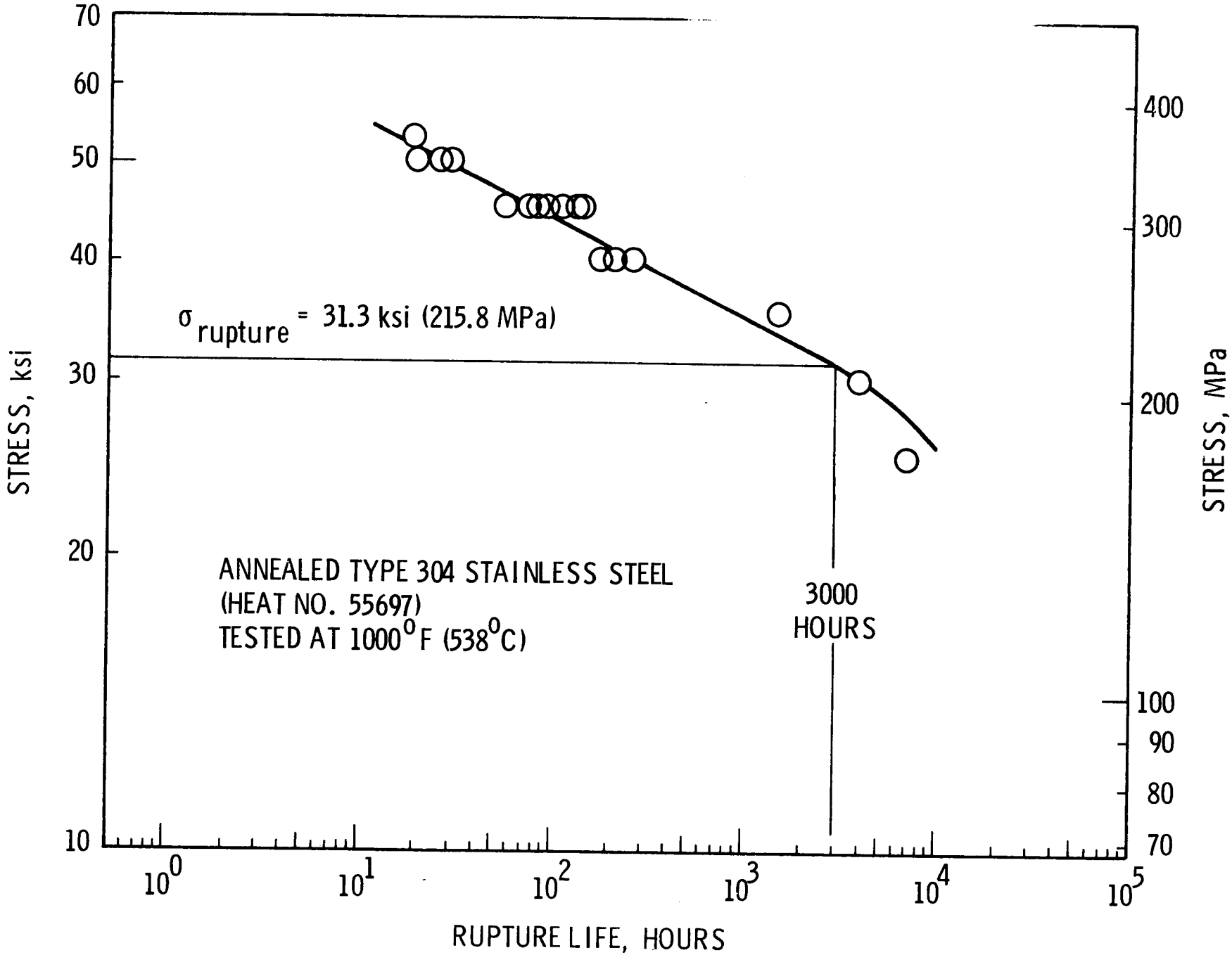
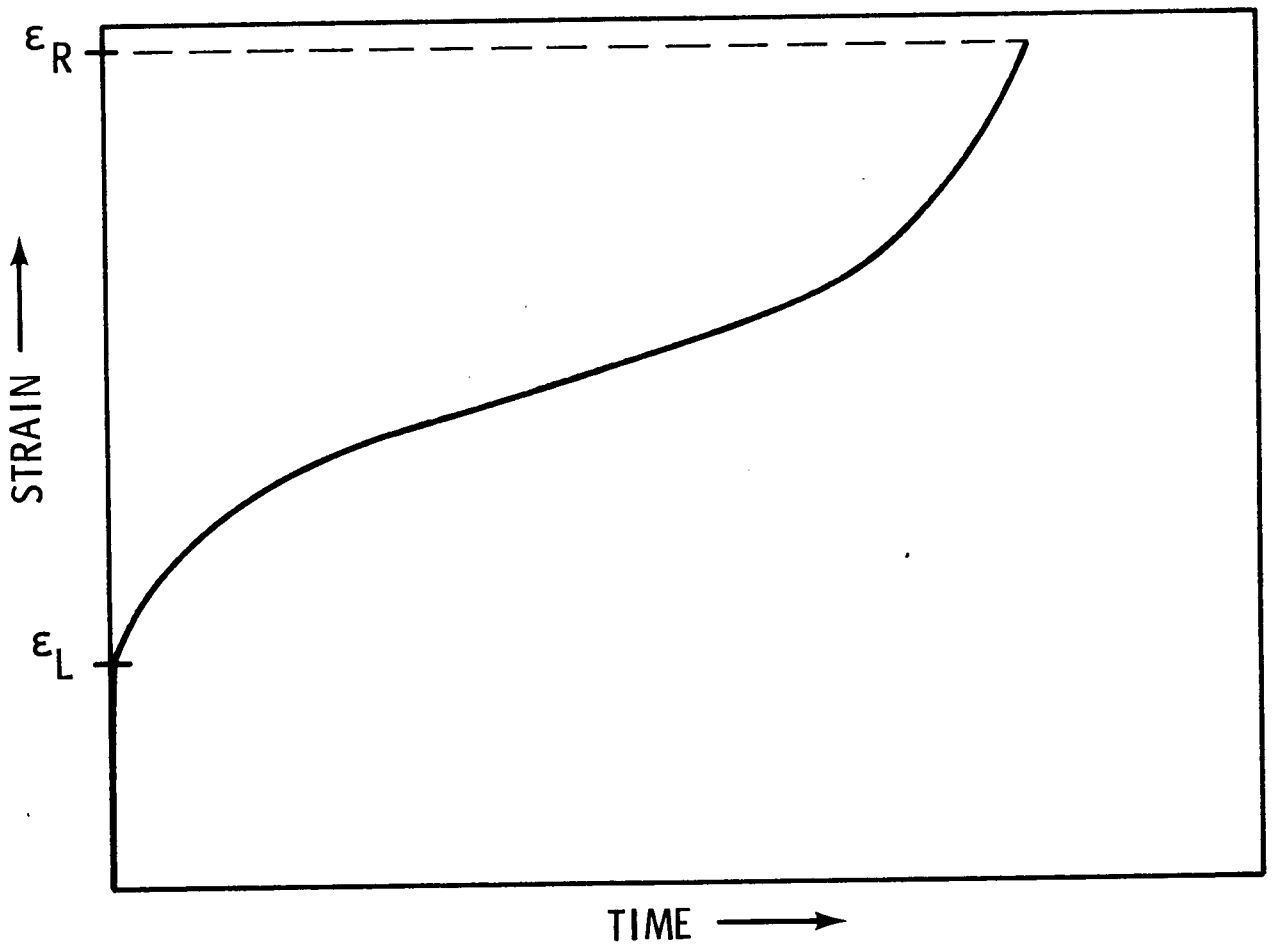


FIGURE A-1 Stress-rupture behavior of Heat 55697 at 1000°F (538°C).



ε_R = TOTAL STRAIN AT RUPTURE

ε_L = TIME-INDEPENDENT STRAIN IMPOSED ON LOADING SPECIMEN

$(\varepsilon_R - \varepsilon_L)$ = CREEP STRAIN

FIGURE A-2 Schematic strain-time diagram for a stress-rupture test.

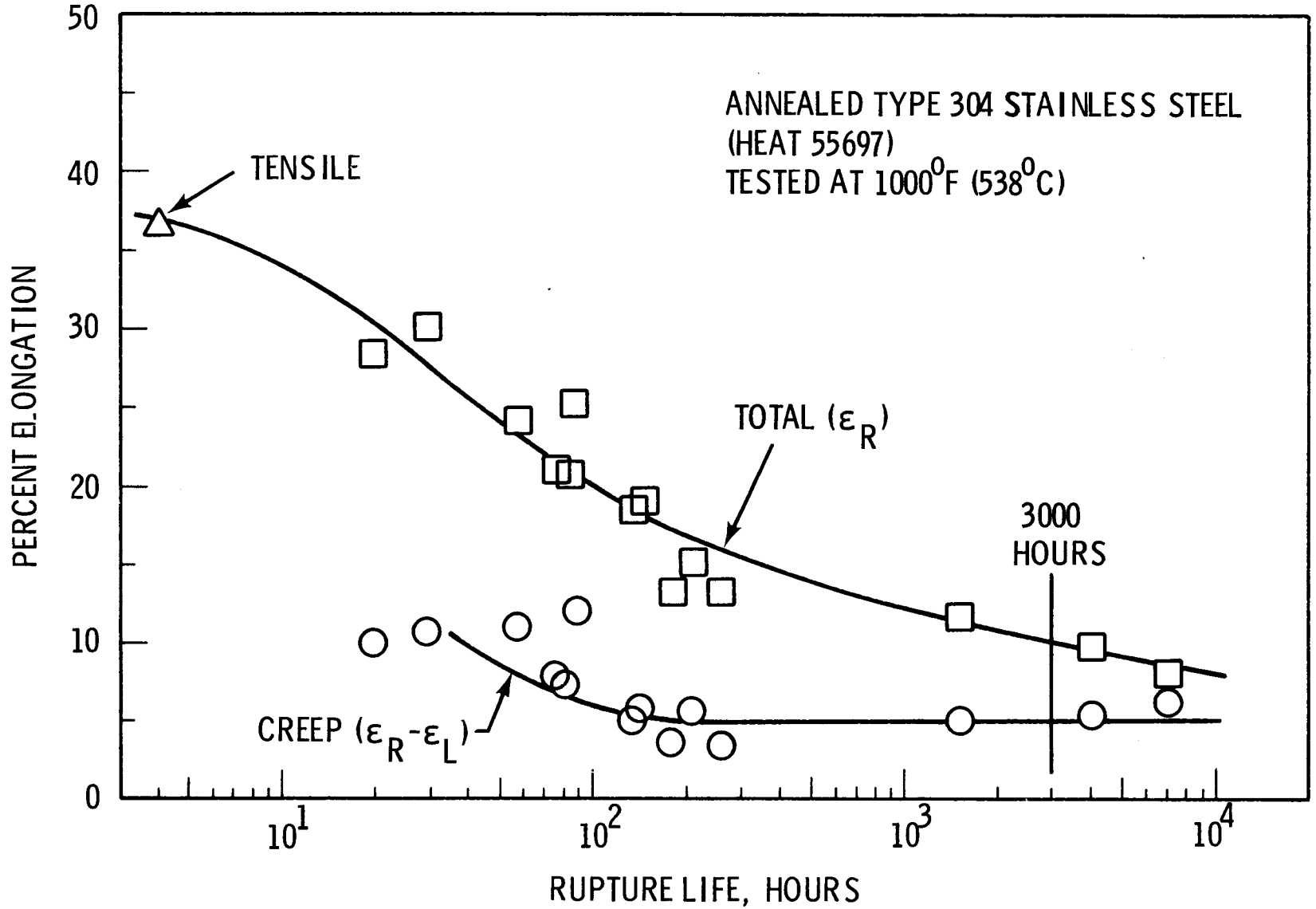


FIGURE A-3 Total strain and creep strain for Heat [illegible] tested at 1[illegible] (538°C).

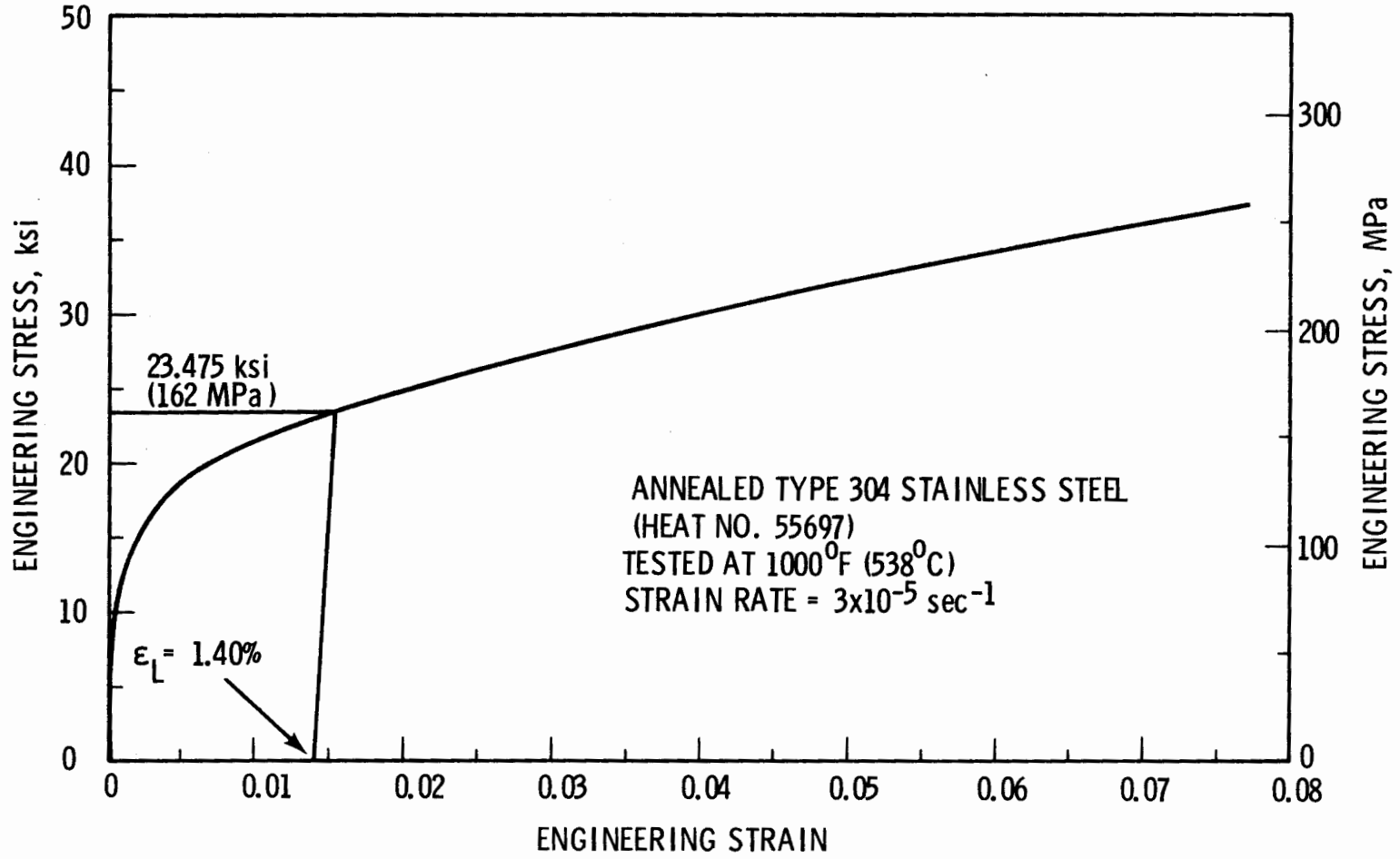


FIGURE A-4 Stress-strain curve for Heat 55697 tested at 1000°F (538°C).