ELEVATED TEMPERATURE FATIGUE OF 2 1/4 CR-1 MO STEEL

K. D. Challenger

Mechanical Engineering Department, Naval Postgraduate School, Monterey, CA 93943, USA

ABSTRACT

The effect of the load wave form on the elevated temperature fatigue life of 2 1/4 Cr-1 Mo steel in air is shown to be best explained by oxidation-fatigue interaction mechanisms. Data are presented that indicate that the form of oxide cracking that occurs is dependent on the loading wave form and that fatigue crack initiation is dependent on the mode of oxide cracking. Fatigue crack initiation is shown to be influenced more by oxidation than is fatigue crack propagation. The effect of oxidation on fatigue crack propagation depends on the time between load reversals, independent of loading wave form and this effect saturates very quickly for temperatures around 525°C. A worst-case predictive equation is presented that assumes that fatigue endurance is dependent only on crack propagation, i.e., initiation of a crack in oxidizing conditions is assumed to occur with only a few cycles.

KEYWORDS

Fatigue; creep-fatigue interaction; oxidation; ferritic steel; steam generators

INTRODUCTION

Fatigue resistance at elevated temperatures involves the complex interaction between fatigue damage, creep damage and oxidation in both the initiation and growth stages of a crack. The development of design correlations for elevated temperature fatigue has, for the most part, assumed that oxidation effects are negligible and that the crack initiation stage can be combined with the crack propagation stage. Thus, elevated temperature fatigue is modeled as creep-fatigue interaction for total endurance life. This has worked reasonably well for the austenitic stainless steels, but has proved inappropriate for 2 1/4 Cr-1 Mo steel.

2 1/4 Cr-1 Mo steel enjoys a wide spread industrial use, operating in air, steam, corrosive gases, impure helium and sodium at temperatures up to 600°C with both sustained and cyclic loads. Its expected use as the structural material for the steam generating plant of liquid metal cooled fast breeder reactors has led to the generation of vast amounts of mechanical property data (see for example Brinkman 1976, 1980). The fatigue data at elevated temperatures exhibits some unexpected characteristics which have led to differences in the interpretation of these data among researchers. The purpose of this paper is to review these data for 2 1/4 Cr-1 Mo, present some new information and summarize our understanding of the damaging mechanisms active during cyclic deformation of this alloy at elevated temperatures.

TOTAL ENDURANCE DATA

For temperatures up to 600°C and at total strain ranges >1% the fatigue life is rather insensitive to the loading wave form, i.e., very little effect of testing frequency or the length of constant strain dwell periods. However, for total strain ranges <1% very large reductions in cyclic life occur when dwell periods are introduced at the compressive strain limit (C); the cyclic life is only slightly reduced with dwells at the tensile strain limit (T); but, combined T + C dwells result in the shortest cyclic life of all loading wave form tested (Brinkman 1976, 1980). Examples of these data are shown in Figs. 1. This is in contrast to the behavior of the austenitic stainless steels (Brinkman 1980) and 1CrMoV steel at 565°C (Ellison 1976) where the shortest lifetimes result when T only dwells are imposed and combined T + C dwells increase the lifetime over T only dwell periods.

For 2 1/4 Cr-Mo steel, less oxidizing environments increase the cyclic lifetime especially at strain ranges below 1%. This has been shown for impure helium (Brinkman 1980), and sodium (Chopra 1980); in fact, T only dwell periods become more damaging than C only dwells in these environments. However, in the impure He(1500/500/50/50 μatm. $H_2/CO/CH_4/H_2O$) some oxidation occurs and C only dwells do reduce the cyclic life, but less than that which occurs in air and T + C dwells are still the most damaging wave form studied (Brinkman 1980, 1983). These results suggest that an oxidation-fatigue rather than a creep-fatigue interaction is controlling the cyclic life in air for 2 1/4 Cr-1 Mo. Tests with dwell periods introduced at zero stress were performed in order to minimize creep damage, but allow oxidation damage to occur (Challenger 1981). The cyclic life was decreased by this zero stress hold by the same amount as that which resulted from C only constant strain dwells provided the zero stress dwell followed the compressive strain limit, see Fig. 1. However, the cyclic life resulting from zero stress dwells following the tensile strain limit is about the same as that resulting from dwells at the tensile strain limit. These results provide additional evidence of an oxidation-fatigue interaction rather than creep-fatigue interaction.

In earlier papers (Challenger 1981a, 1981b), we have shown that in order to successfully model, or predict, the total endurance fatigue, the effects of loading wave form (dwell periods) on crack initiation must be separated from the effects on crack propagation. Further, from a design analysts point of view, the total failure process should be treated as at least two distinct steps occuring in series:

1) Crack initiation
2) Short crack growth in a plastically cycled material

The remainder of this paper discusses the effects of loading wave form on these two steps.

FATIGUE CRACK INITIATION

In air at temperatures in excess of about 427°C, the oxidation rate of 2 1/4 Cr-1 Mo occurs at a rate such that quite quite short dwell periods, e.g., 0.01 hrs., will result in the formation of oxide during the test that will be thick enough to be cracked by the strain ranges used in laboratory tests ($\Delta\varepsilon$>0.35%). The manner in which the oxide cracks depends on the loading wave form (Challenger 1983; Teranishi 1979); compression strain holds or the zero stress hold periods following the maximum compressive strain, result in circumferential oxide cracks and the oxide remains adherent to the specimen surface. Continuous cycling and tensile strain dwells result in both circumferential and longitudinal oxide cracks that produce a spalling of the oxide. The circumferential oxide cracks that occur with compression dwells result in rapid fatigue crack initiation and/or preferential oxidation at the end of the oxide crack that results in an oxide-filled crack penetrating perpendicular to the applied load. This effect is best seen by sectioning a sample in the manner illustrated in Fig. 2; thus, geometrically enlarging the metal-oxide interface. Figure 2 compares the metal-oxide interface for a continuously cycled specimen to one with an 0.01 hr. hold period at the compressive strain limit. Note that the cracks in the oxide of the sample tested with compressive dwells coincide with the oxide filled cracks penetrating into the sample; whereas, the metal-oxide interface remains smooth for the continuously cycled test.

The different oxide cracking behavior has been explained by Challenger (1981). The oxide will crack perpendicular to the maximum tensile strain experienced by the oxide during the test. With a compressive strain dwell (or a zero stress dwell following the compressive strain limit) the sample spends most of its time in compression and since the oxide grows on the compressed sample, the "stress-free" state for the oxide is at the compressive strain limit. Thus, the tensile strain experienced by the oxide is equivalent to the total strain range and this tensile strain occurs in a direction parallel to the longitudinal axis of the specimen resulting in the circumferential oxide cracks. For tensile strain dwell tests the reverse is true, the maximum tensile strain experienced by the oxide will be $\nu\Delta\varepsilon_T$ (ν is Poisson's ratio) and it will exist in a circumferential direction resulting in the longitudinal cracks observed for these tests.

In order to further study the effect of oxide cracks on fatigue crack initiation, several uniform gage samples were tested by first heating them to 538°C, loading to a strain of either +0.002 or -0.002 and held for 250 hrs. (This would produce an oxide thick enough that it would be certain to fracture at a tensile strain of less than 0.004 (Challenger 1981b). After 250 hrs. the test was begun by starting the cycling loading in a direction that would result in a total strain range of 0.004 cycling about the initial zero strain point of the specimen. Continuous cycling at a strain rate of $4x10^{-4}$ sec^{-1} was continued until fracture occurred. The resulting fatigue life is compared to an identical test that did not receive the pretest 250 hr. hold period in Table 1.

Circumferential oxide cracks were observed to form during the tensile side of the first cycle for both tests held in compression for 250 hrs., but no oxide cracking occurred for either the continuously cycled test or the 250 hr. tensile hold test. One of the 250 hr. compression hold tests was interrupted after 2350 cycles, cooled to room temperature and examined in a low power stereo microscope. Figure 3 shows the specimen surface after 2350 cycles. Note the circumferential striations in the specimen surface. Since the oxide spalled during cooling it was not possible to verify that these striations corresponded to oxide cracks, but there is every reason to believe that they do. The same general trend is observed that is seen when a dwell period is introduced into each loading cycle; e.g., a slight reduction cyclic life with the tensile dwell and large reduction in cyclic life with the compression dwell. The reduction in fatigue life due to the 250 hour dwell at a strain of -0.002 is not as great as that which would be expected from even shorter dwells during each cycle; this is probably due to the effects of the dwell periods on the rate of crack propagation (to be discussed in the next section). However, the result clearly indicates that the mode of oxide cracking reduced the fatigue life and that complex loading wave forms are not a prerequisite for reduced fatigue life due to oxidation.

The total amount of oxidation on these samples is less than those samples tested with dwells during each cycle even though the testing time for these tests, approximately 290 hours for 7,000 cycles including the 250 hour pretest dwell, is much longer than the short, 0.01 hr. hold period tests, ~70 hours. This suggests that the oxidation rate on cyclically deformed material is greater than free surface oxidation, a result reported by Skelton (1978) for 1 Cr Mo V steel.

The fact that the test that was interrupted also experienced a significant reduction in cyclic life indicates that cracks must have been present at the time the test was interrupted. Otherwise, the cyclic life should have been unaffected since the oxide completely spalled-off during the interruption. Note that the continuously cycled test has a cyclic life considerably less than that which would be expected from previous tests and yet the 250 hr. test with pretest hold at a strain of -0.002 failed with approximately the same number of cycles as a test with 0.1 hr. dwell at -0.002 strain imposed during each cycle. The difference in the continuously cycled tests results may be due to a difference in specimen geometry. The tests presented in Table 1 employed uniform gage samples 9.5 mm in diameter and a gage length of 30 mm, while all previous tests used hourglass shaped specimens with a minimum diameter of 6.35 mm. Hourglass shaped specimens consistently exhibit longer cyclic lifetimes than do similarly tested uniform gage samples for this alloy (Brinkman 1983). This may be due to earlier crack initiation in the uniform gage specimens as the crack is not forced to initiate in a specific location. Since oxide cracking is believed to assist crack initiation and if crack initiation in the absence of oxide cracking occurs with fewer cycles in a uniform gage specimen, then the effect of oxide cracking on total cyclic lifetime would be expected to be less, consistent with these results.

CRACK PROPAGATION IN HIGH STRAIN FATIGUE

The propagation of "short" fatigue cracks with plastic strain cycling for this alloy has been recently studied by Skelton and Challenger (1984) and Challenger and Vining (1983). In order to separate creep from oxidation effects, tests were performed with constant strain dwell periods in both air and vacuum environments (1.3 MPa) at 525°C (Skelton 1984). Tests were performed at frequencies varying from 0.25 Hz to tests with 30 minutes dwell periods at plastic strain ranges of 0.002, 0.001 and 0.0002. These

experiments used 12.5 mm diameter x 12.5 mm gage length samples with short starter notches (0.25 mm) cut with a jewelers saw as a chord at the center of the gage. Potential drop techniques were used to monitor the crack growth.

The continuously-cycled crack growth rate as a function of crack depth for isothermally annealed steel in vacuum and air are shown in Fig. 4 for plastic strain ranges of 0.001 and 0.002. A relationship of the form given in equation (1) was obeyed where the constants B and q depend on the plastic strain range,

$$da/N = Ba^q \tag{1}$$

For short cracks, the crack growth rate in vacuum is much slower than in air but approaches the crack growth rates in air towards the end of the test (crack depths of ~4 mm). This behavior suggests a marked oxidation contribution to growth in the early stages in air at low crack growth rates.

The effects of a 1/2 hour dwell at various points in the cycle are shown for both air and vacuum in Fig. 5. The results in air indicate that the growth rates were independent of whether a T or C type dwell was imposed and, furthermore, that there was no overall increase in the growth rate compared to the continuously cycled data (strain rate of $5 \times 10^{-5}\mathrm{sec}^{-1}$) at the same strain range. The oxidation component had apparently saturated and, additionally, there is no evidence of creep damage contribution to crack growth. Only with T + C cycling did the crack growth rate increase. It is important to emphasize that the air data in Fig. 4 are from step tests on a single specimen so that these observations are not a result of specimen to specimen scatter. These results on the effect of dwell periods are similar to those reported by Challenger and Vining (1983) where fatigue striation spacing was used as a measure of crack growth rates.

Crack growth results in vacuum (again on a single specimen) were surprising. A 1/2 hour T type dwell caused an increase in the crack growth rate over the continuously cycled tests (compare Figs. 4 and 5), now suggesting that there was a creep damage component present. During the C dwell stage cycling crack growth ceased altogether suggesting that any crack advance that occurred during each cycle was rewelded during the compressive strain dwell. Intermediate crack growth rates were obtained with T + C cycling.

The effect of increasing frequency is small, but for growth rates $<10^{-3}$mm/cycle, the crack growth rate decreases with increasing frequency.

By comparing all these data, Figs. 4 and 5, it is clear that there is a significant oxidation contribution to the fatigue crack growth. This contribution decreases as the crack growth rate increases. When the crack growth rate reaches ~2×10^{-3}mm/cycle, the effect of oxidation appears to cease as the crack growth rates in vacuum approach those in air which indicates that at this rate, the crack is advancing by a purely mechanical means.

These oxidation effects are best interpreted as a spalling/regeneration process that occurs at the crack tip. With each cycle the oxide is expected to grow to a thickness, X, in time, t, according to the expression below:

$$X^2 = Kpt \tag{2}$$

where Kp is the parabolic rate constant.

Using $Kp = 3 \times 10^{-10} mm^2 s^{-1}$ at 525°C (Pinder 1981), this would lead to the cycle-by-cycle contribution to crack growth shown in Table 2.

These increments are additive to the advance that results from purely mechanical means, i.e., those measured in vacuum. However, they are insufficient to account for the large divergence between vacuum and air results at low crack growth rates. This may be due to either crack tip rewelding in vacuum that would result in the very low rates in vacuum or accelerated oxidation at the crack tip due to the strain concentration (Skelton 1978) or both. Since the results in Figs. 4 and 5 show that the 1/2 hour dwell in air does not lead to any increase in the crack growth rate over continuously cycled tests, the oxidation contribution to crack growth must saturate very quickly at 525°C. The most likely point of saturation would be where the oxide layers from each side of the crack impinge upon each other. Thus, the time required for a saturation of the oxidation effect will depend on temperature and crack opening displacement.

SUMMARY

For the testing parameters used to date in laboratory tests of 2¼ Cr -1Mo, very little creep damage is occurring. The oxidation that is occurring appears to strongly influence the fatigue lifetimes of these laboratory tests. When significant creep damage does occur, such as reported by Ellison (1976) for 1 Cr Mo V steel, oxidation effects may be overshadowed; however, at temperatures less than about 525°C oxidation effects should dominate and creep damage can be ignored for 2¼ Cr -1 Mo when tested in air. Oxidation has a more pronounced effect on fatigue crack initiation than it does on crack growth. Thus, a large effect of loading wave form on cyclic lifetime should be anticipated when the cyclic strain range is low, but can practically be ignored for large strain ranges.

From the practical standpoint of a design engineer, the number of cycles required to initiate a crack will be reduced from thousands to zero when even very short dwell periods are introduced such that circumferential oxide cracking occurs. Therefore, predictive models should be based on the number of cycles to propagate a short crack to failure.

Integration of equation (1) predicts that (provided $q \neq 1$):

$$N_f = \frac{1}{(q-1)B} \left(a_o^{1-q} - a_f^{1-q} \right) \tag{3}$$

a_o = initial crack length
a_f = final crack length

Using the air data from Fig. 4, $a_f = 5.5$, and the number of cycles to "initiate" the crack, a_o, ignored, the total endurances predicted are given in Table 3. The total strain ranges for the three different plastic strain ranges were calculated from equation

$$\Delta\varepsilon_T = \Delta\varepsilon_p + \frac{\Delta\sigma}{E} \tag{4}$$

Where $\Delta\varepsilon_T$ is the initial total cyclic range for each plastic strain range.

Notice that for these low strain ranges, the cyclic lifetime is quite insensitive to the initial crack length, i.e., comparatively few cycles are involved in the early stages of crack growth.

These predictions are compared to actual data in Fig. 1. This method of prediction should represent a "lower bound" since crack initiation is ignored. This, however, is not an unreasonable assumption as discussed in the section, Fatigue Crack Initiation.

ACKNOWLEDGMENT

The author wishes to thank the Naval Postgraduate School's Foundation Research Program for their support of this work.

REFERENCES

Brinkman, C.R., J.P. Strizak, M.K. Booker and C.E. Jaske (1976). J. Nucl. Mater., 62, 181.
Brinkman, C.R. (1980). Fatigue, Environment and Temperature Effects, Plenum Press, New York, 241.
Brinkman, C.R. (1983). Private communication.
Challenger. K.D., A.K. Miller and C.R. Brinkman (1981a). J. Eng. Mater. Tech., 103, 7.
Challenger, K.D., A.K. Miller and R.L. Langdon (1981b). J. Mater. Energy Syst., 3, 53.
Challenger, K.D. and P.G. Vining (1983). J. Engr. Mater. Tech., 105, 280.
Chopra, O.K., K. Nateson and T.F. Kassner (1980). 2nd Int. Conf. Liquid Metal Technology in Energy Production, Richland, Washington.
Ellison, E.G. and A.J.F. Patterson (1976). Proc. Inst. Mech. Engrs., 190, 321.
Pinder, L.W. (1981). Corr. Sci., 21, 749.
Skelton, R.P. and J. I. Bucklow (1978). Metal Sci., 12, 64.
Skelton, R.P. and K.D. Challenger (1984). Accepted for publication in Mat. Sci. Eng.
Teranishi, H. and A.J. McEvily (1979). Met. Trans., 10A, 1806.

TABLE 1 Effect of Pretest Dwell on the Continuously Cycled Fatigue Life

Testing Temperature 538°C
Total Strain Range 0.4%
Strain Rate 4×10^{-4} sec $^{-1}$

Test Condition	Fatigue Life, Cycles
Continuous Cycling	14,277
250 hr. dwell at +0.002 strain	11,619
250 hr. dwell at -0.002 strain	6,665
250 hr. dwell at -0.002 strain	7,848*

*Test stopped after 2350 cycles and surface examined; oxide spalling occurred during the cool down to room temperature. After examination, the sample was reheated to 538°C and continuously cycled to failure.

TABLE 2 Oxide Contribution to Crack Growth in Air

Frequency, w Hz	$(Kp/w)^{1/2}$ mm	Remarks
0.25	3.5×10^{-5}	
0.05	7.8×10^{-5}	
0.025	1.1×10^{-4}	
0.017	1.3×10^{-4}	
4.8×10^{-4}	7.9×10^{-4}	1/2 hr. dwell + reversal time

TABLE 3 Predicted Cyclic Lifetimes in Air from Crack Growth Data Using Equation (3)

$\Delta\Sigma p$	$\Delta\sigma$ MPa	$\Delta\epsilon_T$ (Eq'n.4)	q	B $\times 10^3$	Predicted Life Times a_o = 0.2mm	a_o = 0.02mm	a_o = 0.002mm
0.002	446	0.0049	0.72	1.06	3.3×10^3	4.3×10^3	4.8×10^3
0.001	386	0.0036	0.59	0.82	4.5×10^3	5.4×10^3	5.8×10^3
0.0002	312	0.0023	0.90	0.23	1.5×10^4	2.2×10^4	2.8×10^4

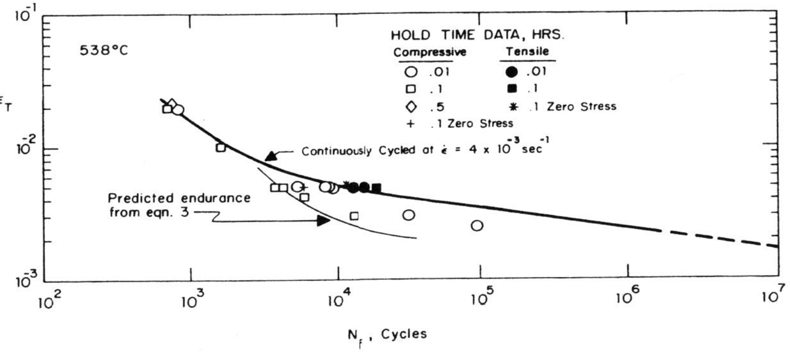

Fig. 1. Effect of cyclic wave form on the cyclic life of 2¼ Cr-1Mo steel at 538°C and a comparison of the cyclic life predicted from the crack growth model (equation 3) with these data.

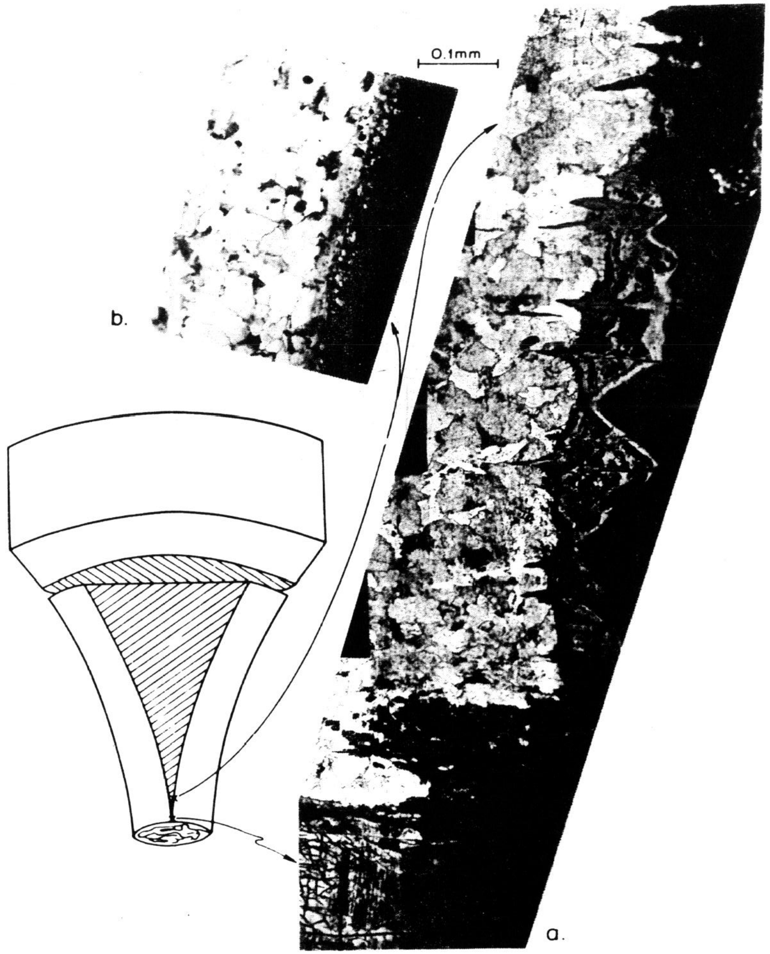

Fig. 2. Tapered cross sectional views of specimens tested at 482°C with a strain range of 0.5%.
a) 0.01 hr constant compression strain hold period, note the coincidence between the oxide cracks and the specimen surface striations.
b) continuously cycled specimen, note the smooth specimen surface.

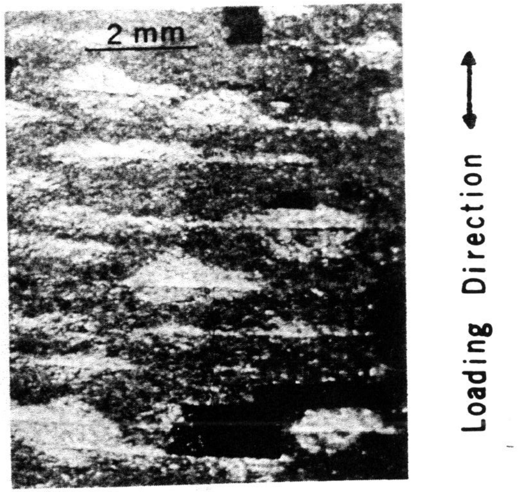

Fig. 3. Surface topography of a specimen that was held at 538°C at a strain of-0.002 for 250 hrs. before continuously cycling to 2350 cycles for surface examination. Total cyclic lifetime for this specimen was 7672 cycles. Note the surface markings perpendicular to the loading direction.

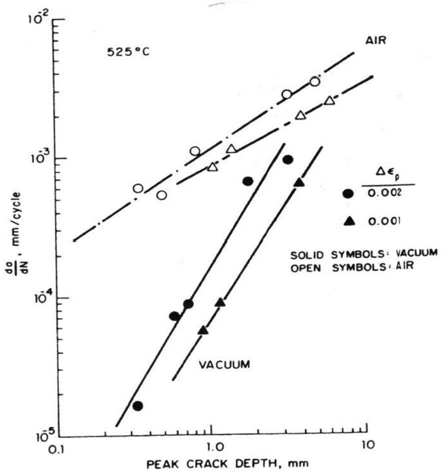

Fig. 4. Continuously cycled high strain fatigue (HSF) crack growth rates in air and vacuum of 2¼ Cr-1Mo steel at 525°C.

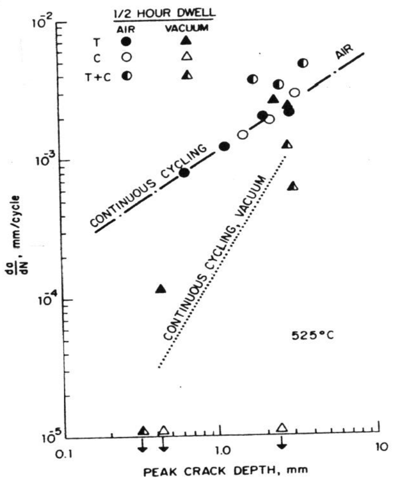

Fig. 5. HSF crack growth rates of 2¼ Cr-1Mo steel at 525°C in air and vacuum with ½ hour constant deflection dwell periods.