HIGH TEMPERATURE FATIGUE DAMAGE MECHANISMS IN ALLOY 800 H

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The dependence of creep fatigue damage interactions on cycle shape and cycle time were investigated at 800°C in air and in high vacuum. The asymmetrical tests led to internal cavitation and reduced the lifetime compared to the corresponding symmetrical tests. In a particular case, the ratio of lifetime in vacuum to that in air decreased from a factor of five to unity.

INTRODUCTION

The lifetime of metals and alloys at high temperatures is mainly affected by creep and fatigue processes, and by corrosive attack. The interaction of these mechanisms often leads to accelerated failure.

For creep fatigue interaction Wareing (1) distinguishes between three failure types:

Type a. Damage is caused only by the propagation of a crack originating at the surface. This is the case if the deformation rate is so high that time dependent mechanisms do not contribute to failure.

Type b. Internal damage (e.g. cavities) enhances type (a) crack propagation and is in turn affected by the presence of the crack. This typical "creep fatigue interaction" results in intercrystal-

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line crack propagation and the reduction in cyclic lifetime.

Type c. Internal damage develops so quickly that the specimen fails by coalescence of cavities before any surface crack can propagate through. Obviously this should result in the shortest cyclic lifetime.

Internal damage is also called creep damage because grain boundary cavitation often occurs during creep experiments. Two types of cavities have been observed. Round (r-type) cavities are found mainly on grain boundaries normal to the stress axis. Because high stresses are required for their nucleation, r-type cavities are associated with second phase particles (Riedel (2) and Raj(3)). Wedge type (w-type) cavities are found at grain boundary triple points. Their formation is due to accommodation problems and stress concentrations during grain boundary sliding (Min and Raj (4)).

These cavities can either grow or shrink depending on the sign of stresses. As a result, asymmetrical tests (slow/fast and hold-time tests) are more likely to cause specimen failures of types (b) and (c) (see Min and Raj (5) and Majumdar and Maiya (6)). The influence of the corrosive attack is expected to depend on the different failure types (a,b and c). The propagation of a surface crack should be strongly enhanced, which leads to longer lifetimes in vacuum than in air tests. On the other hand, corrosion should hardly affect the development of internal damage (Hirakawa and Tokimasa (7)), resulting in similar lifetimes for air and vacuum tests, when type (c) mechanism dominates. In order to better seperate the effects of the main failure mechanisms, creep, fatigue and corrosive attack, we have performed symmetrical and asymmetrical fatigue tests in air and vacuum.

EXPERIMENTAL DETAILS

Material

The chemical composition of alloy 800 H is given in Table 1. The material was solution annealed at 1125°C for 30 min, water quenched and aged at the test temperature of 800°C for 16 h in order to stabilize the carbide microstructure. γ'-precipitation could not be observed. After this heat treatment the grain boundaries were heavily covered with $M_{23}C_6$-type carbides. The average grain size was 70 μm.

Test Conditions

All tests were conducted in air or in high vacuum ($p \leq 10^{-5}$ mbar) on smooth specimens of 7 mm diameter and 11 mm gauge length using a servohydraulic fatigue machine. The strain was measured directly with an axial extensometer. The test temperature of 800°C was produced by induction heating, with a gradient less than ±2.5 K along the gauge length.

TABLE 1 - Chemical Composition of Alloy 800 H (w/o)

Fe	Ni	Cr	Ti	Al	C	Si	Mn	Co
46.5	31.1	20.3	0.32	0.34	0.07	0.46	0.68	0.23

Test Types

The plastic strain amplitude was kept constant during both the strain controlled and the stress controlled tests (Fig. 1).

Strain controlled tests. In symmetrical (S) tests, constant total strain rates from $6 \cdot 10^{-6}$ to $2 \cdot 10^{-2}$ s^{-1} were used. In the slow/fast (S/F) tests, the strain rate in tension was 10 to 500 times lower than that in compression.

Stress controlled tests. In the stress controlled tests the stress was increased very quickly (in ≤ 0.1 s) up to a predetermined stress value and then kept constant, until the specimen had crept to a fixed plastic strain value. As a result the holdtime does not necessarily remain constant during an experiment. In the symmetrical tests (H/T/C), the same stress level was chosen in compression and in tension, thus the holdtimes in tension and compression were equal for each cycle. In asymmetrical tests, the stress levels were varied by up to a factor of two, hence the holdtimes at the lower stress were up to 200 times longer than at the higher level. These tests are referred to as tension hold (H/T) or compression hold (H/C) tests, depending on which holdtime was longer.

Definition of cyclic lifetime.

In the strain controlled tests N_f was defined as a 20 % drop in the saturation stress. In the stress controlled tests an equivalent definition is the reduction of the saturation value of the cycle time by a factor of four, assuming that a power-law for the minimum creep rate with a stress exponent of $n = 6.8$ (Lerch et al.(8)) applies.

Metallographic preparation

As the material was rather ductile, an electropolishing technique was used to remove artefacts caused by mechanical preparation methods before cavities could be identified in the scanning electron microscope.

RESULTS

Fig.2 summarizes the results of all symmetrical (strain and stress controlled) tests. The numerals beside each datapoint in the diagram give the order of magnitude of the cycle time in seconds for each test. For example, 3 stands for a cycle time between 100 and 999 s. S-tests with cycle times less than 15 s lie on a straight line, with deviations to longer N_f-values for lower strain amplitudes. With increasing cycle time the cyclic lifetime N_f decreases. This "frequency effect" becomes more pronounced as the plastic strain amplitude decreases. In vacuum tests (marked Vac) the cyclic lifetime is greater than in air tests by a factor of five.

In Fig.3 all asymmetrical tests are presented. They all show a lower N_f than symmetrical tests having a comparable cycle time. (Exceptions: The S/F-test with $t_{cyc} \leq 10$ s and the one H/C-test have the same cyclic lifetimes as the corresponding symmetrical test.) The longer the cycle time the greater the cyclic lifetime reduction is. Especially the H/T-tests at $\Delta\varepsilon_{pl} = 1$ % show extreme reductions. In this case the cyclic lifetime N_f is not affected by the environment.

Metallographic results

The fracture mode changes gradually from transcrystalline to intercrystalline with increasing cycle times. This transition starts a bit earlier in S/F-tests than in symmetrical tests. In contrast to the S-test results of Schusser (9) the change occurs earlier in vacuum than in air. The H/T-specimens with the drastic reduction in N_f showed very pronounced bulk damage, mostly in the form of wedge-type cracks in grain boundaries normal to the stress axis.

Cavities were also found in other asymmetrical H/T and S/F-tests, but the densities were much lower. Up to now, no bulk damage could be found in S-test specimens. Surprisingly, in standard creep tests at similar stress levels, a much lower density of cavities could be observed.

DISCUSSION

If crack propagation were the only damage mechanism (type a), all data points of Fig.2 and 3 would lie on one line (1). This suggests, that the deviation from this line can be used as a measure for the presence of additional damage mechanisms like corrosive and/or creep effects. This reference line is determined by the "fast" symmetrical tests with $\Delta\varepsilon_{pl} \geq 0.2$ % and $\varepsilon_{tot} \geq 0.001$ s^{-1}, because in this strain rate region no frequency dependence is observed. At smaller strain amplitudes, more cycles are required for crack initiation and the cyclic liftime increases. Fig.4 and 5 show all cyclic lifetimes normalized in respect to that reference line.

The tests with a strain range of 0.3 % show significant reduction in cyclic lifetime with increasing cycle time. This frequency effect is accompanied by a transition from transcrystalline to intercrystalline crack propagation. The tests with $\Delta\varepsilon_{pl} = 1$ % also show this cracking mode transition but the frequency effect is not so distinct. Min and Raj (5) explain this difference by asymmetrical sliding of a grain boundary in front of the crack tip. This effect should be more pronounced at smaller plastic strain amplitudes.

Asymmetrical tests

The asymmetrical tests with longer periods in tension (S/F and H/T) also show significant lifetime reductions with increasing cycle time. In all cases these reductions are more drastic than in the corresponding symmetrical tests, i.e. tests with a similar cycle time and the same plastic strain amplitude. This indicates, that in addition to the frequency effect further damage mechanisms come into action. This is confirmed by the metallographic results that cavitation occurs only in asymmetrical tests. The possible mechanisms for cavity formation are grain boundary sliding and stress concentrations at hard second phase particles in the grain boundaries ((4), Raj and Ashby (10) and (2)).

But in order to affect the cyclic lifetime the cavities have to grow. Because this is possible only under tensile stress (compressive stresses cause cavity shrinkage), net growth during fatigue experiments is possible only if the specimen spends more time in a state of tensile than compressive stress.

This easily explains the fact that the H/C test does not show any cyclic lifetime reduction in comparison with the corresponding symmetrical test. The cavities can sinter during the long compressive hold period. On the other hand, the S/F-test with short cycle time does not show a lifetime reduction, because the total test time was too short to allow enough cavity growth.

Environmental effects

Generally, the environment always has an effect on the cyclic lifetime, but to an extent determined by the nature of the dominant damage mechanism. This becomes apparent in table 2, where the cyclic lifetimes of air and vacuum tests for identical test types are compared with each other.

TABLE 2 Comparison of Cyclic Lifetime in Air and in Vacuum

Test type		N_f in air	N_f in vacuum	N_{fv}/N_{fair}
S	0.3 %	3470	16140	4.7
S/F	0.3 %	1090	3290	3.0
H/T/C	1.0 %	460	2350	5.1
H/T	1.0 %	65	87	1.3

For the fast symmetrical tests (H/T/C and S), where crack propagation dominates, the lifetimes in vacuum exceed those in air by a factor of five. The H/T-tests, by contrast, have rather similar cyclic lifetimes in air and in vacuum. This is strong evidence that only internal damage, which is not affected by the environment, causes failure.

In the range between these two extremes both mechanisms should be expected to contribute to failure. Since the lifetime of the S/F-tests in vacuum is greater than that in air by about a factor of three and hence lies between these two extremes, the assumption appears to be justified.

CONCLUSIONS

From our experiments on Alloy 800 H the following conclusions can be drawn:
All the three failure types (a,b,c) can be produced in fatigue ex-

periments using the proper cycle shape. If the internal damage shall dominate the lifetime (type c) the H/T-cycle type seems to be most suitable. "Creep damage" is more pronounced in H/T-fatigue tests than in standard creep experiments.

ACKNOWLEDGEMENTS

This research is part of the European research project COST 501. The financial support by the Bundesministerium für Forschung und Technologie, Bonn, F.R. Germany is gratefully acknowledged. The alloy investigated was provided by the Vereinigte Edelstahlwerke, Kapfenberg, Austria.

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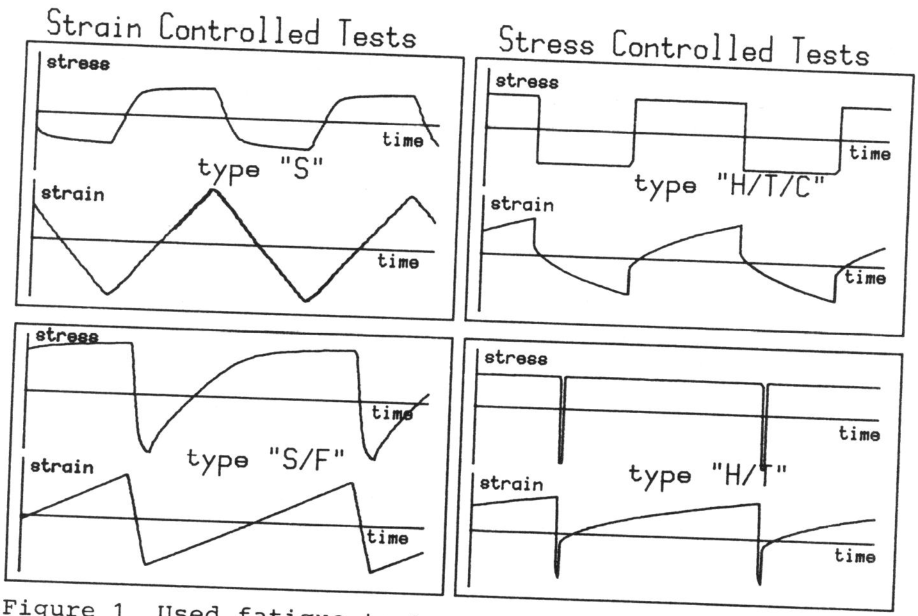

Figure 1 Used fatigue test types (The H/C - type is the inverse of the H/T - type)

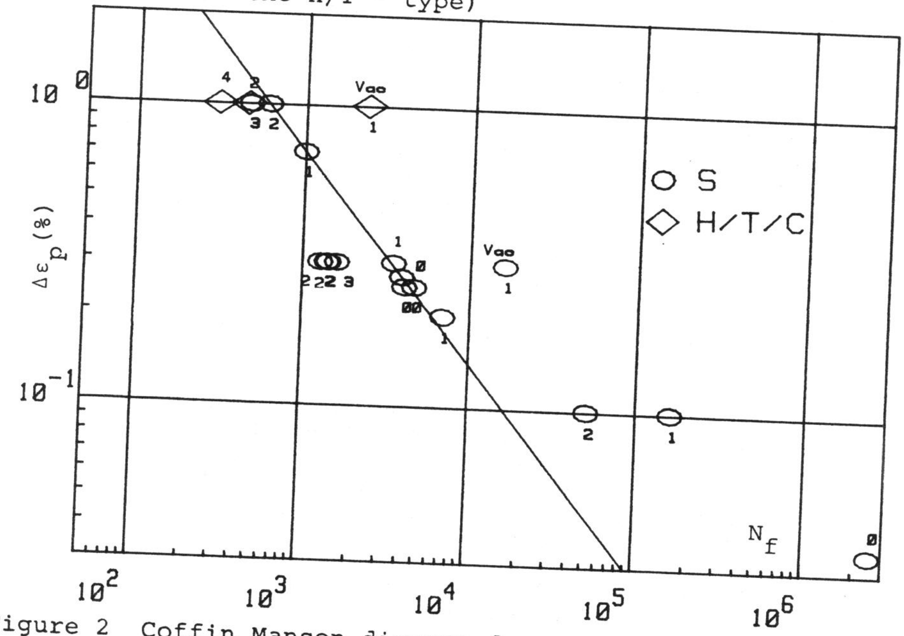

Figure 2 Coffin Manson diagram for symmetrical fatigue tests (S and H/T/C) with reference line

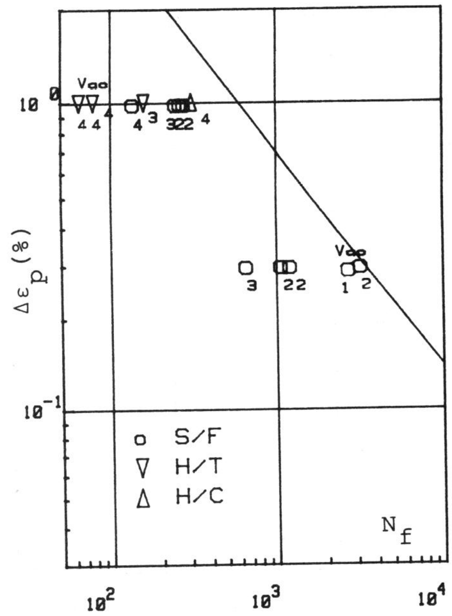

Figure 3 Fig.2 for asymm. tests (S/F, H/T, H/C)

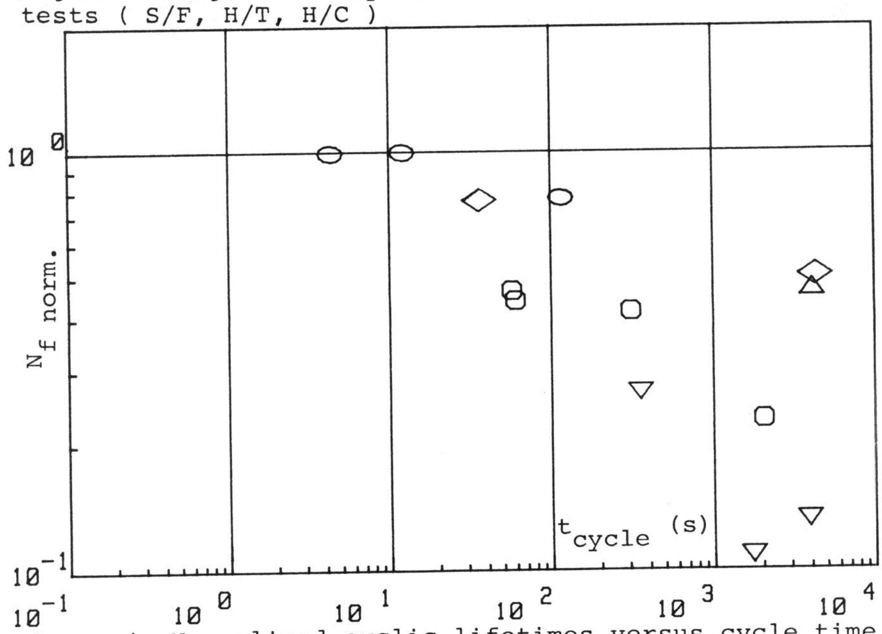

Figure 4 Normalized cyclic lifetimes versus cycle time for ε_p > .5% (Symbols as in figures 2 and 3)

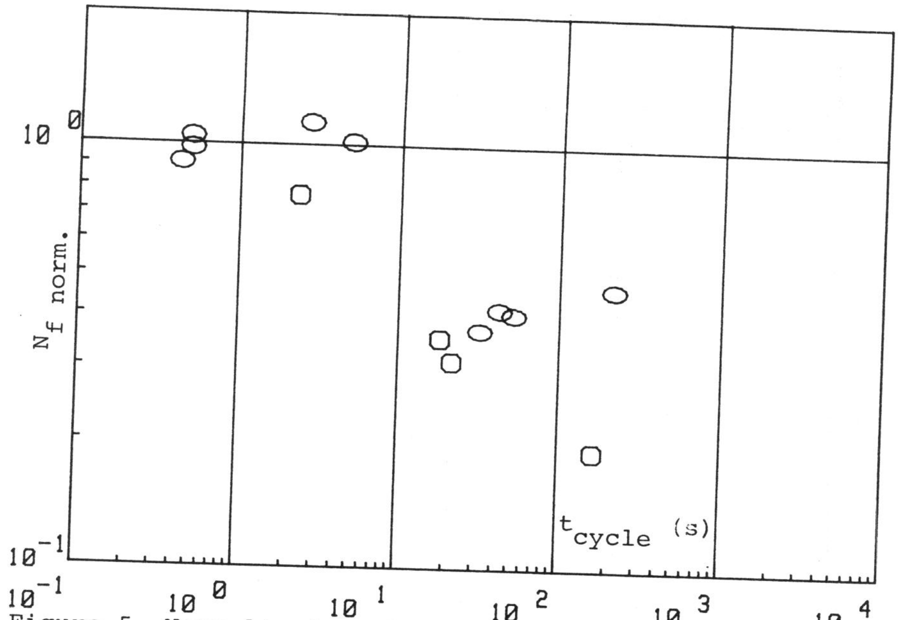

Figure 5 Normalized cyclic lifetimes versus cycle time for ε_p < .5% (Symbols as in figures 2 and 3)