ELEVATED TEMPERATURE, LOW CYCLE FATIGUE OF AISI 316 STAINLESS STEEL

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ABSTRACT

Elevated temperature, low cycle fatigue tests were carried out on AISI 316 stainless steel. The effect on fatigue endurance and crack growth of testing environment, cyclic strain rate and plastic strain range were studied. Environment was shown to exercise a great influence on fatigue lives through its effect on both crack nucleation and crack propagation. The deleterious effect of decreasing the cyclic strain rate was shown to be part of a complex interaction with plastic strain range which influenced both crack initiation and propagation.

KEYWORDS

Low cycle fatigue; environment; fatigue crack nucleation; fatigue crack propagation; cyclic strain rate.

INTRODUCTION

The operation of chemical and nuclear plant at elevated temperatures has led to the consideration of low cycle fatigue as a possible failure mechanism of components subjected to thermally induced stresses. The complex stress-time histories experienced by such components and the wide range of environments in which they operate could allow creep, fatigue and environmental attack to occur simultaneously and possibly in a synergistic manner.

In order to provide design data, the fatigue endurance of smooth specimens has been extensively studied and this has revealed the important influence of temperature, environment, cyclic waveshape and strain rate on fatigue life, (Coffin 1975, Kanazawa 1973, Manson 1975, Wood 1977). Much of this work has led to the evolution of failure prediction equations as in the frequency modified Coffin-Manson equation, (Coffin 1975), and the strain range partitioning approach, (Manson 1975). Whilst this approach has met with some success in predicting relatively short term behaviour in the laboratory, its phenomenological approach makes its extrapolation to long term service conditions uncertain.

In addition to the above there has been a development of a more fundamental, mechanistic approach to low cycle fatigue at low and elevated temperatures, (Laird 1966, Tomkins 1975, Wareing 1979). The concentration in this line of approach on crack growth mechanisms has lent greater confidence to the prediction of long term fatigue behaviour.

It is the aim of the work presented here to follow this latter development and to show that the lifetimes of low cycle fatigue tests are a result of the interaction of different failure processes which complicates the use of any simple constitutive equation.

EXPERIMENTAL

The test material, AISI 316 stainless steel, and the cylindrical specimen design were identical to that used by Morris, (1977) in his study of the creep behaviour of this alloy. After solution treatment at 1050°C for 45 mintues, the specimens were water quenched and prepared for testing by electropolishing their gauge lengths in a perchloric acid-glycerol-ethanol solution. The specimens tested in an inhert environment had stainless steel bellows welded to the specimen shoulders, to contain an environment of low pressure argon around the specimen gauge length.

All tests were fully reversed low cycle fatigue tests at 625°C, under axial strain control. They were conducted in a servohydraulic fatigue testing machine equipped with a clamshell furnace and a programmable waveform generator.

The testing variables examined were tension going strain rate, $\dot{\varepsilon}_t$, compression going strain rate, $\dot{\varepsilon}_c$, plastic strain range and environment, (air or argon). Continuous load- and extension- time histories were recorded for each test until failure occurred, which was defined as a 5% reduction in the maximum tensile load of a fatigue cycle from the maximum recorded during the test. After testing the specimens were sectioned and subjected to extensive fractographic and metallographic examination.

RESULTS

Fatigue Tests in Air

Mechanical properties. An increase in the plastic strain range or a decrease in the cyclic strain rate reduced the fatigue endurance although there were indications that the strain rate effect saturated at both low and high strain rates, (0.00067%/sec. and 0.67%/sec. respectively), fig.1. Tests which employed different strain rates for the tension and compression going phases of the fatigue cycle, ($\dot{\varepsilon}_t \neq \dot{\varepsilon}_c$), resulted in fatigue lives very similar to those recorded for tests conducted using only the tension going strain rate.

A short series of tests was performed at the maximum plastic strain range, 0.817%, using two different strain rates for given fractions of the tests, table 1. Linear damage summation of the tests conducted using the slower strain rate first followed by a faster strain rate resulted in a damage summation of less than unity. The single test in which the order of strain rates was reversed resulted in a damage summation not significantly different from unity.

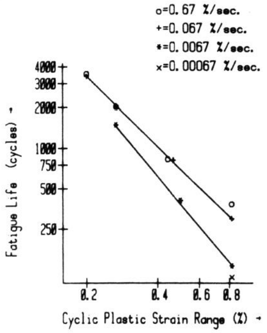

Fig. 1 The effects of plastic strain range and cyclic strain rate on fatigue life in air

TABLE 1 Summary of Mixed Strain Rate Fatigue Tests

Initial Test Strain Rate $\dot{\varepsilon}_1$ (%/sec)	Number of Cycles at $\dot{\varepsilon}_1$	Second Test Strain Rate $\dot{\varepsilon}_2$ (%/sec)	Number of Cycles at $\dot{\varepsilon}_2$	Linear summation damage parameter
0.67	400	-	-	1
0.067	300	-	-	1
0.0067	133	-	-	1
0.67	20	0.067	305	1.067
0.067	10	0.67	320	0.833
0.067	20	0.67	300	0.816
0.0067	5	0.67	305	0.8

Metallography and fractography. In all tests multiple crack initiation occurred with typically more than thirty cracks in the gauge length of each specimen. Coplanar cracks appeared to propagate independently at first before joining up to produce a united crack front, as has been described by Wareing (1977). At the higher strain rates continuous, ductile striations were formed on the fracture surface, fig. 2a.

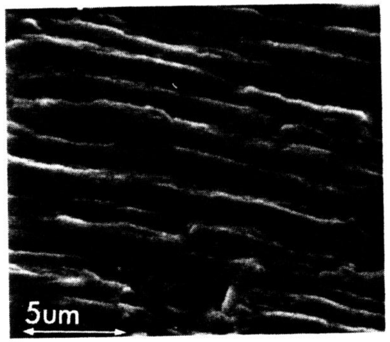

Fig. 2a Typical fatigue striations at high strain rates in air.

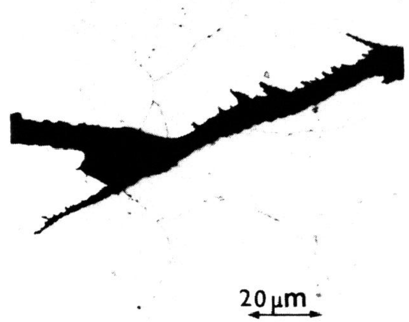

Fig. 2b Crack propagation at high strain rates in air.

The relationship between striation spacing, crack depth and plastic strain range was similar to that proposed by Wareing:

$$S = A(\Delta\varepsilon_p)^{\gamma} a \qquad (1)$$

where S is the striation spacing at a crack length, a, in a test with a plastic strain range, $\Delta\varepsilon_p$, fig 3. A and γ are constants, 2.66 and 1.11 respectively in these tests. The lines shown in fig 3 represent a least squares fit through at least fifty data points and in each case the correlation coeffictent was greater than 0.9. Longitudinal sections of the cracks revealed a rough fracture surface with a high incidence of crack branching fig. 2b.

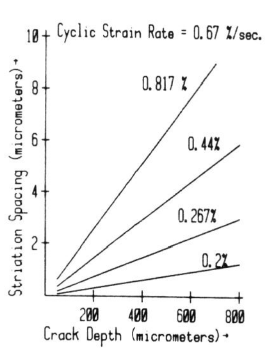

Fig. 3 The effect of plastic strain range on striation spacing in air.

The lower strain rate tests resulted in a mainly intergranular fracture path with very few, non continuous striations; consequently no estimate could be made of any striation spacing/crack depth relationship. There were some signs of internal damage and some grain boundaries adjacent to the free surface showed distinct cavitation, even at strain rates as high as 0.067%/sec, fig 4a.

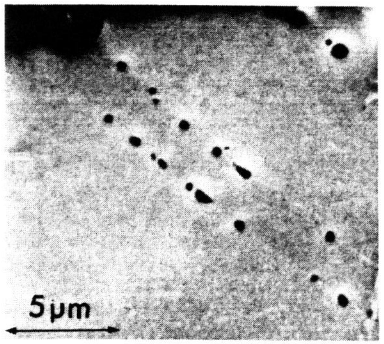

Fig. 4a Cavitation along grain boundaries adjacent to free surface strain rate = 0.067%/sec.

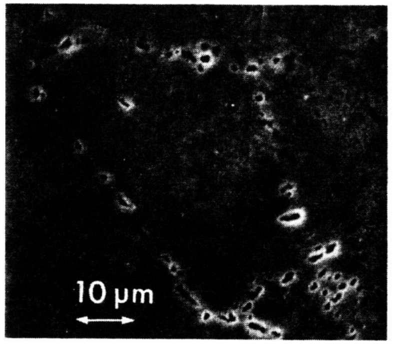

Fig. 4b Typical internal damage at very low strain rates

Transmission electron microscopy revealed a gradual change in dislocation microstructure from a dense square or equiaxed cell structure at high strain rates to a more open network of dislocation tangles at low strain rates, similar to those reported for high stress creep specimens, (Morris 1978). Extensive carbide precipitation had occurred on grain boundaries, along the dislocation cell walls and at the centre of dislocation tangles.

Fatigue Tests in Argon

Mechanical properties. All tests were conducted at the maximum strain range of the air tests, 0.817%, but with a variety of strain rates. Again a decrease in strain rate had a deleterious effect on fatigue edurance, an effect which

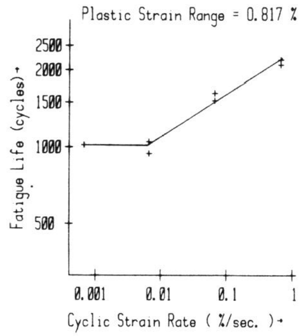

Fig. 5 The effect of cyclic strain rate on fatigue life in argon.

saturated at the lowest strain rate, fig. 5.

Metallography and fractography. In all tests only one or two cracks nucleated, however at the two highest strain rates several cracks had nucleated internally and propagated transgranularly.

At high strain rates surface crack propagation was transgranular resulting in continuous but indistinct striations, fig 6a. Sectioning of the cracks revealed very smooth fracture surfaces with no signs of crack branching, fig. 6b.

At the lowest strain rate crack propagation was totally intergranular. Scanning electron microscopy of a polished and etched section showed extensive internal grain boundary cavitation but little internal cracking, fig. 4b.

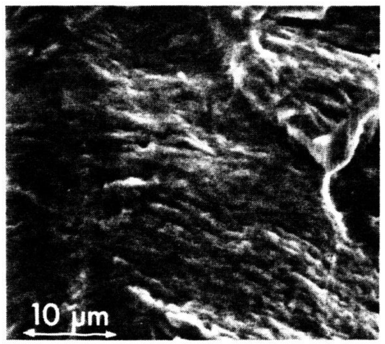

Fig. 6a Typical striations at high strain rates in argon.

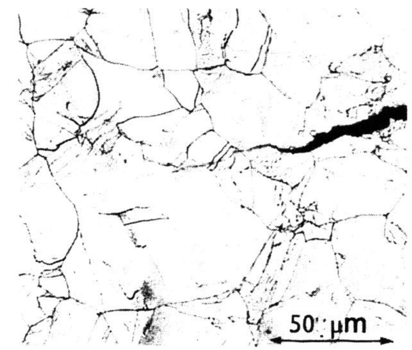

Fig. 6b Crack propagation at high strain rates in argon.

Crack initiation tests. Two series of tests were performed to investigate crack nucleation in an argon environment. In the first a 200 um deep notch was machined into the specimen surface before testing in argon. This reduced the life from the expected 2100 cycles for an unnotched specimen to 365 cycles. In the second series of tests, specimens were fatigued for a given fraction of their expected life in one environment, say air, before testing to failure in the other environment, argon. The results of these tests, table 2, showed a marked difference in fatigue lives depending upon the order in which the tests were carried out. This was best shown by performing a linear damage summation for each test.

TABLE 2 Results of Mixed Atmosphere Tests. Plastic strain range = 0.817%
Strain rate = 0.67%/sec.

Test	First Atmoshpere	Number of Cycles in A	Second Atmosphere B	Number of Cycles in B	Linear Damage Summation
1	Air	400	-	-	1
2	Argon	2100	-	-	1
3	Air	200	Argon	400	0.69
4	Argon	800	Air	410	1.41
5	Argon	1000	Air	415	1.51
6	Argon	1200	Air	365	1.48

DISCUSSION

High Strain Rate Tests

This section includes all tests conducted at a strain rate of 0.067%/sec. Low cycle fatigue has long been considered as a process of rapid crack nucleation followed by slow, strain controlled crack growth. Numerous models of the crack growth process have been proposed, (Laird 1968, Solomon 1972, Tomkins 1968), and these have usually incorporated a crack growth law of the form

$$da/dN = B \ (\Delta\varepsilon p)^{n} \ a \qquad (2)$$

where da/dN is the crack growth per cycle and B and n are material, temperature, strain rate and environment sensitive constants. Integration of this equation between an initial crack length, a_0, and a final crack length, a_f, results in a Coffin-Manson type equation

$$(1/B\ln(a_f/a_o))^{1/n} = N^{1/n}(\Delta\varepsilon p) \qquad (3)$$

where N is the number of cycles spent in crack propagation.

The similarity between the latter type of expression and actual experimentally derived Coffin-Manson type equations has resulted in fatigue endurances being interpreted solely in terms of crack propagation; i.e. crack nucleation was instantaneous or followed a similar equation to the crack growth based life prediction equation.

High strain rate fatigue crack propagation rates could be measured in the present tests because of the one-to-one correspondence between striation spacing (S in eqn. 1) and crack growth per cycle, (da/dN in eqn. 2) at these high strain ranges, (Wareing 1977). The subsequent application of crack growth integration to the results of the highest strain rate tests in air resulted, in all cases, in a required initial crack size of 1-3 um. This initial crack size agreed closely with Wareing's (1977) results on this steel.

In the tests conducted at the second highest strain rate the crack growth rate at a given crack length overlapped those of the highest strain rate except at the highest strain range where, although the overall life was less than the highest strain rate test, the crack growth rate was lower, Fig. 7.

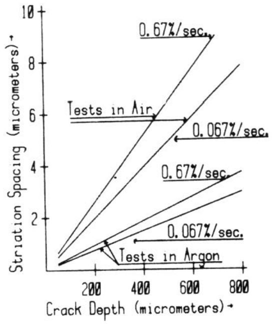

Fig. 7 Effect of strain rate and environment on striation spacing.

Integration of the crack growth equation as above for the latter lower strain rate test and for the mixed strain rate tests which employed the slower strain rate first, indicated that a crack of size ∽50 um was present very early in the test. It is suggested that the larger initial crack produced by these tests was due to grain boudary sliding occurring at the free surface. Here the constraints from other grains would be less than in the interior and any sliding could be rendered irreversible during the compression phase of the fatigue cycle by oxidation of the surface displacement. At the highest strain rate, grain boundary sliding would be less marked and so would be unable to contribute to the production of a long initial crack. Hence in the highest strain range tests at the two higher strain rates, it is suggested that this effect has occurred thus reducing the overall fatigue life of the lower strain rate test despite its lower crack propagation rate.

It would appear that this effect occurred only at high plastic strain ranges because of the similarity in fatigue life at lower strain ranges, but further work similar to that presented here would be required to confirm this.

The two higher strain rate tests in argon, (0.67 and 0.067%/sec) exhibited a linear striation spacing/crack depth relationship, fig. 7, with the striation spacing at a given crack depth being less than that of the corresponding test in air. Integration of these crack growth curves in a manner identical to that used for the air tests resulted in a fatigue life of ∽50% of that observed experimentally. A similar observation by Levaillant, (1979), was interpreted as the result of delayed crack nucleation in a non-agressive environment. The relative densities of fatigue cracks in the present air and argon tests would indicate difficulties of crack nucleation in argon tests. The single test in argon on a specimen with a sharp notch, i.e. with an initial crack of 200 um, resulted in a dramatically reduced life, a life which closely agreed with that predicted from integration of the crack growth curve obtained in argon.

The mixed atmosphere tests were performed to confirm the incubation period for crack nucleation in argon. The results, table 2, indicate that ∽50% of the measured fatigue life in argon was spent in crack nucleation, a figure which agreed with above estimates of the remaining time spent in crack propagation. In test 3, where the specimen was cycled first in air then in argon, the part of the test in air was sufficient to nucleate and grow a crack so that only crack propagation took place in argon, resulting in the low linear damage summation factor.

Apart from the effect of strain rate and atmosphere on crack nucleation, the effects of both these parameters on crack propagation alone would result in a change in fatigue endurance. A change in environment from air to argon reduced the crack growth rate by a factor of ∽3, and this could be the result of either crack rewelding or lack of oxidation contribution to crack growth at the crack tip. Whilst the above experiments did not confirm either theory, the observation of atmospheric attack and penetration in the air tests would suggest that difference in crack growth rates was at least partly the result of oxide formation at the crack tip.

Low Strain Rate Tests

Due to the lack of well formed striations the striation measurement technique could not be applied at strain rates of less than 0.067%/sec. Other workers have made measurements at lower strain rates which led to an adequate prediction of fatigue life from crack propagation considerations alone, Tomkins (1977). This would apparently contradict any strain rate dependent, enhanced nucleation effect, although the small amount and growth data from these tests must leave this question still open.

In air a decrease in testing frequency generally reduces life and it has been argued that this is due either to enhanced environmental attack or to creep processes contributing to crack tip extension. In the present work two tests were carried out in air at a low strain rate, 0.0067%/sec., but incorporating a dwell at the maximum compressive strain in each cycle. Such dwell periods have been acknowledged as capable of sintering creep damage, thus restoring the fatigue life of creep-fatigue tests to that of no hold time fatigue tests. The present dwell periods had no effect on fatigue life. It is therefore concluded that unless the creep damage in the slow strain rate tests was extensively localised near the crack tip and thus stabilised by interaction with the crack and/or the environment, the acceleration of crack growth in air as a result of decreasing the strain rate appeared to be a result of environment-fatigue interaction. This observation agrees with the relative lives of the lowest strain rate tests in argon and air.

Extensive internal damage in the form of grain boundary cativies was found only in the lowest strain rate test in argon, but other slow strain rate tests ($<$ 0.0067%/sec) in air resulted in a small amount of such damage. This bulk cavitation could not be explained by stabilisation of cavities through crack or environment interaction. The mechanics of creep damage nucleation in fatigue and creep has been shown to be a complex phenomenon but one in which grain boundary particle size, spacing and nature plays an important role, (Raj, 1979). This test in argon at a strain rate of 0.00067%/sec had a lifetime of 2000 hours as opposed to the usual 2-200 hours and as such possessed a far coarser precipitate structure which alone showed signs of chi and laves phases. Hence either the ripening of the precipitates or a change in their nature could explain the observation of far more extensive creep damage than in any other test, but such a susceptibility to damage nucleation in "pure fatigue" conditions has to be borne in mind for service lives of many thousands of hours.

CONCLUSIONS

The elevated temperature low cycle fatigue life of AISI 316 stainless steel was determined in air and argon for a variety of cyclic strain rates and strain ranges. In addition a series of tests was performed to investigate the crack nucleation behaviour of this alloy.

Fatigue life was shown to be the sum of time spent in nucleating a surface crack and growing it to macroscopic dimensions. Both these processes were influenced by the strain rate, strain range and environment of a particular test and results obtained from tests in which these factors varied showed that simple linear damage summation laws could be in error.

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