Dynamic Strain Ageing Effects in Low Cycle Fatigue

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ABSTRACT

This paper presents the role of Dynamic strain ageing (DSA) on Low Cycle Fatigue (LCF) properties of a type 304 stainless steel at 823K. Dynamic strain ageing in low cycle fatigue is manifested as a peak in the saturation stress, a minimum in plastic strain, an increase in work hardening rate and is at times associated with serrated stress-strain hysteresis loops. Dynamic strain ageing also increases the inhomogeneity of deformation during fatigue. Persistent slip bands are observed in the interior of the material with an associated intergranular cracking.

1. INTRODUCTION

It is well established that two time dependent damage processes, creep and oxidation, contribute significantly to the loss of low cycle fatigue life at slow strain rates and elevated temperatures. Many empirical life-time prediction laws have been proposed to account for these damage modes [1-5]. Recent studies on austenitic stainless steels [6, 7] have shown that the life reduction can also occur in the sub-creep temperature regime where oxidation effects are also expected to be negligible. Loss of life with decreasing strain rate in the sub-creep range was related mainly to another time dependent process, dynamic strain ageing. Tsuzaki *et al.* [8] have recently studied dynamic strain ageing during fatigue deformation in type 304 stainless steel and have demonstrated that the lower critical temperature for the appearance of serration during fatigue deformation was much lower than during monotonic tensile deformation. This has been attributed to the longer arrest time of dislocations in fatigue deformation in comparison with that in monotonic deformation.

The authors have recently examined the grain size dependence of low cycle fatigue behaviour of type 304 stainless steel at 923K [9]. Serrations characteristic of dynamic strain ageing were observed in both fine (75 μm) and coarse (310 μm) grain size specimens. For 75 μm grain size specimen, a peak in saturation stress and a minimum in plastic strain range was observed at an intermediate frequency of 0.01 Hz at all strain ranges investigated. These features were not observed in coarse grain size (310 μm) specimens. The effects of DSA on LCF properties of type 304 stainless steel are further examined in this paper at a lower temperature of 823K on coarse grain (310

μm) samples. The experimental conditions were chosen carefully to minimise the influences of creep and oxidation.

2. EXPERIMENTAL DETAILS

The material examined conformed to the AISI 304 stainless steel specification and its chemical composition (wt %) was as follows:

C = 0.042, Cr = 18.2, Ni = 9.2, Si = 0.38, Mn = 1.65, P = 0.024 and S = 0.003.

135 mm long, 24 mm diameter bars machined from 25 mm thick mill annealed plate were given a solution treatment at 1573 K/1½h followed by water quenching. These samples were subjected further to the recommended solution anneal of ½h at 1323 K for this grade of stainless steel. The combination of these treatments produced a linear intercept average grain size of 310 μm .

Fully reversed total axial strain controlled LCF tests were conducted on 50 mm gauge length, 10 mm diameter cylindrical samples in an Instron 1343 servo hydraulic system equipped with radiant heating facility. Tests were carried out at 823 K using a triangular wave shape over a frequency (ν) range of 10^{-2} to 1.0 Hz at a strain range ($\Delta\epsilon_t$) of 0.8%. The effect of strain range on DSA was investigated by carrying out tests in strain ranges varying from 0.5 to 2.5% at a frequency of 0.01 Hz. Fracture surfaces of the tested specimen were examined in a Philips PSEM 501 scanning electron microscope. The peak DSA temperature has been established as 823 K by strain controlled LCF tests, conducted at the intervals of 50 K, in the temperature range of 623–973 K [10]. The temperature at which the maximum in tensile stress amplitude and minimum in the plastic strain range per cycle recorded, was taken as peak DSA temperature. The tensile tests conducted over the same temperature range and at a strain rate of $3 \times 10^{-4} \text{ s}^{-1}$ exhibited a yield strength peak, a maximum work hardening rate and a minimum in fracture and uniform elongation at 823 K; all these effects are considered as characteristics of DSA [10].

In this investigation, the number of cycles corresponding to the 20% drop in maximum stress in tension is reported as the fatigue failure life, N_f . The reasons for adopting this as a failure criterion are published elsewhere [11].

TABLE 1. Low Cycle Fatigue Properties of AISI 304 Stainless Steel at 823 K

S. No.	$\dot{\epsilon}$ (s^{-1})	ν (Hz)	$\Delta\epsilon_t$ (%)	$\Delta\epsilon_p$ (%)	$(\Delta\sigma/2)_1$	$(\Delta\sigma/2)_{max}$	N_f	Remarks
1	1.6×10^{-2}	1.00	0.80	0.567	159	227	2390	—
2	1.6×10^{-3}	0.10	0.80	0.562	147	262	715	—
3	1.6×10^{-4}	0.01	0.80	0.555	112	271	429	Serrated stress-strain hysteresis loops

3. RESULTS AND DISCUSSION

I. Influence of DSA on Fatigue Properties and Serrated Flow

Table 1 summarises the influence of ϵ ($\dot{\epsilon} = 2 \times \text{strain range} \times \nu$) on LCF properties, such as N_f , tensile stress amplitude of the first cycle $(\Delta\sigma/2)_1$, maximum tensile stress amplitude $(\Delta\sigma/2)_{max}$ and the plastic strain range $(\Delta\epsilon_p)$ per cycle. $\Delta\epsilon_p$ was actually measured from the width of the hysteresis loop at zero mean stress corresponding to the cycle exhibiting $(\Delta\sigma/2)_{max}$. N_f decreased markedly with decreasing cyclic deformation rate ($\dot{\epsilon}$). There was an increase in $(\Delta\sigma/2)_{max}$ and a very slight decrease in $\Delta\epsilon_p$ with decreasing $\dot{\epsilon}$. For all the strain rates studied the material was found to undergo cyclic hardening.

Cyclic stress-strain response of the material tested with $\dot{\epsilon} = 1.6 \times 10^{-3} s^{-1}$ is shown in Fig. 1a. The stress to enforce the total strain limit with successive reversals increased rapidly, before reaching a peak value in a relatively few cycles; after which it remained stationary for a short period until micro-cracks developed. The failure of the specimen was indicated by a rapid decrease in the peak stress (Fig. 1b). It should be noted that the hysteresis loops for this strain rate remained smooth both in strain hardening and saturation stages and even towards the end of the test.

A summary of the tensile stress response derived from the hysteresis loops recorded for different strain rates is presented in Fig. 2. All tensile peak stresses of successive cycles have been normalised with respect to the peak tensile stress of the first cycle. The most important conclusions that emerge from this figure are:

1. At higher strain rate [$\dot{\epsilon} = 1.6 \times 10^{-2} s^{-1}$] the alloy displayed a short period of strain hardening followed by a long period of a nearly steady state response.

Fig. 1a: Stress strain hysteresis loop showing cyclic hardening (Grain Size: 310 μm , Temp: 823K, Frequency: 0.1Hz).

2. At the lower strain rates, the period of strain hardening dominated the stress response.

3. The number of cycles required to attain the $(\Delta\sigma/2)_{max}$ increased with decreasing $\dot{\epsilon}$.

Fig. 1b: Cyclic stress-strain loop showing drop in stress after saturation with continued cycling.

Fig. 2: Normalised cyclic hardening curves illustrating the influence of strain rate at 823 K ($\Delta\epsilon_t = 0.8\%$)

4. The large normalised cyclic hardening at the lower $\dot{\epsilon}$ and the $\dot{\epsilon}$ dependence of $(\Delta\sigma/2)_{\max}$ indicated the negative strain rate sensitivity.

In addition to the negative strain rate sensitivity, markedly unstable plastic flow has been noted in the plastic regions of the stress strain hysteresis loops at the lowest $\dot{\epsilon}$ employed. The actual serrated flow observed in the lowest $\dot{\epsilon}$ test ($\Delta\epsilon_t = 0.8\%$, $\nu = 0.01$ Hz) could be ascertained from the compiled diagram (Fig. 3) depicting the strain range dependence of the serrated flow. Fig. 3 comprises the magnified portions of the

Fig. 3 Stress strain hysteresis loops in the compression peak portion indicating serrated flow at different total strain ranges (Grain Size: $310 \mu\text{m}$, Temperature 823K, Frequency 0.01Hz)

compressive peak regions of 5th cycle in each strain range test. The magnitude of serrations depended on $\Delta\epsilon_t$ to a large extent; and the regular serrations characterised by an abrupt rise in the flow stress followed by discontinuous drops back to or below the general level of the stress-strain curve i.e. type A serrations [12] occurred only at higher strain ranges.

The serrated flow in fatigue deformation, the negative strain rate sensitivity of the cyclic hardening and the associated decrease in N_f at the sub-creep temperatures have been attributed earlier to DSA effects [6, 13]. Results shown in Table I and Fig. 2 reveal that there is increase in cyclic stress when strain rate is changed from 1.6×10^{-2} to $1.6 \times 10^{-3} \text{ s}^{-1}$. However the stress strain hysteresis loops remain smooth at strain rates of $1.6 \times 10^{-3} \text{ s}^{-1}$ (Fig. 1). These results indicate that DSA commenced earlier than it manifested in the form of serrations in cyclic stress strain curves.

The higher work hardening rates during DSA in monotonic tension have been generally shown to be associated with increased rates of dislocation accumu-

lation /14-18/. In fact, the abnormally high rate of cyclic hardening (Fig. 2) observed in the lower $\dot{\epsilon}$ tests could be associated with increased rates of dislocation storage due to DSA. At the temperature employed in this study DSA during tensile deformation of austenitic stainless steels has been considered to result from the locking of mobile dislocations by substitutional solute atoms. Chromium has been identified as the species responsible for dynamic strain ageing. Vacancies produced during deformation enhance the diffusion of solute atoms to effect locking of dislocations temporarily held up at penetrable obstacles /19-21/. All the more, the conditions conducive to rapid dislocation arrest might be achieved quickly during fatigue deformation by enhancing the solute diffusion, since reverse straining is known to create vacancies at a rate greater than unidirectional deformation /22-24/. The enhancement of dynamic strain ageing in fatigue deformation has also been considered to arise from the longer arrest time of dislocations in fatigue deformation in comparison with the monotonic tensile deformation /8/. The longer arrest time in fatigue deformation is considered to be related to cell shuttling motion of dislocation. Increased amount of dislocation storage and solute locking of dislocations could also be invoked to explain the stronger effect of DSA on $(\Delta\sigma/2)_{\max}$ than on $(\Delta\sigma/2)_1$ (see Table 1) because it is the strain hardening process which is affected to a maximum extent by dislocation generation and immobilisation. It may be mentioned that DSA would further enhance inhomogeneity of deformation during fatigue due to solute locking of the slow moving dislocation between the slip bands. Also grain boundary regions may act as preferred sites for DSA to occur /17, 28-30/, further-leading to inhomogeneity of deformation.

II. Influence of DSA on LCF Fracture

The representative micrographs of the fracture surface obtained in scanning electron microscopy studies for three strain rates are presented in Figs. 4 and 5. At the strain rate of $1.6 \times 10^{-2} \text{ s}^{-1}$ (Fig. 4a) and $1.6 \times 10^{-3} \text{ s}^{-1}$ (Fig. 4b) the fatigue crack propagated transgranularly by the ductile striation formation mechanism. Crack growth occurred by mixed trans and intercrystalline mode at the lowest strain rate of $1.6 \times 10^{-4} \text{ s}^{-1}$ (Fig. 5). Incidentally, the occurrence of serrated flow (Fig. 3) coincided with the formation of intergranular cracking (Fig. 5). Apparently this intergranular damage does not seem to have resulted from either oxidation or creep

Fig. 4a: Ductile striations on fracture surface ($\dot{\epsilon} = 1.6 \times 10^{-2} \text{ s}^{-1}$)

effects. It should be noted that, in the event, where negative strain rate sensitivity exists and no serrated flow occurs ($\dot{\epsilon} = 1.6 \times 10^{-3} \text{ s}^{-1}$), crack propagation remains transgranular (Fig. 4b).

Fig. 4b: Ductile striations on fracture surface ($\dot{\epsilon} = 1.6 \times 10^{-3} \text{ s}^{-1}$)

Figs. 5a and 5b show that the impingement of slip bands on grain boundary is responsible for the occurrence of intergranular brittle decohesion. Conditions that lead to intergranular cracking by impingement of slip bands during high cycle fatigue of polycrystalline copper have been analysed theoretically and experimentally by Mughrabi *et al.* /25/. It was concluded that the intergranular decohesion due to intersection of persistent slip bands (PSB's) stems from the long range stresses associated with the piling up of the matrix-slip band interface dislocations. According to Stroh /27/, the largest tensile stress at the head of a dislocation pile up occurs across a plane making an angle of 70.5°

with the pile up. Mughrabi *et al.* /25/ consider that in real situations these planes lie almost in the grain boundary itself. The intergranular decohesion observed (Fig. 5a and b) in this study could be attributed to

(a)

(b)

Fig. 5 Intergranular decohesion due to impingement of slip bands ($\dot{\epsilon} = 1.6 \times 10^{-4} \text{ s}^{-1}$)

pile ups associated with PSB-matrix interface dislocation /25/. Such localised deformation bands are considered to be mainly a surface phenomenon and confined to the surface grains /26/. But, the observations in the present investigations (Fig. 5) suggests that the persistent slip can even occur in the bulk of the specimen.

III. Influence of DSA on LCF Life

DSA can influence the N_f in several ways. The reasons which lead to the reduction in N_f can be qualitatively derived from the following factors:

1. Use of the net tensile hysteresis energy in a fatigue cycle has been proposed by Ostergren /4/ as

energy measure of fatigue damage. The equation is:

$$\left[\left(\frac{\Delta\sigma}{2} + \sigma_m \right) \Delta\epsilon_p \right] N_f^\beta = \text{constant}$$

where $\Delta\sigma/2$ is the stress amplitude associated with a given $\Delta\epsilon_p$, σ_m = tensile mean stress, N_f = number of cycles to failure and β - slope of Coffin-Manson plots.

Since for the different strain rate conditions in this investigation, $\Delta\epsilon_p$ is approximately equal and $\sigma_m = 0$, it can be expected that increased fatigue damage per cycle may result partially from the increase in $\Delta\sigma/2$.

2. It is well known that endurance at higher strain ranges is mainly governed by crack propagation stage. From the crack growth rate (dA/dN) expression of Smith *et al.* /31/:

$$\frac{dA}{dN} \propto \Delta\sigma \cdot \Delta\epsilon$$

where $\Delta\sigma$ = stress range and $\Delta\epsilon$ is the applied strain range. From this it can be calculated that, for a given $\Delta\epsilon$, dA/dN will increase with an increased stress response at the lower strain rates due to DSA.

3. Higher stresses reduce the critical crack size for final fracture, reducing N_f /32/.

4. The combined action of the crack tip stress field and the stress concentration at the front of pile-up leads to intergranular cracking. Reduction in N_f occurs due to accelerated crack propagation in the event of intergranular cracking.

4. CONCLUSIONS

1. Fatigue life decreases with decreasing strain rate in the sub-creep temperature regime due to DSA effects.

2. Maximum cyclic tensile stress response increases markedly under the influence of DSA and stress response dependency on strain rate is negative.

3. The occurrence of serrated flow is not a necessary prerequisite for characterizing the DSA process.

4. DSA in LCF enhances the plastic deformation inhomogeneity.

5. The LCF fracture mechanism is altered due to DSA. Preferentially, brittle intergranular decohesion results from the occurrence of inhomogeneous plastic deformation.

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