Strain-rate effect on high temperature low-cycle fatigue deformation of AISI 304L stainless steel

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The effect of strain rate on the behaviour of high temperature low-cycle fatigue is investigated for AISI 304L stainless steel. Regardless of the test temperature of 873 or 973 K, the fatigue life is saturated in the strain-rate range of slower than $4 \times 10^{-3}\,\mathrm{sec}^{-1}$. Also it is interesting to note that serrated flow, which is evidence of the occurrence of dynamic strain ageing, is clearly observed in the load–elongation hysteresis loops for strain rates that are slower (at 873 K) and faster (at 973 K) than $4 \times 10^{-3}\,\mathrm{sec}^{-1}$. Since the combination of temperature and strain rate is concerned with the phenomenon of dynamic strain ageing, it is considered that the above-mentioned saturated fatigue life at 873 K is caused by dynamic strain ageing and that the hardening effect due to dynamic strain ageing abnormally increases the fatigue life. However, even though the behaviour of fatigue life under strain rates slower than $4 \times 10^{-3}\,\mathrm{sec}^{-1}$ at 973 K has nothing to do with the dynamic strain ageing, it has been found that the failure life is also saturated in this slower strain-rate range. This behaviour is considered to be caused by the effect of creep, because the deformation under the low strain-rate activates the recovery process and as a result it causes saturation of the inelastic strain range.

1. Introduction

Nowadays, high temperature low-cycle fatigue (HTLCF) has become an important topic in safe-life design consideration [1]. Investigation of the influence of the cycling frequency or the effect of the strain rate on HTLCF behaviour is one of the objectives of this consideration. To predict the safe service life of a structure very effectively, the effect of frequency on the development of damage should be considered. One of the significant results for the expression of the fatigue life (N_f) depending on the frequency (f) can be phenomenologically represented as follow [2, 3]:

$$N_f f^{(k-1)} = \text{constant} \qquad (1)$$

This relation shows that, with decreasing frequency, k decreases from 1 as in pure fatigue, via $1 > k > 0$ as in the interaction between creep and fatigue, to 0 as in pure creep. Many materials [2–5] have been shown to be in general agreement with this empirical relation in the regime of intermediate frequency. However, some experimental results show that this equation may not be good in the regime of frequency both at very high and low values. The experimental investigation of Udimet 700 at 760° C by Organ and Gell [6] shows that, in the high-frequency range, a maximum fatigue life is observed and then the fatigue life is decreased with increasing frequency, whereas it is expected from Equation 1 that N_f should not be affected by the frequency. In the regime of low frequency, the equation predicts a continuously decreasing fatigue life with frequency as $k \to 0$. However, Manson [7, 8] and other investigators [9, 10] show a saturated behaviour of fatigue life with decreasing frequency. Likewise, observing these contradictory experimental results at the regime of low frequency, del Puglia [11] raised the question of the application of the frequency-modified fatigue life Equation 1 at low frequency, whether N_f is decreased as $k \to 0$ or saturated with decreasing frequency.

This present investigation is focused on understanding the fatigue behaviour under low frequency or low strain rate, and proposes another suggestion for a life prediction method using the concepts of creep and environmental damages.

2. Experimental procedures

Specimens of Type 304L stainless steel, whose chemical composition is shown in Table I were prepared with a gauge length of 6 mm and diameter 4 mm. A drawing of the specimens is shown in Fig. 1; they were machined from hot-rolled slabs of 11 mm thickness in such a way that the loading axis of the specimens was parallel to the rolling direction of the slab. After machining, specimens were solution-annealed at

TABLE I Chemical composition of AISI 304L stainless steel

Element	C	Si	Mn	P	S	Ni	Cr	Mo	Cu
Amount (wt %)	0.026	0.287	0.80	0.027	0.032	10.14	18.09	0.40	0.498

0022–2461/88 $03.00 + .12

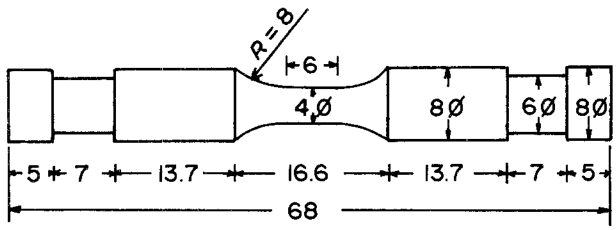

Figure 1 Drawing of the specimen used for fatigue tests (unit : mm).

1373 K for 1 h and aged at 1053 K for 50 h in order to prevent the possible change of microstructure during fatigue testing as reported earlier [12]. The surfaces of the specimens were polished like a mirror using emery papers upto No. 4000.

Fatigue tests were carried out with an Instron Model 1362, which is the electromechanically driven closed-loop machine. The testing set-up is similar to that described previously [13]. The specimen was heated using a radiant heating device and the temperature was maintained within the accuracy ± 2 K as measured on the top and bottom of the gauge section. To avoid the experimental oxidation effect at high temperatures, fatigue tests were performed in a high-purity argon atmosphere. The axial strain was controlled with an extensometer set on the specimen holder situated outside the furnace. The equivalent strain in the gauge length was calibrated using an elastic–plastic analysis, which is based on the measured cyclic load–elongation hysteresis loop. A push-pull symmetrical triangular waveform was used to control the applied total strain amplitude of 2.0% at the test temperature of 873 K (0.48 T_m) and 973 K (0.54 T_m).

For the failure criterion, the concept of critical fatigue life (N_c) is used and the concept has been reported earlier [13, 14] as the fatigue cycle corresponding to the beginning of unstable crack growth. All the measurements of cyclic stress and plastic strain were made from the hysteresis loop obtained at $\frac{1}{2}N_c$. For this investigation, Equation 1 is rewritten by replacing N_c for N_f and strain rate ($\dot{\varepsilon}$) instead of frequency (f), and it is expressed as

$$N_c \dot{\varepsilon}^{k-1} = \text{constant} \qquad (2)$$

This modified equation expresses the influence of creep damage on the fatigue behaviour up to the life of the beginning of unstable crack growth. Because the effect of strain rate with a constant strain amplitude has the same dimensions to that of the frequency as the reciprocal of time for one cycle, it is well known that a slow deformation rate is responsible for creep damage.

3. Experimental results and discussion

3.1. The behaviour of fatigue life and the characteristics of the hysteresis loop with strain rate at 873 K and 973 K

The critical fatigue lives were measured for various strain rates at 873 K (0.48 T_m) and 973 K (0.54 T_m) and they are shown in Figs 2 and 3, respectively. In both cases, near to the strain rate of 4×10^{-3} sec^{-1}, the slope of the plot for N_c against $\dot{\varepsilon}$ is observed to be changing drastically. In other words, for strain rates slower than 4×10^{-3} sec^{-1}, irrespective of test temperature (i.e. at 873 K or 973 K), Type 304L stainless steel is observed to have the tendency of saturating fatigue life with respect to strain rate, and a similar result has been reported for Type 304 stainless steel [10]. Using the concept of Equation 2, from the plots in Fig. 2, k is measured to be 0.47 at the faster and 0.83 at the slower strain rate than 4×10^{-3} sec^{-1} at 873 K. Similarly, at 973 K (Fig. 3), k is calculated to be 0.27 at the faster and 0.85 at the slower strain rate than 4×10^{-3} sec^{-1}.

The increasing tendency of the value of the exponent k, from 0.47 to 0.83 at 873 K and from 0.27 to 0.85 at 973 K with decreasing strain rate, respectively, is just the opposite to the suggestion of Coffin [2, 3, 15], who considers that the frequency dependence of his work is mainly caused by the environmental effect. He has pointed out that as the frequency decreases the time for the environmental effect has been increased, to give a longer period per cycle to enhance the oxidation of the material and to damage it. As a result, he claims that a decreasing tendency of the exponent k with frequency has to be observed. However, concerning the effect of oxidation, another criticism has been made [11]; since the oxidation occurs in a relatively short time, a longer period per cycle does not necessarily cause much more environmental damage. On the other hand, Manson [7, 8] has suggested that the fatigue behaviours reported by Coffin [2] were confined

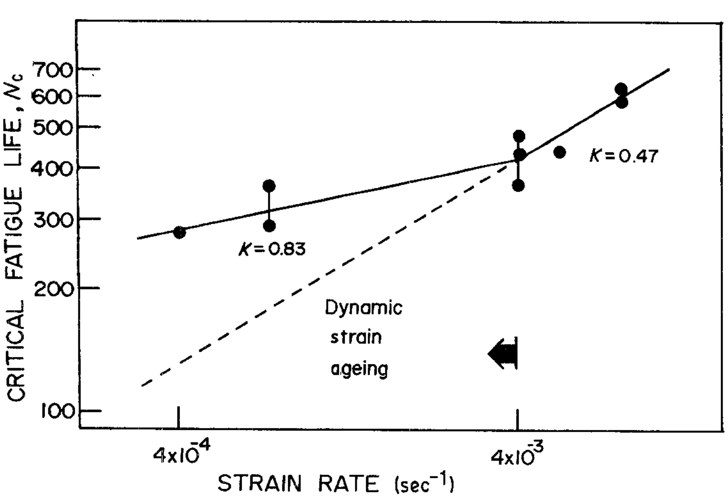

Figure 2 Strain-rate dependence of fatigue life at 873 K. DSA is observed under the strain rates slower than 4×10^{-3} sec^{-1}. $\pm 2.0\%$, $T_t/T_c = 1$, $N_c\dot{\varepsilon}^{k-1}$ = constant.

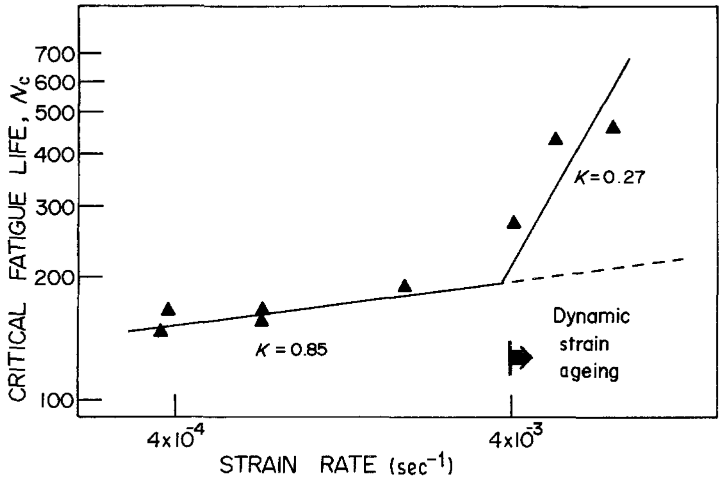

Figure 3 Strain-rate dependence of fatigue life at 973 K. DSA is observed at strain rates above $4 \times 10^{-3}\,sec^{-1}$. $\pm 2.0\%$, $T_t/T_c = 1$, $N_c\dot{\varepsilon}^{k-1}$ = constant.

to certain test conditions, i.e. continuous cycling with symmetrical triangular wave forms. Even in a vacuum environment, Manson [7] has found that a long hold-time at the maximum strain period might cause more damage than continuous cycling. From this result, he confirmed that the influence of the creep phenomenon on fatigue life is at least as important as that of environment. Also the experimental results by Coffin [15], even in a vacuum at very low frequency, show that the fatigue life is decreased with frequency. This may indicate that the creep effect rather than the environmental one is the dominant process under a very low frequency regime or at a low strain rate. Anyhow, both creep and environmental processes are known to be not only time-dependent but also thermally activated, and are affected by the strain rate during HTLCF.

In addition to this, dynamic strain ageing (DSA) is also one of the time-dependent and thermally activated processes [16, 17]. The process of DSA during fatigue deformation at high temperature near $0.5\,T_m$ for Type 304 stainless steel is known to be produced by the interaction between chromium atoms and dislocations [16]. It is known that one of the pieces of evidence for the phenomenon of DSA is the serrated flow which can be shown in the stress–strain curve [17]. For this work, typical characteristics of the hysteresis loops for load and elongation are shown in Fig. 4 for the best temperatures of 873 K (Figs 4a and b) and 973 K (c and d). The loops obtained during push-pull fatigue tests with a symmetrical strain rate of $8 \times 10^{-3}\,sec^{-1}$ are shown in Figs 4a and c and those of $3.8 \times 10^{-4}\,sec^{-1}$ are in Figs 4b and d. When all of the hysteresis loops are checked with strain rates, serrations are clearly observed in the loops obtained at strain rates slower than $4 \times 10^{-3}\,sec^{-1}$, both in tension and compression for the test temperature of 873 K. On the other hand, for that of 973 K, serrations are only observed at strain rates faster than $4 \times 10^{-3}\,sec^{-1}$. This temperature effect produces the opposite effect on the serration behaviour, depending on the strain rate. From these results, one can understand that DSA is affected by both temperature and strain rate. Also for this reason, DSA is called a time-dependent and thermally activated process, which can influence the behaviour of fatigue life at high temperatures just as creep or environmental damage can do.

It has been already mentioned that the values of the exponent k in Equation 2, i.e. the slopes shown in Figs 2 and 3, vary from 0.47 to 0.83 at 873 K and from 0.27 to 0.85 at 973 K with decreasing strain rate. Considering the variation of k, i.e. the fatigue life behaviour and the characteristics of the hysteresis loop at the same time, one may understand that the variation of k is influenced by the effect of hardening due to DSA. In other words, comparing the experimental results in the regime of DSA with the dotted line which is extrapolated from the results of that without DSA as shown in Figs 2 and 3, the fatigue life is observed to be increased in the regime of DSA. This abnormal behaviour of fatigue life may be caused by the effect of hardening due to DSA. However, at 973 K for the range of the strain rates slower than $4 \times 10^{-3}\,sec^{-1}$, the behaviour of fatigue life tends to be saturated without any evidence of DSA, i.e. the life is not decreased with strain rate.

3.2. The effect of creep in the regime of no DSA

The inelastic strain component is measured with the strain rate and is shown in Fig. 5. Little change of the inelastic strain amplitude is observed at 873 K and this may be related with DSA. However, the inelastic strain amplitude at 973 K is observed to be increased and saturated near to the strain rate of $4 \times 10^{-3}\,sec^{-1}$ with decreasing strain rate. Considering the condition of no DSA under strain rates slower than $4 \times 10^{-3}\,sec^{-1}$ at 973 K, the behaviour of the inelastic strain component under these conditions is considered to be due to creep damage rather than the environmental kind. A similar behaviour of the inelastic strain range variation with strain rate was observed and is concluded to be the creep component by Manson [8] in the application of the strain range partitioning method.

Another example of creep damage which can influence the deformation behaviour of HTLCF may be related to the variation of mean stress with strain rate for the test temperatures of 873 K and 973 K. A more general relationship between flow stress and strain rate at constant temperature and strain is called the mechanical equation of state [18]. This relation is

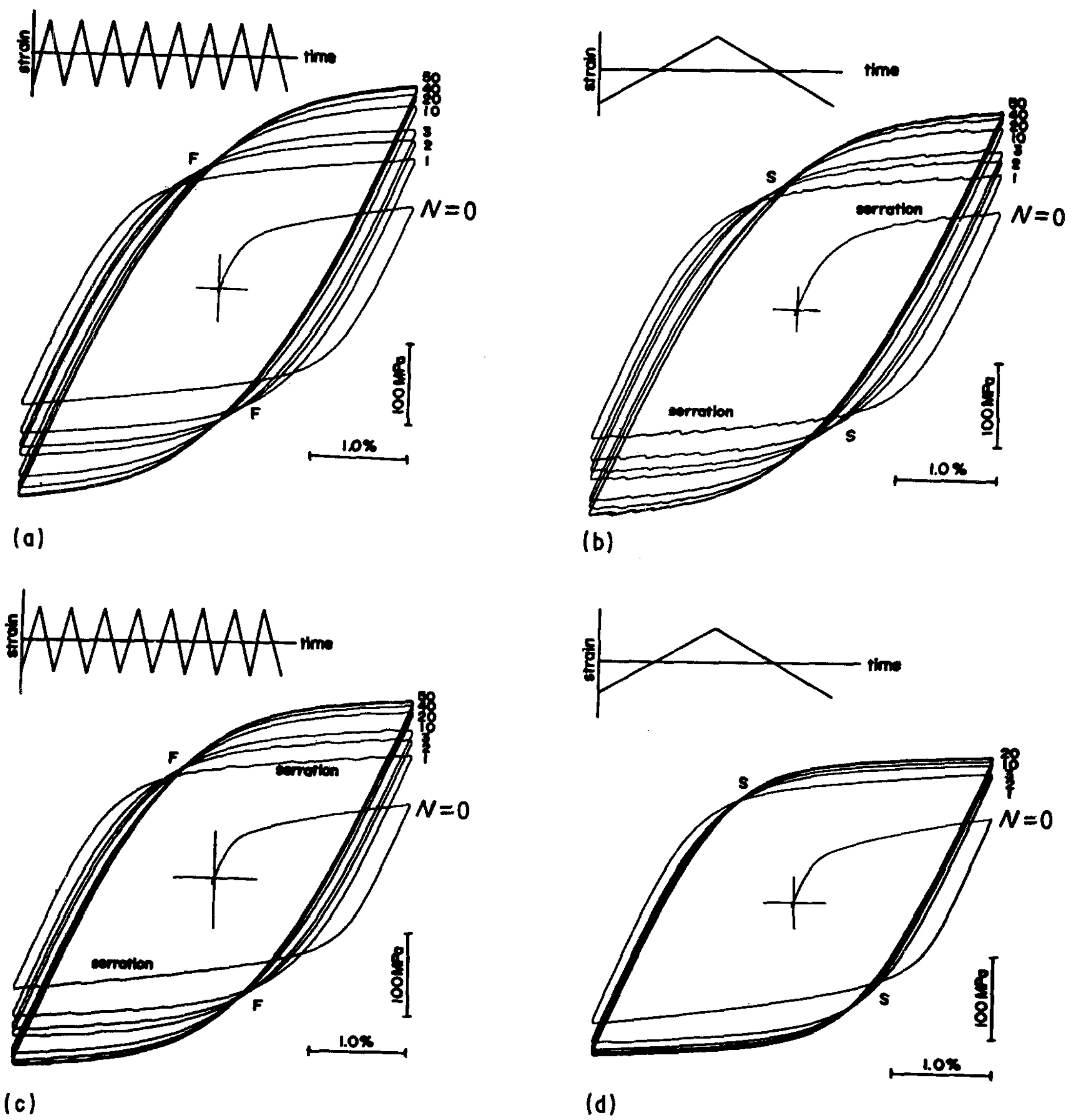

Figure 4 The stress–strain hysteresis loops obtained in strain-controlled fatigue deformation with an applied total strain amplitude of 2.0% at (a, b) 873 K and (c, d) 973 K. (a) The fast–fast equal ramp (F) with $8.0 \times 10^{-3}\,\text{sec}^{-1}$ shows a smooth stress–strain curve at 873 K. (b) The slow–slow equal ramp (S) with $3.8 \times 10^{-4}\,\text{sec}^{-1}$ shows serrated flow in the hysteresis loop at 873 K. (c) The fast–fast–equal ramp (F) with $8.0 \times 10^{-3}\,\text{sec}^{-1}$ shows serrated flow in the hysteresis loop at 973 K. (d) The slow–slow equal ramp (S) with $3.8 \times 10^{-4}\,\text{sec}^{-1}$ shows a smooth stress–strain curve at 973 K.

known to be applicable to a tensile test (or creep test) in the temperature range above $0.5\,T_m$, confirming that the deformation is stable and uniform [19, 20]. However, for strain amplitude-controlled fatigue tests, it is generally accepted and observed in this study that the level of stresses in tension and compression is saturated at $\frac{1}{2}N_c$, where the analysis of data has been made. It is also known that the development of dislocation microstructure, or cell structure, within the matrix is stable and homogeneous at this saturation stage [21, 22]. For this reason, it is thought that the relation is applicable to the fatigue tests of this work.

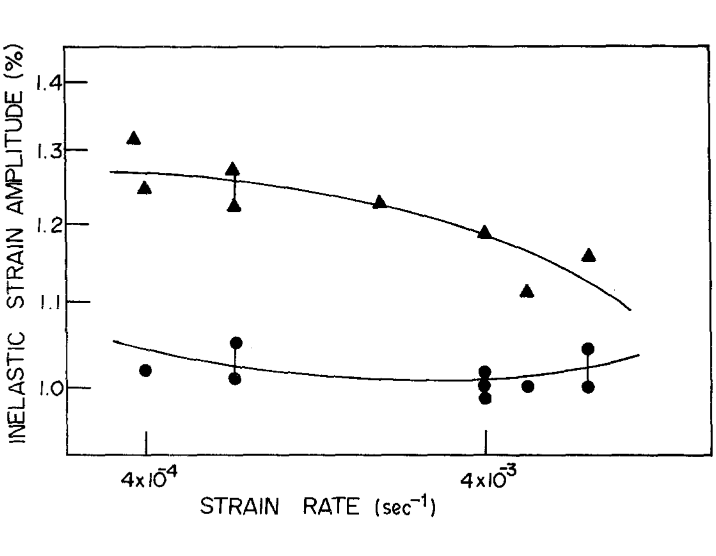

Figure 5 The variation of the inelastic strain component with the strain rate at (●) 873 K, (▲) 973 K. $\pm 2.0\%$, $T_t/T_c = 1$.

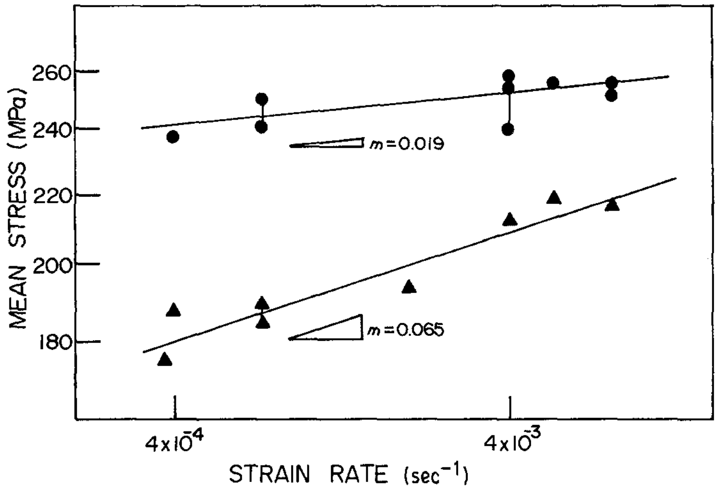

Figure 6 Strain-rate dependence of the mean stress during symmetrical cycling near $0.5\,T_m$: (●) 873 K, (▲) 973 K. $\pm 2.0\%$, $T_t/T_c = 1$, $\sigma_m = A\dot{\varepsilon}^m$.

Using the concept of the mechanical equation of state [18], the variation of mean stress σ_m (MPa) with strain rate $\dot{\varepsilon}$ (sec^{-1}) can be expressed as

$$\sigma_m = A\dot{\varepsilon}^m \quad (3)$$

where A is the coefficient and m is the index of strain-rate sensitivity. It is also known that, as the index of strain rate sensitivity (m) increases, the recovery process is becoming more dominant [19, 20]. When the value of m is calculated by the least-squares method at each test temperature of 873 and 973 K, a higher m value of 0.065 is obtained at 973 K than that of 0.019 at 873 K, and this is shown in Fig. 6. This shows that the recovery process is more dominant at 973 than at 873 K.

Observing the more significant decrease of the mean stress with decreasing strain rate at 973 K and, again, considering the condition of no DSA for strain rates slower than $4 \times 10^{-3}\,sec^{-1}$ at 973 K in this work, this increased and saturated value of the inelastic strain amplitude under these test conditions is considered to be due to the recovery process caused by the effect of slow strain rate. The effect of very slow strain rate on the fatigue deformation is expected to decrease the cyclic work-hardening characteristics and therefore to increase the amount of deformation per cycle near the high temperature of $0.5\,T_m$, in the regime of no DSA. As a result, it is thought that the fatigue life is saturated due to creep damage in the slower strain rate range of $4 \times 10^{-3}\,sec^{-1}$ at 973 K.

4. Conclusions

By examining the fatigue behaviour of Type 304L stainless steel near at $0.5\,T_m$, the following conclusions can be drawn.

1. For strain rates slower than $4 \times 10^{-3}\,sec^{-1}$ at 873 K, serration is clearly observed both in tension and compression. On the other hand, for the test temperature of 973 K, serration occurs only at strain rates faster than $4 \times 10^{-3}\,sec^{-1}$.

2. The fatigue life is shown to be saturated with decreasing strain rate both at 873 and 973 K. According to the strain-rate modified equation, k is found to be increasing with decreasing strain rate and this observation is known to be the opposite to the result of Coffin. The change in the value of k with strain rate is considered to be caused by dynamic strain ageing. In other words, the abnormally increased fatigue life is observed in the regime of dynamic strain ageing.

3. Under strain rates slower than $4 \times 10^{-3}\,sec^{-1}$ at 973 K, the saturated fatigue life without any dynamic strain ageing is considered to be caused by the effect of creep.

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