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Hold-time effects on high temperature fatigue crack growth in Udimet 700

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Crack growth behaviour under creep–fatigue conditions in Udimet 700 has been studied, and the crack growth data were analysed in terms of the stress intensity factor as well as the J-integral parameter. Crack growth behaviour is shown to depend on the initial stress intensity level and the duration of hold-time at the peak load. For stress intensities that are lower than the threshold stress intensity for creep crack growth, the crack growth rate decreases with increase in hold time even on a cycle basis, da/dN, to the extent that complete crack arrest could occur at prolonged hold times. This beneficial creep–fatigue interaction is attributed to the stress relaxation due to creep. For stress intensities greater than the threshold stress intensity for creep crack growth, the growth rate on a cycle basis increases with increase in hold time. For the conditions where there is no crack arrest, the crack growth appears to be essentially cycle-dependent in the low stress intensity range and time-dependent in the high stress intensity range. Both the stress intensity factor and the J-integral are shown to be valid only in a limited range of loads and hold-times where crack growth rate increases continuously.

1. Introduction

There has been an increasing concern in recent years that conventional fatigue and creep rupture data are inadequate in predicting life of a component or a structure, since many of these components are subjected to neither pure cyclic load nor pure static load,* but a combination of the two. It has been shown in many materials [1] that combined creep–fatigue conditions are more damaging in terms of service life of a component than either type alone. Most of these investigations, however, have been performed using round bar specimens with homogeneous stresses. Since failure in these types of specimens includes both a crack nucleation stage as well as a crack propagation stage with relative times in each stage varying with stress amplitude and material, it is difficult from the above studies to determine to what extent the combined load affects the crack propagation stage as compared to the crack nucleation stage. Knowledge of this is important since in many cases the nucleation stage is circumvented because of the existence of stress-concentrated regions due to sharp defects, bending moments or thermal gradients.

There has been some recent work [2–10] using a fracture mechanics-type of specimen to determine the extent of creep–fatigue damage on the crack propagation stage. The general conclusion from these studies is that depending upon the material and the temperature of the test, crack growth rates increase with the introduction of static hold periods during cycling. Some authors have attributed this to creep effects and some others to environmental interactions. At the temperatures where the hold-time effects are observed, both creep deformation as well as environmental interactions occur to a varying degree depending upon the alloy and its processing history, and it becomes difficult to separate creep effects from

*The terms static load and creep load are used here interchangeably although it is recognized that creep damage generally implies a process involving plastic deformation and does not necessarily include any environmental interactions that may occur under static load.

environmental effects unless the tests are performed in an inert environment. Where the effects are due to creep or environmental interactions or a combination of the two, it should be recognized that these effects are important and should be taken into consideration in design of high temperature components.

Previous analyses of high temperature crack growth behaviour under static [11] and cyclic [12] loads in Udimet 700 showed that there is a large disparity in the crack growth behaviour under the two types of loads. In particular, the threshold stress intensities, crack growth rates, as well as the fracture mode are significantly different under the two types of loads. This implies that superposition of static hold periods during cycling should significantly affect the fatigue crack growth behaviour in this alloy. In fact, it was shown earlier [13, 14] using unnotched specimens that at the temperature of 760° C and above, hold-time decreases significantly the fatigue life of this alloy. The purpose of the present investigation was to determine to what extent hold-time influences the crack propagation stage and how these effects depend on the applied load. Additionally, since hold-time effects, which are essentially non-linear, are superimposed in each cycle, it becomes pertinent to study the validity of fracture mechanics parameters in predicting the crack growth behaviour in this alloy under these complex loading conditions.

2. Experimental details

The details of heat treatment and chemical composition of the alloy used are listed in Table I. Modified compact tension specimens [15] (1T dimensions but with 12.7 mm thickness) were loaded in a MTS machine. The crack growth behaviour at various cyclic loads was studied at 850° C. Selected tensile hold periods were superimposed in each cycle without altering the loading and the unloading rates. Displacement of the specimen during cycling was measured using a strain gauge extensometer via quartz rods affixed across the notch of the specimen. Load–displacement plots at various crack lengths were obtained using an x–y recorder. Crack lengths were measured periodically during the tests by means of a travelling microscope. Crack growth rates were determined from the plots of crack length versus number of cycles. Since both cyclic and static loads are combined, the crack growth data were analysed both on a cycle basis, $\mathrm{d}a/\mathrm{d}N$, as well as on a time basis, $\mathrm{d}a/\mathrm{d}t$, to determine which is the controlling factor for crack growth under combined load conditions. The results are compared with those under cyclic as well as with those under static loads.

3. Results and discussion

3.1. Crack growth behaviour: Stress intensity factor

The calculation of the stress intensity factor was made for a/W values up to 0.7 from the expression [15]:

$$K = \frac{P}{BW^{0.5}}\,[29.6(a/W)^{0.5} - 185.5(a/W)^{1.5} + 655.7(a/W)^{2.5} - 1017.0(a/W)^{3.5} + 638.9(a/W)^{4.5}] \quad (1)$$

and for $a/W > 0.7$ from the relationship [16]:

$$K = \frac{P}{2B}\;\frac{W+a}{(W-a)^{1.5}}\quad 4.0 + \frac{(W-a)}{(W+a)} \quad (2)$$

where P is the applied load, a is the average crack length, W is the specimen width, and B is the thickness. The stress intensity factor range, ΔK, is obtained from $\Delta K = K(1-R)$ where R is the ratio of minimum load to maximum load.

Crack growth rates on a cycle basis, $\mathrm{d}a/\mathrm{d}N$, are plotted as a function of stress intensity factor range for various loads and hold times, Fig. 1. The dashed line represents the fatigue crack growth rate data at zero hold time obtained earlier [12].

TABLE I (a) Composition of Udimet 700 (wt %)

C	Mn	Si	Cr	Ni	Co	Fe	Mo	Ti	Al	B	Zr	S	P	Cu
0.06	0.10	0.10	15.1	Bal.	16.6	0.23	4.95	3.48	4.15	0.025	0.04	0.003	0.01	0.1

TABLE I (b) Heat treatment of Udimet 700

Anneal at 1180° C for 4 h and air-cool
Intermediate anneal at 1080° C for 4 h and air-cool
Age at 845° C for 24 h, furnace-cool to 760° C, hold at 760° C for 16 h and furnace-cool.
ASTM grain size 3

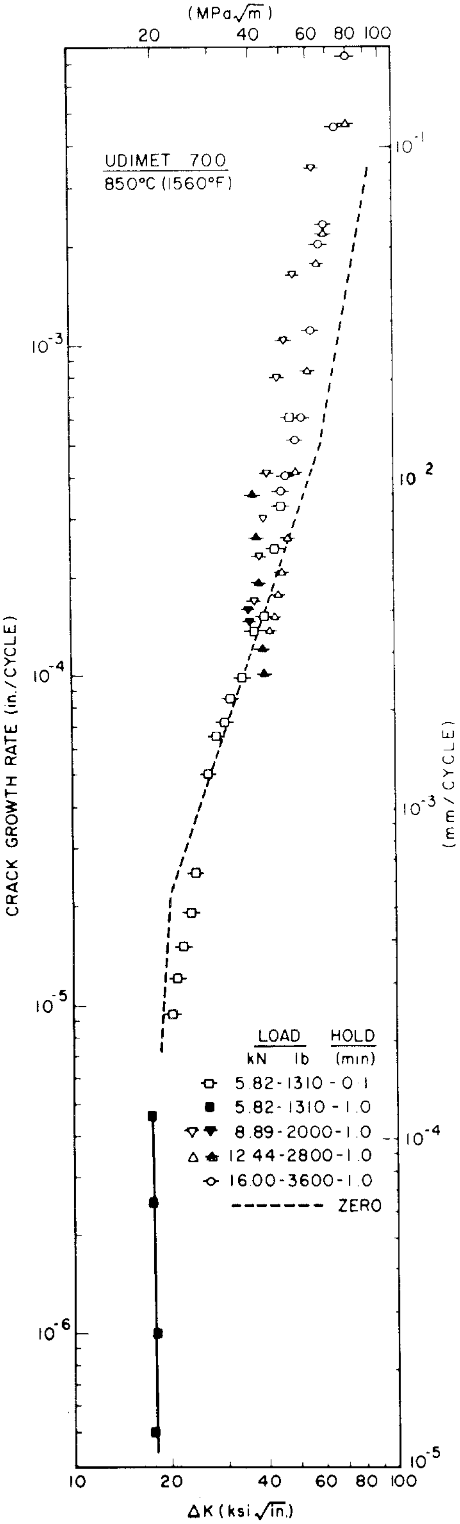


Figure 1 Crack growth rates, da/dN, as a function of the stress intensity factor range for various loads and hold-times.

Crack growth behaviour in the Udimet alloy under creep–fatigue conditions depends on the load, or more likely on the initial stress intensity, as well as on the hold-time. At low loads* and small hold-times, corresponding to 5.82 kN and 0.1 min hold-time, the crack growth rates are not very much different from the zero hold-time data. There is a trend, however, towards decreased growth rates at low ΔK values (<25 MPa m$^{1/2}$) and increased growth rates at high ΔK values (>45 MPa m$^{1/2}$) as compared to the zero hold-time data. The decreased growth rates are probably due to an increased threshold stress intensity value with hold-time. Such an increase in threshold stress intensity with hold-time was also observed earlier [8] in Alloy 718. This increase in threshold stress intensity could be attributed to the stress relaxation that could occur during hold-time which could effectively decrease the stress intensity at the crack tip.

That the observed effects of hold-time at low ΔK values are not just part of the scatter in the crack growth data, but part of a general trend in the behaviour, may be noted from the results at 1 min hold-time at the same load. For this case, crack growth occurs initially but the growth rate decreases rapidly with time to the extent that there is complete arrest in the crack growth after some number of cycles. Because of the increased hold-time, creep relaxation may be large to the extent that the decrease in the stress intensity value due to stress relaxation could be much more than the increase in the stress intensity value due to increase in crack length, thereby contributing to the crack arrest. It may be noted here that such a crack arrest phenomenon associated with a creep relaxation process was also observed earlier [17] under pure static load in Type 308 weld metal at 590° C, especially at low ΔK values.

Coming back to the 0.1 min hold-time data, the increase in growth rates at ΔK values greater than 45 MPa m$^{1/2}$, as compared to the zero hold-time curve, is probably related to the earlier onset of the Stage III crack growth process. Since Stage III has been shown [12] to be related to the superposition of the creep crack growth process over the fatigue crack growth process, increased creep contribution should be expected during a hold-time test. The net result is larger growth rates, at least

*For convenience, we shall refer to the loads that are below 7.12 kN as low loads, those that are above 13.35 kN as high loads and those that are in between as intermediate loads.

on a cycle basis, as compared to the zero hold-time test.

The effects of hold-time on the crack growth behaviour at high loads is somewhat different from that observed at low loads. At loads corresponding to 16 kN and hold-times corresponding to 1 min crack growth rates increase continuously and the growth rates on a cycle basis are larger than those at zero hold-time. This means that for these loads crack growth occurs both during cycling as well as during hold-time. The crack growth during hold-time is expected since the applied load is high enough to be in the range where creep crack growth occurs in this alloy.

A complex combination of the effects at low loads and of those at high loads are noted during the hold-time tests at intermediate loads. For a specimen tested at 12.44 kN with 1 min hold-time, the crack growth rate decreases initially as in a specimen tested at a low load, but increases later as the crack length increases. For this case, although the creep relaxation slows down the crack growth process initially, the fatigue crack growth rates are large enough to overcome this relaxation, contributing to the upward turn in the crack growth rates. As the ΔK value increases further, the creep crack growth process also occurs thereby increasing the growth rates over those at zero hold-time. The specimen that did not fail during the hold-time test at the low load was precracked again and tested at 8.89 kN with 1 min hold-time and the results are shown in Fig. 1. The growth rates for this specimen also decrease initially and increase later as the crack length increases. The data points corresponding to the decreasing part of the growth rates are filled to distinguish them from those corresponding to the increasing part of the growth rates. The crack growth rates in the 8.89 kN specimen are slightly larger than those in the other hold-time tests especially in the high ΔK region. Since the specimen was at 850° C for nearly 700 h before the second precracking, higher growth rates in this specimen could be due to possible changes in the microstructure due to an overageing process. We shall show later that in the high ΔK range where creep effects are important, crack growth rates are very sensitive to any changes in the microstructure.

With all these complexities associated with hold-times during fatigue, it is obvious that stress intensity factor is not a valid parameter to predict the growth rates in the whole ΔK range. On the other hand, if one neglects the initial decreasing growth rate region as well as the data corresponding to the overaged specimen, then the rest of the crack growth rate data fall close together indicating that ΔK may be valid in this high ΔK range.

Finally, we may note that for ΔK less than 40 MPa $m^{1/2}$ the crack growth rates under hold-time test (provided hold-times are not long) are similar to those under fatigue tests. This indicates that to the extent that the creep relaxation process is negligible, the crack growth process in this ΔK range is essentially cycle-dependent.

The crack growth rate data are next analysed on a time basis, da/dt, and are shown in Figs. 2 and 3 along with data for pure static load, assuming R to be equal to zero for static tests. Therefore ΔK and K will be equal under static load. Note that the threshold stress intensity for creep crack growth is much larger than that for fatigue crack growth. Also, Fig. 2 shows the two extreme cases of crack growth behaviour corresponding to both low and high loads. At low loads, the disparity between the zero hold-time and 0.1 min hold-time data is much larger on a da/dt basis than on da/dN basis, indicating further that crack growth in this ΔK range is not time-dependent but cycle-dependent. With increasing hold-time to 1 min at this load, crack growth rates again decrease, ultimately contributing to a complete crack arrest. The results at high loads differ from those at low loads. At the high loads, corresponding to 16 kN, the crack growth behaviour becomes increasingly time-dependent with increasing ΔK as well as with increasing hold-time. This can be seen in Fig. 2 where the data for all the hold-times including that of fatigue converge to the pure static load data. The convergence occurs earlier with increasing hold-time.

Fig. 3 shows the crack growth behaviour at the intermediate loads. The crack growth rates for 1 min hold again decreases and then increases. The disparity between the zero hold-time curve and 1 min hold curve is again large in the low ΔK end, but they both converge to the static curve at the high ΔK end. Thus the crack growth process may be initially cycle-dependent but soon converges to a time-dependent process with increase in ΔK value.

3.2. Creep–fatigue interaction

The term creep–fatigue interaction has been used extensively in the literature [1] to characterize the

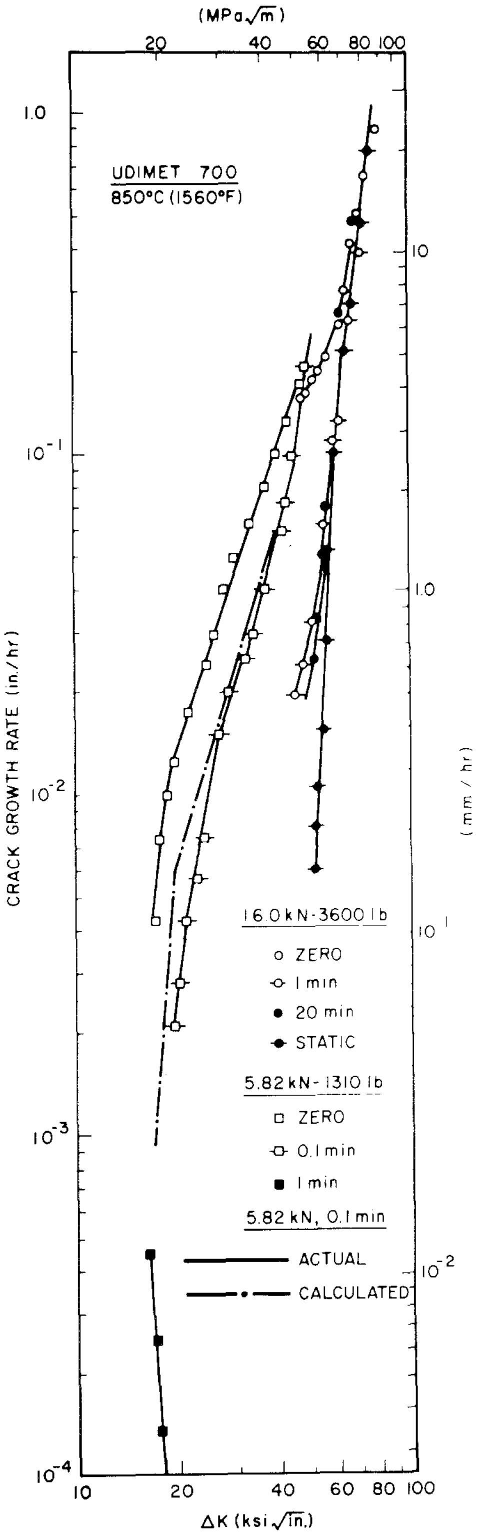


Figure 2 Crack growth rates, da/dt, as a function of stress intensity factor at low and high loads.

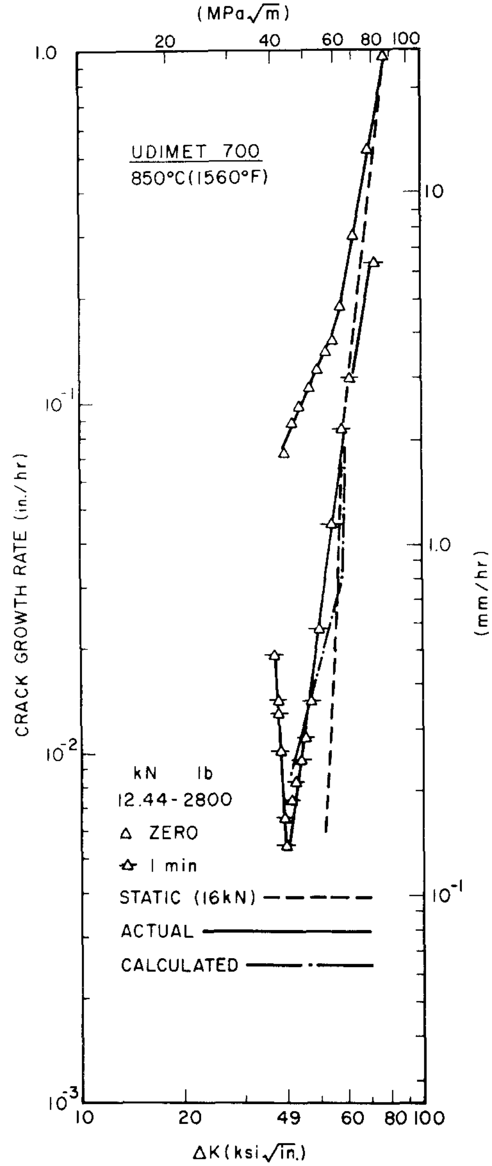


Figure 3 Crack growth rates, da/dt, as a function of stress intensity factor at intermediate loads.

damage that occurs under creep–fatigue conditions. Specifically, the interaction is said to exist only if the damage under creep–fatigue conditions is different from that obtained by the simple fractional addition of creep and fatigue damages determined separately. Thus, if there is no interaction, crack growth under creep–fatigue conditions can be calculated for a given ΔK value from creep and fatigue crack growth data using the relation

$$\frac{\Delta a_{cf}}{\Delta t_{cf}} = \frac{\Delta a_f}{\Delta N_f} \cdot \frac{\Delta N_f}{\Delta t_{cf}} + \frac{\Delta a_c}{\Delta t_c} \cdot \frac{\Delta t_c}{\Delta t_{cf}} \quad (3)$$

where

$\Delta a_f/\Delta N_f$ = crack growth rate under fatigue,
$\Delta N_f/\Delta t_{cf}$ = frequency during a hold-time test,
$\Delta a_c/\Delta t_c$ = crack growth rate under creep,
$\Delta t_c/\Delta t_{cf}$ = ratio of hold-time to total time in each cycle,
$\Delta a_{cf}/\Delta t_{cf}$ = crack growth rate under creep–fatigue conditions.

A similar relation could also be written for crack growth rates on a cycle basis. To keep the environmental effects nearly the same, it is necessary to obtain all the requisite data under the same environment.

The crack growth rates calculated on the basis of the addition rule is also represented in Fig. 2 and 3 for the low and intermediate loads. The data corresponding to 16 kN is used for the static component. For the 0.1 min hold-time at low load, the observed values are slightly lower than the calculated value but the two curves converge to each other with increasing ΔK. The lower crack growth rates could be due to stress relaxation during hold-time. Although the differences between the observed and calculated values are small and are within the scatterband observed for these tests, the differences may be real since these become magnified with increasing hold-time to 1 min. In support of this is the fact that the crack growth ultimately stops under a 1 min hold-time test at 5.82 kN, which indicates that a creep–fatigue interaction definitely exists, and in the present case, it is beneficial to the life of a component.

For the intermediate load in Fig. 3, the decrease in growth rates with ΔK indicates the existence of creep–fatigue interaction which is again beneficial. With further increase in ΔK, the growth processes become time dependent since the fatigue contribution becomes much smaller than the creep contribution at least for long hold-time tests. We may note that the beneficial interactions of the above type do not exist with further increase in loads to say 16 kN since Fig. 2 shows that for all hold-times there is a continuous increase in growth rates with ΔK. On the other hand, it should be possible to arrest the crack growth processes completely under creep–fatigue conditions even at a high initial ΔK value of 40 MPa $m^{1/2}$ by increasing the hold-time in each cycle, since in the limit of infinite hold-time (static test) crack growth does not occur for these ΔK values. We could, in fact, say that the beneficial creep–fatigue interaction exists for all cases where the initial ΔK value falls in between the threshold stress intensity value for creep crack growth and that for fatigue crack growth. Of course, the extent of this interaction depends on the ΔK value and the imposed hold-time in each cycle. In order to observe these interactions, it is necessary that the threshold stress intensities for creep and fatigue crack growth be significantly different. Furthermore, the temperature should be sufficiently high to insure that the stress relaxation occurs rapidly enough to contribute to the crack arrest. Indeed, such beneficial interactions were observed recently [8] in Alloy 718 specimens tested at 760° C especially at low loads and long hold-times.

3.3. J-integral parameter

Since the linear-elastic parameter is valid only for a limited range of ΔK values and hold-times, it is of interest next to analyse the crack growth rate data in terms of non-linear fracture mechanics par-

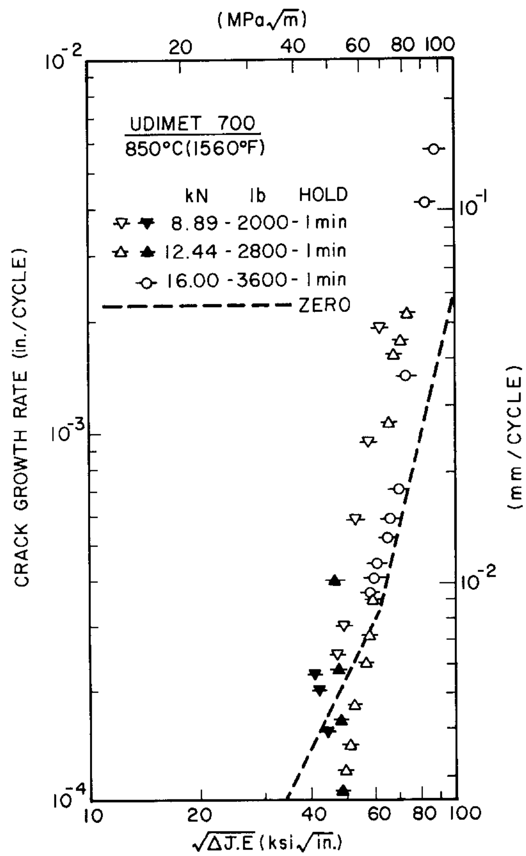


Figure 4 Crack growth rates, da/dN, as a function of J-integral parameter.

ameter, J-integral, [18] to see if the range of validity extends beyond that for ΔK. Using the load–extension curves and the Merkle–Corten estimation procedure for ΔJ-integral discussed in detail elsewhere [12], ΔJ is determined for various loads and crack lengths. The crack growth rates on a cycle basis are represented as a function of $\sqrt{(\Delta J \cdot E)}$ in Fig. 4 along with ΔJ-data for zero hold-time [12]. Comparison of Figs. 1 and 4 shows that hold-time effects are similar both in terms of ΔK as well as in terms of ΔJ. In particular, the initial decreasing growth rate region, as well as the relatively higher growth rates for the specimen loaded at 8.89 kN can both be seen also in terms of ΔJ. The only difference between the two figures is that ΔJ data fall slightly to the right of ΔK data.

3.4. Crack growth under creep

Since the crack growth behaviour both under fatigue as well as under creep–fatigue conditions converges to that under static load with increasing ΔK, it is important to discuss here a few aspects of the crack growth behaviour under static load that are pertinent to the present results. Most importantly, it will be shown that the crack growth behaviour in the creep range is very sensitive to small changes in the microstructure.

Fig. 5 shows creep crack growth behaviour in the Udimet alloy at 850° C. The data were obtained using specimens that underwent the same heat treatment but in two different batches. The dashed curve represents the average of several data points obtained earlier [11] using compact tension specimens but with side grooves. The solid line represents the data from Fig. 2. To insure that the disparity in the crack growth rates of the two batches is not due to the effect of side grooves, *per se,* one of the batch 2 specimens was side grooved, and its crack growth data are also shown in Fig. 5 as data points. Side grooves change the character of the creep crack growth curve but not to the extent that it can account for the disparity between the two batches. Difference in the hardness values of the two batches was found to be negligibly small. This could imply that any small variation in the microstructure could result in large variations in the creep crack growth rates. This behaviour is not specific to Udimet 700 alone since similar effects were noted earlier [19] for Alloy 718. Higher growth rates observed for the specimen that was left at 850° C for a long time

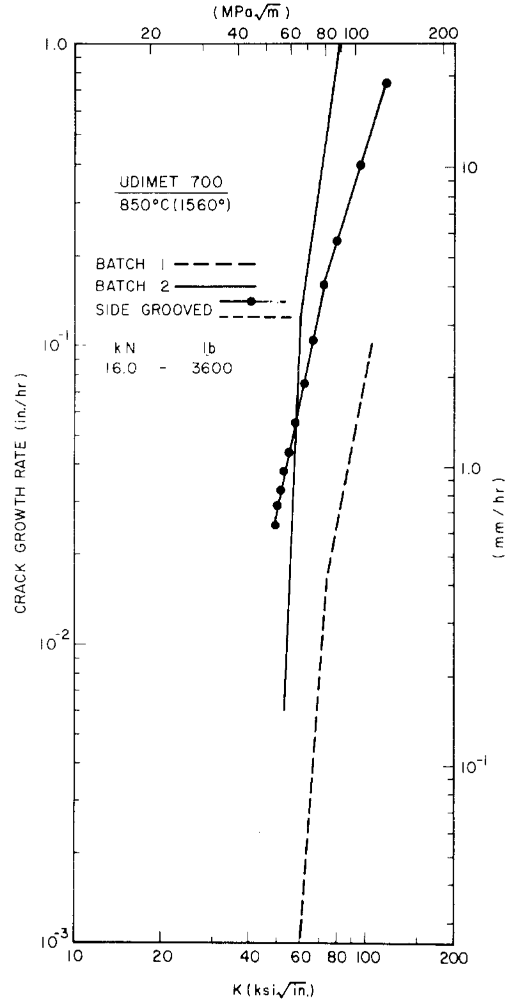


Figure 5 Creep crack growth rates, da/dt, as a function of stress intensity factor.

before the test (Fig. 1) could be due to microstructural changes that might have occurred in the specimen, as discussed in the text. In contrast to creep and creep–fatigue crack growth, fatigue crack growth is rather insensitive to small variations in the microstructure, since data from specimens from both batches fell on a single curve [12]. This sensitivity of the creep crack growth process to microstructure may be related to its intergranular fracture mode.

These results raise some important questions concerning the usefulness of the fracture mechanics approach to characterize the crack growth behaviour for design purposes. For example, the results imply that close control of the chemistry as well as the processing treatment is required to utilize the fracture mechanics techniques for design. Also, in terms of the evaluation of the fracture mechanics parameters itself, the results imply that great caution has to be applied in comparing the crack growth data obtained by different investigators. In particular, it becomes difficult to attribute the spread in the crack growth rate data as characteristic of a particular parameter when microstructural variations could also contribute to such a spread.

3.5. Fracture surfaces

Fracture surfaces were examined under a scanning electron microscope to evaluate the micromechanisms of crack growth in the Udimet alloy under creep–fatigue conditions. Fig. 6 shows the surface characteristics at various crack lengths in a specimen tested at 8.89 kN with a 1 min hold-time. Fig. 6a shows the region of transition between the room temperature pre-crack and high temperature crack. Although the crack propagation is essentially transgranular with the fine cleavage-type steps characteristic of fatigue crack growth, there is some intergranular crack growth in the small region close to the pre-crack. This region coincides

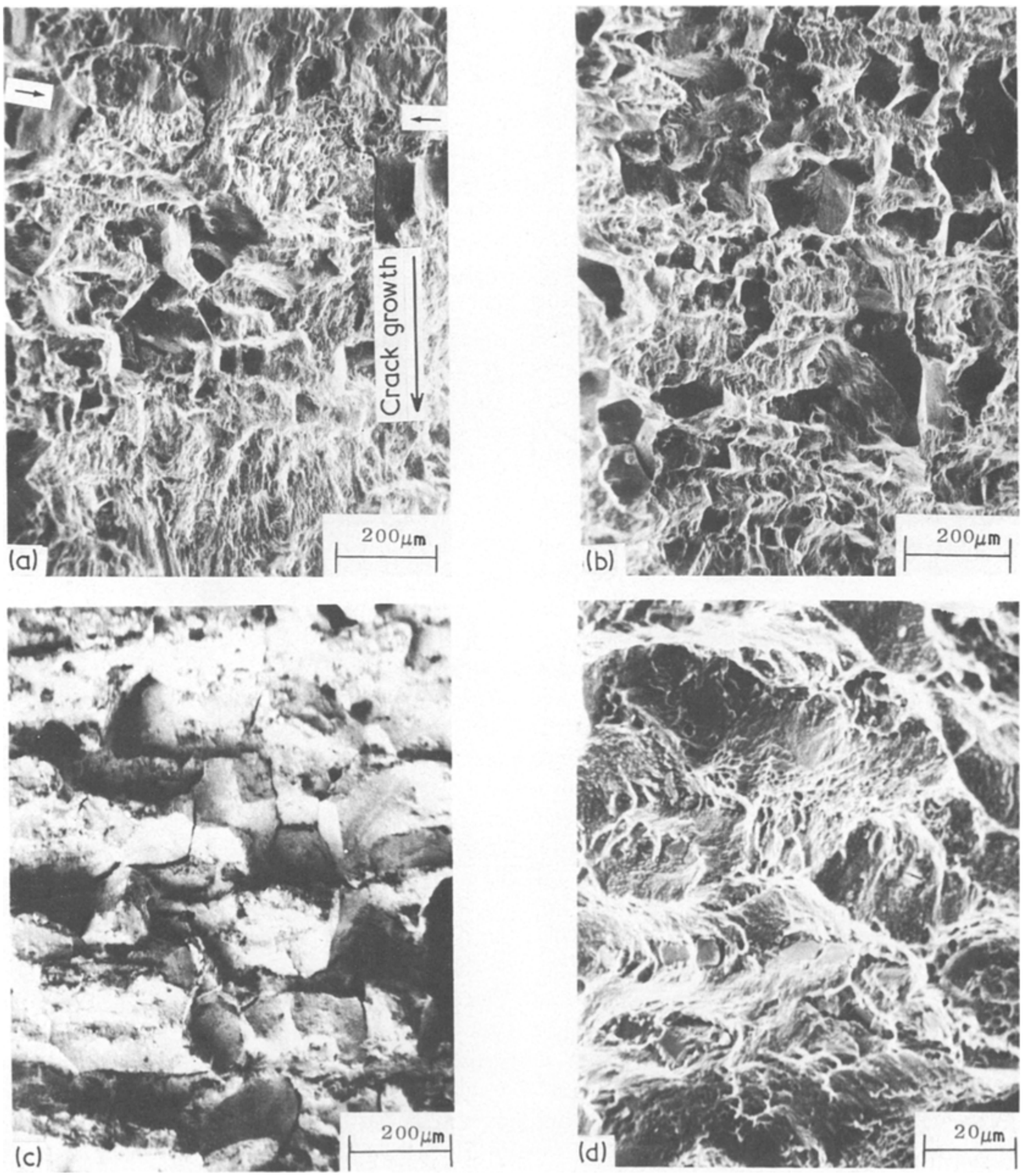


Figure 6 Changes in the crack growth mode with increase in crack length for a specimen tested at 8.89 kN with 1 min hold.

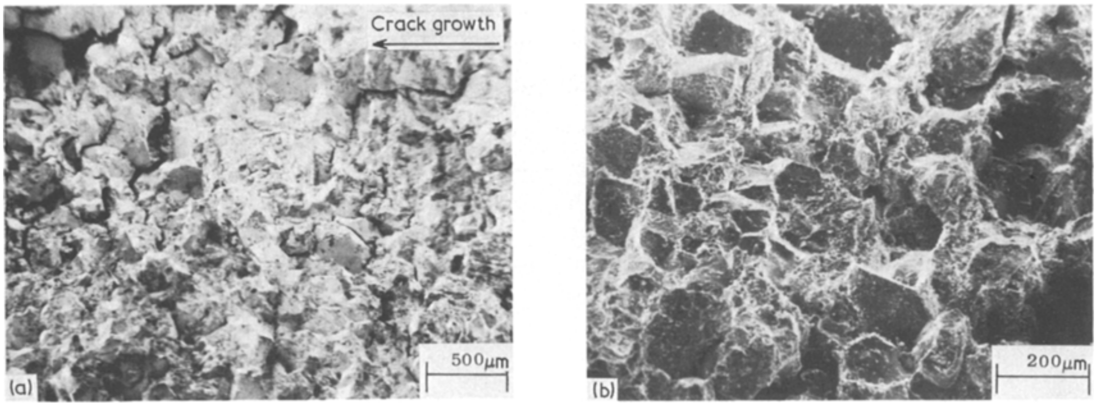


Figure 7 Fracture surfaces of specimens tested at 16 kN under (a) cyclic load with 1 min hold, (b) static load.

with that where a decrease in crack growth rate, Fig. 2, was observed. Also in this region, some cracks intersecting the fracture surface could be noted, Fig. 6a. With increase in ΔK, the creep crack growth process becomes superimposed over the fatigue crack growth, resulting in a mixed mode of crack growth as shown in Fig. 6b. At still higher ΔK crack growth becomes essentially intergranular, as shown in Fig. 6c. In the region of fast fracture, Fig. 6d, the crack growth mode again changes back to transgranular with microvoid coalescence occurring simultaneously with a cleavage mode of crack growth. Thus creep effects dominate both in the low ΔK range (first intergranular growth) where time may be the controlling factor, as well as in the high ΔK range (second intergranular growth) where stress magnitude may be the controlling factor.

For the specimen tested at high load, 16 kN, with 1 min hold-time, the fracture surface (Fig. 7a) shows that the crack growth is mostly intergranular with transgranular fracture occurring intermittently. For the same load and the same stress intensity value, the fracture is completely intergranular for the specimen loaded in creep, Fig. 7b. Again, a large number of cracks intersecting the fracture surfaces can be seen in both figures.

The effects of hold-time on the fatigue crack growth behaviour became further evident when the side surfaces normal to the fracture surface were examined under the microscope. Figs. 8a and b show the normal surfaces at relatively low magnification of the specimens tested under static and cyclic loads at 16 kN respectively. The brighter areas in these figures correspond to the edges of the fracture surfaces. Under a static load, a number of microcracks could be seen well below the fracture surface. From these observations, it can be hypothesized that creep crack growth in this alloy occurs by first nucleating the microcracks ahead of the main crack, and then linking of these cracks with the main

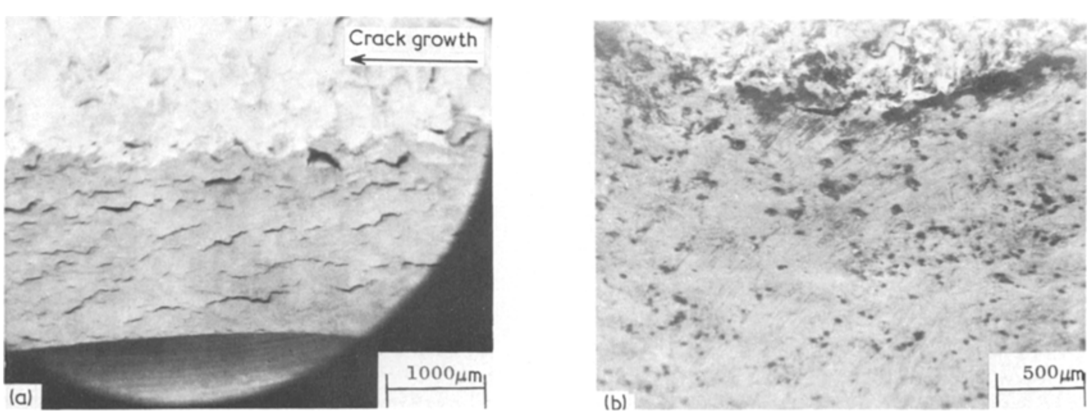


Figure 8 Surfaces normal to the fracture surfaces of specimen tested at 16 kN under (a) static load, (b) cyclic load.

crack. Formation of such microcracks under fatigue, on the other hand, is very limited (Fig. 8b) and occurs very close to the fracture surface, especially in the high ΔK range where creep effects become important.

The surface normal to the fracture surface for a specimen loaded at the same load but with 1 min hold-time is shown in Fig. 9 as a function of crack length. Fig. 9a, in particular, shows the surface close to the pre-crack. Microcrack formation occurs from the beginning but the density of these, however, is less than that under creep but more than that under fatigue. With increase in ΔK, the specimen side surface looks essentially like that under static load, Fig. 9b. In the fast fracture area, Fig. 9c, microcrack formation is essentially non-existent except for one or two, which might have formed before the final fracture occurred. Increased tendency for the formation of microcracks with increased ΔK as well as with increased hold-time are consistent with the crack growth rate data, Fig. 2, that high ΔK as well as hold-times increase the contribution from the creep crack growth process.

With the existence of such a large number of microcracks ahead of the main crack, it is rather surprising to find a reasonable correlation of crack growth rate data at the high ΔK end. In fact, since the presence of these microcracks is neglected in both the crack length measurements as well as for stress intensity calculations, the actual stress field existing ahead of the main crack may be very different from that given by the linear elastic formulation. On the other hand, correlation of the crack growth data on the linear elastic basis may imply that the actual stress field may be proportional to that determined by the formulation. The proportionality factors, however, could change with the change in the type of load (tension versus bending) as well as with change in the specimen geometry. It becomes imperative, therefore, that further evaluation of the fracture mechanics parameters should involve use of different specimen geometries and different types of load.

Finally, on the basis of the available evidence, we could speculate why the side-grooved specimen had lower growth rates, at least in the high ΔK range, than the specimen without side grooves, Fig. 5. The presence of side grooves, for example, could limit the nucleation of these microcracks close to the fracture plane. This nucleation, in turn, may be further restricted by the decreased number of available grain boundary junctions present in the region of side grooves. These factors together reduce the number of available microcracks that could link up with the main crack, thereby reducing the overall crack growth rates.

4. Summary and conclusions

Crack growth behaviour under creep–fatigue conditions was studied using compact tension specimens and the results were analysed using the stress intensity factor as well as J-integral

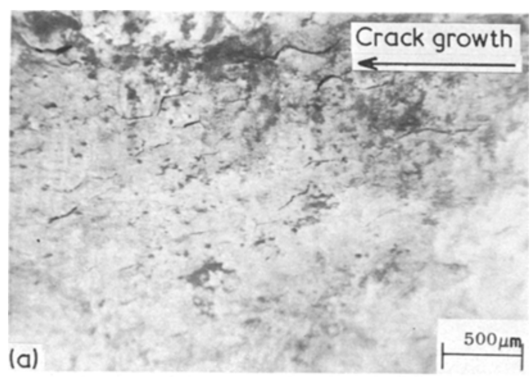
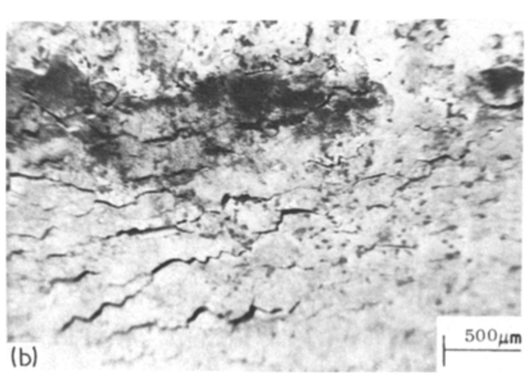
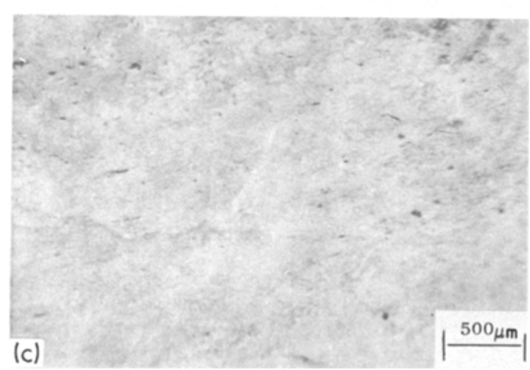


Figure 9 Surface normal to the fracture surface of a specimen tested at 16 kN under cyclic load with 1 min hold.





parameters. Complex creep–fatigue interactions were observed depending on the initial stress intensity value and the duration of the hold-time in each cycle. The threshold stress intensity for creep crack growth appears to be much ligher than that for fatigue crack growth. For the stress intensity values that fall in between the threshold stress intensities for creep and fatigue, crack growth rates under creep–fatigue conditions in general decrease with hold-time, both on a cycle basis as well as on a time basis, to the extent that for large hold-times even complete crack arrest could occur. On the other hand, if the stress intensity exceeds that of the threshold value for creep, hold-time increases the crack growth rates on a cycle basis due to crack growth occurring during hold-time. Because of these complexities, neither the stress intensity factor nor the J-integral is valid for all the ΔK range and hold-times. Limited application of these factors may exist in the high ΔK range where there is continuous crack growth during hold-time tests. It is further shown that crack growth under creep and creep–fatigue conditions is very sensitive to small variations in the microstructure of the alloy. This may be related to the sensitivity of the intergranular crack growth process to the variation in the microstructure along the grain boundaries.

Detailed fractographic analysis showed that the tendency to form intergranular cracks increases with stress intensity as well as with hold-time. Crack growth under these conditions occurs by first nulceation of microcracks ahead of the main crack and linking of these cracks with the main crack. Some of the limitations involved in the characterization of such crack growth by the fracture mechanics parameters are described in detail.

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