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On the growth of intercrystalline wedge-cracks during creep-deformation

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From a study of the crack length distribution and of the increase in crack density versus strain, an expression has been derived for the growth of intercrystalline wedge-cracks in a 20 Cr-35 Ni-steel during creep-deformation. The square root of the crack length is found to increase proportionately with the strain. The growth rate versus strain at 750°C is found to change only slightly over the stress interval studied, 13 to 22 kp/mm², and is highest at the lowest stress. By combining the expression for the crack growth with the rupture criterion, known earlier, an expression is obtained for the rupture strain.

1. Introduction

During creep-deformation at temperatures above 0.5 Tm intercrystalline micro-cracks are formed in most metals and alloys, usually leading to an intercrystalline fracture [1]. Two different crack morphologies have been observed, pores and wedge-cracks [1]. The former type of crack, favoured by high temperatures and low stresses, seems to grow by vacancy-condensation whereas the latter type grows mechanically by the creep-deformation [1].

From results in two earlier investigations [2, 3] on a 20 Cr-35 Ni-steel it is possible to derive an expression for the growth of the wedge-cracks versus the strain. In [2] a relationship $\log Z = -k_1 - k_2\sqrt{c_{\max}}$ was derived between the total number of cracks Z and the largest crack length $c_{\max}$, the cracks being of the wedge-type. The relationship was obtained by a statistical consideration from the crack length distribution obeying the expression $\log x = k_1 + k_2\sqrt{c}$, where x is the fraction of cracks having a length c. One way to derive the expression is described in the Appendix, where it is also pointed out that Z in the general case should be divided by a volume-dependent quantity $k_V \geq 1$, giving the relationship $\log Z = \log k_V - k_1 - k_2\sqrt{c_{\max}}$. Results given in [3] suggest that the total number of cracks Z (or the crack density) increases exponentially with the strain ϵ, thus $\log Z = k_3 + k_4\epsilon$. By combining these expressions for $\log Z$, a relationship $\sqrt{c_{\max}} = k(\epsilon - \epsilon^*)$ is obtained for the growth of the largest crack versus the strain (where $k = -k_4/k_2$ and $\epsilon^* = -(k_1 + k_3 - \log k_V)/k_4$). In the present investigation the stress-dependence of k and ϵ^* were also studied.

2. Experimental method

The experimental method is based upon slow hot tensile-testing and upon a determination in the light microscope of the crack length distribution in the ruptured specimens. From the tensile-test curve it is possible to study the increase in Z with strain, using a method that will be described below. The calculation of k and ϵ^* from the measurements indicated will be described in Section 3.

The slow hot tensile tests were performed at 750°C in an atmosphere of pure argon, resulting in stress-strain curves shown in Fig. 1. The stress was calculated from $\sigma = FL/A_0 L_0$ and the strain from $\epsilon = (L - L_0)/L_0$, the elongation being homogeneous without necking (F = tensile force, L = attained gauge length, L_0 = original gauge length and A_0 = original cross sectional area.) Three different elongation rates were used, 0.02, 0.1 and 0.5 mm/min, corresponding to strain-rates of 6.67×10^{-6}, 3.33×10^{-5} and 1.67×10^{-4} sec^{-1}. The specimen dimensions were ϕ 5 × 50 mm.

It has been shown for the present test material by measurements in the light microscope that the decrease in stress in the later stage of the tensile test (as calculated from $\sigma = FL/A_0 L_0$ with no correction for the crack area) is proportional to the total crack length per unit area and also to the total number of cracks in the specimen, the

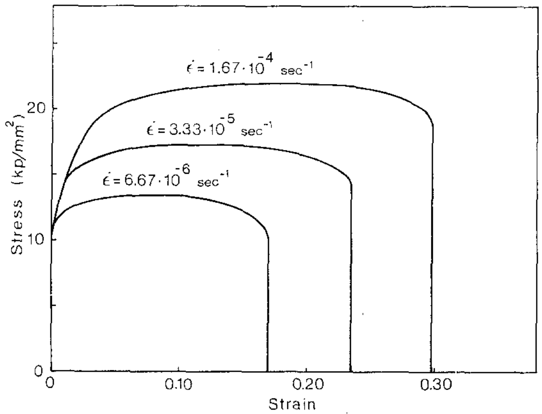

Figure 1 Stress-strain curves from slow hot tensile testing at 750°C.

average crack length being constant [2]. It has also been shown by relaxation tests that the true stress during this stage remains constant and equal to the steady state stress attained earlier [4]. The (apparent) decrease in stress, $\Delta\sigma$, therefore is obviously caused by the reduction in load carrying area from the crack formation, A_c, given by $\Delta\sigma/\sigma = A_c/A$. ($\sigma$ = steady state stress, A = total cross sectional area.)

The conditions necessary for the proportionality between Z and $\Delta\sigma/\sigma$ to hold are that no crack formation of significance for the stress (as calculated from $\sigma = FL/A_0 L_0$) takes place until the steady state stress is attained and that the true stress remains constant during the later stage of the tensile test, where extensive crack formation takes place. The latter condition implies that no recrystallization or change in recovery rate must take place during the crack formation. Since the test-conditions in the present investigation are similar to those in [2] it is reasonable to assume that the porportionality between Z and $\Delta\sigma/\sigma$ holds in the present investigation also. Therefore, in the present investigation the increase in Z with strain is studied from the tensile test curve, using the proportionality between Z and $\Delta\sigma/\sigma$.

The crack length distribution was determined by using the method introduced by Lindborg [3] for measuring the crack lengths. This involved counting the number of grain facets which each individual crack covered. When the cracks terminated along a grain facet the crack length was taken to the nearest integer of facets. Most cracks, however, had stopped at or close to a grain boundary junction.

The detailed chemical analysis of the test material, denoted alloy A, is given in Table I. Also included in this table are alloys referred to later in the text. The test-material was solution annealed at 1120°C for 30 min, quenched in water and subsequently heat-treated at 750°C for 100 h. This last heat-treatment eliminated the possibility of any significant influence of ageing effects occurring during the testing period, which varied from 0.5 to 7 h. The grain-size, as measured by the intercept method, was found to be about 50 µm.

3. Results

By plotting $\Delta\sigma/\sigma$ versus ϵ on a semi-log diagram the relationship $\log \Delta\sigma/\sigma = k_5 + k_6\,\epsilon$ was found to exist, see Fig. 2. The term $\Delta\sigma/\sigma$ was evaluated from Fig. 1, $\Delta\sigma$ being the decrease in stress in the later part of the tensile test and σ the steady-state stress attained before the decrease in stress starts. The relationship, which confirms that Z increases exponentially with the strain (Z being proportional to $\Delta\sigma/\sigma$ [2]), is found to hold at least from $\Delta\sigma/\sigma \sim 0.01$ and up to the onset of the final fracture. At crack densities corresponding to values of $\Delta\sigma/\sigma$ lower than 0.01 the present experimental method is not sensitive enough. However, according to the measurements of Lindborg the exponential relationship should hold down to at least 0.02 of the crack density at rupture, see Fig. 7 of [3], corresponding to a value of $\Delta\sigma/\sigma$ of about 0.003 ($\Delta\sigma/\sigma$ at rupture varying from $\sim$ 0.12 to $\sim$ 0.20, see Fig. 2).

The crack length distributions for the ruptured specimens are shown in Fig. 3 and are found to be the same for the three stresses studied. About 3500 cracks were measured on each specimen, the largest crack lengths observed covering, 20, 16 and 16 grain facets for σ = 13 5, 17.4 and 22.0 kp/mm^2, respectively. The scatter-band at shorter crack lengths is small, but then increases somewhat, obviously because the number of cracks in each crack length class decreases. In order to improve the counting statistics, cracks larger than 9 grain facets have been grouped together into length classes, which span a larger interval than 1 grain facet. Thus, the following four extended length classes were introduced: 10-11, 12-13, 14-15-16 and 17-18-19-20 grain facets. (The number of cracks in these extended classes was divided by the number of unit length classes grouped together.) It is found that the crack length distribution obeys the relationship $\log x = k_1 + k_2 \sqrt{c}$ up to the largest crack

TABLE I Chemical analysis (%)

Alloy	C	Si	Mn	Cr	Ni	Mo	Ti	Al	B	Fe
A	0.047	0.46	0.53	18.7	34.4	—	—	—	—	Bal.
B [6]	0.008	0.13	0.52	19.7	34.8	—	0.01	0.006	—	Bal.
C [7]	0.008	0.13	0.52	19.7	34.8	—	0.01	0.006	0.010	Bal.
D [8]	0.058	0.23	1.70	18.0	33.6	0.87	0.70	—	0.006	Bal.
E charge a	0.042	0.58	1.06	21.0	34.0	—	0.31	0.31	—	Bal.
E charge b	0.036	0.32	1.12	20.3	33.6	—	0.47	0.49	—	Bal.
E charge c	0.040	0.45	1.12	21.1	33.8	—	0.52	0.58	—	Bal.

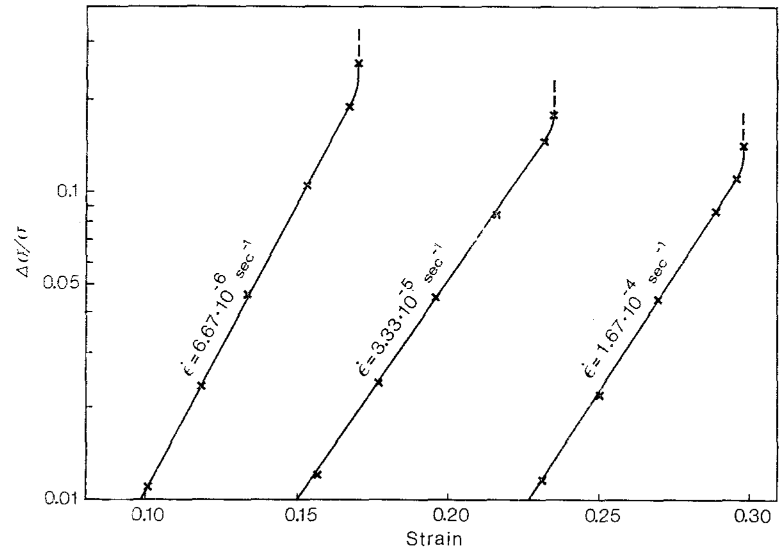

Figure 2 Relationship between $\Delta\sigma/\sigma$ and ϵ as evaluated from Fig. 1, $\Delta\sigma$ being the decrease in stress in the later part of the tensile test and σ the steady-state stress attained before the decrease started.

length observed. By plotting the crack length distribution on different $\log x$ - c^n-diagrams the error in $n = 0.5$ was found to be less than $\pm$ 0.1.

From the proportionality $\Delta\sigma/\sigma = k_Z.Z$ and the relationship $\log \Delta\sigma/\sigma = k_5 + k_6\,\epsilon$ and $\log Z = \log k_V - k_1 - k_2\sqrt{c_{\max}}$ the following expression is obtained for the growth of the largest crack length:

$$\sqrt{c_{\max}} = k(\epsilon - \epsilon^*) \quad (1)$$

with

$$k = -\frac{k_6}{k_2} \quad (2)$$

and with

$$\epsilon^* = \frac{\log k_2 + \log k_V - k_1 - k_5}{k_6} \quad (3)$$

The growth rate coefficient k can be evaluated from the slope coefficients k_2 and k_6 using Equation 2 k_2 and k_6 being evaluated from the slopes of the lines in Figs. 2 and 3. The term ϵ^* can be estimated from a backwards extrapolation in a $\sqrt{c}$-ϵ-diagram from the largest crack length observed at the rupture strain and using the calculated values for the growth rate coefficient k, see Fig. 4. The results of the evaluations are given in Table II. In Fig. 4 a full line is drawn down to a strain corresponding to a $\Delta\sigma/\sigma$-value of 0.003 according to Fig. 2, that is down to the crack length for which the expression $\sqrt{c_{\max}} = k(\epsilon - \epsilon^*)$ at least should hold.

4. Discussion and conclusions

4.1. Crack growth

In the present investigation an expression $\sqrt{c_{\max}} = k(\epsilon - \epsilon^*)$ has been derived for the growth of the largest crack length. It is reasonable to assume that the relationship is valid for the growth of all the individual cracks in the specimen, although with different k- and

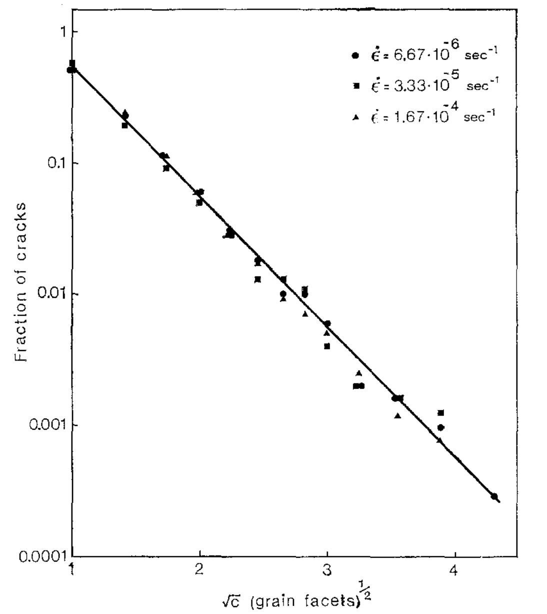

Figure 3 Crack-length distribution in ruptured specimens as plotted on a log x-$\sqrt{c}$-diagram, where x is the fraction of cracks having a length c.

TABLE II Evaluation of k and ϵ^*

Stress (kp/mm²)	k_2 (grain facets)$^{\frac{1}{2}}$	k_6	k (grain facets)$^{\frac{1}{2}}$	ϵ^*
13.5	− 1.00	18.4	18.4	− 0.056
17.4	− 1.00	14.3	14.3	− 0.044
22.0	− 1.00	15.1	15.1	0.035

ϵ^*-values. With the present experimental technique, it has only been possible to evaluate the growth of the largest crack length from a length of about 5 grain facets, see Fig. 4. However, since the crack length distribution obeys the same expression for all length classes it seems likely that the crack growth will obey a relationship $\sqrt{c} = k(\epsilon - \epsilon^*)$ also at shorter crack lengths.

It is generally accepted that the formation of wedge-cracks is caused by grain-boundary strain rather than by total strain [1]. Assuming that the proportionality between grain boundary strain and total strain [1] also holds during the crack formation, the expression for the crack growth can be written as $\sqrt{c} = k_{gbs}(\epsilon_{gbs} - \epsilon_{gbs}^*)$, where index gbs indicates grain-boundary strain. The crack growth rate versus grain-boundary strain, $dc/d\epsilon_{gbs}$, then can be written $dc/d\epsilon_{gbs} = 2k_{gbs}^2(\epsilon_{gbs} - \epsilon_{gbs}^*)$. The increase in crack growth rate with ϵ_{gbs} may suggest that the

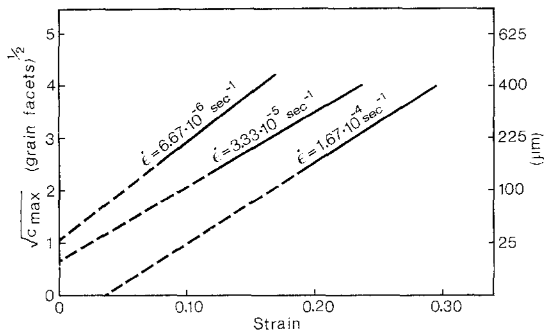

Figure 4 Calculated growth of largest crack length versus strain. The μm-scale is calculated using an average grain facet length of 25 μm.

resistance of the grain boundaries against the crack propagation is broken down by the grain-boundary sliding, e.g. by the formation of pores or other defects. The above expression can also be written in the form $dc/d\epsilon_{gbs} = 2k_{gbs}\sqrt{c}$, which alternatively might suggest that the crack growth rate is controlled by the stress concentration at the crack tip, this being proportional to $\sigma\sqrt{c}$. However, this latter interpretation is not supported by the present observations on the stress-dependency of the crack growth rate which will be discussed below.

The term ϵ^* for the largest crack length at the two lowest stresses has (small) negative values. The reason for this may be a more or less stepwise growth of the cracks, suggested by the observation that most of the cracks are arrested at or close to grain-boundary junctions [3]. If the cracks grow fast over the first grain facet but slowly over the first grain-boundary junction, this will result in negative ϵ^*-values for small nucleation-strains. It is also possible that the cracks first nucleated grow with a higher k-value than the final one along the first grain boundaries, the first cracks probably being nucleated in grain boundaries particularly prone to crack formation. The increase in ϵ^* with increasing stress (or increasing strain rate), shown by Fig. 4, suggests that the nucleation strain increases with the stress (or the strain rate). This conforms with observations by Shapiro and Dieter, who found that the nucleation strain for wedge-cracks increases with the stress during hot-working conditions for high purity nickel [5].

The term k has been found to change only slightly over the stress interval studied, being highest at the lowest stress. The stress dependency of k, k_{gbs} and ϵ_{gbs}^* will be analysed from the stress dependency of the ratio grain-boundary strain to total strain, $\epsilon_{gbs}/\epsilon_{tot}$, which first will be described. For steels which are very similar to the present test material (alloys B, C and D in Table I), $\epsilon_{gbs}/\epsilon_{tot}$ at 700°C first decreases sharply with increasing stress, approximately proportional to $\sigma^{-(2\text{ to }3)}$, and then above a certain stress level remains constant or possibly decreases slowly [6-8]. The critical stress where $\epsilon_{gbs}/\epsilon_{tot}$ levels out is found for all these steels to be within the range of 11 to 13 kp/mm^2.

Since many observations indicate that $\epsilon_{gbs}/\epsilon_{tot}$ for the same stress remains constant over a temperature range in the order of 100°C [1] it is most likely that in the present investigation (made at 750°C to obtain lower stresses in the tensile test) at least the two highest stresses are above the critical value where $\epsilon_{gbs}/\epsilon_{tot}$ levels out. The almost constant value of k at the two highest stresses therefore suggests that the crack growth rate versus grain-boundary strain is independent or little dependent upon the applied stress. If this result is valid at lower stresses also, k will increase with decreasing stresses approximately in the same proportion as $\epsilon_{gbs}/\epsilon_{tot}$, whereas k_{gbs} will remain approximately constant. This might be the reason for the somewhat higher k-value at the lowest stress studied. Another conclusion is that the term ϵ_{gbs}^* within the stress interval studied increases with increasing stress (or strain rate).

4.2. Creep ductility

An earlier investigation [2] and an investigation now in progress [4] suggests that the final fracture for the present alloy in creep-testing obeys a Griffith-type relationship (whereas in the present tensile-test rupture started before this condition was fulfilled on approaching the spring-constant of the tensile test machine). Combining the rupture criterion, which can be written $\sigma\sqrt{c_{max}} = k_r$, with the relationship found for the crack growth versus the total strain an expression is obtained for the rupture strain, ϵ_r, valid within the range of wedge-crack formation.

$$\epsilon_r = \frac{k_r}{k} \cdot \frac{1}{\sigma} + \epsilon^* . \tag{4}$$

Using the relationship for the crack growth versus grain-boundary strain, the expression for ϵ_r alternatively can be written:

$$\epsilon_r = \frac{1}{\epsilon_{gbs}/\epsilon_{tot}} \left[\frac{k_r}{k_{gbs}} \cdot \frac{1}{\sigma} + \epsilon_{gbs}^* \right] \tag{5}$$

Taking k_r and k_{gbs} generally to be constant, ϵ_{gbs}^* to increase with the stress and $\epsilon_{gbs}/\epsilon_{tot}$ to change in the way described above, the rupture strain will increase according to Equation 5 up to the stress where $\epsilon_{gbs}/\epsilon_{tot}$ levels out. This behaviour is shown by a Ti-stabilized 20 Cr-35 Ni-steel (alloy E in Table I), see Fig. 5. For this material rupture takes place by wedge-crack formation over the stress interval indicated in the diagram. At still higher stresses the progress of the rupture strain will depend upon the increase in ϵ_{gbs}^* compared to the decrease in $(k_r/k_{gbs}) \cdot (1/\sigma)$. However, the rupture strain for most commercial alloys, no doubt, will be influenced also by time-dependent phase transformations, overlapping the true stress-dependency. In Equation 5 the influence of time-dependent phase transformations will be reflected by changes in the parameters $\epsilon_{gbs}/\epsilon_{tot}$, k_r, k_{gbs} and ϵ_{gbs}^*.

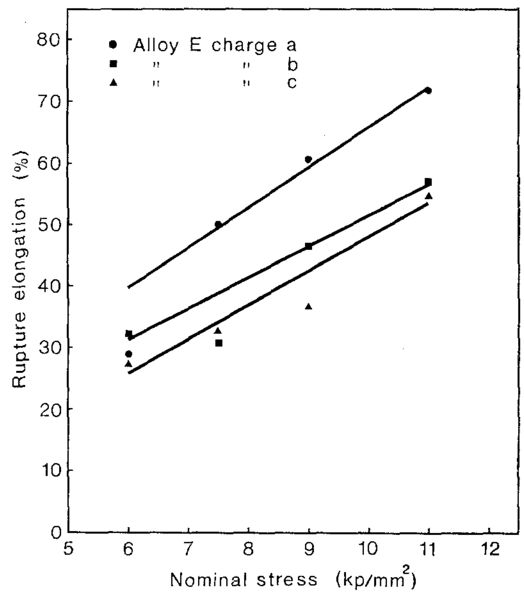

Figure 5 Rupture strain versus nominal stress at 700°C for three different charges of a Ti-stabilized 20 Cr-35 Ni-steel.

It is of interest to see that, in principle, the same expression for ϵ_r as in Equation 5 can be derived from some results obtained earlier by J. A. Williams in an investigation on a coarse-grained Al-Mg-alloy [9]. Using the Cottrell model, Williams found that the final fracture obeyed the criterion $\sigma h = 2\gamma$, h being the height of the crack at the initiation point and γ the effective surface energy. The height of the cracks at the initiation point were found to grow proportionately with the grain-boundary strain, $h = k\epsilon_{gbs}$. Combining these expressions will give an equation of the same form as Equation 5 (taking ϵ_{gbs}^* to be zero). We find that Williams expressions $\sigma h = 2\gamma$ and $h = k\epsilon_{gbs}$ in principle are identical to the authors corresponding expressions $\sigma\sqrt{c} = k_r$ and $\sqrt{c} = k\epsilon_{gbs}$ (taking ϵ_{gbs}^* to be zero) if a relationship $\sqrt{c} = kh$ is supposed to exist.

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Appendix

The number of cracks N in a crack length class c can be written $Z \cdot x = N$, where Z is the total number of cracks and x the fraction of cracks having the length c. If the expression for the crack-length distribution, $[x = f(c)]$ is valid for the largest crack length c_{max}, the largest crack length will be related to Z by $Z \,.\, f(c_{max}) = k_v$, where k_v is the number of cracks in the largest crack length class. The present investigation suggests that the expression obtained for the crack-length distribution, $x = \exp(k_1 + k_2\sqrt{c})$, is valid for the largest crack length. Therefore, for the present alloy, the relationship between c_{max} and Z can be written $\log Z = \log k_v - k_1 - k_2\sqrt{c_{max}}$. Results in an earlier investigation on the present alloy indicate that the crack-length distribution is independent of crack density, see Fig. 8 of [3]. Consequently, k_1 and k_2 remain constant during the crack formation. A plot of $\sqrt{c_{max}}$ versus $\log(\Delta A/A)$, where $\Delta A/A = k \cdot Z$, forms a straight line, see Fig. 3 of [2]. Thus, also k_v will remain constant during the crack formation. It seems likely that k_v will depend upon the specimen volume, probably being one up to a certain material volume, then increasing as the volume increases above this critical value.

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