INTERGRANULAR CREEP DAMAGE AND FRACTURE CRITERIA

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ABSTRACT

An analysis of intergranular cavitational damage in high temperature creep of copper, some copper alloys and some heat resistant steels has been performed using the methods of quantitative microscopy. Extensive measurements of the topologic as well as metric parameters of both the grain boundary damage and deformation processes have shown that in the course of creep accumulation of damage takes place, finally reaching a critical value characterizing the so called fracture state. An analysis of the fracture state has shown, that the essential fracture criterion can be expressed as $\tilde{A}_f = \tilde{\xi}$, where $\tilde{A}_f$ is the area fraction of cavities on cavitated grain boundaries and $\tilde{\xi}$ is a constant. The fracture state can also be defined by a critical value of deformation process controlling fracture development.

KEYWORDS

High temperature creep; creep strain; grain boundary sliding; cavitation; intergranular creep fracture; quantitative microscopy; damage and deformation parameters; fracture criteria; copper and copper alloys.

INTRODUCTION

The high temperature creep failure in metallic materials is a result of damage accumulation at the grain boundaries in a broad range of external conditions. The damage consisting in the nucleation, growth and mutual coalescence of intergranular voids (cavities) in many materials, typically represented by copper and copper alloys, is easily observable at the surface as well as in the interior of creep specimen (Sklenička, Saxl and Čadek, 1977). The methods of quantitative metallography makes it possible to estimate fairly accurately and exhaustively by the global as well as local characteristics of cavitation (Saxl, Sklenička and Čadek, 1979). For the appearence and growth of damage the deformation processes and stress assisted diffusion are responsible (Ashby and Raj, 1975; Greenwood, 1978; Edward and Ashby, 1979). Thus, besides local damage parameters it is reasonable to examine also the parameters of local strains making it possible to look for a possible correlation between strain components and damage (Sklenička and others, 1980).

In the present paper the main results of the extensive study of creep cavitation in copper, some copper alloys and some heat resistant steels are summarized.

EXPERIMENTAL PROCEDURES

Constant load creep tests were performed in purified dried hydrogen or argon (copper and copper alloys) or in air (steels) and the cavitation measured on sections parallel to the tensile stress axis. The experimental materials and testing conditions are given in Table 1.

TABLE 1 Materials and test conditions

Material	Grain* size $(N_L)_\beta$ [1/mm]	Temperature T [K]	Stress σ [MPa]
Cu (99.96%)	1.8	673 → 973	6 → 100
Cu2.2%Fe	1.5	773, 873	7.5 → 30
Cu2.5%Al	5.9	873	4 → 20
Cu10%Zn	3.1	773 → 973	20
Cu20%Zn	5.6	673 → 973	8 → 80
Cu30%Zn	2.1	773 → 973	3 → 40
Steel 0.5Cr0.5Mo0.3V	44.6	873	70 → 200
Steel 20Cr32NiTiAl	8.2	1023 → 1173	10 → 80

*$(N_L)_\beta$ - the number of intercepts of grains per unit length of test line.

The methods of quantitative measurements employed are based on the lineal analysis along grain boundary traces in the plane of section and are described in detail elsewhere (Saxl, Sklenička and Čadek, 1979).

RESULTS

The damage accumulation may be at the best seen from the time dependence of the areal fraction of cavities in grain boundaries $(A_{gb})_\alpha$, Fig. 1. The first traces of damage are observable already in the first tenth of time to fracture t_f, i.e. already during the primary stage of creep. The steep initial increase of $(A_{gb})_\alpha$ is followed by the region of moderate "parabolic" growth that occupies the considerable part of t_f (the critical extent of cavitation). Not sooner than in the close proximity of the failure the damage rate again increases up to the fracture. It may be assumed that just the above mentioned moderate rate of approach to the above "critical" extent of cavitation is responsible for the usually observed great scatter of fracture times (especially at very slow creep rates, Fig. 1).

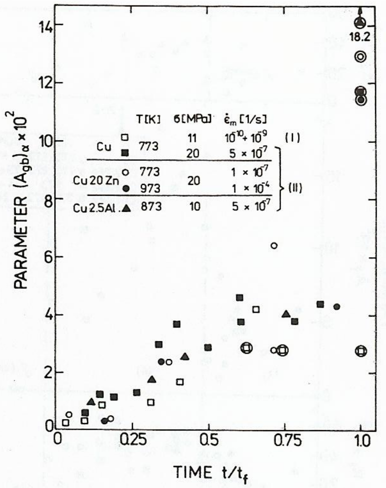

Fig. 1. Time dependence of parameter $(A_{gb})_\alpha$ for copper and copper alloys in the regions of diffusional (Region I) and GBS (Region II) fracture.

The inhomogeneity of cavitation may be described by the parameter $\varkappa$, giving the fraction of boundary facets with observable cavitation. The homogeneity of cavitation increases during creep; typical maximum values of $\varkappa$ range from 0.2 to 0.5 for copper, copper alloys and an austenitic steel but the value as low as 0.1 may be found for a bainitic CrMoV steel. By means of $\varkappa$ another quantity perhaps characterizing the local cavitation better than $(A_{gb})_\alpha$ can be introduced, namely the parameter $(\tilde{A}_{gb})_\alpha = \varkappa^{-1}(A_{gb})_\alpha$ - the areal fraction of cavities in cavitated facets only.

The values of damage parameters measured in the homogeneously stressed parts of fractured specimens (the omission of indices α, gb leeds to simpler notation A_f, $\tilde{A}_f$, $\varkappa_f$) of copper and its alloys are summarized in Fig. 2. The range of creep rates can be tentatively divided into three regions namely Region I: $\varkappa_f \sim 0.2$, $A_f \sim 0.05$, Region II: $\varkappa_f \sim 0.4$, $A_f \sim 0.1$ to 0.15 and finally Region III: $\varkappa_f, A_f \longrightarrow 0$ and transgranular features on the fracture surface. This division can be closely tied to the mechanism of damage development.

It can be shown, Fig. 3, that the mean diameter of cavity sections in the grain boundaries is approximately equal to or slightly exceeds the total grain boundary sliding (GBS) vector $\bar{p} \doteq 2\bar{u}$ in Region II, but considerably exceeds $\bar{p}$ in Region I. The vector $\bar{u}$ means the mean boundary displacement in the tensile stress axis direction measured on the boundary facets with $u \neq 0$, η is the fraction of sliding boundaries; thus $\eta \bar{u} (N_L)_\beta = \varepsilon_{gb}$ is the contribution of GBS to the total creep strain (Sklenička and others, 1975). Thus GBS is the damage growth controlling process in Region II, whereas in Region I the estimate of the cavity growth rate is in good agreement with the theory of diffusion controlled cavity growth. On the other

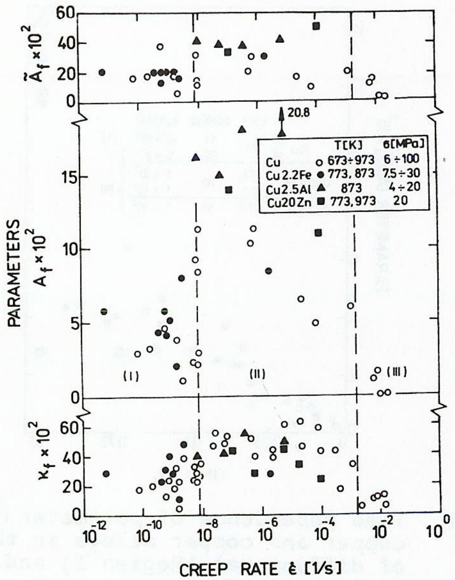

Fig. 2. Fracture values of parameter $\varkappa_f$, A_f and $\tilde{A}_f$ for copper and copper alloys crept under various conditions. Region I: diffusional cavitational fracture; Region II: GBS cavitational fracture; Region III: transition from intergranular to transgranular fracture.

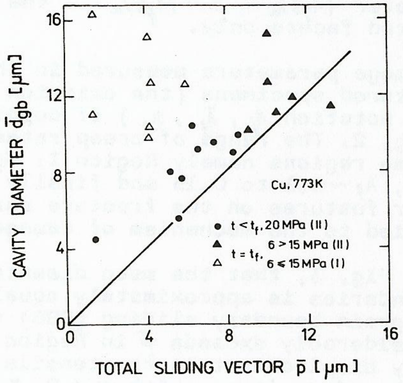

Fig. 3. Relation between the metric parameters of cavitation and of GBS. Copper crept in the regions of diffusional (I) and GBS (II) fracture.

hand, the nucleation of cavities seems to be in both Region I and II of intergranular fracture due to GBS. This may be seen from the proportionality between density of cavities and extent of GBS as described quantitatively by parameter η (Fig. 4).

A recent analysis of the possibilities to nucleate cavities on the grain boundary inclusions (Raj and Ashby, 1975) stressed the principal role of GBS in the creation of necessary stress concentrations; however, a relatively high sliding rate is required. A detailed measurement of sliding initiation and development in polycrystalline copper has shown that these severe requirements are met only at the early activated facets (as to GBS) and only immediately after the activation (Saxl and others, 1976). At homologous temperatures above 0.5 slidings on about 60% of facets are initiated within first tenth of time to fracture and it can be reasonably supposed that just from this fraction of facets the cavitated one are chosen. Consequently, it may be supposed that all cavity nuclei are formed approximately at the beginning of the creep exposure (Raj, 1978b). This assumption simplifies the prediction of the time to fracture considerably.

The results of an analysis of critical intragranular cavitational damage of an austenitic steel of the 20Cr32Ni type (Incoloy 800) and a 0.5Cr0.5Mo0.3V bainitic steel are summarized in Fig. 5. One can see that the parameter $\tilde{A}_f$ does not depend significantly on experimental material and creep conditions.

A great attention has been paid to the cavity distribution in the creep specimens which is in many respects critical. The significant deviation from the uniformity of strain and consequently the uniformity of cavitation along the gauge length has been observed only in the early stages of creep exposure and then again in the proximity point of fracture. More important are variations of cavitation on the specimen cross section, because the surface layers are considerably preferred. Nevertheless, even in this case the inhomogeneity of cavitation decreases with the time of creep exposure, but in any case care must be taken to obtain truly representative average of any quantity. Probably, the most important is the cavity dis-

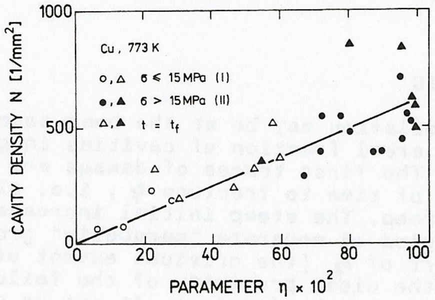

Fig. 4. Relation between density of cavities on grain boundaries N and the relative frequency of sliding boundaries η. Copper crept in the regions of diffusional (I) and GBS (II) fracture.

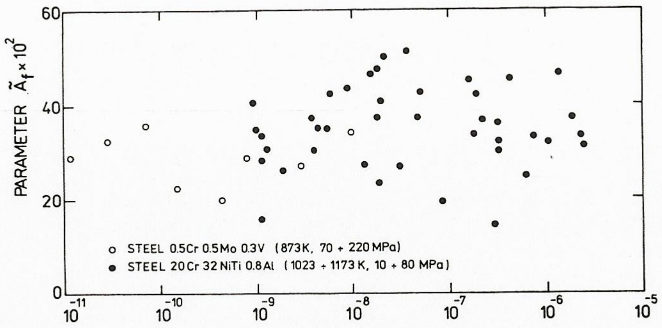

Fig. 5. Fracture values of the parameter $\tilde{A}_f$ for some heat resistant steels.

tribution on cavitated facets, especially with respect to the description of the process of cavity coalescence. The evaluation of coordination number $\bar{c}$ (the mean number of cavity contacts with its neighbours) and of the contiguity c_T of cavity sections (Gurland, 1968) has been quite helpful in this respect. It has been shown that cavities tend to form chains (gradually increasing in length), which finally manifoldwise interconnect border lines of facets thus forming "channels" enabling the main crack to propagate (Saxl and others, 1979).

The noticable feature of boundary damage is the occurence of intercrystalline surface cracks observed in all materials investigated. These microcracks develop from surface cavities in the very early stages of exposure (below 0.2 t_f) and then they widen but do not deepen, thus remaining confined to the surface layer. This process may be explained by the GBS in the surface layer (Sklenička and others, 1975), the rate of crack widening $\dot{s}$ (s is the surface crack width measured in the direction of tensile stress axis) approximately equals the sliding rate $\dot{u}$ (parabolic time dependence) and $\ddot{s}<0$ up to about 0.8 to 0.9 t_f . Only after grain boundaries in the interior are sufficiently (critically) weakened by the cavitation some of surface cracks spread throughout the whole cross section of the specimen; their widening rate again increases according to the $\dot{s} \sim s$ law.

All the information obtained leeds to the conclusion that till approximately 0.8 t_f the accumulation of damage proceeds uniformly according to the time-laws, that can be derived either from theory or from experimental observation performed in any part of the specimen crept under one-axis-loading and properly chosen external conditions. In this region it holds for time dependence of an arbitrary parameter of damage ω that $\ddot{\omega}<0$ (for ω one can choose either A_f, $\varkappa_{\bar{s}}$ or a parameter characterizing the deformation process that controlls the fracture).

DISCUSSION

The results obtained suggest that a reliable prediction of the fracture characteristics t_f and ε_f can be based on the limit state of uniform (parabolic, $\ddot{\omega}<0$) accumulation of damage. For this state of damage, we propose the term "fracture state" (Sklenička, Saxl and Čadek, 1977). This state can be formally defined as one in which there is a maximum probability of interaction of damage on distances comparable with the cross section dimensions of the creep specimen. In fact, before this state is reached the damage takes place in isolated micro-volumes only (the growth of isolated cavities, their coalescence and formation of cavity chains on individual grain boundary facets, the growth of individual surface microcracks), while the last period of creep life is characterized by interlinkage of the damage in macroscopic scale (the propagation of several "critical" cracks from which one - the magistral crack - causes the final failure). On the basis of performed measurements the most representative and at the same time experimentally the most suitable parameter seems to be the relative cavitated area of the facets on which cavitation have taken place $\tilde{A}_f$, that in the essential involves the amounts of damage at which the critical value of grain boundary cohesion enables the blunted surface cracks to propagate. Comparison of Fig. 2 to Fig. 5 shows the possibility to express the essential fracture criterion in the form

$$\tilde{A}_f = \tilde{\xi} \, ,$$

where $\tilde{\xi}$ is a constant roughly independent of creep conditions and the crept material and ranging from 0.1 to 0.4. This is a value considerably lower than 1 i.e. the value of $\tilde{A}_f$ on grain boundaries forming the fracture surface; $\tilde{A}_f$ close to 1 is frequently assumed in the predictions of time to fracture (Ashby and Raj, 1975). Probably such a high value of $\tilde{A}_f$ would be appropriate for bicrystals only (Raj, 1978a). Using the above criterion expressions for the calculation of time to fracture and strain to fracture have been developed for both types of intergranular creep fracture i.e. diffusional and

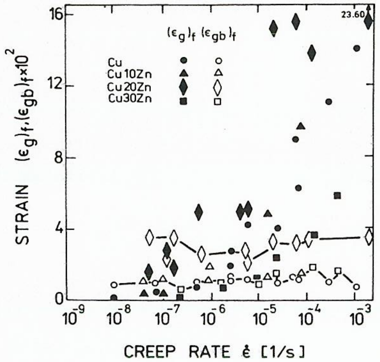

Fig. 6. Fracture values of the total creep strain components $(\varepsilon_g)_f$ and $(\varepsilon_{gb})_f$ in the regions of GBS (II) fracture. Copper and some copper alloys.

GBS ones (Sklenička, Saxl and Čadek, 1977; Sklenička and others,1980). Both the calculated time and strain to fracture are in very good agreement with those determined experimentally.

The fracture state can as well be defined by a critical value of strain resulting from the deformation process that controls fracture development. Thus, for instance, for copper and its alloys in Region II, where the fracture is controlled by GBS, the strain component ε_{gb} is constant, Fig. 6. On the contrary, the strain component ε_g , caused intercrystalline deformation as well as total creep strain ε, ranges in the ratio 1 : 20. The possibility to characterize the state of damage by the total creep strain is realistic only in the case when either the process of deformation controlling fracture contributes to the total creep strain quite dominantly, or when the contributing deformation processes are linearly interrelated (this is the case of the so called zone grain boundary sliding with $\varepsilon_{gb}/\varepsilon_g \simeq$ $\simeq$ const.). It is only under these conditions that the parameter of deformation process can be uniquelly correlated with the state of damage.

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