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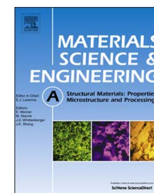

Formation of a dislocation back stress during creep of copper at low temperatures

Rolf Sandström

Materials Science and Engineering, KTH Royal Institute of Technology, SE-100 44 Stockholm, Sweden

A R T I C L E I N F O

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A B S T R A C T

Copper gives creep strain versus time curves at 75 °C that look very similar to those recorded at much higher temperatures. Thus, for example, an extended secondary stage where the strain rate is constant is observed. Considering the high creep exponent that can be up to 75, one would expect a creep curve with rapidly increasing strain rate but that is not found. The difference to creep of pure metals at high temperatures is so large that we can talk about an entirely new material class with respect to creep. To explain the observations a recently developed dislocation model (Sandstrom, 2017) for cell structures is used. A new creep model is presented where a back stress based on the dislocations in the cell walls is introduced. Unbalanced sets of dislocations without matching dislocations of opposite signs are formed in the cell walls. Since the unbalanced content is not exposed to static recovery, it forms a stable back stress. It is shown that the computed back stress can fully explain the observations and reproduce both creep curves and results for slow strain rate tensile tests.

1. Background

Pure copper is extensively used in various technical applications at temperatures up to 200 °C, but rarely above [1]. A large number of investigations of creep testing of copper have been published [2–6]. Most of this creep testing has been performed above 200 °C and often much above this temperature. At high temperatures copper is described as a class II material with respect to creep, where dislocation climb is the rate controlling mechanism [7]. At very high temperatures diffusion creep and grain rotation can be involved. The exact controlling mechanisms at very high temperatures are still controversial [8].

Also in the temperature range 75–200 °C, extensive creep deformation is observed. In the present paper this will be referred to as low temperatures. The lower temperature limit 75 °C is given because few tests have been carried out below this temperature. Below 200 °C climb cannot be the controlling mechanism, because it would give creep rates that are orders of magnitude below the observed ones. This is directly evident from the established expression for the dislocation climb mobility [9]. This mobility is proportional to the self-diffusion coefficient, and since it is negligibly small at low temperatures, so is the climb rate. Instead, it is assumed that glide is governing the creep deformation. A combined expression for the climb and glide mobility of dislocations has been formulated [9]. It has been demonstrated that this mobility can describe the creep rate both in the climb and glide controlled regimes, see for example [9].

Recorded creep versus time curves (commonly referred just to as creep curves) look very similar in the glide controlled and climb controlled regimes. The creep curves have typically well-developed primary, secondary and tertiary stages. At lower temperatures the creep exponent is very high, in the range 20–75 and this should give a strongly accelerating rate during creep testing, but that is not observed. This will be analysed in detail in the next section.

In fact, the creep behaviour below 200 °C is very different from that at higher temperatures. Logarithmic creep is frequently observed for steels at room temperature [10,11]. For this type of creep, the strain is approximately proportional to the logarithm of time. This implies that the creep rate eventually becomes so low that the creep deformation stops and rupture does not take place unless the initial stress is close to the tensile strength. Since the creep deformation in copper does not slow down, it is clearly not logarithmic creep that is observed.

The creep properties of pure copper can be much improved by addition of 50 ppm of phosphorus. Oxygen free copper with P additions is referred to as Cu-OFP. Cu-OFP has both much better creep strength and creep ductility than oxygen free copper without phosphorus (Cu-OF) at temperatures below 200 °C [12,13]. Since most of the creep testing of pure copper below 200 °C has been performed for Cu-OFP, the analysis in the present paper will primarily refer to that type of material. Cu-OFP is planned to be used in canisters for final disposal of spent nuclear fuel in Finland and Sweden [14]. The copper may be exposed to creep for thousands of years at temperatures below 100 °C. It is of course critical that the correct approaches are used in the design of the canisters and that the procedures that are employed are fully understood.

E-mail address: rsand@kth.se.

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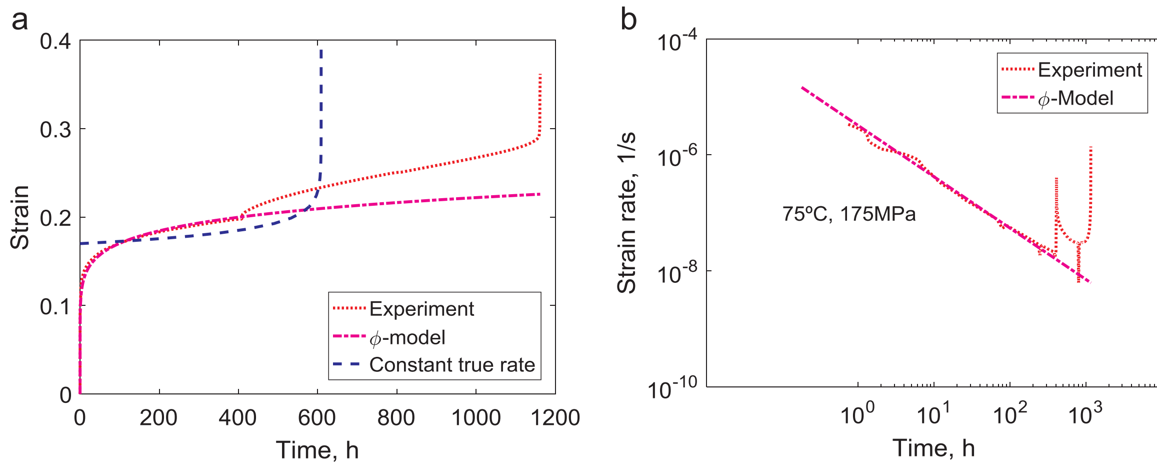

Fig. 1. Creep strain curve for Cu-OFP at 75 °C and 175 MPa. The empirical ϕ-model is fitted to the data; a) Creep strain versus time; b) Creep strain rate versus time. Experiments, see [15].

The absence of an accelerating rate during creep testing implies that a new type of creep behaviour is involved. It is the purpose of the paper to describe and analyse this type of creep and present a model that explains the effect.

2. Creep of copper at low temperatures

Copper shows creep deformation at as low temperature as 75 °C without any sign of logarithmic creep. This has been demonstrated in a number of investigations, see for example [9,15]. An example of a creep curve (creep strain versus time curve) is given in Fig. 1a. Well-developed primary, secondary and tertiary stages are observed. Rupture took place after about 1200 h. The cusp in the experimental creep curve is due to the necessity of reloading the specimen when the strain capacity had been exceeded due to the fairly large total creep strain.

Another characteristic feature of the creep curve is illustrated in Fig. 1b. In the double logarithmic diagram the experiment gives a straight line at least before the cusp in the creep curve. This strain rate versus time behaviour is referred to as the ϕ-model [15,16] but in some form it was already proposed by Bailey [17]. The creep data has been fitted to the ϕ-model to demonstrate how closely the data follow this empirical model. The ϕ-model has successfully been applied to a number of other materials to describe primary creep. The best documented materials in this respect are probably the martensitic creep resistant 9Cr steels P91 and P92, see for example [18].

Thus, it is illustrated in Fig. 1a and b that the creep data of copper at low temperatures show many features that are well established at higher temperatures. However, there are important differences and that will be shown now. Let us try to describe the creep rate with a Norton equation, see for example [19].

$$\frac{d\varepsilon}{dt} = A_0(\sigma_{\text{appl0}}e^{\varepsilon})^n \tag{1}$$

In this equation we have the conventional stress dependence in the form of a power law with a creep exponent n. ε is the creep strain, σ_{appl0} the nominal applied stress and A_0 is a constants. The factor $\exp(\varepsilon)$ takes into account the reduction of the specimen cross section during testing. For the data in Fig. 1, $\sigma_{\text{appl0}} = 175$ MPa and $n = 72$. The remaining constant A_0 is fitted to one point on the creep curve (the time t = 600 h). The integrated creep curve using Eq. (1) is illustrated in Fig. 1. To take into account the primary creep to some extent, the creep curve is starting at $\varepsilon_0 = 0.17$. Obviously, the conventional expression in Eq. (1) for the secondary creep rate can in no way represent the creep curve in Fig. 1. These conclusions are not affected by choosing other constants in Eq. (1).

With the high creep exponent the factor $\exp(n\ \varepsilon)$ has a dramatic effect on the strain rate. It gives a strong and continuous bending upwards, which is fully inconsistent with the experimental data. The effect is not small. For example, at a strain of 0.15, $\exp(n\ \varepsilon) = 49{,}000$.

Obviously, the conventional power law cannot represent the stress dependence of the creep rate. The expected increase in creep rate with increasing strain does not appear in Fig. 1 or in other experimental curves. A few more examples will be given later in the paper. It seems that there is a back stress that prevents the full applied stress to have effect. In the next section a model for such a back stress will be formulated. If this back stress is taken into account it is possible to describe the creep curves. It must be concluded that we have a creep behaviour that is different from that of either class I or class II. If we call logarithmic creep for class III, we can call the described behaviour class IV.

3. Model

The basis of our understanding of stationary creep is the recovery creep theory. According to this theory a balance between work hardening and recovery is reached in the stationary stage, i.e. the number of dislocations that is generated by work hardening is the same as the number that is annihilated by recovery. During primary creep the dislocation density in Cu-OFP increases [20]. The recovery rate is more strongly dependent on the dislocation density than the work hardening rate. When the dislocation density has reached a certain value a balance between the two processes is obtained. For many alloys such a balance exists for a significant fraction of the creep life resulting in a constant strain rate. This is referred to as the stationary or secondary stage. A basic model for primary and secondary creep of copper has previously been developed [15]. It has been based on the following equation for the development of the dislocation density ρ.

$$\frac{d\rho}{d\varepsilon} = \frac{m\rho^{1/2}}{bc_L} - \omega\rho - 2\tau_L M\rho^2/\dot{\varepsilon} \tag{2}$$

ε is the strain, m the Taylor factor, b Burger's vector, τ_L the dislocation line tension, and M the dislocation mobility. c_L is a work hardening constant and ω is a constant that controls the amount of dynamic recovery. The three terms on the right hand side of Eq. (2) represent work hardening, dynamic recovery and static recovery. Dynamic recovery is a strain controlled process, where dislocations are forced together and in this way reduce the total dislocation content. Static recovery is a time dependent slow mechanism where dislocations of opposite sign attract each other and eventually annihilate. In Eq. (2), it is assumed that the dislocations are homogeneously distributed in the material. Later dislocations in a cell structure will be considered. For some parts of Eq. (2), the derivation is difficult to find in the literature. For this reason, the derivation of the whole equation has been given in a recent paper [21].

Following the basis of the creep recovery theory, it means that the

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