1

Macroscopic Characteristics of Strain of Metallic Materials at High Temperatures

The deformation of a metal specimen begins with the application of a load. There are two kinds of high-temperature strain, namely, deformation under constant stress σ (i.e. creep) and deformation under constant strain rate $\dot{\varepsilon}$. Physical distinctions between these two processes are not essential. In this book we shall use the definitions "high-temperature strain" and "high-temperature creep" almost as synonyms.

In Fig. 1.1 one can see the dependence of strain upon time, $\varepsilon(t)$, when the applied stress remains constant. In the general case the curve contains four stages: an incubation, primary, steady-state and tertiary stages. The steady-state stage is the most important characteristic for metals, because it takes up the greater part of the durability of the specimen. Correspondingly, the minimum strain rate during the steady-state stage, $\dot{\varepsilon}$, is an important value because it determines the lifetime of the specimen. The tertiary stage is associated with a proportionality of the creep strain rate and the accumulated strain. It is observed to a certain extent in creep resistant materials. The tertiary stage is followed by a rupture.
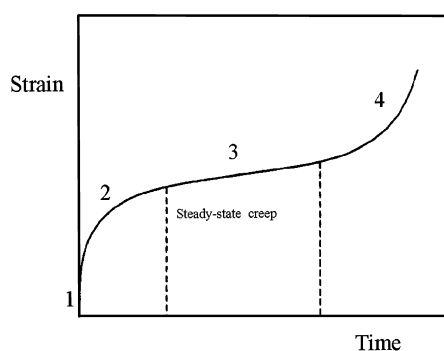

Fig. 1.1 The typical curve of creep.

Thus, the following stages are observed:

1. The incubation deformation. For this stage the strain rate $\dot{\varepsilon} \neq \text{const}$; $\ddot{\varepsilon} > 0$.
2. The primary stage, during which $\dot{\varepsilon} \neq const$; $\ddot{\varepsilon} < 0$. The creep rate decreases when the strain increases.
3. The steady-state strain. The plastic strain rate is a constant value. $\dot{\varepsilon} = \text{const}$.
4. The tertiary stage. $\dot{\varepsilon} \neq \text{const}$; $\ddot{\varepsilon} > 0$. The tertiary creep leads to a rupture.

High-temperature strain is a heat-activated process. An elementary deformation event gets additional energy from local thermal excitation. It is generally agreed that above $0.5\,T_{\mathrm{m}}$ (T_{m} is the melting temperature) the activation energy of steady-state deformation is close to the activation energy of self-diffusion. The correlation between the observed activation energy of creep, Q_{c}, and the energy of self-diffusion in the crystal lattice of metals, Q_{sd}, is illustrated in Fig. 1.2. More than 20 metals show excellent correlation between both values.

The measurement of the dependences $\dot{\varepsilon}(\sigma, T)$ was the first step in the investigation of the problem under consideration. The functions $\sigma(\dot{\varepsilon}, T)$ and the rupture life (durability) $\tau(\sigma, T)$ have also been studied. For the dependence of the minimum strain rate $\dot{\varepsilon}$ upon applied stress σ several functions have been proposed by different authors. The explicit function $\dot{\varepsilon}(\sigma, T)$ is still the subject of some controversy. The power function, the exponent and the hyperbolic sine have been proposed.

The following largely phenomenological relationships between $\dot{\varepsilon}$, σ and T are presented in various publications.

$$\dot{\varepsilon} = A_1 \exp\left(-\frac{Q}{kT}\right)\sigma^n \tag{1.1}$$

$$\dot{\varepsilon} = A_2 \exp\left(-\frac{Q - v\sigma}{kT}\right) \tag{1.2}$$

$$\dot{\varepsilon} = A_3 \exp\left(-\frac{Q}{kT}\right)\sinh\left(\frac{\alpha\sigma}{kT}\right) \tag{1.3}$$

where $A_1, A_2, A_3, n, v, \alpha$ are constant values; Q is the activation energy of the process; k is the Boltzmann constant and T is temperature.

If we suppose that constants A_1, A_2, Q, n, v do not depend upon temperature then it is easy to obtain

$$Q = -k\left[\frac{\partial \ln \dot{\varepsilon}}{\partial\left(\frac{1}{T}\right)}\right]_\sigma \tag{1.4}$$
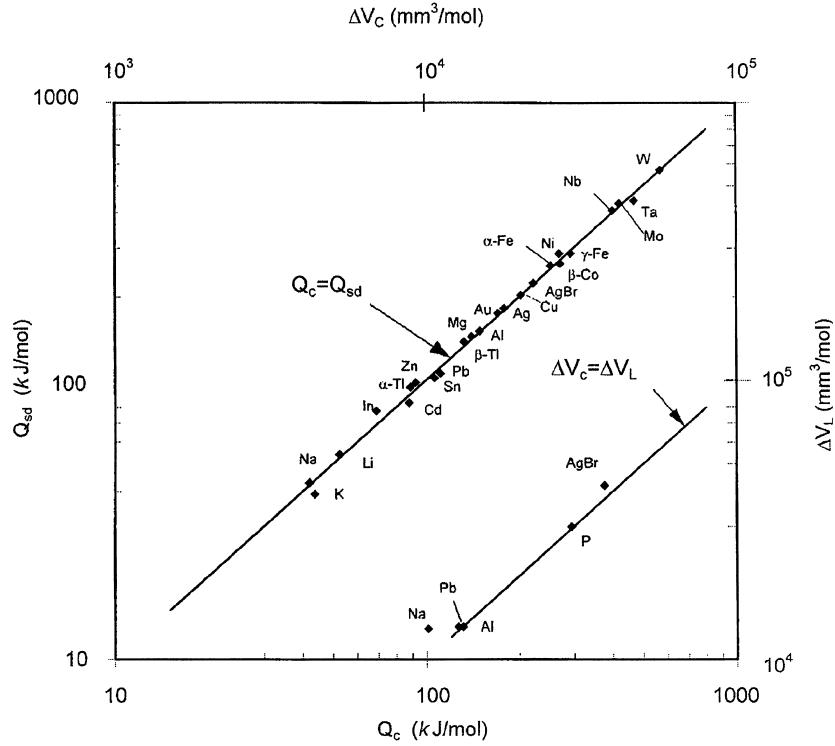
Fig. 1.2 Comparison of the activation energy of creep, Q_c, and the activation energy of self-diffusion, Q_{sd}, for pure metals. The activation volume, ΔV_c is also shown. Data of Nix and Ilshner [7].

Thus, the activation energy can be found from experimental curves of $\ln \dot{\varepsilon}$ vs. $1/T$.

If A_2 and Q do not depend upon stress

$$v = kT \left(\frac{\partial \ln \dot{\varepsilon}}{\partial \sigma} \right)_T \tag{1.5}$$

where v is an activation volume. The latter value can be calculated from the dependence of $\ln \dot{\varepsilon}$ on σ.

Transmission electron microscopy is used, in particular, for the study of crept metals. Investigators have observed the formation of subgrains in different metals. Grains in polycrystalline materials as well as in single crystals disintegrate during high-temperature deformation to smaller parts called subgrains or cells.

First, we show an electron micrograph of subgrains and sub-boundaries in crept nickel, Fig. 1.3. One can see a clean area in the center of (a), i.e.

a subgrain or cell, surrounded by dislocation aggregations. The cell walls separate relatively dislocation-free regions from each other. Subgrains are also seen at the borders of the picture.

Aggregations of dislocations in sub-boundaries seem to be more or less ordered. We observe regular dislocation lines elongated in the same direction. The dislocation lines form low-angle sub-boundaries unlike the large-angle boundaries between crystallites (grains). Thus, the subgrains are misoriented to each other. The misorientation is of the order of tens of angle minutes i.e. of milliradians.
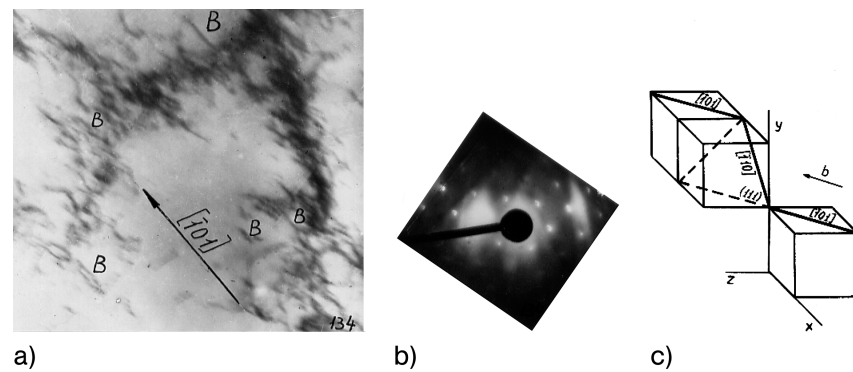

a)   b)   c)

Fig. 1.3 Subgrain in nickel tested at 1073 K, stress 20 MPa. (a) Bright-field image. Screw dislocations along [$\bar{1}01$] are denoted as B. (b) Electron diffraction pattern. (c) Scheme of the arrangement of dislocations inside the boundary.

In Fig. 1.3(a) the so-called diffraction contrast is observed. It is created by separate dislocations in sub-boundaries. Strictly speaking, the electronic beam generates an interference contrast due to stresses near the dislocation line. In Fig. 1.3(c) the screw sub-boundary dislocations are shown to be elongated in the directions of the face diagonals of the cubic face-centered crystal lattice.

Several theories of dislocation mechanisms of high-temperature deformation were proposed in early studies on the problem. According to the theories of one group a glide of dislocations along slip planes occurs during the creep process and this is followed by a climb of edge dislocations at the rate-controlling distances [9, 10]. The climb velocity depends upon the flux of vacancies in the crystal lattice.

Another group of theories consider creep as a diffusion controlled motion of screw dislocations with jogs [11]. The jog is known to be a bend, a double kink at the dislocation line. The jog cannot move further without diffusion of the lattice vacancies or interstitial atoms. Only thermal equilibrium generation of jogs was considered. The probabilities of the heat generation of alternating

jogs that have opposite signs (vacancy-emitting and interstitial-emitting) are equal to each other. Thus, from Barrett and Nix's [11] point of view a screw dislocation contains both types of thermally generated jogs, equally spaced and alternate along the dislocation line. They emphasize that the average spacing between jogs was never measured directly.

Attention has been devoted in the literature to other theories. Some investigators developed a model for creep based on the Frank dislocation network [12]. Concepts of internal stresses were discussed in subsequent publications as well as steady-state substructures and possible values of n in the power law (1.1). The dislocation theories of creep have been considered in detail in a review [7].

I would like to emphasize certain shortcomings in these studies and in the state of the problem under consideration.

1. The researchers pay special attention to the functional connections between the external parameters of deformation: i.e. between the strain rate and stress. For example, principal concern is paid to the numerical value of the steady-state stress exponent, n, in the power law (1.1). However, the same experimental data can satisfy both Eq. (1.2) and Eq. (1.1). The more so when graphs are plotted usually in logarithmic coordinates. Moreover, Eq. (1.3) becomes Eq. (1.2) if the stresses are not small enough. According to my point of view, an analysis of the dependences $\dot{\varepsilon}(\sigma, T)$ or $\sigma(\dot{\varepsilon}, T)$ cannot allow one to conclude unequivocally about the physical mechanism of the phenomenon under consideration.

2. Some properties of dislocations as defects of the crystal lattice are the basis for various dislocation models of high-temperature deformation. It would be much better to use the real parameters of the structure which could be measured experimentally. On the contrary, some parameters of theories, which have been proposed, cannot be measured.

3. It is surprising that though the substructural elements have been observed in many studies on various metals, none of the previous strain rate equations contains these parameters directly. It appears that very little systematic data for correlation between the structure and the creep behavior have been reported. Dimensions and misorientations of substructural elements have not been measured sufficiently.

4. No attempts have been made to calculate or even to estimate the strain rate of metals and solid solutions based on the test conditions, observed structure and material constants.

5. Some authors introduce equations, which contain 3–5 or more so-called fitting parameters. Varying these parameters enables one to obtain a satisfactory fit between experimental and calculated deformation curves. However, one should not draw any conclusion about the correctness of a physical theory from this fit.

6. The various directions of research are somewhat separated from each other.

7. The physical nature of the creep deformation behavior of industrial superalloys has not been investigated sufficiently. Quantitative physical theories are still being worked out.

I consider that the essence of the problem of the physical fundamentals of high-temperature strain consists in structural evolution under specific external conditions. My approach to the problem is based on the concept that the effect of applied stresses upon the crystal lattice at high temperatures results in distinctive structural changes and these specific changes lead to the definite macroscopic behavior of a material, especially, to the strain rate and to the stress resistance.

A key to the problem is the response of the structural elements of a material. In some way the situation is in accordance with the Le Chatelier rule. The changes in a metallic system which take place under the influence of external conditions are directed so as to relax this influence. The formation of an ordered dislocation structure is just an evolution process which tries to act against applied stresses. The point is that the high temperature conditions give the possibility of supplying the dislocation rearrangement with energy and which results in the substructure formation.

That is why our aim is first to investigate quantitatively and in detail the interaction of dislocations with each other, the formation of subgrains, the interactions between dislocations and particles in superalloys, and only then to conclude a physical mechanism for the process.

The nature of microscopic processes should be revealed as a result of experiments that enable one to observe the events on the atomic, microscopic scale, and not on the basis of the general properties of crystal lattice defects nor on the basis of mechanical tests.

This approach enables us to find unequivocal and explicit expressions for the high-temperature steady-state strain rate. These expressions contain substructural characteristics, physical material constants and external conditions. The essence of this approach is defined as the physics of the processes, which are the structural background and the kinetic basis of the macroscopic deformation of metals and solid solutions in the interval (0.40–0.70) T_{m}, where T_{m} is the absolute melting temperature. Superalloys operate at higher temperatures.

Thus, the planned path can be shown schematically as follows. Systematic investigations of the structure of metals strained at high-temperature. $\Rightarrow$ The determination of the physical mechanism of strain, which should be based upon experimental data. $\Rightarrow$ Calculation of the macroscopic strain rate on the basis of this mechanism. $\Rightarrow$ Comparison with experiment.

This plan demands first an efficient structural investigation and detailed proofs of correctness of physical models.

An *in situ* investigation of metals is necessary in order to address the problem of the physics of the high-temperature deformation.