SOME RESULTS OF AN EXPERIMENTAL STUDY OF SHORT-TIME CREEP IN UNIAXIAL TENSION

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Creep rates of metals may vary within very wide limits, depending on the levels of stress and temperature; consequently appreciable changes in the dimensions of a part or its failure in some way or other may take place after a time varying from one second to tens of years. The processes that occur during the shorter times may be described as dynamic, and their investigation requires special methods.

The term short-time creep will be used to describe creep that takes place in approximately $1-10^3$ sec. The determination of such creep is rather arbitrary and approximate in character, its significance consisting not so much in indicating the duration of the process as in defining a certain range of stresses and temperatures in which measurable creep deformation will have taken place after a given time. The latter expression may seem very indeterminate. In practice the designer must take into account the deformations that make up a certain appreciable amount of elastic deformation, for example 10%. Thus a lower limit is established for what is called "significant" deformation, this limit corresponding roughly to the amount of deformation that can be reliably measured by ordinary methods. As regards the upper limit, this is the deformation, of the order of 1–2%, which is usually regarded as the maximum permissible in design. The region of larger deformations is usually of interest only in connection with conditions of fracture, which takes place at a uniform deformation of the order of 6–10%.

In the last ten years there has been a large number of publications containing the short-time creep characteristics of various metals and alloys; these investigations were undertaken primarily to satisfy practical requirements, and their principal object was to provide the initial data for estimating the strength of various parts. However, short-time creep also has a certain importance in the mechanics of materials under creep. At high temperatures and short periods of loading, the structural changes taking place during creep in the majority of materials do not have any significant effect on the creep rate, and the only variable structural parameter is that of embrittlement, the effect of which becomes apparent when the deformation reaches a certain value. Hence a theory of short-time creep may be based on a simpler model than those which are usually employed to describe long-and medium-term creep.

§1. A characteristic feature of short-time creep is the absence of the initial sections of the creep curves; this may be interpreted as the absence of hardening. If the equation for creep accompanied by hardening is written in the form

$$\dot{e}e^{\alpha} = f(\sigma) \qquad (1.1)$$

then α must be regarded as zero.

Thus creep under constant stress takes place at a constant rate $\varepsilon = \dot{e}$ as long as $\dot{e}$ is fairly small and embrittlement, leading to acceleration of the creep, does not occur.

The absence of hardening at fairly high stresses and temperatures has been observed in tests on many materials (D16, AMg6, VM-65-1, OT-4, VT-14, EI-811, EI-696 and certain others). Bernett [1] has noted this fact in the alloy N-155.

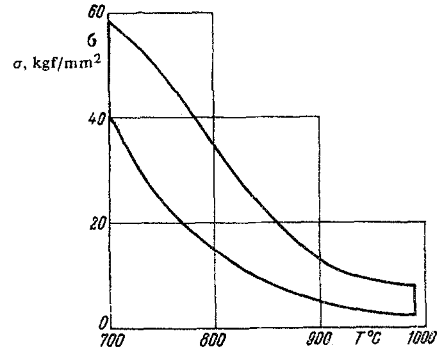

Fig. 2. Region where there is no hardening for steel EP-164.

The usual power and exponential relationships between rate of deformation and stress are very well satisfied for short-time creep. The choice of the form of the relationship which is best suited to each individual case is determined by considerations of convenience in use [2]. We shall take

$$\varepsilon = \varepsilon_n\left(\frac{\sigma}{\sigma_n}\right)^n \quad \text{or} \quad \varepsilon = \varepsilon_e \exp\left(\frac{\sigma}{\sigma_e}\right) \qquad (1.2)$$

Here ε_n, σ_n, ε_e, σ_e are constants at a given temperature.

The validity of relationships of the type (1.2), which establish the independence of the creep rate from the previous history of the specimen, is confirmed by the following experiments.

1. **Repeated relaxation** [3]. A specimen (material AMg6M, temperature 227° C) is loaded to a certain stress σ_0; then the apparatus is switched over to the relaxation regime and the σ-t curve is recorded; after a certain time (tens of seconds) the stress is again raised to the value $\sigma_1 \approx \sigma_e$ and the relaxation recorded. These cycles are repeated three to five times, the ratio of the final stress in the cycle to the initial stress being 0.3–0.5. The results are as follows: 1) if the initial

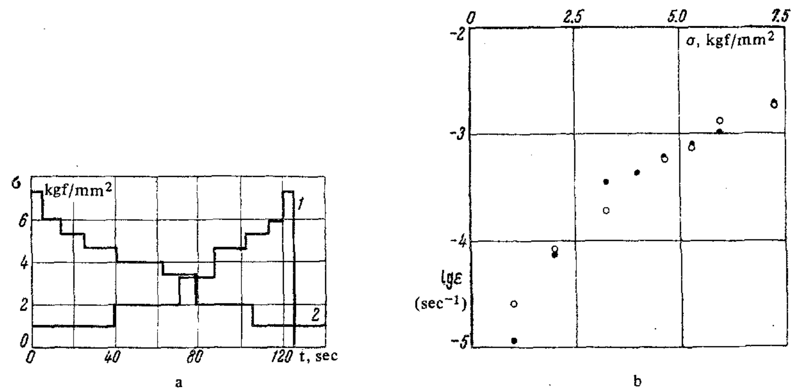

Fig. 1. Titanium alloy VT-14, 800° C. a) change of stress with time during the testing of two specimens; b) the creep rates corresponding to (a) (open circles No. 1, full circles No. 2).

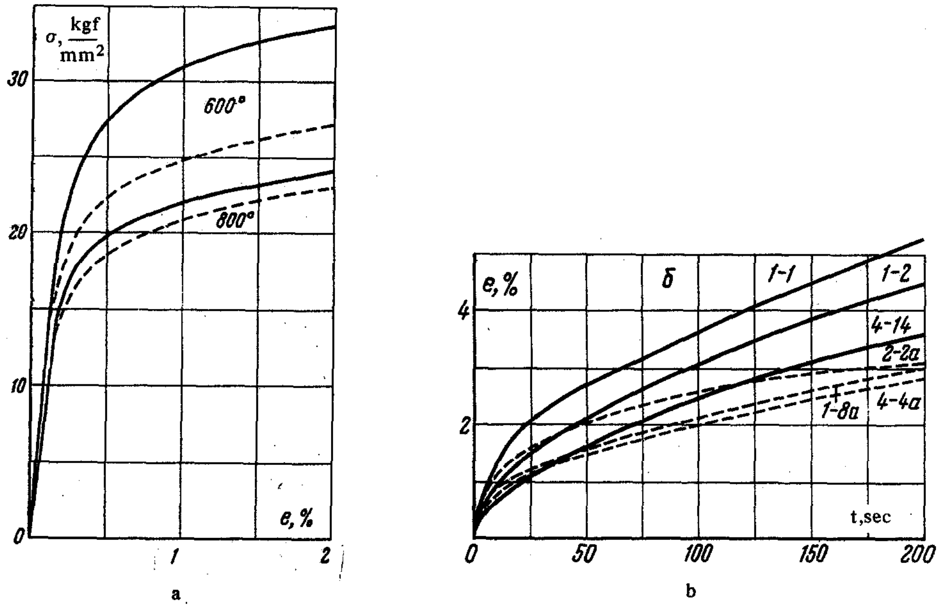

Fig. 3. Steel EI-654, 60° C. Full lines represent specimens treated by the first regime and broken lines those treated by the second regime. a) Load-extension curves at a loading rate of ~30 kgf/mm^2 sec. The specimens were cut from a disk 120 mm in diameter at a distance of 40 mm from the center. b) Creep curves for σ = 36 kgf/mm^2. The specimens were cut from the same disk 1–1, 1–2, 1–8a at distances of 40 mm from the center; 2–2a at 25 mm; 4–14, 4–4a at 53 mm.

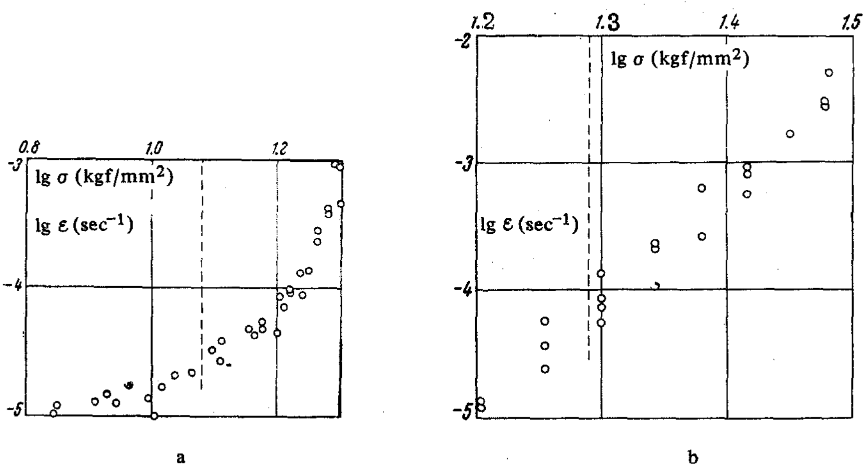

Fig. 4. Dependence of creep rate on stress. The broken lines indicate the limit of proportionality on the transient-deformation curve. a) Aluminum alloy AMg6M, 200° C (five specimens); b) Steel EI-654, 700° C (five specimens).

Properties of Alloys

Alloy	Base	Heat-treatment conditions	Vickers hardness after heat treatment
VT-14 sheet 2 mm	Ti	Annealed in air at 850° C for 1 hr	330
EP-164 disk 100 mm diameter	Fe	1) Quenched in water from 1130° C (held 2 hr) 2) Aged at 750° C for 16 hr + 700° C for 8 hr	270—304
EI-696 rod 40 mm diameter	Fe	1) Quenched from 1150° C (held 2 hr), cooled in air 2) Aged at 760° C for 20 hr	329—338
EI-811 sheet 2 mm	Fe	Quenched from 1000° C (held 30 min), cooled in air	205—208
Kh18N10T sheet 2.5 mm	Fe	Heated to 1070° C (held 20 min), cooled in air	136—134
AMg6M sheet 2 mm	Al	Tested in as-received state	

stresses of the cycle are identical, the relaxation curves coincide exactly; 2) the equation of the relaxation curve obtained, for example, for the exponential relationship Eq. (1.2) in the form

$$\sigma(t) = -\sigma_e \ln\left[\exp\left(-\frac{\sigma_0}{\sigma_e}\right) + \frac{\varepsilon_e E}{\sigma_e}\, t\right] \qquad (1.3)$$

describes the experiment for any cycle equally well (E is the elastic modulus and σ_0 is the initial stress of the cycle). The relevant curves are given in [3].

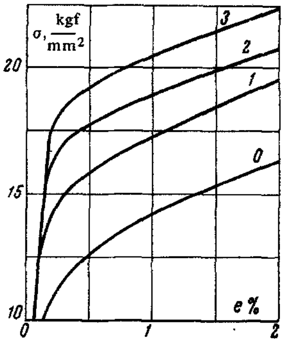

Fig. 5. Steel Kh18N10T, 800° C. Load-extension curves at a loading rate of 10–15 kgf//mm² sec. 0) Undeformed specimen; 1) preliminary creep deformation 0.28% (after 58 sec); 2) 0.52% (80 sec); 3) 1.0% (3240 sec). Preliminary creep stress 12.0 kgf/mm², temperature 800° C.

2. Stepwise loading. By varying the loading in stages we obtain a creep rate which is independent of the order of the steps and which is determined only by the stress acting at a given moment (the experiment is carried out at constant temperature). Figure 1a shows examples of such stepwise programs carried out on two specimens of titanium alloy VT-14 at T = 800° C. The relationships between the creep rate and stress are given in Fig. 1b.

The test described with stepwise loading may be conveniently taken as a basis for determining the constants in Eqs. (1.2). The method of determining these quantities has been set out in detail in [4], which also gives the mechanical and electrical diagrams of stressing device with a rapid-action lever unit for carrying out the respective programs.

It should be noted that the independence of the short-time creep rate from the previous history of the specimen extends with certain limitations to the temperature history. If the temperature is raised in steps

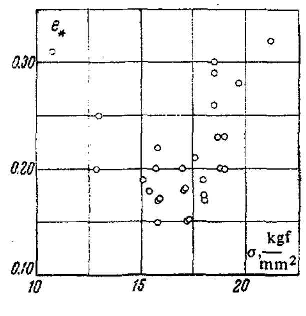

Fig. 6. Steel Kh18N10T, 800°C. Dependence of elongation at fracture on stress.

during testing, the creep rate will be determined by the instantaneous values of temperature and stress. Such an experiment has been made on the alloy D16T at 240–380°C and on alloy VT-5 at 600–900° C.

By combining the results of short-time creep tests with existing test data for medium- and long-term creep, the general conclusion may be drawn that with rise of stress and temperature the quantity α in Eq. (1.1), which characterizes the degree of hardening during creep, decreases and becomes zero at fairly high values of σ and T. (In this it is not assumed that values of α may not again increase with further rise in temperature beyond the range investigated). This conclusion agrees with physical concepts about the increasing role played by recovery with rise in temperature [5].

The available experimental evidence leads to the conclusion that in many materials the region of σ–T in which α may be regarded as zero approximately coincides with the region in which $\varepsilon = 10^{-5}$–10^{-2} sec^{-1}. Figure 2 shows this region for the steel EP-164. This does not rule out the possibility that there may be conditions for creep without hardening outside the region defined.

§2. We will now consider certain deviations from the normal behavior described above.

Copper and certain copper alloys in the temperature range 400°–1000° C and at stresses which give a creep life of the order of tens to hundreds of seconds exhibit considerable hardening. In tests with stepwise loading the "deformation-time" curves lead to conclusions about the theory of hardening in its simplest form in which the accumulated creep deformation is taken as a measure of the hardening.

Rapid hardening occurs during the creep of the austenitic steel Kh18N19T. After austenitization of this steel at 1070°C for 20 min, hardening at 800°C is so great that the creep curves approximate only poorly to a power function. The creep of specimens cut from the same (cold-rolled) sheet and tested under the same conditions but without preliminary heat treatment takes place without hardening.

It should be noted that the heat-treatment regime indicated does not give a completely austenitic material. This appears only at testing temperatures of the order of 90°C. In this case "negative" creep is observed at small loads, no doubt as a result of the $\alpha \to \gamma$ phase transformation. Positive creep at fairly high stresses is not accompanied by hardening.

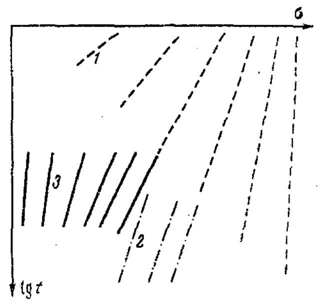

Fig. 7. Temperature dependence of creep (schematic): 1) According to Zhurkov; 2) according to Dorn; 3) short-time creep.

We will now consider aspects of the short-time creep of the titanium alloy VT-5-1. In the as-received state this alloy has undergone low-temperature annealing. Creep curves of this alloy without preliminary heat treatment are straight lines when plotted in the coordinates e–t. Annealing for times from 10 min to 36 hr in air at 800°C substantially changes the shape of the curves. A region of delayed creep appears which is similar in certain respects to the incubation period in the creep of iron containing carbon [6]. Then the rate begins to rise, increasing by 5–30 times relative to the minimum attained in the incubation region, and it remains constant up to the tertiary stage which begins at deformations exceeding 5–7%. The longer the annealing time the more marked is this effect, which is evidently connected with the gas saturation of the surface layers during annealing.

§3. For purposes of calculation it is often assumed that plastic deformation at high temperatures consists of "transient" and creep deformations. We shall give certain additional experimental data to show that the forms of plastic deformation mentioned are clearly governed by different mechanisms.

The stainless steel EI-654 was heat treated in accordance with two regimes. The first consisted of: 1) heating in a salt bath to 820°C and holding for 10 min; 2) heating in a barium bath to 1050°C, holding for 5 min and cooling in water; 3) repetition of the first operation;

4) repetition of the second operation. The second regime consisted of heating to 1050°C in air and cooling in water. Disregarding the analysis of the structures obtained after heat treatment by these two regimes, we will turn to the test results. Figure 3a gives the deformation curves determined at a loading rate of about 30 kgf/mm^2 sec at 600°C for specimens treated by the two regimes. The resistance to transient plastic deformation of the material subjected to single quenching after heating in air is much the lower (~20% of the $\sigma_{0.2}$ stress). At the same time the creep curves of these two sets of specimens (Fig. 3b) differ only slightly, the material treated in accordance with the second regime exhibiting a somewhat greater creep resistance. This effect evidently stems from the fact that the same structural changes during heat treatment may have different effects on the resistance of the material to transient plastic deformation on the one hand and to short-time creep on the other.

It should be noted that with rise in temperature to 80°C no difference is found between the sets of specimens treated by the two regimes in creep tests, but on rapid extension there is a small difference.

An important question concerns the effect of these two forms of plastic deformation on one another. The designer must be clear about this if he is to choose a sound design method for parts which in service undergo both transient and creep deformations. In this case the sequence of these deformations is unimportant provided that the total plastic deformation remains fairly small. In this field only a few isolated experimental results are available. It was shown in [7] that in steel 30 KhNMA a preliminary small transient deformation has no effect on subsequent comparatively slow stress relaxation (duration of test tens of hours). Such a result has been obtained in many investigations into the effect of the amount of preliminary deformation of subsequent creep; reducing the initial deformation to 1–2% has no effect on the creep curve (see, for example, [8]). The effect of a small transient deformation on the short-time creep of the aluminum alloy AMg6M was investigated in [3], where it was found that if any effect exists it is masked by the scatter of the experimental data.

On the other hand it often appears that in treating the test data by means of Eq. (1.2) it is necessary to take values of the constants σ_n', n' for $\sigma \le \sigma'$ and σ_n'', n'' for $\sigma \ge \sigma'$. The value of σ' is close to the limit of proportionality. However, there are also cases in which the limit of proportionality lies within the stress range investigated and yet the constants σ_n, n remain invariable over the whole range. An example is illustrated in Fig. 4.

The effect of creep deformation on the transient deformation curve has been studied even less.* The hardening effect of creep deformation has been noted in [9].

The present authors, with A. V. Dolgov, have carried out the following experiment. A series of specimens of the austenitic steel Kh18N10T was subjected to creep deformations up to 1% at a stress of 12 kgf/mm^2 and a temperature of 800°C. Under these conditions the creep curves indicated hardening. Then the specimens were unloaded, cooled, held for a period of several days, and then subjected to rapid failure (at a rate of ~10–15 kgf/mm^2 sec) in an electromagnetic machine [3]. The curves obtained in this way, which are practically the same as transient-deformation curves, are presented in Fig. 5. The resistance to transient plastic deformation of specimens which have undergone preliminary creep deformation is higher, and becomes higher the greater the extent of plastic deformation. Specimens of the titanium alloy VT-14 were tested under the same conditions at 600°C, when creep occurs with practically no hardening. No effect of the preliminary creep was found on the transient-deformation curve.

It may be assumed that in the case when creep deformation hardens the material and increases its creep resistance, it also impedes any plastic deformation; when creep occurs without hardening (ordinary short-time creep) it does not affect the subsequent transient deformation. However, this assumption requires verification. In particular, it is necessary to determine the part played by aging in the tests on steel Kh18N10T described above.

*This paragraph is concerned with experiments in which both transient and creep deformations occur at the same temperature.

With rise in temperature the creep deformation greatly exceeds the transient plastic deformation. Consequently the question under discussion in practice loses its significance at fairly high temperatures, except for cases of very short-time incidental stresses beyond the "transient" limit of proportionality.

Thus, under conditions of short-time creep, it may as a rule be assumed in design that small transient plastic deformation and creep deformation have no effect on one another, although the use of this hypothesis in a general form for creep with hardening may lead to errors.

The hypothesis described has formed the basis of a method for constructing transient-deformation curves [3]. This method consists in determining the load-extension curves at a constant rate of stress increase; then, from a knowledge of the dependence of the creep rate on stress, it is possible to distinguish the component of deformation which is associated with creep.

It may be noted that the hypothetical transient curve obtained in this way at comparatively low temperatures often differs little from the initial deformation curve if the latter is determined at a fairly high rate of leading. At higher temperatures the difference may be considerable.

§4. By carrying out the reverse operation, i.e., by adding to the transient curve the creep deformations calculated in accordance with the given law of stress variation with time, we obtain a curve showing the change of deformation with time. A good check experiment is provided by extension at a constant rate of loading (a fan of σ–e curves is calculated for rates of leading varying by two to three orders of magnitude). It is also possible to do this with more complex loading conditions.

§5. Two features of short-time creep may be noted.

1. It is well known that unloading after fairly prolonged creep causes the recovery of part of the accumulated deformation. This effect does not occur in short-time creep either in the usual case (without hardening) or in the case where hardening takes place. This has been confirmed on steel Kh18N10T, copper, and a copper alloy, the specimens being held after creep for a time comparable with the creep time.

2. It is also known that in the creep of structural materials at temperatures of the order of half the melting point, there is marked anisotropy of the creep characteristics (see, for example, [10]). In the field of short-time creep the anisotropy decreases with rise in temperatures at which there is still an adequate load-carrying capacity. This is evidently connected with the rapid recrystallization of the material.

§6. The embrittlement parameter (designated by ω) characterizes the structural change during creep which weakens the material and so leads to its fracture. For a "pure" undeformed specimen $\omega = 0$ and at the moment of failure $\omega = 1$.

Features of the failure of the alloy D16 in short-time creep have been noted in [13] and estimates of the value of ω during creep have been given. It is shown that if we take $\omega = \varkappa e$ ($\varkappa$ being a constant) and consider that the stress acting in the specimen increases as a result of weakening of the specimen by microcracks and macrocontraction, many details of failure are adequately described.* In particular these assumptions lead to a constant deformation at failure and to a linear summation of the damage observed in the experiment. By an appropriate choice of the value of $\varkappa$ the tertiary portion of the creep curve and the creep life can be calculated.

A general phenomenological scheme of creep failure has been constructed in [11], which is based on the assumption that creep is a combination of at least two processes–plastic deformation and crack formation. The interaction of these is reduced to a simple but fairly general scheme. (The scheme in [3] may be derived as a particular case of this.)

*It should be noted that these assumptions are in accordance with certain microstructural observations of creep failure [12, 13].

The kinetic equations of creep and fracture are written in the form

$$\varepsilon = \varepsilon_n \left(\frac{\sigma}{\sigma_n}\right)^n (1-\omega)^{-r} ,$$
$$\omega = k\sigma^m (1-\omega)^{-q} . \qquad (6.1)$$

Without dwelling on the details of this model, we will note only certain features of short-time creep and fracture observed in tests on a large number of structural materials. In doing so we shall indicate the relevant simplifications of the general Eqs. (6.1).

1. The fact noted in [3] that the deformation at failure is independent of the nominal stress (creep life) for alloy D16 is true for the majority of materials tested. This means that in Eqs. (6.1) we should take n = r = m = q. However, in individual cases comparatively small changes in stress may cause sharp and nonlinear changes in e_* (see Fig. 6 for the results of tests on steel Kh18N10T at 800°C).

2. In contrast with long-term creep, in which a rise in temperature causes rapid embrittlement and there is a fall in the plasticity of commercial alloys, short-time creep is characterized by a gradual change to greater plasticity with rise in temperature. This corresponds to a reduction in the value of k with temperature. There are, of course, certain exceptions to this general type of behavior. In particular, commercially pure copper, which is fairly plastic at 500–600°C, exhibits completely brittle fracture at 800°C (the test was carried out in air).

3. For the majority of materials there is a temperature range in which embrittlement during short-time creep can generally be neglected (k = 0).

§7. The parameters which appear in Eqs. (1.2) are functions of temperature. In determining the temperature dependence of these parameters it was intended only to find suitable empirical formulas which could be used to interpolate the test data; any extrapolation of these data must be regarded as speculative. The existing theoretical relationships are associated as a rule with certain hypothetical elementary creep mechanisms which can be realized in pure form only under special conditions. Thus the exponential relationships used by Zhurkov [14]

$$\varepsilon = \varepsilon_0 \exp\left(-\frac{U_0 - \gamma\sigma}{kT}\right) \qquad (7.1)$$

and by Dorn [15]

$$\varepsilon = \varepsilon_1 \exp\left(-\frac{U_0}{kT}\right) \varphi(\sigma) \qquad (7.2)$$

are confirmed in the relevant ranges of stress and temperature.

If the creep test data are plotted in the coordinates σ against lg ε (Fig. 7), then according to Zhurkov the slope of the straight lines obtained decreases with rise in temperature, whereas according to Dorn it remains constant. In the region of σ-T where commercial alloys exhibit short-time steady-state creep, the change in the relationship σ–lg ε with temperature may be more complex, in spite of the fact that a comparatively narrow range of deformation rates is involved. At low temperatures in the range being considered the slope of the straight lines σ–lg ε may remain constant, but at fairly high temperatures it always increases with temperature.

Hence the temperature dependence of the parameters ε_e and σ_e may be taken as follows:

for low temperatures

$$\sigma_e = \text{const}, \qquad \varepsilon_e = \varepsilon_{e1} \exp\frac{T}{\vartheta_1} , \qquad (7.3)$$

then

$$\varepsilon = \varepsilon_{e1} \exp\left(\frac{\sigma}{\sigma_e} + \frac{T}{\vartheta_1}\right); \qquad (7.4)$$

for high temperatures

$$\sigma_e = \frac{T_0 - T}{\beta}, \qquad \varepsilon_e = \varepsilon_{e2} \exp\frac{T}{\vartheta_2} , \qquad (7.5)$$

then

$$\varepsilon = \varepsilon_{e2} \exp\left(\sigma \frac{\beta}{T_0 - T} + \frac{T}{\vartheta_2}\right). \qquad (7.6)$$

The ranges for the determination of the constants are found in each case by experiment.

By way of example we give the values of these constants for two steels:

EP-164 in the range 700–875°C

$$\sigma_e = 3{,}33 \text{ kgf/mm}^2, \ \varepsilon_{e1} = 3.63\cdot 10^{-30} \text{ sec}^{-1}, \ \vartheta_1 = 15.80^\circ \text{ C} ;$$

EI-811 in the range 650–800°C

$$\varepsilon_{e2} = 8.32\cdot 10^{-17} \text{ sec}^{-1}, \qquad \beta = 240 \text{ mm}^2 \cdot \text{deg/kgf} ,$$
$$T_0 = 1128^\circ \text{ C}, \qquad \vartheta_2 = 32.7^\circ \text{ C} .$$

These values of the constants will be valid in the range of creep rates 10^{-5}–10^{-2} sec^{-1}.

The obvious inaccuracy of the exponential creep law (1.2) at stresses close to zero makes it preferable to use the power relationship at very high temperatures (for the given material). The power law, written in the form (1.2), contains three constants, one of which may be chosen at random. If we always take $\varepsilon_n = 10^{-4}$ sec^{-1}, then σ_n will be comparable with the working stress σ.

Usually n decreases monotonically with temperature, but in certain cases the value of n may become constant. Thus, in the case of titanium alloy VT-14, n can be taken as 3 in the temperature range 700–1150° C.

It should be noted that the short-time creep rates are very strongly dependent on temperature. For example, the values of the constants for steel EP-164 give a change in rate of an order of magnitude for a temperature change of 30°C when the mean temperatures are around 800° C. In estimating the scatter of the experimental data the accuracy of measurement and of the temperature control in the experiment should be taken into account.

Clearly the temperature dependence of creep is stronger than that of transient plastic deformation; in this connection the stress range in which the creep rate varies from 10^{-5} to 10^{-2} sec^{-1} moves with rise in temperature along the transient-deformation curve in relation to the limit of proportionality.

§8. In conclusion we give information about the alloys whose properties have been used above (table). The heat treatment was carried out by O. F. Stankevich and P. V. Romanov.

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