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Power law steady state creep in α/β titanium alloys

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Secondary creep behaviour for two α/β titanium alloys, IMI318 and IMI685, is presented for temperatures in the range 670 to 825 K. The relationship between secondary creep rate, $\dot{\epsilon}_s$, and stress, σ, is shown to take the form $\dot{\epsilon}_s = A\sigma^n \exp(-Q_c/RT)$ with n, the stress exponent, and Q_c, the activation energy for creep, having values that are much larger than can be explained by conventional recovery creep models. The anomalies, however, can be explained if $\dot{\epsilon}_s$ is related to the effective stress $(\sigma - \sigma_0)$, with σ_0 an experimentally determined friction stress. The exponent now has a value of about 3.5 and the activation energy for creep at constant $(\sigma - \sigma_0)$ is close to what is expected for diffusion in the matrix. It is suggested, therefore, that a recovery model is appropriate to describe the rate controlling process. Finally, it is shown that if the effective stress is normalized by the proof or yield stress of the alloy then the results for IMI318 and IMI685 fall on a universal line with data previously determined for nickel- and iron-based alloys.

1. Introduction

In a review on creep in α/β titanium alloys Rosenberg [1] pointed out that relationships describing the stress and temperature dependence of steady state creep rates can be placed conveniently into three categories. At high temperatures, i.e. $T > 0.6\,T_m$ (where T_m is the absolute melting temperature), a simple power law is found to be adequate:

$$\dot{\epsilon}_s = A\sigma^n \exp(-Q_c/RT) \qquad (1)$$

with the exponent n typically having a value of 3 to 4 and Q_c a value close to the activation energy for self diffusion. Such behaviour is similar to that observed in pure metals and can be accounted for in terms of conventional recovery creep models. At low temperatures, i.e. $T < 0.4\,T_m$, an exponential dependence of creep rate on stress has been reported but behaviour can be complicated by the occurrence of solute atom drag and strain ageing effects. The third regime, and the one of most significance for engineering applications of creep resistant titanium alloys, is concerned with the temperature range 0.4 to $0.6\,T_m$. The observed behaviour is considered to be the result of a transition from low to high temperature modes and is typified by a functional relationship similar to Equation 1 but with exponents generally larger than 4 and creep activation energies larger than those expected for self or solute atom diffusion. Such behaviour has led to uncertainties about rate controlling mechanisms and hence to the introduction of a number of conflicting models to describe the experimental observations. In assessing the situation Paton and Mahoney [2] consider that the failure to identify clearly the mechanisms determining creep resistance has curtailed the development of alloys with improved creep strength.

The occurrence of high creep activation energies and large stress exponents is not restricted to titanium alloys. Similar anomalies have been reported for precipitation hardened alloys in general and the γ' strengthened nickel-base alloys in particular. However, recent work on these materials has shown that the anomalous behaviour can be removed if the creep rates are related to an effective stress $(\sigma - \sigma_0)$, with σ_0 a friction stress, rather than to the applied stress alone as in Equation 1 [3, 4]. It has been suggested that a more appropriate equation to describe the stress

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and temperature dependence of creep rates is:

$$\dot{\epsilon}_s = A^*(\sigma - \sigma_0)^p \exp(-Q_c/RT). \qquad (2)$$

The exponent p has a value in the range 3 to 4 and activation energies for creep at constant $(\sigma - \sigma_0)$ are close to those for self or solute atom diffusion. Essentially these complex alloys behave under constant $(\sigma - \sigma_0)$ in a similar manner to pure metals at constant σ. The major difference, then, between the complex alloy and the pure metal is the magnitude of the friction stress term. Once allowance has been made for this it would appear that the rate controlling event is the same for both classes of material evaluated to date.

So far, however, titanium alloys have not been evaluated by the σ_0 approach. The present investigation was initiated to rectify this situation and to determine whether the anomalous behaviour in titanium alloys at intermediate temperatures also can be explained by friction stresses. To this end creep tests and σ_0 measurements were carried out on IMI318† and IMI685† within the temperature range 670 to 825 K. The former is essentially a solid solution strengthened alloy which was considered in order to evaluate the effects of microstructural variations on creep behaviour. The latter was included because it contains silicon which is an important constituent of many creep resistant titanium alloys.

2. Experimental procedure

Details of compositions, heat treatments and microstructures are given in Table I together with the codes and symbols used to identify each of the alloys in the graphs presented below.

Commercially IMI685 is β processed and has an acicular microstructure consisting of alpha laths in prior β grains. The IMI318 alloy is usually used at temperatures below those being considered here and for such applications it is forged and heat treated in the α/β phase field. In this study it is evaluated in the three heat treated conditions shown. The β318 was produced by heat treatment above the β transus temperature to be microstructurally similar to β685.

The basic creep data were obtained under constant stress at temperatures in the range 670 to 825 K. Strain measurement was by LVDT transducers located in extensometers which were attached to ridges on the specimens and gauge lengths of 25 mm allowed strain discrimination of about 1×10^{-6}. Friction stress measurements were made during steady state creep and involved progressive reductions in applied stress. After each reduction, $\Delta\sigma$, an incubation period, Δt, occurred and further stress decrements were not made until this period of zero creep had terminated. The reductions continued until it was apparent from a graph of $\Sigma\Delta\sigma$ against $\Sigma\Delta t$ that a limiting stress σ_0

TABLE I Details on the titanium alloys

Alloy	Composition	Heat treatment	Microstructure	Identification code and figure symbols
IMI318	6.19% Al 3.94% V 0.16% O 0.07% N 0.05% Si 0.09% Fe 0.04% C	970 K for 2 h and air cooled	equiaxed α grains with retained β at the boundaries	α318 △
		1220 K for 1 h, air cooled, 1020 K for 2 h, air cooled	equiaxed α and transformed β grains	α/β318 +
		1300 K for 0.5 h, air cooled, 770 K for 8 h, air cooled	acicular α in prior β grains of 0.3 mm diameter	β318 □
IMI685	6.01% Al 5.03% Zr 0.49% Mo 0.25% Si 0.15% O 0.06% N 0.04% Fe	1320 K for 0.5 h, air cooled, 850 K for 24 h, air cooled	acicular α in prior β grains of 0.3 mm diameter	β685 ○ ●

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†Trade name Imperial Metal Industries.

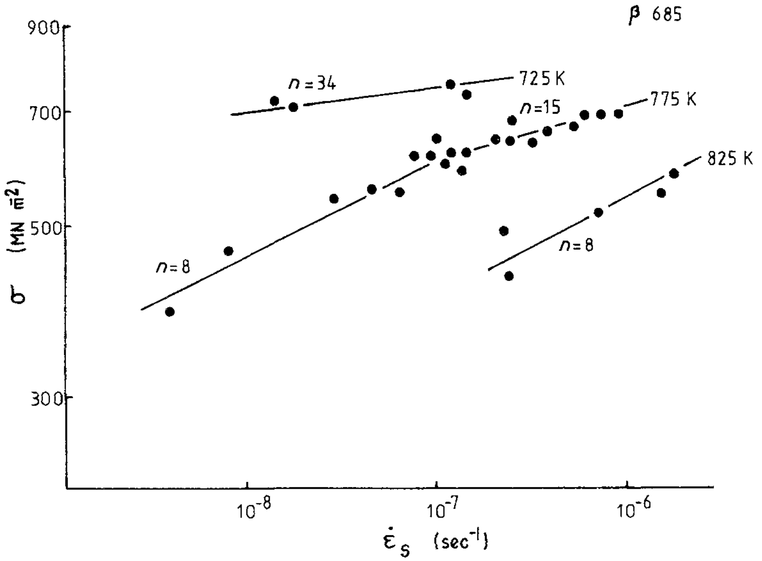

Figure 1 Stress dependence of creep rates in IMI685.

was being approached. Full details on the experimental procedure have been published in previous papers [3, 4].

3. Experimental data

3.1. Stress dependence of creep rate

The β685 was creep tested nominally at 725, 775 and 825 K and the resultant steady state creep rate data are recorded in Fig. 1. It is evident from the log–log plot that for a given range of creep rates the exponent *n* in Equation 1 decreases with increasing temperature. At constant temperature the results demonstrate that there is a tendency for *n* to increase as the stress increases. Thus at 775 K, *n* is approximately 8 for stresses up to a transition stress of about 600 MN m^{-2} and 15 for higher stresses.

The α and α/β318 were tested at 675, 725 and 775 K and the resultant data are recorded in Fig. 2. The changes in *n* with temperature and stress are similar to those already noted for IMI685. Furthermore, it is evident that for these structures producted by heat treatment below the β transus, creep behaviour is little affected by changes in the morphology and distribution of the α and β phases.

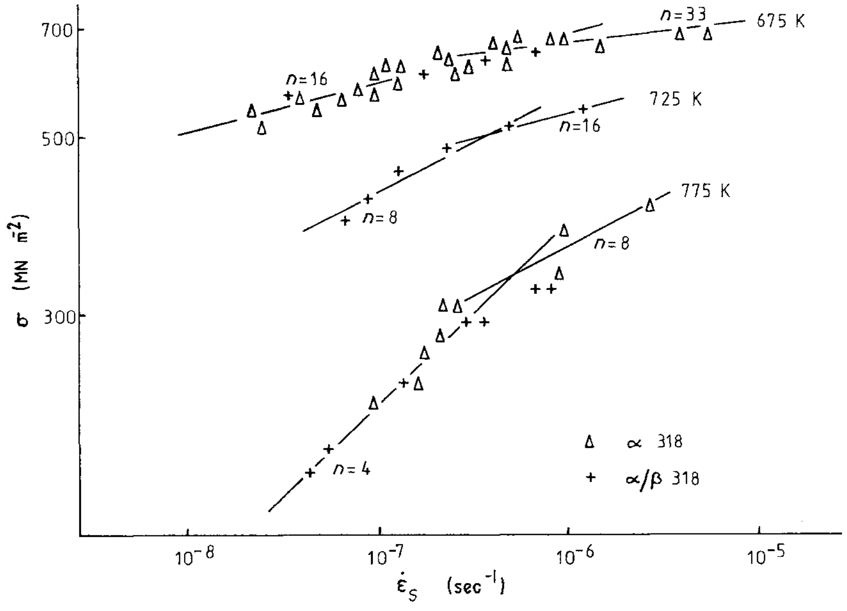

Figure 2 Stress dependence of creep rates in IMI318.

The β318 was tested at 675 and 775 K. Creep behaviour was highly variable at the lower temperature with, for instance, tests failing on loading at 672, 710 and 761 MN m^{-2} but lasting 78 and 220 h, respectively, at 695 and 648 MN m^{-2}. It was evident from the fractured specimens that some of the failures were relatively "brittle" in nature with evidence of facet formation.

At the higher temperature fractures were more ductile and creep behaviour more predictable. The steady state creep rates obtained at this temperature are presented in Fig. 3. Superimposed on the graph are some of the data on α318 and β685 for Figs. 1 and 2. The three sets of data have been deliberately plotted together because of an interesting trend in behaviour that is revealed. Taking α318 at 725 K, β685 at 825 K and β318 at 775 K it is evident that there is an exact superimposition and indeed that the results for these three distinct materials appear to lie on a continuous curve as if a single set of data. Two observations spring immediately to mind from this curve. In the first place, the β microstructure in IMI318 gives a 50 K creep advantage over the α microstructure. Secondly, compositional differences provide a further 50 K advantage for β685 over β318. The second curve plotted in Fig. 3 reinforces these observations. It is made up of α318 data at 675 K and β685 data at 775 K. Once again the two sets merge into one continuous curve which can be represented by several straight line segments with exponents in the range 8 to 34. The superimposition in this case confirms the 100 K creep strength differential between these two alloys.

3.2. Friction stress measurements

Friction stress measurements were made during the secondary creep stage on α318 at 675 K and β685 at 775 K – temperatures at which the respective creep rates at a given stress superimpose. The range of friction stresses that were measured together with the applied stresses at which the tests were conducted are tabulated in Table II.

Within the measurement variability it would appear that over the range of stresses considered, the σ_0 values are virtually independent of stress.

Following previous work [3–6] the friction stress measurements were used to replot the steady state creep rate data in terms of the effective stress $(\sigma - \sigma_0)$. The resultant graph for both materials is reproduced in Fig. 4. Comparing Fig. 4 with Fig. 3, it is evident that there has been a significant change in the form of the stress dependence. From Fig. 3, it can be seen that when creep rates are expressed in terms of stress alone, the stress exponents for α318 at 675 K and β685 at 775 K vary in the range 8 to 34. However, when the rates are expressed in terms of $(\sigma - \sigma_0)$ a single stress exponent, with a value of about 3.5 is appropriate for the entire creep rate range evaluated.

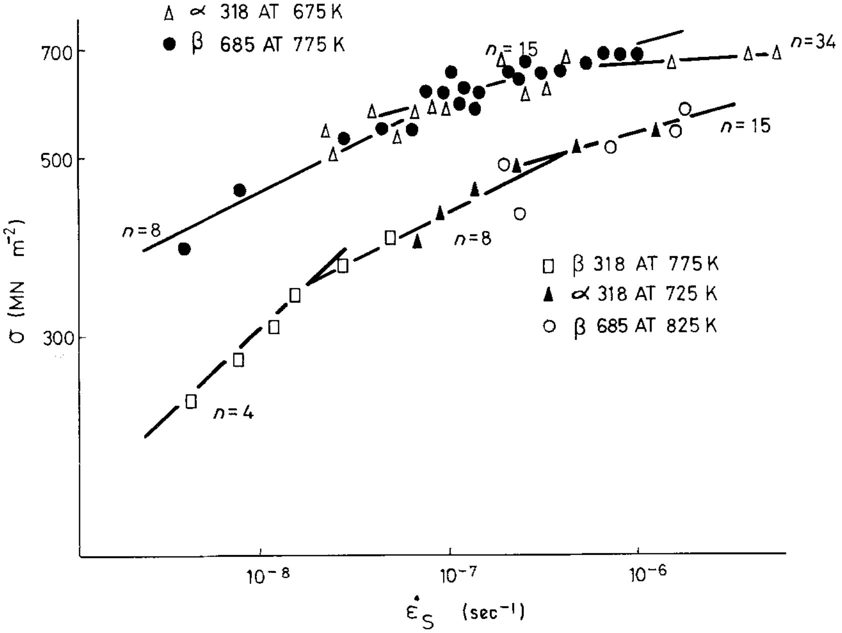

Figure 3 General stress dependence of creep rates in IMI685 and IMI318.

TABLE II Stress dependence of σ_0

β685 at 775 K		α318 at 675 K	
σ(MN m^{-2})	σ_0(MN m^{-2})	σ(MN m^{-2})	σ_0(MN m^{-2})
556	401	556	417
571	417	587	417
	425		401
618	401	627	432
	401		440
	425	648	417
648	409	679	425
658	364		
695	409		
	340		

TABLE III Creep activation energies for β685

σ(MN m^{-2})	Q_c at constant σ (kJ mol^{-1})	Q_c^* at constant $(\sigma-\sigma_0)$ (kJ mol^{-1})
570	361	75
618	361 364 405	143 146 90
648	276	
658	259	105
		Average $Q_c^* = 112 \pm 28$

3.3. Activation energies

Activation energies for creep at constant stress were determined for both α318 at 675 K and β685 at 775 K by means of temperature change experiments. Essentially the temperature initially was increased rapidly by about 20 K and creep rates were measured just before and after the change was affected. Subsequently the temperature was returned to its original value and creep rate measurements were again made. The limited data on IMI318 suggest creep activation energies at constant stress of about 250 kJ mol^{-1}. The more extensive results on IMI685 are recorded in Table III.

For both materials the measured values at constant σ are considerably larger than activation energies for self diffusion (150 kJ mol^{-1}) or solute atom diffusion in α titanium (85 to 127 kJ mol^{-1}) [2].

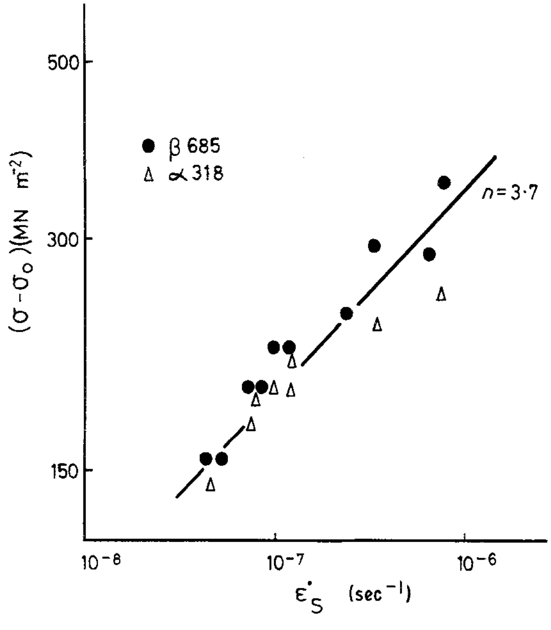

Figure 4 Effective stress dependence of creep rates.

However, in the σ_0 work on precipitation hardened nickel-base alloys reported earlier it has been shown that anomalously high creep activation energies are the result of large friction stresses that change significantly in magnitude with temperature. If activation energy determinations are carried out at constant $(\sigma-\sigma_0)$ as opposed to constant σ then values much closer to those expected for bulk diffusion are obtained [3, 4]. In view of the similarity between the titanium alloys and the nickel base materials, in terms of high stress exponents and activation energies, it was decided to carry out the same experiment on β685 to determine whether a similar temperature compensation is applicable. The base temperature for the experiment was 775 K and temperature increments of about 20 K were again used. Immediately following the attainment of a stable temperature both creep rate and friction stress were measured. All temperature change work was carried out during the steady state stage.

Equation 2 was used to correct for the change in $(\sigma-\sigma_0)$ with temperature. The calculated activation energies for β685 at constant $(\sigma-\sigma_0)$ are recorded in Table III. The pattern of behaviour observed with the nickel-base alloys is repeated and the corrected creep activation energy has an average value of 112 ± 28 kJ mol^{-1} and can be related to typical diffusion activation energies for the titanium alloy system.

4. Discussion

The experimental data demonstrate clearly that apparently anomalous steady state creep behaviour in titanium alloys at intermediate temperatures can be explained by the σ_0 view of creep. In a similar manner to that reported previously for f c c and b c c metals and alloys [3–6], secondary creep rates can be related directly to the effective stress $(\sigma - \sigma_0)$ through the expression:

$$\dot{\epsilon}_s = A^{*\prime}(\sigma - \sigma_0)^p. \tag{3}$$

The exponent p has a value of approximately 3.5 for the two titanium alloys which is typical for the range of alloys evaluated to date in this way.

Measured activation energies at constant stress in both α318 and β685 are considerably higher than those published for diffusion in the titanium alloy system. However, measurements made on β685 are consistent with those made on other alloy systems, with similar anomalies, in demonstrating that a value close to that for self or solute atom diffusion can be obtained if allowance is made for the change in σ_0 with temperature.

The strong similarity between h c p β685, and by inference α318, and the range of b c c and f c c alloys previously tested, in terms of both stress and temperature dependence and the σ_0 correction, implies that the creep rate determining mechanisms are similar in the three alloy systems. The view that the rate controlling process is essentially independent of crystal structure, alloy composition and microstructure has been proposed before [4] when it was shown that creep rate data, obtained over a range of T/T_m from 0.5 to 0.8 on a variety of b c c and f c c alloys, could be made to lie on a universal line if plotted according to the relationship:

$$\dot{\epsilon}_s = B\left(\frac{\sigma - \sigma_0}{\sigma_p}\right)^p \tag{4}$$

The exponent p was found to have a value of approximately 3.5, the constant B appeared to be independent of all material structural and compositional parameters and σ_p was the 0.05% proof or yield stress at the creep temperature.

Using the 0.05% proof stress for α318 and the yield point for β685 at the respective creep temperatures and a strain rate of $8.3 \times 10^{-4}\,\text{sec}^{-1}$, it was straightforward to demonstrate that the creep rate data for both alloys are consistent with Equation 4 with B taking a value of $2 \times 10^{-5}\,\text{sec}^{-1}$. The resultant graph is shown in Fig. 5 and includes not only the results for α318 and β685 but also data for several iron and nickel base alloys.

In essence, therefore, creep in the titanium alloys evaluated is not anomalous but can be accounted for by a dislocation recovery model used previously to describe creep in pure metals and simple alloy systems. The rate controlling event is considered to be the growth of link lengths in a three-dimensional dislocation network [5] with the driving force for growth equated with the effective stress $(\sigma - \sigma_0)$ [6].

An implication of this model is that the essential differences between complex alloys and pure metals and between different alloys of the

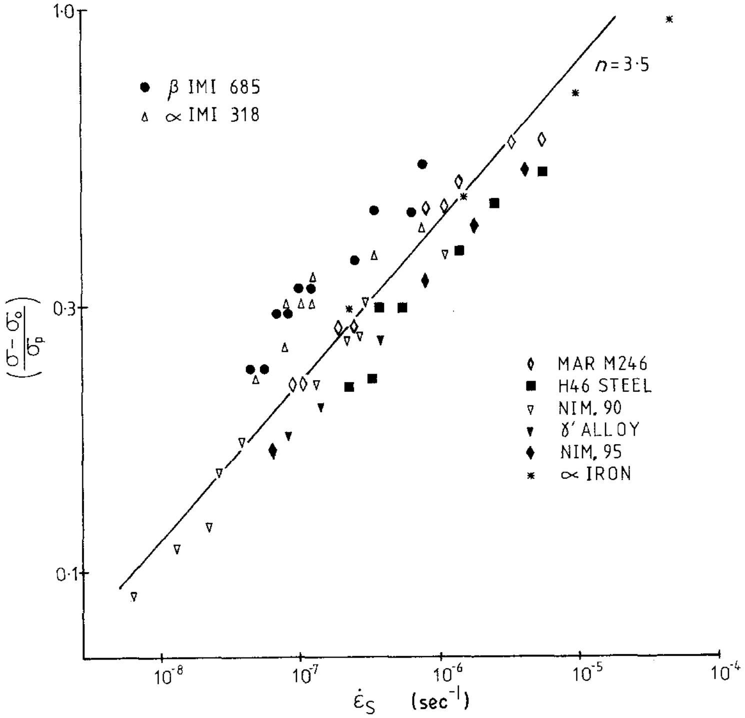

Figure 5 Normalized stress dependence of creep rates in titanium, iron, and nickel based alloys.

same base metal, e.g., IMI318 and IMI685, are determined by σ_0 and the proof or yield stress of the material. Hence the role of alloying elements should be evaluated in terms of their effect on these parameters. For instance with silicon bearing titanium alloys there are a number of possiblities. Thus Paton and Mahoney [2] have demonstrated that silicide precipitation on mobile dislocations can occur and certainly the high exponents, n, and activation energies at constant stress in IMI685 are typical of precipitation hardened materials. The marked effect that precipitates have on σ_0 has been demonstrated for γ' strengthened nickel-base alloys [6]. However, against this silicide precipitation argument must be weighed the evidence that high Q and n were also obtained in α318 where precipitation of silicides would not have played a part. Winstone *et al.* [7] in fact consider that silicide precipitation reduces creep resistance and suggest that silicon acts in solution through a strain-ageing reaction. This is an attractive alternative with a well demonstrated influence on proof stresses. Assadi *et al.* [8] have evaluated in detail the effect that silicon additions can have on strain ageing in a series of titanium–aluminium, zirconium and molybdenum alloys including IMI685. They demonstrate a significant strain ageing response in IMI685 at temperatures in the range 600 to 800 K which encompasses the conditions used in the present investigation. Once again, there is the problem of the similar creep response of α318 but this could be accounted for by their observation that strain ageing does occur in non-silicon bearing alloys, albeit of much reduced magnitude and temperature range, due probably to the presence of oxygen and substitutional elements such as aluminium.

At the present time, the data available are not sufficiently comprehensive to distinguish between these possibilities or to suggest a viable alternative. The value of the work is that it provides an insight into creep of titanium alloys by removing some of the uncertainty about the rate determining mechanism. It demonstrates that what was considered previously to be anomalous behaviour can be rationalized by the introduction of the σ_0 term. Furthermore, it shows that the main influence of alloying elements is on this σ_0 term and on the proof of yield stress of the material. However, it does not and indeed cannot identify the mechanisms by which alloying elements, such as silicon, change σ_0 since this requires more specific experiments involving alloys in which alloy element content is varied in a controlled manner.

5. Conclusions

(a) Secondary creep behaviour of Ti–6Al–4V (IMI318), in three heat treated conditions, and the β processed IMI685 has been evaluated over a range of stresses and temperatures. When behaviour is measured in terms of the applied stress alone the relationship $\dot{\epsilon}_s = A\sigma^n \exp -(Q_c/RT)$ is found to apply. The exponent n has values in the range 4 to 34. The activation energy for creep, Q_c, is greater than relevant activation energies for self or solute atom diffusion. Such behaviour cannot be accounted for by conventional creep models.

(b) Creep behaviour can be rationalized by relating the secondary creep rates to $(\sigma - \sigma_0)$ with σ_0 a measurable friction stress. The appropriate relationship is now $\dot{\epsilon}_s = A^*(\sigma - \sigma_0)^{3.5} \exp -(Q_c/RT)$ with Q_c, the activation energy for creep at constant $(\sigma - \sigma_0)$, similar in value to those for self or solute atom diffusion. Behaviour can now be related to the recovery creep models used to describe creep in pure metals.

(c) Titanium alloys fall on a universal creep line when plotted with f c c and b c c metals and alloys according to the relationship:

$$\dot{\epsilon}_s = B\left(\frac{\sigma - \sigma_0}{\sigma_p}\right)^{3.5}$$

with σ_p the proof or yield stress of the alloy. Alloying elements influence behaviour by modifying σ_0 and the proof stress.

References

1. H. W. ROSENBERG, Proceedings of the 2nd International Ti Conference, Cambridge, Mass., May 1972, "Titanium Science and Technology", Vol. 4 (Plenum Press, New York, London, 1973) p. 2127.
2. N. E. PATON and M. W. MAHONEY, *Met. Trans. A* 7 (1976) 1585.
3. K. R. WILLIAMS and B. WILSHIRE, *Met. Sci. J.* 7 (1973) 176.
4. W. J. EVANS and G. F. HARRISON, *ibid.* **10** (1976) 307.
5. P. W. DAVIES and B. WILSHIRE, *Scripta Metall.* 5 (1971) 475.
6. W. J. EVENS and G. F. HARRISON, *Met. Sci. J.* 13 (1979) 641.
7. M. R. WINSTONE, R. D. RAWLINGS and D. R. F. WEST, *J. Less Common Met.* 39 (1975) 205.
8. A. TAHMASSEBI KHADEM ASSADI, H. M. FLOWER and D. R. F. WEST, *Met. Technol.* 6 (1979) 8.

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