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The temperature and strain rate sensitivity of α-titanium (*)

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TENSILE stress-strain curves are presented for annealed commercially-pure α-titanium tested at five temperatures from 77° to 288 °K at six strain rates covering more than six orders of magnitude from 10^{-3} to about $2500s^{-1}$. The variation with strain rate and temperature of the flow stresses at 0.2% and 5% plastic strain is determined and the results are analysed in terms of the theory of thermally-activated flow. A marked increase in the rate sensitivity of the flow stress is found in tests at all temperatures as the strain rate is increased above about $50s^{-1}$.

Krzywe rozciągania naprężenie-odkształcenie przedstawiono dla wyżarzonego technicznie czystego α-tytanu badanego w pięciu temperaturach od 77° do 288°K i przy sześciu prędkościach odkształcenia, obejmujących zakres ponad sześciu rzędów wielkości od 10^{-3} do około $2500s^{-1}$. Określono zmianę naprężeń plastycznych w funkcji prędkości odkształcenia i temperatury przy 0.2% i 5% odkształcenia plastycznego, a wyniki przeanalizowano w ramach teorii termicznie aktywowanego płynięcia. W doświadczeniach zauważono znaczny wzrost wrażliwości naprężenia płynięcia na prędkość odkształcenia we wszystkich temperaturach, gdy prędkość odkształcenia przewyższała $50s^{-1}$.

Кривые растяжения напряжение-деформация представлены для прокаленного технически чистого α-титана исследуемого в пяти температурах от 77° до 288°K и при шести скоростях деформации, охватывающих интервал свыше шести порядков величин от 10^{-3} до около 2500 $сек^{-1}$. Определены изменения пластических напряжений в функции скорости деформации и температуры при 0,2% и 5% пластической деформации, а результаты анализируются в рамках теории термически активированного течения. В экспериментах обнаружен значительный рост чувствительности напряжения на скорость деформации во всех температурах, когда скорость деформации превышает 50 $сек^{-1}$.

1. Introduction

To a greater or lesser extent all metals and alloys show some dependence of the flow stress on temperature and strain rate. Several models have been proposed to describe this behaviour. In fcc metals, for example, the rate-controlling mechanism is thought to be the thermally-assisted intersection of dislocations [1, 2] while in bcc metals a rate-dependent frictional resistance to dislocation motion arising either from the intrinsic resistance of the lattice [3, 4] (the Peierls stress) or from an interaction between dislocations and interstitial impurity atoms [5, 6] (solid solution hardening) has been postulated. Such models lead to a thermodynamic rate equation of the form

$$\dot{\varepsilon} = \dot{\varepsilon}_0 \exp\left\{\frac{-G(\tau, T)}{kT}\right\}, \tag{1.1}$$

where G is the free energy of activation and $\dot{\varepsilon}_0$ corresponds to the limiting strain rate reached when the applied stress is high enough for the given mechanism to no longer resist the motion of dislocations.

In principle it should be possible to determine the stress and temperature dependence of such parameters as the activation energy, enthalpy and volume using experimental data obtained in constant temperature and strain rate tests and hence to distinguish between

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the various possible rate-controlling mechanisms. In practice, however, although in many investigations results have been obtained over a wide range of temperatures, almost always the strain rates applied have been low and limited to a range of not more than 3 orders of magnitude. At any given temperature, therefore, the range of stresses that can be measured is small. Consequently it becomes difficult to separate the effects of temperature and strain rate on the various rate parameters and hence to assess how closely the assumption of a single rate-controlling mechanism fits the experimental data.

In two relatively recent investigations [7, 8], however, both concerned with metals of bcc structure, tests have been performed over a wide range of both temperature and strain rate. The results, when analysed in terms of the theory of thermally activated flow for a single rate-controlling mechanism assuming a constant pre-exponential factor in the standard rate equation did not support theoretical predictions that the activation energy, enthalpy and volume should be simple functions of stress alone.

Very much less information is available on the behaviour of metals and alloys of cph structure than on those of fcc and bcc structures. In tests at room temperature results have been obtained for α-titanium over a wide range of strain rates under torsional loading by TSAO and CAMPBELL [9] and LAWSON and NICHOLAS [10] and over a more restricted range of strain rates under combined (tension/torsion) loading by RANDALL and CAMPBELL [11]. Several investigations have been carried out by CONRAD and his co-workers [12, 13, 14] at conventional strain rates and temperatures up to 800°K on α-titanium specimens having different interstitial contents. The results were analysed in terms of the theory of thermally-activated flow and it was concluded that the rate-controlling mechanism is the thermally-activated motion of dislocations on first order prism planes past interstitial obstacles on these planes. These experiments suffered, however, from the limitation previously referred to in that the range of strain rates used was small.

The present investigation was undertaken in order to provide data for a metal of cph structure as comprehensive as that previously obtained for metals of bcc structure [7, 8]. Tensile tests have been performed on specimens of annealed, commercially pure, α-titanium at six strain rates covering more than six orders of magnitude, from 10^{-3} to over $2500\,s^{-1}$, at five temperatures over the range 77° to 288°K. The variation with temperature and strain rate of the flow stress at plastic strains of 0.2 and 5.0% has been determined and the results are compared with the behaviour expected on the assumption of a single rate-controlling mechanism.

2. Experimental details

A standard Instron screw-driven loading machine was used for tests at the two lowest rates of strain, 10^{-3} and $10^{-1}\,s^{-1}$, a hydraulically-operated loading machine [15] for tests at the intermediate rate of $45\,s^{-1}$ and a drop-weight impact loading machine [16] for tests at the highest rates, nominally 500, 1500 and $2500\,s^{-1}$. Tests below room temperature were carried out by controlling the flow of liquid nitrogen into a tufnol container surrounding the specimen.

Specimens were machined from annealed, 1/4 inch diameter, commercially-pure titanium rod having the following composition of metallic impurities (ppm):

Table 1

Ag	Al	Ca	Cu	Fe	Mg	Mn	Mo	Si	Sn	V
< 1	70	3	20	200	< 1	3	30	5	100	20

Typical interstitial impurity levels for titanium rod from the same source were (ppm):

Table 2

Carbon	Nitrogen	Oxygen
300	100	1500

The interstitial content expressed in terms of the equivalent atomic fraction of oxygen [12] was approximately 6×10^{-3}. Specimens were stress relieved at 500°C for one hour and furnace cooled under a vacuum of 10^{-4} torr after machining. The mean grain size was about $300 gr/mm^2$.

3. Results

3.1. Stress-strain curves

Engineering stress-strain curves averaged from two or more tests at each temperature at a strain rate of $10^{-3}s^{-1}$ are presented in Fig. 1. The curves shown agree very closely with earlier results for a 99.9% pure titanium, tested at a rate of about $10^{-4}s^{-1}$, obtained by GEIL and CARWILE [17]. Both sets of results show the same type of yield behaviour and both show an increased range of work-hardening and an increased ductility at the lowest temperatures.

Engineering stress-strain curves averaged from three tests at each temperature at a strain rate of $45s^{-1}$ are presented in Fig. 2. The flow stresses are raised and the strains to fracture reduced compared with those obtained at $10^{-3}s^{-1}$. No significant effect of temperature on ductility is observed. At the lowest temperatures a very slight drop in load at the yield point is beginning to appear.

A more marked effect of temperature on the yield behaviour was found at the highest strain rates. Typical sets of oscilloscope traces for tests in the impact loading machine are presented in Figs. 3 and 4. Timing pips interrupt each trace at 2μs intervals. The upper curve in each set records the variation of stress in the specimen during the course of the test and the lower curve is an integrated signal used in calculating the specimen strain. While no yield drop is found in the room temperature test at $800s^{-1}$ (Fig. 3a), a small, but distinct, yield drop is observed at about the same strain rate when the temperature is lowered to 77°K (Fig. 3b). The magnitude of this drop in load is significantly increased when the strain rate is raised to $2400s^{-1}$ at 77°K (Fig. 3c).

More unusual behaviour was found in tests at room temperature at a rate of $1900s^{-1}$ (Fig. 4a) where a stress oscillation, or "double-yield" effect, was observed at the start of

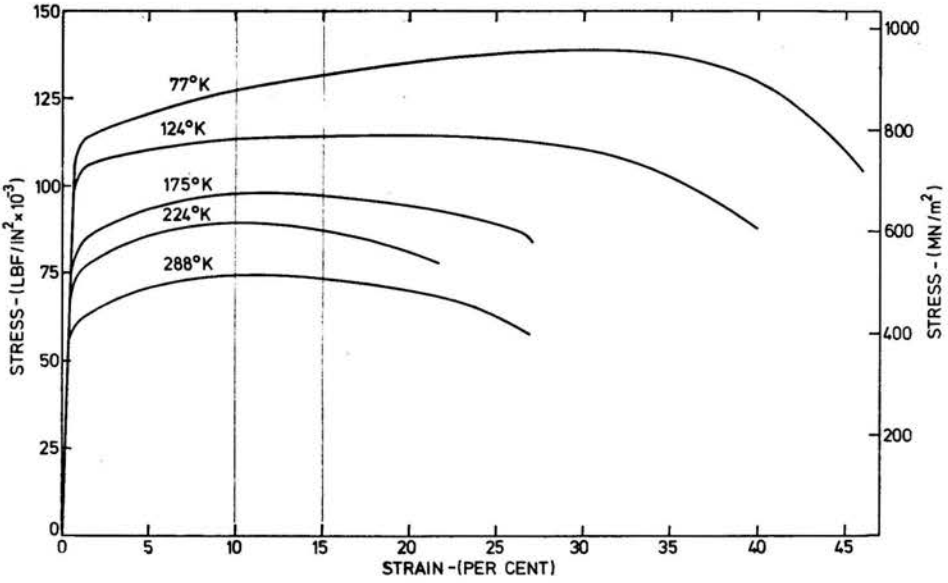

FIG. 1. Effect of temperature on stress-strain curves for specimens tested in the instron loading machine (mean strain rate: 10^{-3} s^{-1}).

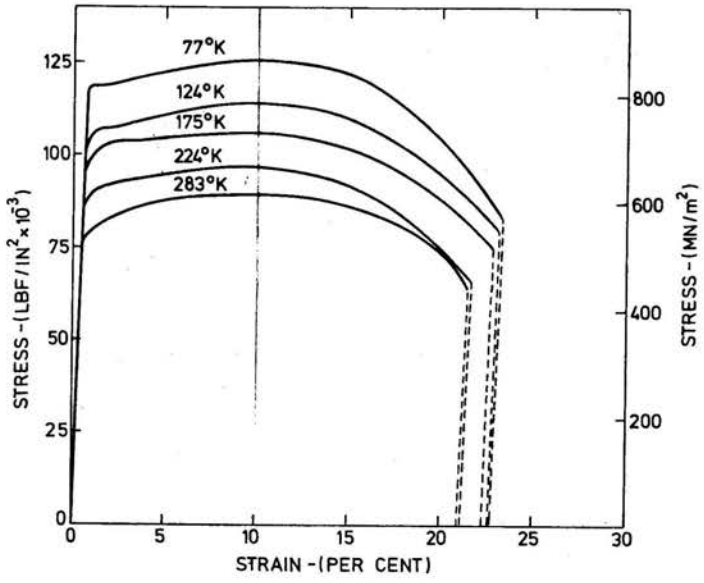

FIG. 2. Effect of temperature on stress-strain curves for specimens tested in the hydraulically operated loading machine (mean strain rate: .44 s^{-1}).

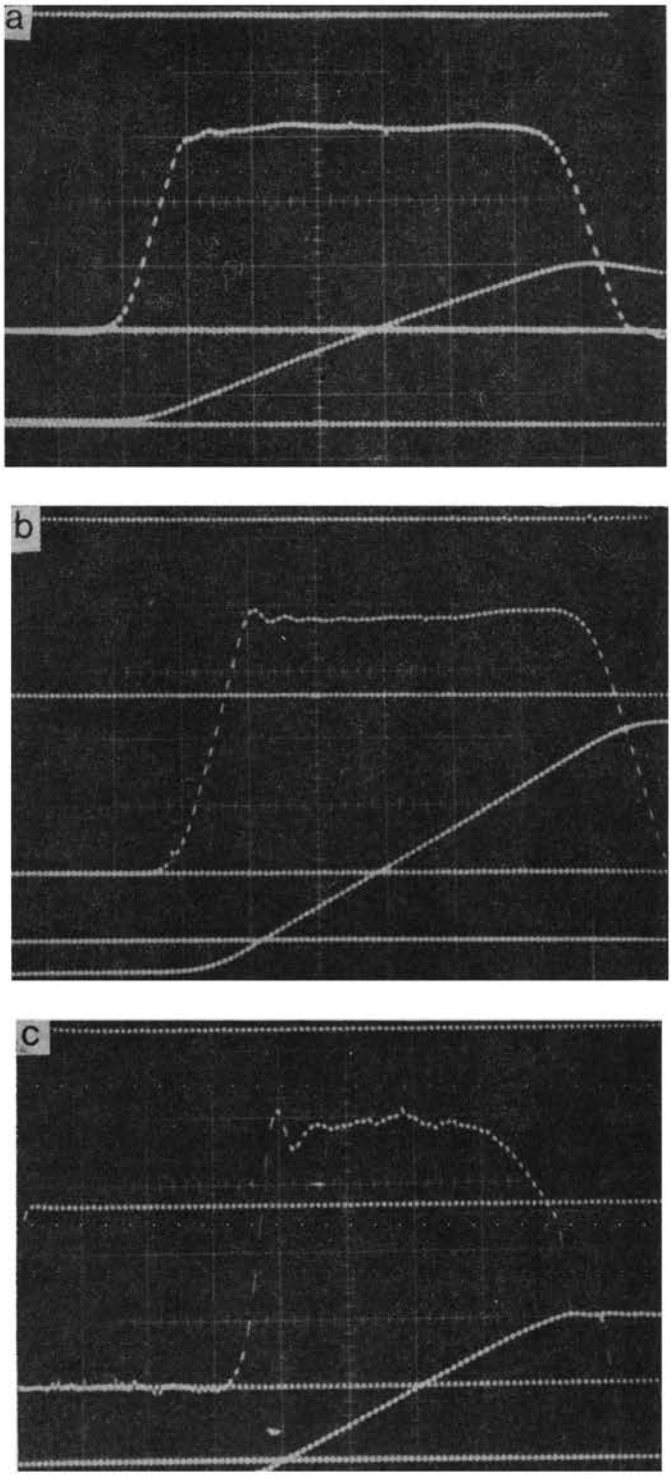

FIG. 3. Oscilloscope traces for impact tests showing increasing yield drop with increasing strain rate and decreasing temperature (Timing pips at 2 μs intervals)
a) 288 °K, 800 s^{-1}, b) 77 °K, 500 s^{-1}, c) 77 °K, 2400 s^{-1}.

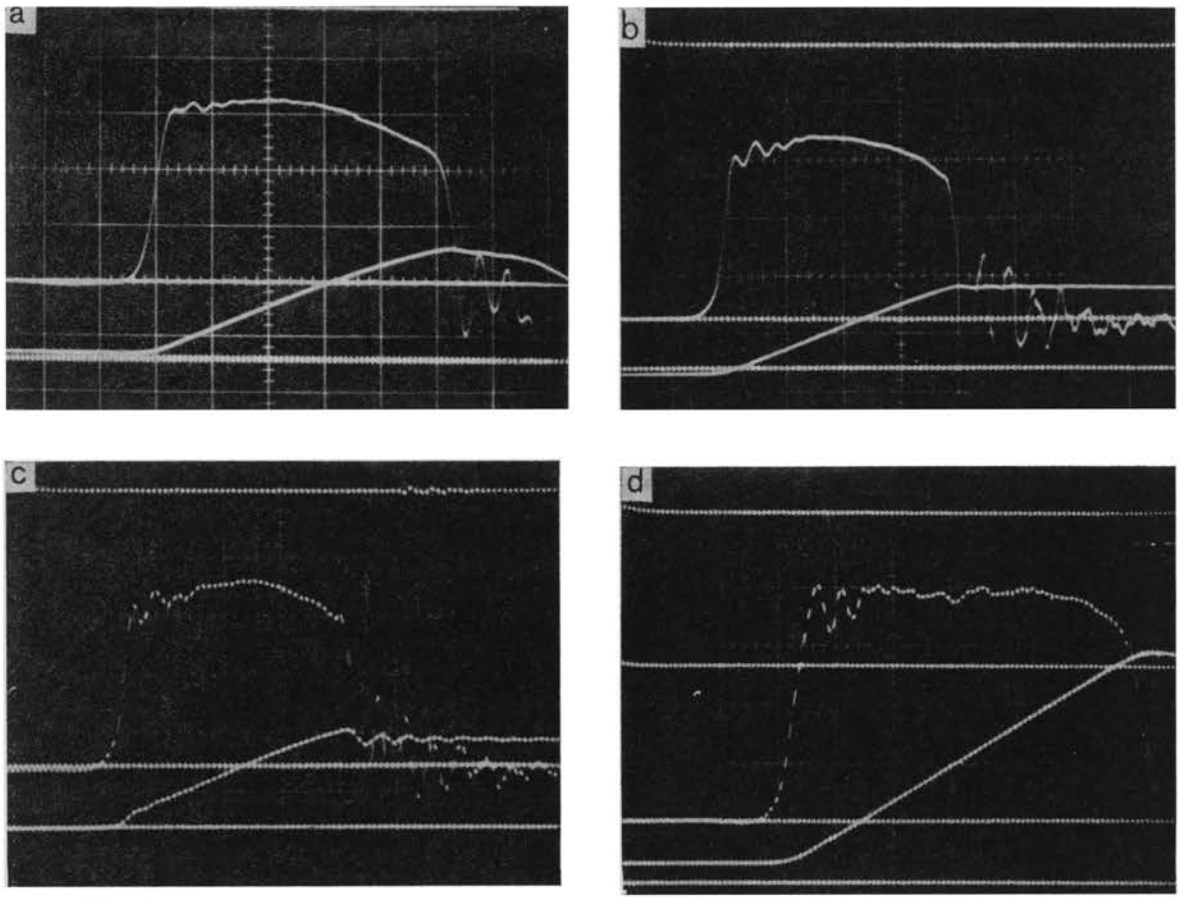

FIG. 4. Oscilloscope traces for impact tests showing double yield effect (Timing pips at 2 μs intervals)
a) 288 °K, 1900 s^{-1}, b) 288 °K, 2500 s^{-1}, c) 288 °K, 2500 s^{-1}, d) 77 °K, 1375 s^{-1}.

plastic flow. This observation might have been ignored but for the fact that on increasing the strain rate to 2500 s^{-1} in tests at room temperature (Figs. 4b and 4c) the "double-yield" effect was found to be both more marked and highly repeatable. It appears to be an intermediate stage in the transition from uniform yielding to a fully developed yield drop. Although no explanation has been found for this behaviour, in the most extreme cases, of which Fig. 4d for a test at 77°K and 1375 s^{-1} is an example, the stress traces observed closely resemble those obtained previously in tests on pure iron single crystals [18] where the load drops were thought to be due to mechanical twinning. Metallographic examination of the specimens tested in this investigation showed extensive twinning to have occurred in tests at 77°K, particularly in specimens deformed at the highest strain rates. At room temperature, however, twins were present in very much smaller numbers in specimens tested at the highest rates and could rarely be detected after tests at the lowest rates.

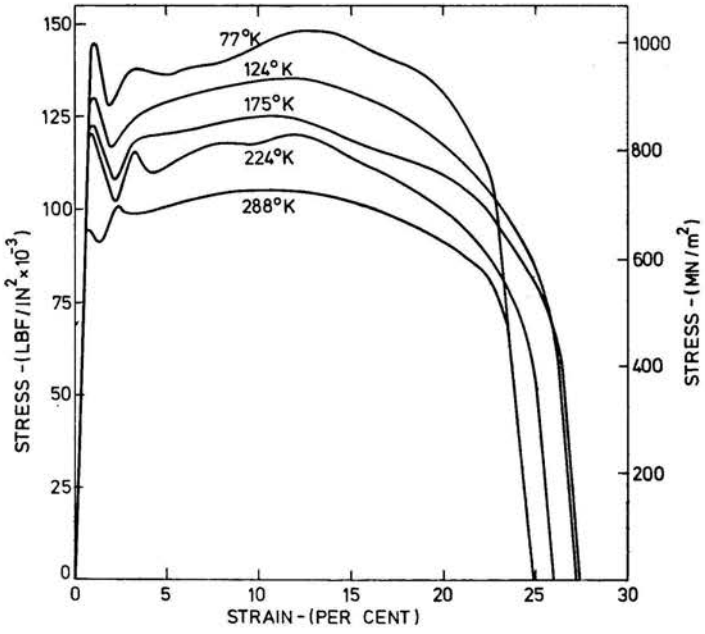

FIG. 5. Effect of temperature on stress-strain curves for specimens tested in the impact loading machine (mean strain rate: 2475 s^{-1}).

Averaged engineering stress-strain curves for tests at each temperature at the highest rate of strain are shown in Fig. 5. As the temperature is lowered the "double-yield" effect at room temperature is replaced by a more marked single yield drop at the lowest temperatures. The effect of strain rate on the stress-strain behaviour at constant temperature is shown in Fig. 6 for tests at room temperature. Curve *B* in this figure relates to tests at a nominal strain rate of 500 s^{-1}. At this rate the specimen is unloaded in the impact test before it has strained sufficiently to break. Curve *B*, therefore, does not go through to fracture. Similar results obtained at 77°K are shown in Fig. 7.

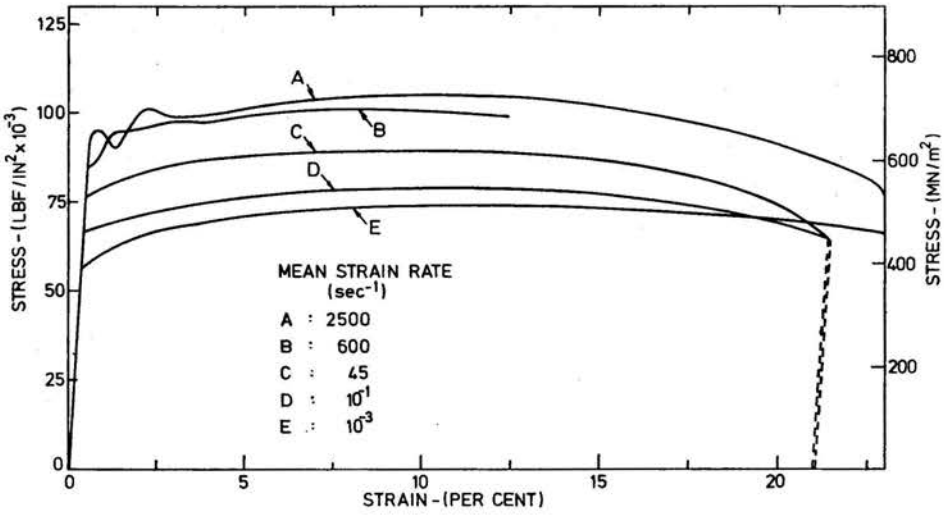

FIG. 6. Effect of strain rate on stress-strain curves for specimens tested at 288°K.

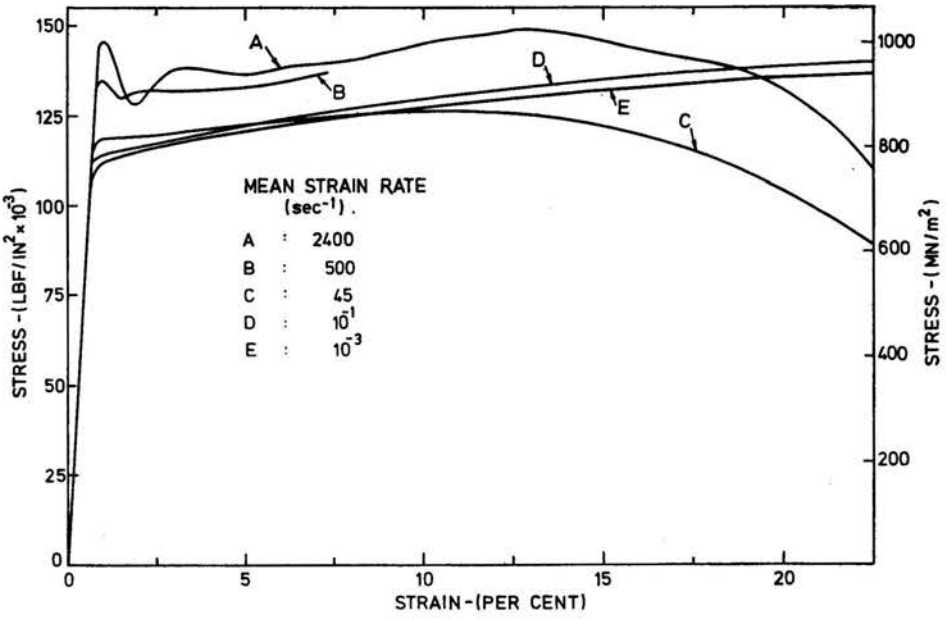

FIG. 7. Effect of strain rate on stress-strain curves for specimens tested at 77°K.

3.2. Strain-rate sensitivity curves

The semi-logarithmic rate sensitivity λ, defined by the expression

$$\lambda = \{\partial\sigma/\partial \log \dot{\varepsilon}_p\}_{\varepsilon_p, T}, \tag{3.1}$$

may be determined from the slopes of the curves of Fig. 8 which show the variation of the 0.2% proof stress with the logarithm of the mean plastic strain rate at each test tem-

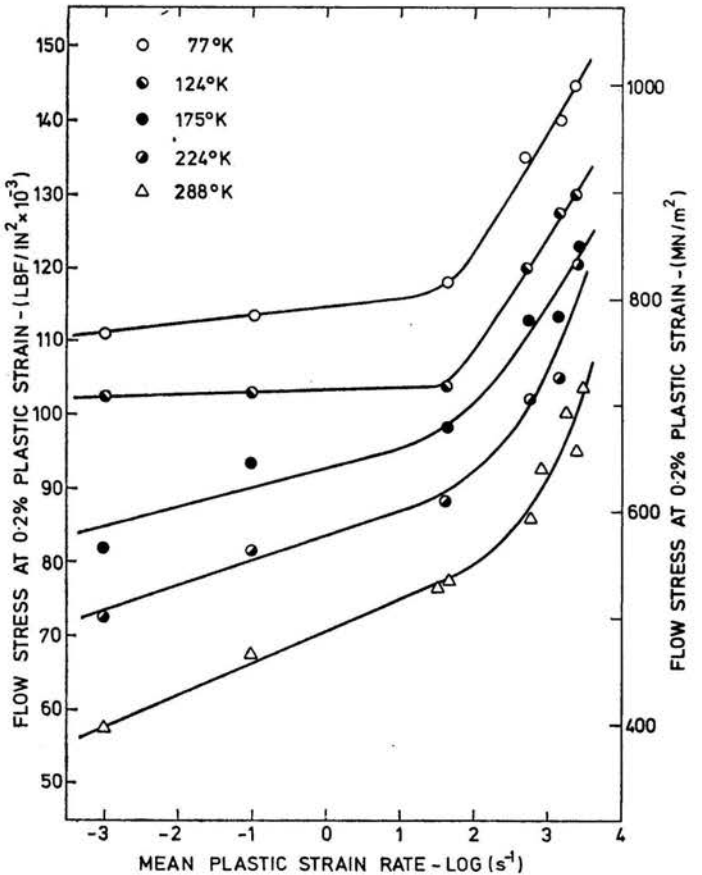

FIG. 8. Rate sensitivity of the flow stress at 0.2% plastic strain.

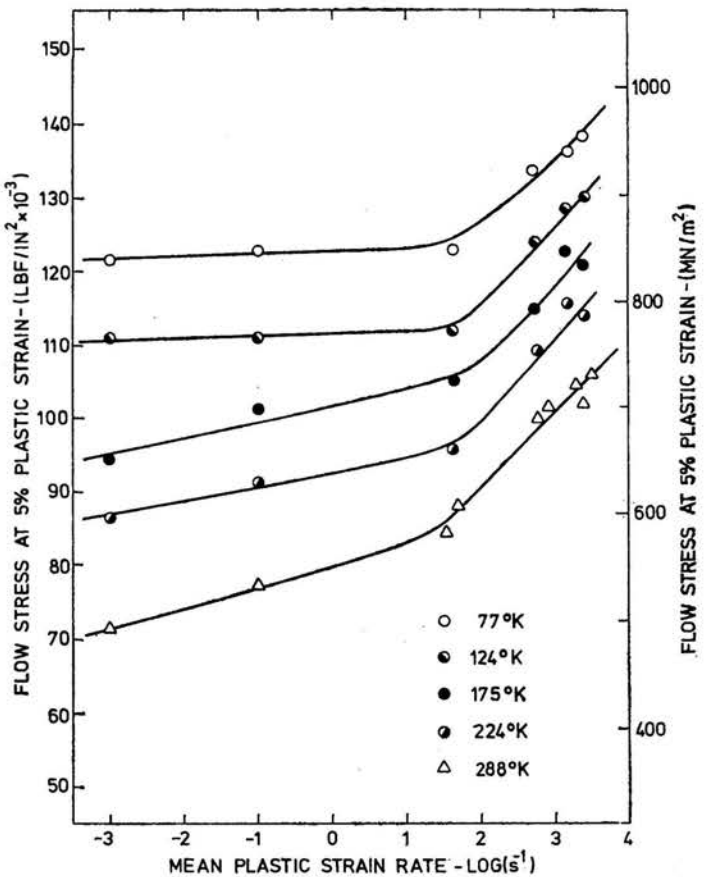

FIG. 9. Rate sensitivity of flow stress at 5% plastic strain.

perature. Two distinct regions are apparent; at strain rates below about $50s^{-1}$ λ is found to be relatively independent of strain rate and to decrease with temperature from about $30MN/m^2$ at room temperature to about $8MN/m^2$ at 77°K, while at strain rates above about $50s^{-1}$ much larger values of λ are obtained, of the order of $110MN/m^2$. At room temperature λ appears to increase continuously with strain rate in the high strain rate region. At lower temperatures the evidence is less clear and the results might well be described by two constant values of λ, one for low rates and the other for high rates.

In the two previous investigations where commercially-pure α-titanium has been tested at room temperature at strain rates above $50s^{-1}$ conflicting results have been obtained. LAWSON and NICHOLAS [10] found a linear dependance of the flow stress at 10% shear strain on the logarithm of the shear strain rate, corresponding to a value for λ of about $50MN/m^2$, for all strain rates from 10^{-4} to $5000s^{-1}$, there being no region of increased rate sensitivity. TSAO and CAMPBELL [9] however, found λ for the flow stress at 10% shear strain to increase from about $40MN/m^2$ at low rates to about $120MN/m^2$ at high rates, the transition occurring at an equivalent tensile strain rate of about $400s^{-1}$. Although this is rather higher that the transition strain rate in the present tests there were insufficient results in the earlier investigation for a detailed comparison to be made.

In tests at the highest strain-rates the 0.2% proof stress lies in the region of non-uniform yield associated with the yield drop or the "double-yield" effect. To check whether this is the cause of the increased rate sensitivity shown in Fig. 8 at strain rates above $50s^{-1}$, rate sensitivity curves are shown in Fig. 9 for the flow stress at 5% plastic strain. This lies well outside the region of non-uniform yield in all tests. The results of Fig. 9 clearly confirm the existence of two regions of rate sensitivity and give values of λ of $20MN/m^2$ at room temperature falling to about $2MN/m^2$ at 77°K in the low rate sensitivity region and about $75MN/m^2$ in the high rate sensitivity region. These are all a little lower than the corresponding values for the flow stress at 0.2% plastic strain.

3.3. Temperature sensitivity curves

The temperature sensitivity of the 0.2% proof stress at constant strain rate is shown in Fig. 10, results being presented for five of the six applied strain rates. Apart from one result, that at 224°K at the highest strain rate, the experimental points all lie reasonably closely on smooth curves. Also shown in the figure are results obtained by CONRAD and JONES [14] at a strain rate of $10^{-4}s^{-1}$ for two grades of α-titanium having a finer grain size than the material used in the present investigation and interstitial contents, expressed in terms of the equivalent atomic fraction of oxygen, of 10×10^{-3} and 2×10^{-3}, compared with 6×10^{-3} for the present material.

Temperature sensitivity curves for the flow stress at 5% plastic strain, presented in Fig. 11, again show a behaviour very similar to that for the flow stress at 0.2% plastic strain. Results at strain rates of $50s^{-1}$ and below converge rapidly with fall of temperature, while those at impact rates converge only very slowly.

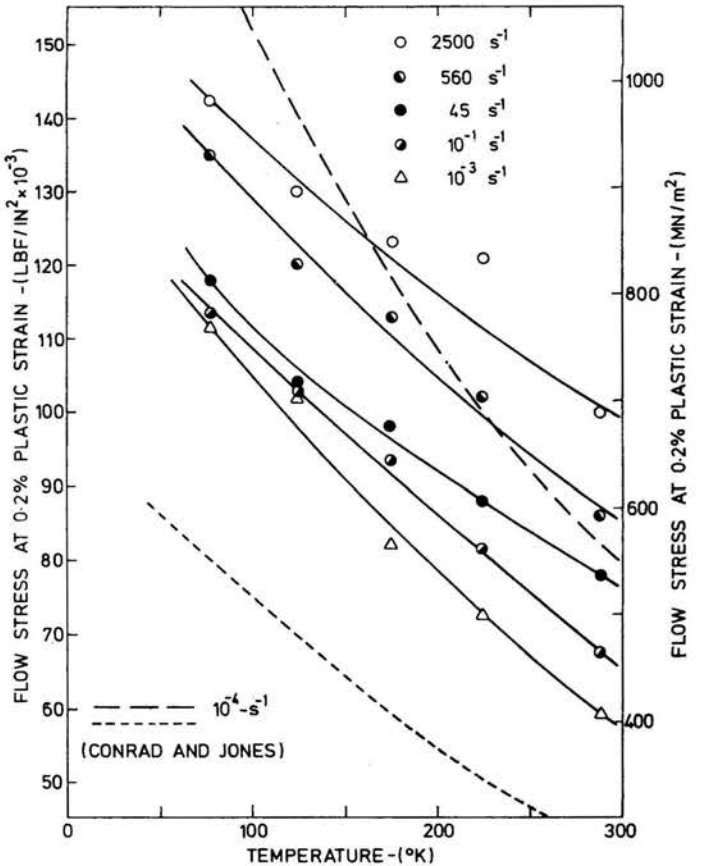

FIG. 10. Temperature sensitivity of the flow stress at 0.2% plastic strain.

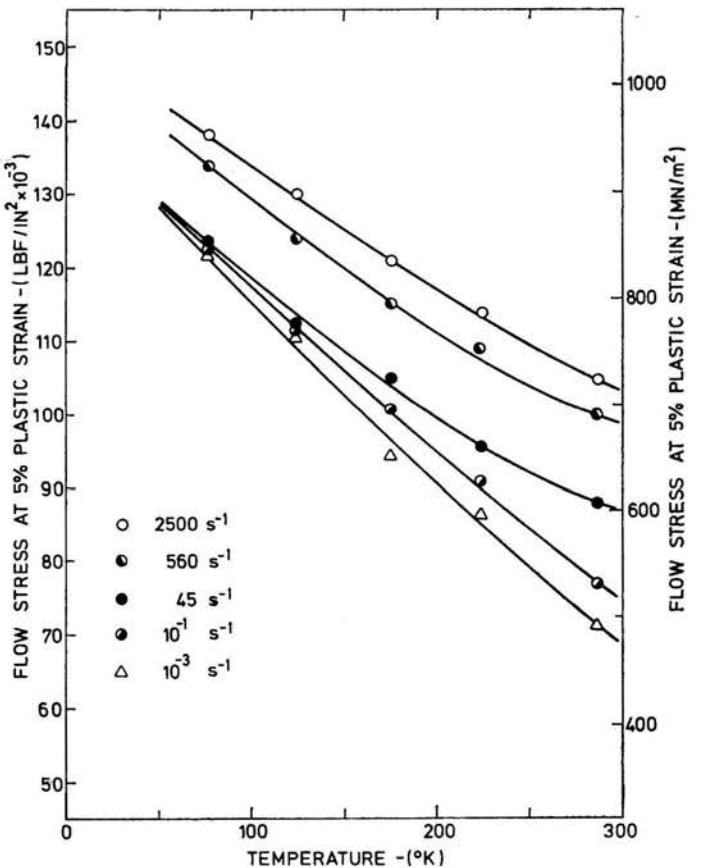

FIG. 11. Temperature sensitivity of the flow stress at 5% plastic strain.

4. Discussion

Having presented the experimental data it is required to determine how closely they fit the assumptions of the theory of thermally activated flow, in particular that of a single rate-controlling mechanism. It is usual when analysing experimental results in this way to divide the applied stress into two parts, an athermal stress σ_A, representing the long range internal stress field, and an effective, or local, stress σ^* which, with the aid of thermal fluctuations, enables dislocations to surmount the short-range barriers; it is this latter, thermal, component of the applied stress which is used in the thermal activation rate analysis. Since the athermal stress is normally considered to be independent of strain rate it may be taken as the stress in the temperature and strain rate insensitive region. Evidence presented by MONTEIRO *et al* [19], however, suggests that for α-titanium the behaviour is much more complex. Unfortunately the present data do not extend to high enough temperatures or low enough strain rates for it to be possible to study the behaviour in the "athermal" region so no estimate of the athermal stress could be made and the total applied stress has had to be used in the analysis.

A more serious limitation to the experimental data arises, however, if, as seems to be indicated by the strain rate sensitivity curves of Figs. 8 and 9, two regions of behaviour are covered in which the rate controlling mechanisms are different. It is then unlikely that sufficient results will be available in each region taken separately for any firm conclusions to be drawn.

The thermodynamic relation $\Delta H = \Delta G - T(\partial \Delta G/\partial T)$, when applied to the standard rate equation $\dot{\varepsilon} = \dot{\varepsilon}_0 \exp(-\Delta G/kT)$, gives for the activation enthalpy the expression

$$\Delta H = -\left\{\frac{\partial \ln(\dot{\varepsilon}/\dot{\varepsilon}_0)}{\partial(1/kT)}\right\}_\sigma, \tag{4.1}$$

while differentiation of the standard rate equation gives

$$V^* = 2kT \Big/ \left\{\frac{\partial \sigma}{\partial \ln(\dot{\varepsilon}/\dot{\varepsilon}_0)}\right\}_T, \tag{4.2}$$

where the activation volume V^* is defined by the thermodynamic relation $V^* = -(\partial \Delta G/\partial \tau)_T$.

Of the several theories that have been proposed to describe the short range barriers to flow that due to FLEISCHER [5, 6], in which the frictional stress results from the interaction between dislocations and interstitial impurities, is thought to best describe the experimental results for α-titanium obtained by CONRAD and his co-workers [12, 13, 14]. In this theory each of the activation parameters, ΔG, ΔH and V^*, is a unique function of stress. A major difficulty in comparing the experimental data with the predictions of the theory arises from an uncertainty regarding the behaviour of the pre-exponential factor $\dot{\varepsilon}_0$. It is usual to assume that $\dot{\varepsilon}_0$ is independent of temperature and strain rate and has a value in the range from $10^6 s^{-1}$ for impure polycrystals to $10^{11} s^{-1}$ for pure single crystals and pure polycrystals [20]. It then follows, by the Eq. (4.1), that, for ΔH to be a unique function of stress alone, a plot of $\ln(\dot{\varepsilon})$ against $(1/kT)$ at constant stress should give a set of straight lines of slope $-\Delta H$ (or $-\Delta G$) with a common intercept on the strain rate axis at $\dot{\varepsilon} = \dot{\varepsilon}_0$.

As is shown in Fig. 12, however, when the present results are plotted in this way two distinct regions of behaviour are found, zone *A* for strain rates less than about $50 s^{-1}$, and zone *B*, for strain rates greater than about $50 s^{-1}$.

In view of the limited number of results it is not clear how closely the experimental points in each zone at each stress level follow a linear relationship. If a linear relationship is assumed, however, the corresponding intercepts on the strain rate axis are found to vary over the ranges 10^{11} to $10^{17} s^{-1}$ for results in zone *A* and 10^{4} to $10^{7} s^{-1}$ for results in zone *B*. These compare with previously quoted values for $\dot{\varepsilon}_0$ in α-titanium of $3 \times 10^{8} s^{-1}$ [14] and $1.5 \times 10^{10} s^{-1}$ [21]. Further they imply that $\dot{\varepsilon}_0$ is strongly dependent on the applied stress. Results similar to those in zone *A* of Fig. 12, taken from a previous investigation

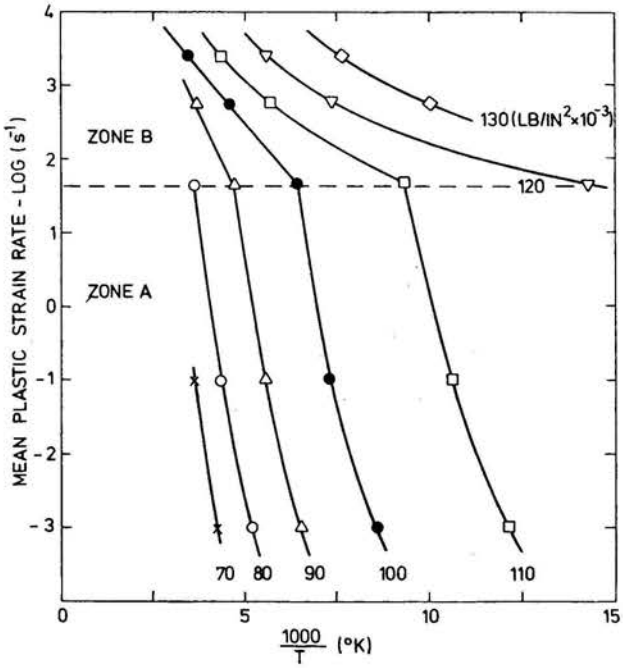

FIG. 12. Variation of the logarithm of strain rate with reciprocal temperature at 0.2% plastic strain.

[8] into the mechanical behaviour of niobium, are shown in Fig. 13. In view of the close similarity between the two sets of results it seems likely that, had the present tests been extended to lower strain rates, i.e. as close to the athermal region as in the earlier investigation, the small curvature found in zone *A* of Fig. 12 would have been confirmed. However, to explain even this small curvature on the basis of a temperature dependent $\dot{\varepsilon}_0$ would require $\dot{\varepsilon}_0$ to vary from 10^{7} to $10^{24} s^{-1}$. Since such a variation in $\dot{\varepsilon}_0$ would seem to be unreasonably large the results in zone *A* may perhaps be rationalised most simply by assuming $\dot{\varepsilon}_0$ to be independent of temperature and ΔH a function of both stress and temperature. Because of the change apparent in the rate-controlling mechanism at strain rates above $50 s^{-1}$ it is not possible to determine whether the results in zone *A* would have extrapolated to a common intercept on the strain rate axis. The results in zone *B*, however, might well be thought

to do so, the strain rate at the common intercept being not greatly different from that previously quoted for $\dot{\varepsilon}_0$ in α-titanium [14]. This would imply that, for zone B if not for zone A, $\dot{\varepsilon}_0$ may be taken to be independent of stress as well as of temperature.

On the basis of this assumption the Eq. (4.2) may be used to determine the activation volume from the slopes of the strain rate sensitivity curves of Fig. 8 at any given temperature and strain rate. The stress dependence of the activation volume calculated in this way

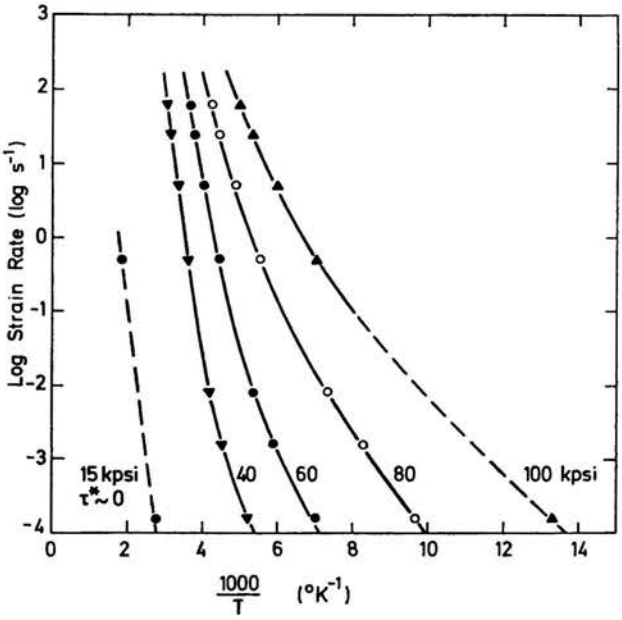

FIG. 13. Variation of the logarithm of the strain rate with reciprocal temperature for niobium at the lower yield stress (after BRIGGS and CAMPBELL, ref. [8]).

at three different temperatures is shown in Fig. 14. Experimental data obtained at strain rates below $50s^{-1}$ give activation volumes which are independent of stress but decrease slightly with decreasing temperature, from 28 to 23 (V^*/b^3), where b is the Burger's vector, as the temperature falls from 288° to 77°K. A similar behaviour was found previously for molybdenum [8] but to a more marked extent, the activation volume being reduced by almost half, from 12.8 to 6.6 (V^*/b^3), as the temperature fell from 292° to 150°K. It is common to find a rapid increase in V^* as $\sigma^* \to 0$, i.e. as the applied stress approaches the athermal level [14, 21]. Lack of experimental data at stresses close to the athermal level may explain why such behaviour is not found here. When results obtained at strain rates above $50s^{-1}$ are used in the calculation the activation volume is found to decrease very rapidly with increasing stress to a value of the order of 2 to 4 (V^*/b^3) and then to become again relatively independent of stress while remaining slightly dependent on temperature.

The activation enthalpy may be calculated from the Eq. (4.1), i.e. from the slopes of the curves of Fig. 12, again on the assumption that $\dot{\epsilon}_0$ is independent of stress and temperature.

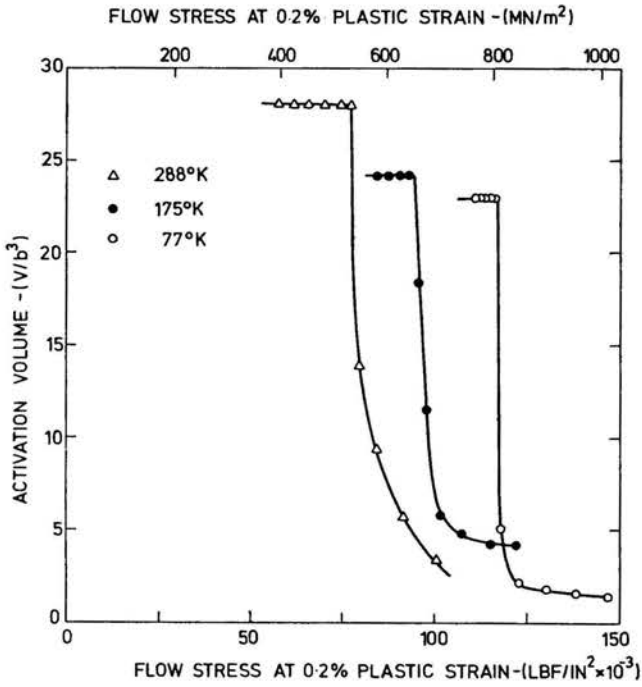

FIG. 14. Variation of activation volume with flow stress at 0.2% plastic strain.

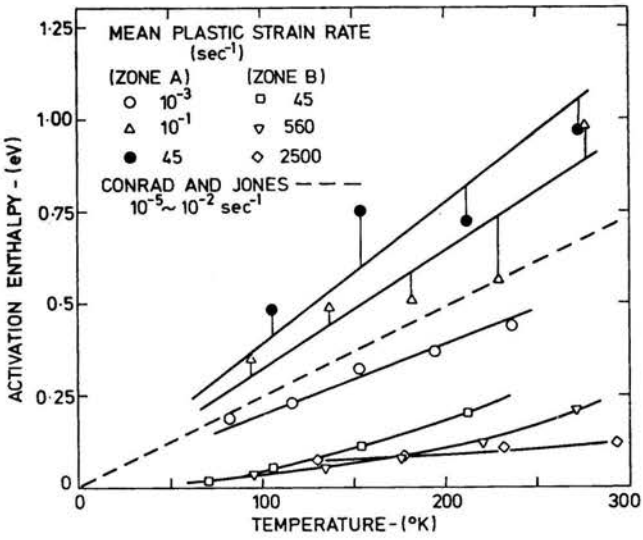

FIG. 15. Effect of temperature on the activation enthalpy at 0.2% flow stress.

Results obtained in this way are compared in Fig. 15 with those of CONRAD and JONES [14] for strain rate change experiments at rates around $10^{-4}s^{-1}$. Results from zone *A* give an approximately linear relationship between the activation enthalpy and temperature, similar to that found by CONRAD and JONES except that here the enthalpy increases

more rapidly with temperature the higher the strain rate. A smaller temperature dependence is shown in zone B where significantly lower values of enthalpy are obtained.

Since ΔH may be given by the expression

$$\Delta H = -TV^*(\partial\sigma/\partial T)_{\dot{\varepsilon}/\dot{\varepsilon}_0} \tag{4.3}$$

and since it is found experimentally that $(\partial\sigma/\partial T)_{\dot{\varepsilon}}$ increases with falling temperature, a linear relationship between ΔH and T supports the observation that V^* falls with falling temperature.

An alternative presentation of the results of Fig. 15 to show the effect of temperature on the activation enthalpy at constant stress is given in Fig. 16. It is clear that, for tests

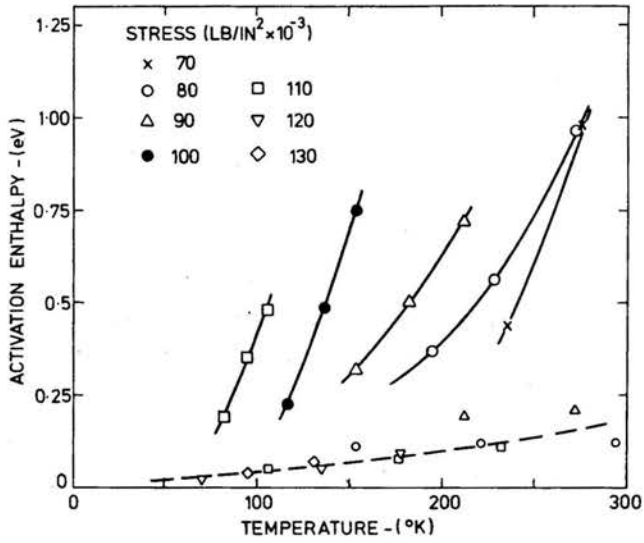

FIG. 16. Effect of temperature on the activation enthalpy at 0.2% flow stress.

at strain rates below $50s^{-1}$, ΔH at constant stress is strongly temperature dependent. A much smaller temperature dependence is shown, however, when ΔH is determined from the results of tests at higher strain rates.

In several previous investigations at strain rates in excess of $10^3 s^{-1}$ [7, 22] a very rapid increase in the semi-logarithmic rate sensitivity with increasing strain rate, corresponding to a linear relationship between stress and strain rate, has been found. Such behaviour has been explained in terms of a viscous-damping process as the rate-controlling mechanism [7, 23, 24]. Although the present results also show a rapid increase in the semi-logarithmic rate sensitivity, the transition strain rate, about $50s^{-1}$, is some two orders of magnitude below that found in the earlier investigations. Nor in the present tests are there sufficient experimental results at the highest strain rates to determine how closely a linear relationship between stress and strain rate is followed. An extension of the work to higher strain rates is required therefore, before any firm conclusions concerning the rate-controlling mechanism in this region may be drawn.

5. Conclusions

Tensile stress-strain curves for α-titanium have been obtained at strain rates from 10^{-3} to over $2500 s^{-1}$ at temperatures from 77 to 288°K. From these curves the temperature and strain rate dependence of the flow stresses at 0.2% and 5.0% plastic strain have been determined. For tests at all temperatures a marked increase in the semi-logarithmic rate sensitivity is found at strain rates above about $50 s^{-1}$. The results are analysed in terms of the theory of thermally activated flow on the basis of a single rate-controlling mechanism in each region of rate sensitivity. Because of the limited number of results when the two regions are considered separately only tentative conclusions may be drawn. Assuming, however, that the pre-exponential factor in the standard rate equation is independent of stress and temperature, the activation enthalpy in the region of low rate sensitivity is found to be a function of both stress and temperature while the activation volume is found to be independent of stress but to decrease slightly with temperature. These results conflict with theoretical predictions that the activation parameters, ΔG, ΔH and V^*, should all be unique functions of stress. Similar discrepancies between theory and experiment have been reported previously [7, 8] in tests on bcc metals.

In the high strain rate region much lower values are obtained for both the activation enthalpy and the activation volume. Their stress and temperature dependence, however, has not been clearly determined. An extension of the experimental data to higher strain rates is required before firm conclusions regarding the rate-controlling mechanism in this region may be drawn.

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