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THE EFFECT OF TEMPERATURE, STRAIN RATE AND STRUCTURE ON THE FLOW STRESS OF AN IRON-2% MANGANESE ALLOY

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ABSTRACT

The temperature, strain rate and structure dependence of the flow stress in a polycrystalline iron-2% manganese alloy was investigated between 77° and 370°K. It was possible to identify the low temperature data from 77° to 160°K with Dorn and Rajnak's theory of the Peierls mechanism of plastic deformation. Above 160°K one or more thermally activated mechanisms must be operative, however, these were not identified in this investigation.

I. INTRODUCTION

The mechanistic basis for the rapid decrease in the flow stress of BCC metals with an increase in temperature over the low temperature range is yet being debated. Various dislocation mechanisms have been suggested in the past but the agreement between the various proposed theories and the experimental results have not been definitive. In the present paper, the authors will demonstrate that the plastic behavior of polycrystalline iron containing 2 wt.% manganese plus interstitials agrees excellently with predictions based on the Peierls mechanism from 77° to 160°K whereas from 160° to about 370°K other as yet unidentified thermally activated mechanisms take over. Undoubtedly, the overlap of these mechanisms with the Peierls mechanism contributed to the difficulties of rationalizing the low temperature behavior of iron and its alloys.

II. EXPERIMENTAL TECHNIQUES AND RESULTS

Although extensive data are now available on the effect of temperature and strain rate on the flow stress of various BCC metals and alloys, it was, nevertheless, deemed necessary for the objectives of the present study to obtain highly accurate data of the type most pertinent for checking the theories on a series of well-documented work hardened states. Iron containing 2 wt.% manganese was used in this investigation in order to obtain a lower ductile to brittle transition temperature than that for iron.[1] The additional elements present were 0.004%C, 0.05%O, 0.004%S, 0.003%P, 0.006%N and 0.001%Si.

The as received 3/8" x 3/4" bars were cold rolled to 0.100" thickness, recrystallized under argon for 30 min at 1073°K, further cold rolled to 0.063" thickness, and finally recrystallized under argon for 30 min at 1000°K. This treatment gave a stable reproducible equiaxed grain structure having an average grain diameter of about 90 microns. Prior to the final recrystallization treatment, the sheet was machined into tensile specimens 1/4" wide having a 1.625" long gage section. All specimens were tested in controlled temperature baths on an Instron Testing Machine. For convenience in the theoretical analyses the data will be given in terms of the shear stress for flow, τ, the shear strain rate, $\dot{\gamma}$, and the shear strain, γ. A series of three standard strain-hardened states were obtained by prestraining specimens at 300°K and at a shear strain rate of 7.68×10^{-5}/sec to three stress levels of 6.89×10^{8}, 10.3×10^{8} and 14.5×10^{8} dynes/cm^2 respectively. Immediately following this prestrain the temperature bath was changed

and each specimen was pulled in tension at either $\dot{\gamma} = 7.68 \times 10^{-5}$ or 3.06×10^{-3}/sec. In terms of the accuracy that was achieved, the values obtained did not deviate more than $\pm 0.1 \times 10^{8}$ dynes/cm^2 and $\pm$ 1°K from the reported values of stress and temperature and not more than $\pm$ 5% in the reported values of the strain rate. The tensile stresses were converted to shear stresses using a factor of 1/2. The shear stress for the initiation of flow at the test temperature was determined by taking a $\gamma = 6.16 \times 10^{-3}$ offset from the modulus line at which point good accuracy could be achieved in determining $\dot{\gamma}$ and the associated flow stress τ, with only a negligible increase in the prestrained state. The experimental data, recorded in Fig. 1, obtained by this technique were shown to be highly reproducible.

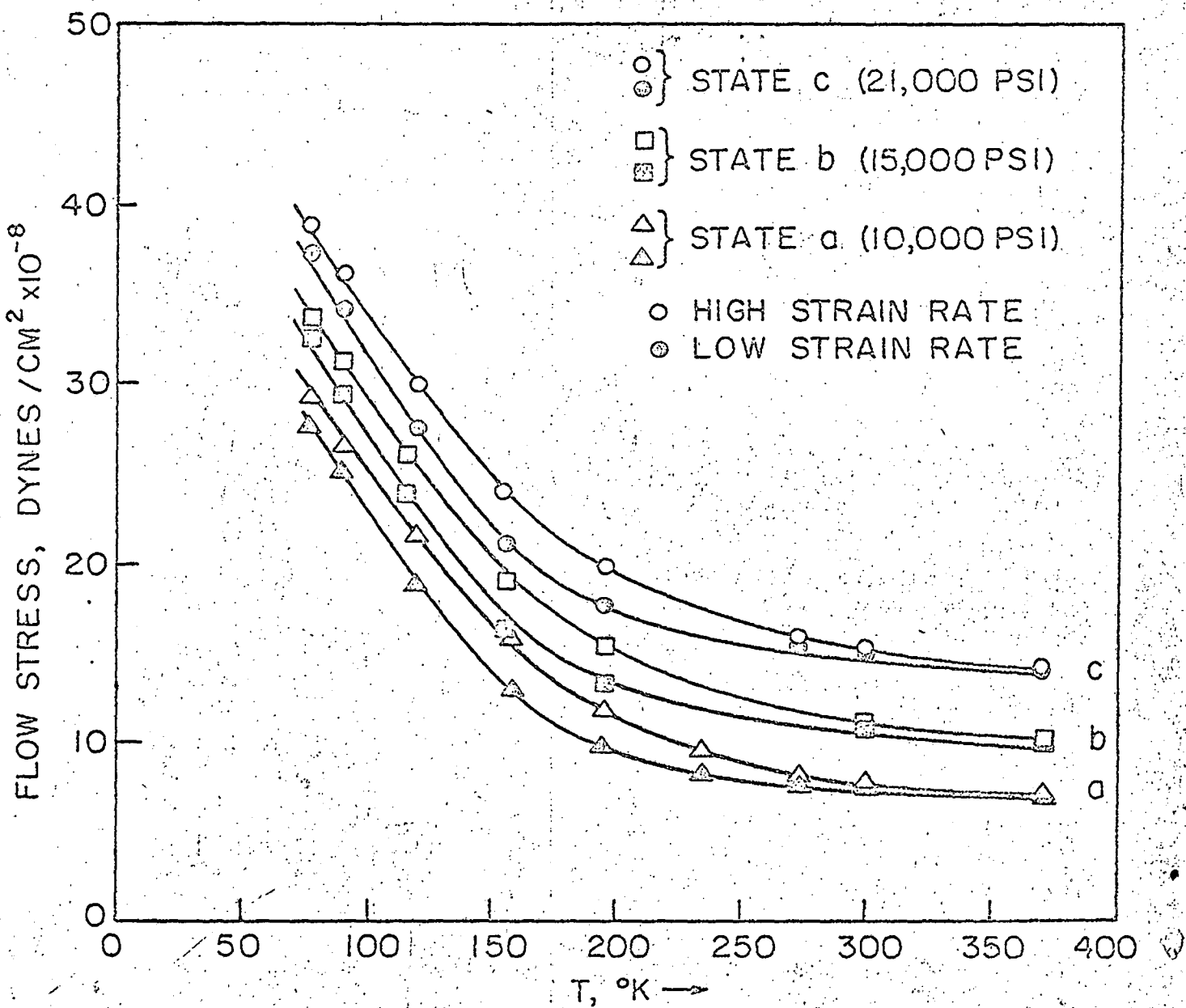

FIG. 1 STRESS vs. TEMPERATURE FOR DIFFERENT PRESTRAINED STATES AND STRAIN RATES.

III. DISCUSSION

The increase in the flow stress for each of the various strain-hardened states with decreasing temperature and increasing strain-rates, as shown in Fig. 1, firmly establishes the fact that the deformation of polycrystalline iron containing 2 wt.% manganese plus interstitials is controlled by thermally activated dislocation mechanisms below about 370°K. From the data, it appears that the deformation becomes athermal at about 370°K; however, this could not be shown conclusively because recovery and strain aging effects take place above 370°K. The general trends of these data are in nominal agreement with those that have been reported by other investigators on the deformation of other BCC metals and alloys.[2-6]

It is now generally admitted that each low-temperature thermally activated mechanism can be characterized by the generalized relationship

$$\dot{\gamma} = \dot{\gamma}_o e^{-\frac{U}{kT}} \qquad (1)$$

where $\dot{\gamma}$ is the shear strain rate, $\dot{\gamma}_o$ is a frequency factor, U is the energy that must be supplied by a thermal fluctuation, and kT has its usual meaning of the Boltzmann constant times the absolute temperature. For deformations at low temperatures, the activation energy U is a function of the thermal stress

$$\tau^* = \tau - \tau_A \qquad (2)$$

where τ is the applied stress and τ_A is the stress that is required to overcome the athermal dislocation interactions; furthermore both U and τ_A decrease with an increase in temperature in proportion to the decrease in the shear modulus of elasticity with an increase in

temperature. Whereas, $\dot{\gamma}_o$ is always dependent on the significant structural details of each mechanism, for some mechanisms such as intersection of dislocations, interactions with solute atoms, interactions with clusters of atoms, U also depends on the substructural details. However for other mechanisms, such as the Peierls process, U is a physically determined function of τ^* and is therefore independent of the structure.

Unfortunately the theoretically predicted τ vs T relationship is frequently about the same for several different mechanisms. A more critical distinction between mechanisms is obtained from the activation volume, v, defined by

$$v = kTB = kT\left(\frac{\partial \ln\dot{\gamma}}{\partial \tau^*}\right)_T = -\left(\frac{\partial U}{\partial \tau^*}\right) \simeq kT\left(\frac{\Delta \ln\dot{\gamma}}{\Delta \tau^*}\right)_T \qquad (3)$$

The activation volumes deduced from the data in Fig. 1 in this way are given in Fig. 2 in terms of $\tau - \tau_s$ where τ_s is the flow stress at 300°K for the prestrained condition. The low activation volumes of about $30b^3$ (where b is the Burgers vector) for the higher stress levels disqualifies the intersection mechanism since its activation volume should have the much higher value of $v \simeq Lb^2$ where L is the mean dislocation spacing; furthermore, the insensitivity of the activation volume to the prestrained work-hardened condition, which should decrease with straining if intersection were rate controlling, also serves to disqualify intersection as the operative mechanism over the entire range of $\tau - \tau_s$.

Although Stein, Low and Seybolt[7] have suggested that the low temperature thermally activated mechanism in single crystals of iron

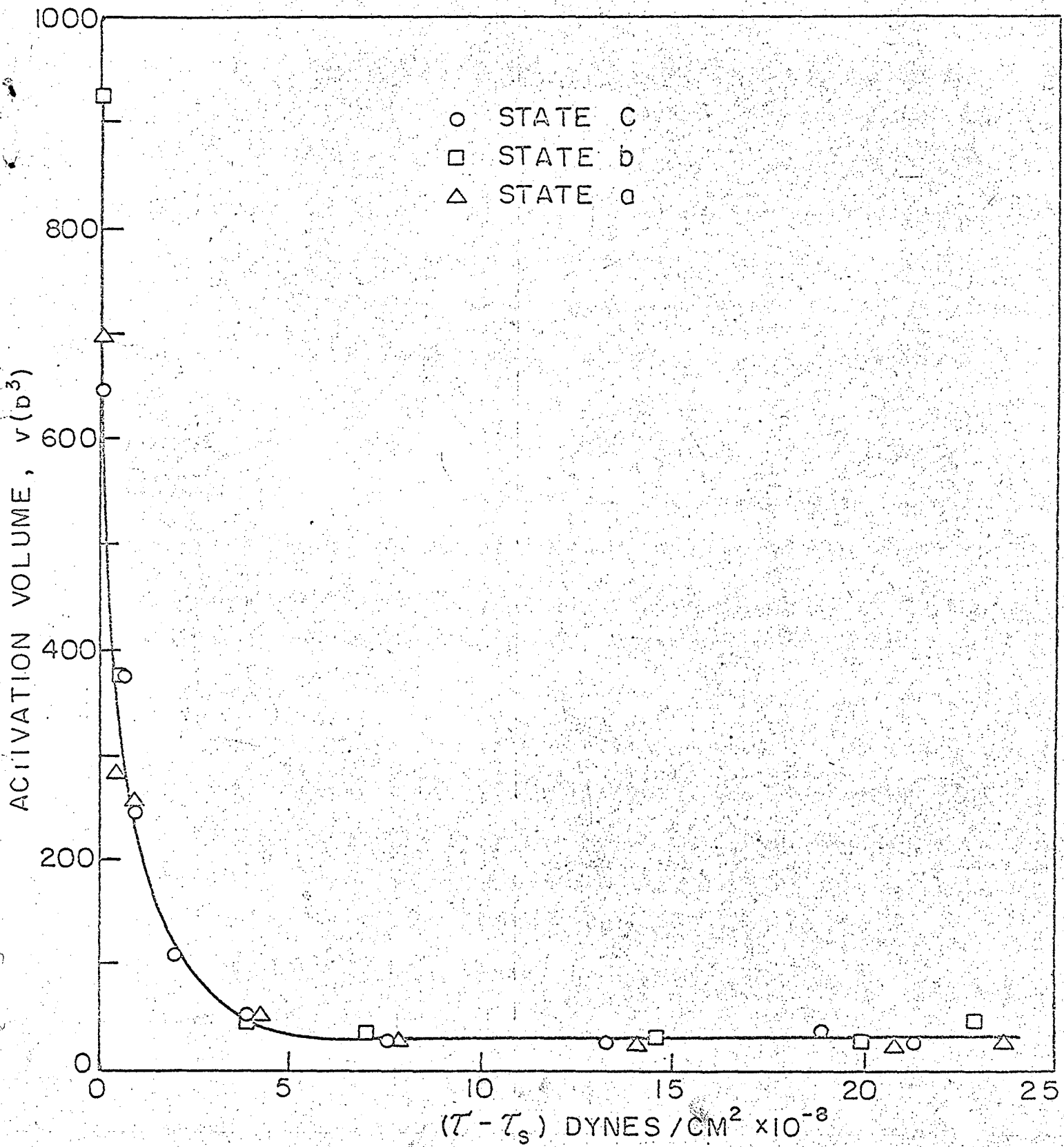

FIG. 2 ACTIVATION VOLUME vs. THE FLOW STRESS EXCESS OVER THE PRESTRESS.

arises from interactions of interstitial solute atoms, and Mordike and Haasen[6] attribute it to intersection of dislocations with clusters of interstitials, many investigators among them Conrad and Hayes,[2] Heslop and Petch[8] and Basinski and Christian[9] have suggested that the Peierls process is operative. The major strength of this view point centers about the small values of the experimentally determined activation volumes and their insensitivity to deformation and impurity content. For example, employing a much higher purity iron than used in this investigation Arsenault[10] obtained albeit by creep tests, about the same trend in activation volume with $\tau - \tau_s$ as that recorded here. The major argument against acceptance of a Peierls mechanism concerns the unacceptedly high activation volumes that are obtained at the lower values of $\tau - \tau_s$. This strongly suggests that if the Peierls mechanism is controlling over the low temperature range, it is overlapped and superseded by some more difficult mechanism at temperatures above about 160°K. In order to present the case for this hypothesis we will assume its validity and check all factors bearing on the Peierls mechanism in the range of its assumed validity, following which we will return to discuss alternate possibilities. For this purpose, we will adopt the Peierls theory recently suggested by Dorn and Rajnak[11] because it has received excellent experimental verification over the entire range of τ^*, including the low stress range so critical to this evaluation in two cases, namely that for prismatic slip of Ag_2Al[12] and for $\{123\}\langle 11\bar{1}\rangle$ slip in the BCC-CsCl type lattice of AgMg.[13] These results were uncontaminated by the presence of another mechanism.

For the Peierls process

$$\dot{\gamma}_o = \rho ab\nu \frac{L}{w} \tag{4}$$

where ρ is the density of dislocations, a is the distance between adjacent Peierls valleys, ν is the Debye frequency, L is the length of dislocation advanced a distance a by the nucleation of a single pair of kinks and w is about the critical separation of the pair of embryonic kinks for the saddle-point free energy for nucleation of the pair. Since w is only mildly dependent on the stress,[11] it can be taken as substantially a constant in the logarithmic expression of Eq. 4. The thermal flow stress τ^* for the Peierls process is given by[11]

$$\tau^*/\tau_p = f_1 \left(\frac{U_n}{2U_k}\right) \simeq f_1 \left(\frac{T}{T_c}\right) \tag{5}$$

where τ_p is the Peierls stress, $U = U_n$ is the activation energy for nucleating a pair of kinks, U_k is the energy of an isolated kink, T is the test temperature, and T_c is the critical temperature where $\tau^* = 0$ and is defined by

$$\dot{\gamma} = \dot{\gamma}_o e^{-\frac{2U_k}{kT_c}} \tag{6}$$

and consequently depends on $\dot{\gamma}$. Both U_k and τ_p change with temperature in the same way as the shear modulus but this effect is small and has been neglected in these calculations. The theoretically established universal τ^*/τ_p vs T/T_c curve for the Peierls mechanism where the Peierls hill is assumed to be sinusoidal is shown by the solid line in Fig. 3.

As shown in the figure $T_c \equiv T \{\text{at } \tau^* \equiv \tau - \tau_A = 0\}$ and $\tau_p \equiv \tau^* \{\text{at } T \equiv 0\} \equiv (\tau - \tau_A) \{\text{at } T = 0\}$. Since a second mechanism is

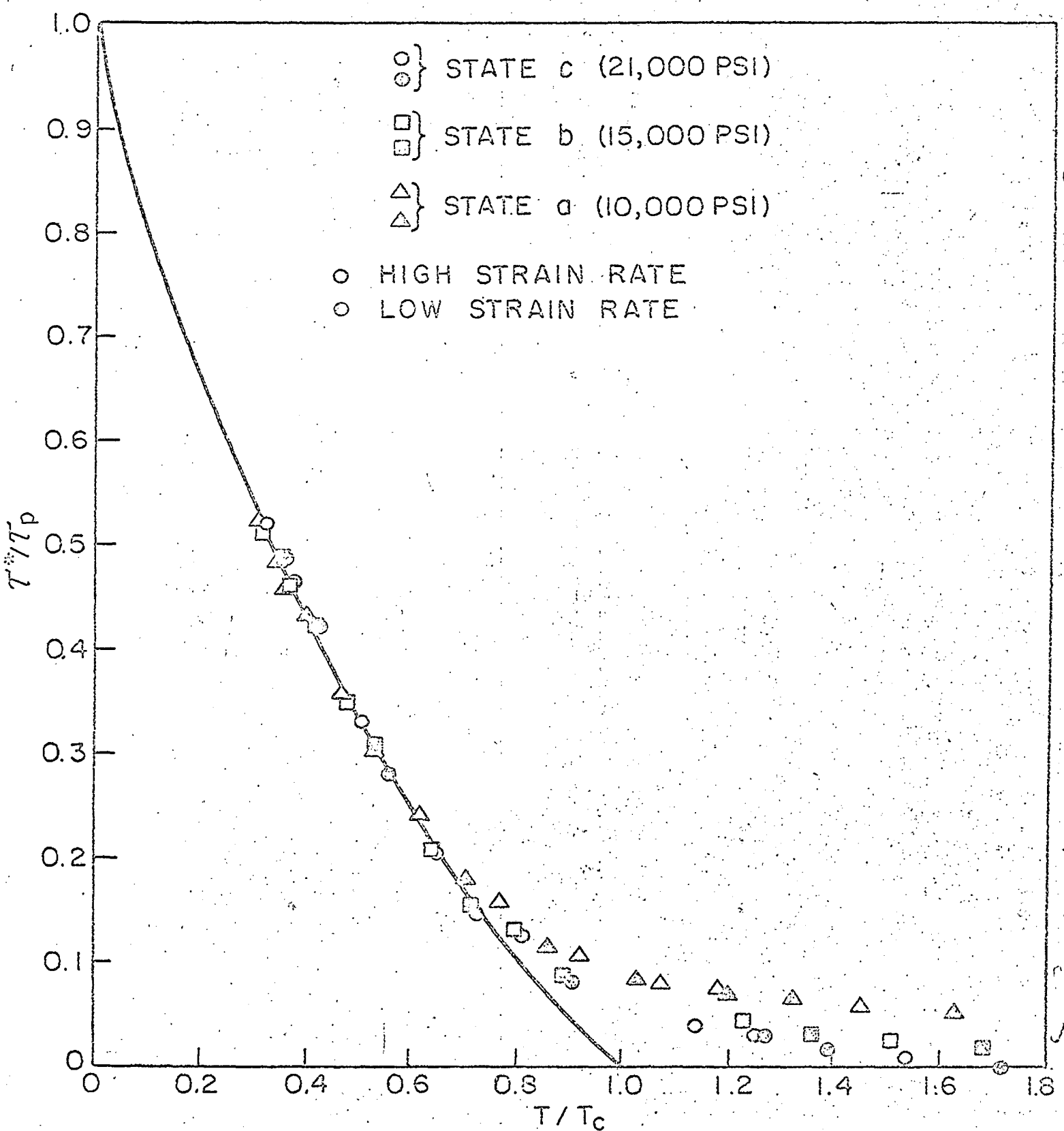

FIG. 3 TEMPERATURE DEPENDENT COMPONENT OF THE FLOW STRESS vs. TEMPERATURE IN DIMENSIONLESS COORDINATES.

presumed to intervene at the higher test temperatures, it is not possible to determine T_c directly from the experimental data, and since brittle fracturing was obtained below 77°K is was not easily possible to obtain $\tau - \tau_A$ at T = 0. For this reason the alternate proceedure of fitting the experimental results to the data was adopted. As shown in Fig. 1, τ_A for state c is closely equal to $\tau_A = \tau \{370°K\} = 13.8 \times 10^8$ dynes/cm^2. Using this value and correlating the lower temperature data of Fig. 1 with the theoretical curve of Fig. 3, τ_p was estimated to be 48.2×10^8 dynes/cm^2. The data for the remaining states were also correlated using this same value of τ_p. The significant data of the correlation are given in Table I.

Using these values all of the experimental datum points were plotted on Fig. 3, revealing excellent agreement of the data with theoretical expectations from T/T_c equals about 0.30 to about 0.75. Above 0.75, however, the experimental results deviate considerably from the expectations based on the Peierls process. The values of U_k in the last column were obtained from

$$\dot{\gamma}_1/\dot{\gamma}_2 = \frac{e^{-\frac{2U_k}{kT_{c_1}}}}{e^{-\frac{2U_k}{kT_{c_2}}}} \tag{7}$$

as required by Eq. 6 and are accurate to about 15%. ρL was estimated from Eqs. 4 and 6, w was obtained by assuming $v \simeq wb^2$ and using a value of $30b^3$ for v (see Fig. 2) and ν was taken to be about 1.9×10^{13} sec^{-1}. The kink energies of about 0.3 ev agree very well for the different states as they should when the Peierls mechanism applies; and

TABLE I

τ_A and T_c Assuming Peierls Process

For $\tau_p = 48.2 \times 10^8$ dynes/cm^2

State	$\tau_s \times 10^8$ dynes/cm^2	$\dot{\gamma}$ sec^{-1}	T_c °K	$\tau_A \times 10^{-8}$ dynes/cm^2	$\rho L \times 10^{-8}$ cm^{-1}	$U_k \times 10^{11}$ ergs
a	6.89	5.12×10^{-5}	227	4.15	3.8	5.29
		2.04×10^{-3}	255			
b	10.3	5.12×10^{-5}	220	8.96	77.5	5.52
		2.04×10^{-3}	245			
c	14.5	5.12×10^{-5}	215	13.8	25.4	5.26
		2.04×10^{-3}	240			

ρL is about the right order of magnitude and it increases with an increase in the dislocation density as suggested by the Peierls mechanism. The Peierls theory requires that

$$\frac{2\pi U_k}{a\Gamma_o} = 5.67\sqrt{\frac{\tau_p ab}{\pi\Gamma_o}} \tag{8}$$

where Γ_o is the line energy. Introducing the values obtained for U_k and τ_p reveals that the line energy is 7.15×10^{-4} ergs/cm which is in agreement with commonly accepted value of $\Gamma_o \simeq 0.5Gb^2$.[16] Finally, we compare the experimentally determined activation volumes with the theoretically established relationship[11]

$$\left(\frac{\tau_p}{2U_k}\right) v = f_2(\tau^*/\tau_p) \tag{9}$$

where the theoretical function is given by the solid line in Fig. 4 and the points refer to the correlated experimental data. The agreement with the dictates of the Peierls mechanism is again excellent for τ^*/τ_p of 0.1 to 0.5 but experimental data deviate seriously from the theoretical expectations for $\tau^*/\tau_p < 0.1$, i.e. for temperatures somewhat above about 160°K. The excellent agreement between the experimental data from 77° to slightly above 160°K with all of the intimate details of the Peierls process vindicates the thought sponsored so effectively by Conrad and also promoted by many other investigators that the low temperature deformation mechanism of iron and also other BCC metals is controlled by the Peierls mechanism. However, the lack of correlation between experiment and the Peierls theory from slightly above 160°K to 370°K implies that at least one other thermally activated mechanism must operate in this range.

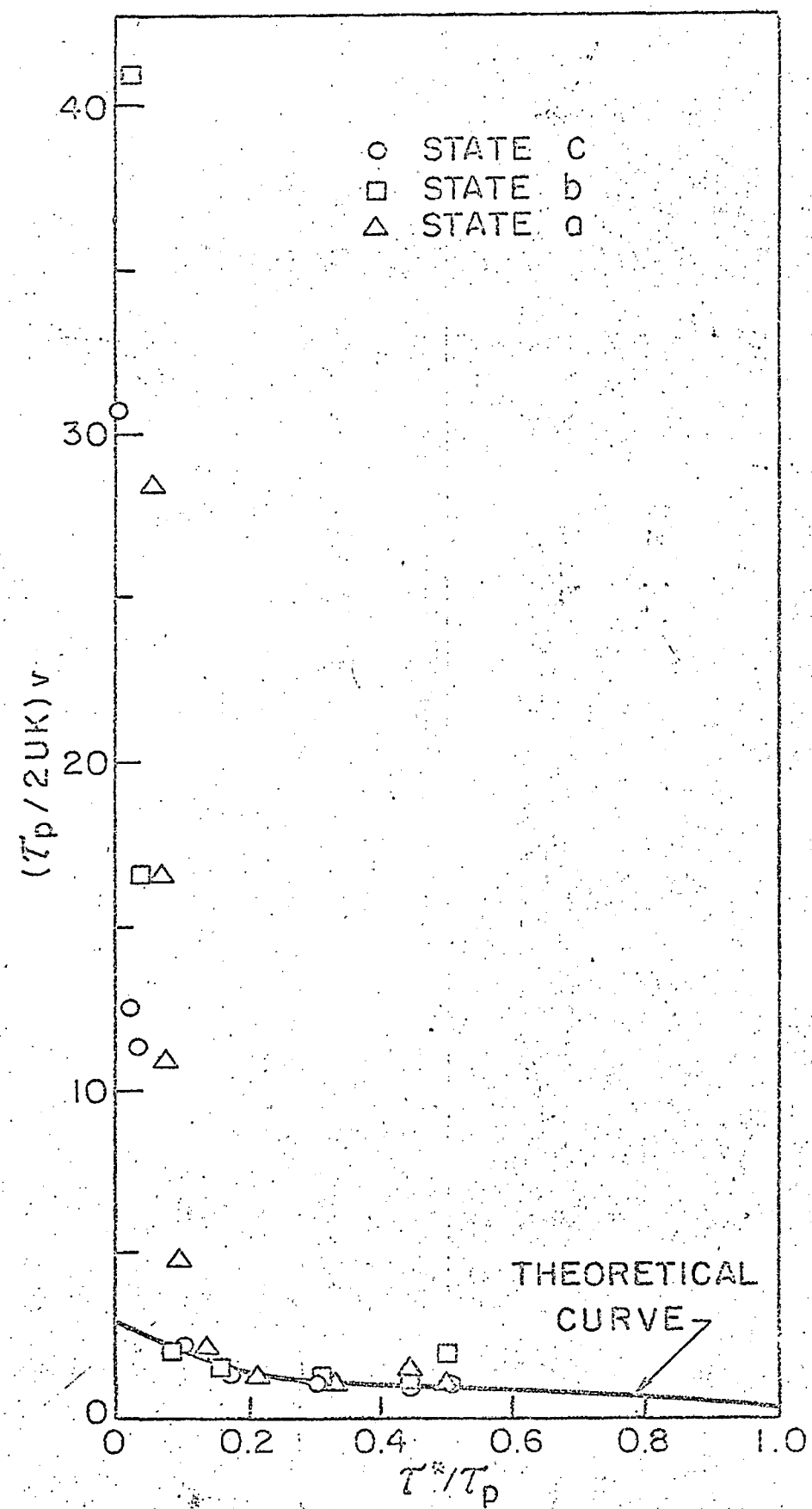

FIG. 4 ACTIVATION VOLUME vs. TEMPERATURE DEPENDENT COMPONENT OF THE FLOW STRESS IN DIMENSIONLESS COORDINATES.

The interstitial atom interaction suggested by Stein, Low and Seybolt,[7] the clustered atom reaction suggested by Mordike and Haasen,[6] the motion of jogged screw dislocations suggested by Schoeck,[14] and the Snoeck effect suggested by Schoeck[15] or combinations of these are the serious contenders for the operative mechanism over the intermediate temperature range. The resolution of this issue appears to be most difficult and will require additional investigations and analyses particularly involving studies of ultra high purity metals. Assumptions were made so that it is conceivable that such a process might display a mild temperature dependence. At present due to the meagre data available, it is rather difficult to draw definite conclusions about this mechanism.

CONCLUSIONS

The temperature, strain rate and structure dependence of the flow stress in iron-2% manganese alloy has been investigated. The following conclusions have been drawn:

1. There are at least two thermally activated mechanisms operating between 77 to 370°K, which can be resolved by means of activation volume calculations.
2. Below 160°K the large temperature dependence of the flow stress can be explained on the basis of Dorn and Rajnak's theory of the Peierls mechanism of plastic deformation.
3. The calculated values of the activation energy for the process and the line energy of a dislocation are in good agreement with the theory.
4. From 160° to 370°K, other unidentified mechanisms having much larger activation volumes appear to be operative.

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FIGURE CAPTIONS

Fig. 1 Stress vs temperature for different prestrained states and strain rates

Fig. 2 Activation volume vs the flow stress excess over the prestress

Fig. 3 Temperature dependent component of the flow stress vs temperature in dimensionless coordinates

Fig. 4 Activation volume vs temperature dependent component of the flow stress in dimensionless coordinates

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