High strain-rate response of commercially pure vanadium

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Abstract

To understand the plastic flow behavior of commercially pure vanadium under high strain rates, uniaxial compression tests of cylindrical samples are performed using UCSD’s enhanced Hopkinson technique. True strains exceeding 50% are achieved in these tests, over a range of temperature from 77 to 800 K at the strain rates of 2500 and 8000 s⁻¹. The microstructure of the deformed and undeformed samples is observed by an optical microscope. The initial microstructure (the initial dislocation density, the grain size and its distribution) is found to have a strong effect on the yield stress and the initial stages of the flow stress. This effect becomes more significant with decreasing temperature. Deformation twins are observed. Their density is seen to increase with decreasing temperature. Adiabatic shearbands occur in vanadium at low temperatures. Finally, an experimentally based micromechanical model is developed for the dynamic response of this material. The model predictions are compared with the results of other high strain-rate tests which have not been used in the evaluation of the model parameters, and good agreement between the theoretical predictions and experimental results is obtained. In addition, the results of a series of low strain-rate tests are presented and briefly discussed. © 2000 Elsevier Science Ltd. All rights reserved.

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1. Introduction

Vanadium, V, is a body-centered-cubic (bcc) crystal, belonging to Group V-A of the Periodic Table. Since its discovery by Manuel Del Rio in 1801 (Kinzel, 1950; Elvers and Hawkins, 1996), it has found broad usage in many fields. For example, it is an important element in steel and Ti al-

loy, and it has applications in the nuclear industry, in medicine, and in superconductor research. Like other refractory metals, the mechanical properties of vanadium are strongly dependent on its purity and hence on its production method. Its flow stress is a function of temperature, strain rate, and its microstructure, e.g., the dislocation density and distribution. For these reasons, efforts have been made to understand the influence of strain rate, temperature, and the microstructure on the mechanical properties of this material. Compared to other refractory metals, some of the remarkable properties of vanadium are:

1. it has more plasticity and its brittle–ductile transition temperature is rather low;

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2. its stacking-fault formation energy seems to be the lowest among the pure refractory metals with a bcc lattice (Zubets et al., 1978);
3. its slip geometry and dislocation structure exhibit distinguishing features, as compared with those of the other bcc refractory metals, because the dislocations in vanadium are more prone to dissociation (Noskova and Dolgopolov, 1978).

In general, at low strain rates (usually \(<1 \text{s}^{-1}\)), a three-stage workhardening response is commonly observed in single-crystal vanadium, as in other crystals. When a single crystal (99.98% purity), oriented along the [1 1 0]-direction, is tested in uniaxial tension at low strain rates, the deformation in stage I is by slip on the \((1 1 2)[1 1 1]\) system. A second system, \((2 1 1)[1 1 1]\), becomes activated in stage II. The slip geometry and the dislocation structure vary according to the deformation temperature. When single crystals of vanadium were deformed in compression along the [1 1 1]-direction at 77 K (Edington and Smallman, 1965), the specimens deform initially by twinning on the \((1 2 1)[1 1 1]\)- and \((2 1 1)[1 1 1]\)-twin systems. The polycrystal V-notched vanadium specimen impacted in a Charpy machine (Glough and Pavlovic, 1960), shows mechanical twins at \(-78^\circ\text{C}\) and lower temperatures. Twins occur on the \((1 1 2)\)-planes. Having investigated the dislocation configuration and density, produced by plastic deformation in vanadium in the 77–673 K temperature range, Edington and Smallman (1964) find that, in the material with a constant grain size \((2d = 0.0013 \text{cm})\), the dislocation density \(\rho_d\) is proportional to the strain, and that \(\sqrt{\rho_d}\) is proportional to the flow stress, over the entire considered temperature range.

Like other bcc metals, the deformation mechanism in vanadium is rate-dependent. Three dislocation mechanisms have been proposed as the rate-controlling ones in bcc metals at low temperatures. They are believed to be responsible for the strong temperature dependence of the yield and flow stress in these materials. These mechanisms are:
1. the overcoming of the Peierls–Nabarro stress;
2. the nonconservative motion of jogs in screw dislocations;
3. the overcoming of the interstitial precipitates (Schoeck, 1961; Conrad and Frederick, 1962; Conrad and Hayes, 1963a,b).

Wang and Bainbridge (1972) have suggested that the dominant rate-controlling mechanism in the deformation of high-purity vanadium at temperatures between 200 and 293 K, is most likely the interaction between the dislocations and the interstitial impurities, while at temperatures below 200 K, the deformation is probably dominated by the Peierls mechanism. It is known that impurities have a profound effect on the plastic behavior of bcc metals. In general, the yield stress is critically dependent on the purity of the metal. The material’s temperature and strain-rate sensitivity vary with purity, especially at high temperatures (Christian and Masters, 1964a,b).

To our knowledge, all results on the plastic deformation of vanadium have focused on the low strain-rate response. There is little work on the high strain-rate response of vanadium, with few modeling efforts. The present paper reports the results of systematic high strain-rate experiments on commercially pure vanadium over a broad temperature range. Based on these experiments, a micromechanically based model is developed and the corresponding predictions are compared with the experimental observations, arriving at good correlation. In addition, the microstructure of the undeformed and deformed samples is examined using optical microscopy.

2. Experiments

In the present work, a commercially pure vanadium is obtained in the form of a 6.45 mm diameter rod from Electronic Space Products International (ESPI). This polycrystal vanadium is made by the electron-beam-melted method. Its impurities are shown in Table 1.

Cylindrical specimens are cut from this rod. They have a 5 mm nominal diameter and are 5 mm high. All samples are annealed at a constant temperature of 1000°C for 1 h in a vacuum of approximately 10^{-5} Torr. Then, they are cooled to room temperature. Metallographic examination of an annealed sample reveals an average grain size of
approximately 120 μm. Compression tests are carried out at strain rates of 2500 and 8000 s⁻¹ over a range of temperatures from 77 to 800 K. Room-temperature experiments at high strain rates of 13 500, 20 000, and 30 500 s⁻¹ are also performed to check the model predictions. All these dynamic tests are performed using a split Hopkinson pressure bar with a momentum trap (Nemat-Nasser et al., 1991). This novel Hopkinson bar is also enhanced by a furnace which heats the sample while only minimally affecting the incident and the transmission bars (Nemat-Nasser and Isaacs, 1997).

To reduce the end friction on the samples during a dynamic loading, the sample ends are first polished using waterproof silicon carbide paper, 800–4000 grit, and then they are greased for low- and room-temperature tests. A molybdenum powder lubricant is used for the high-temperature tests. It is known that the oxidation of vanadium becomes serious at temperatures exceeding 250°C (Elvers and Hawkins, 1996). Therefore, an argon atmosphere is used in the heating furnace in order to prevent oxidization.

3. Experimental results and discussion: microstructure and plastic flow

3.1. Initial microstructure

It is well known that the plastic response of most materials depends on the strain rate and temperature, as well as on the microstructural features such as the grain size, second-phase particles, and the dislocation density and its distribution.

3.2. Grain-size distribution

To examine the effect of grain size on the flow stress, samples of vanadium are sectioned, polished, and then etched in a 12.5 ml HNO₃, 2.5 ml HF, 25 ml H₂O solution for 3 min. The grain-size distribution is then examined by optical microscopy. A nonuniform distribution is observed for both as-received and annealed samples. The average grain size in the central one-third portion of the sample is about 10 μm for the as-received and 76 μm for the annealed cases. For the remaining portion of the sample, the corresponding sizes are 17 and 165 μm, respectively. For the purpose of comparison, the average grain sizes of 14 μm for the as-received and 120 μm for the annealed samples, are used. Fig. 1 shows the microstructure of an annealed sample, taken at a two-third diameter, within the sample.

3.3. The effect of grain size on plastic deformation

Fig. 2 compares the flow stress of the as-received and annealed samples, over the indicated temperatures, at a strain rate of 2500 s⁻¹. As is seen, only the high-temperature response seems to be affected by the average grain size for strains greater than 20%; see also Fig. 3. The initial dislocation density in the as-received material may be as high as 10^{12} cm⁻², while for the annealed samples, it may be as low as 10⁶ cm⁻² (Edington and Smallman, 1964). Hence, the initial low-temperature flow stress of the as-received material is considerably higher than that of the annealed one, because of greater dislocation-impurities and dislocation–dislocation interactions. At room- and high-temperatures, this difference seems to disappear. The effect of the average strain size is seen to become evident for a 700 K initial temperature, as shown in Figs. 2 and 3; this difference is believed to be due to the effect of the grain size on the long-range resistance to the motion of dislocations.

Fig. 4 displays the flow stress of annealed vanadium over the temperature range of 77–800 K, at a strain rate of 2500 s⁻¹. It is seen that a further increase in temperature above 700 K does not
affect the flow stress at this strain rate, indicating that the response is essentially temperature-independent for temperatures greater than 700 K.

3.4. Adiabatic shearbands and deformation twins

Adiabatic shearbands are observed in most bcc metals which are deformed at low initial temperatures and high strain rates, to large strain rates. While vanadium shows greater plasticity than other bcc refractory metals, it does shearband when deformed to large plastic strains, as shown in Fig. 5(a) and (b), for the as-received and annealed samples, respectively.

Fig. 6 shows the microstructure of the sample shown in Fig. 5(b). The arrow indicates the loading direction; the loading direction is the same in all figures. Twinning is observed at low temperatures and high strain rates, as is evident in Fig. 6. The density of twins decreases sharply as the test temperature is increased; see Fig. 7(a) and compare with Fig. 6. The higher temperature produces
serrated twins, and this is also evident in Fig. 7(b) which corresponds to the microstructure that results when a sample is first deformed to about $\gamma = 0.18$ at a 77 K initial temperature, and then is heated to a room temperature of 296 K before being loaded at the same 2500 s$^{-1}$ strain rate, by an additional 20% strain. The deformation mechanism in vanadium is believed to change from that controlled by the Peierls resistance to the dislocation motion at temperatures below about 200 K, to that controlled by the predominant interaction between the dislocations and the interstitial impurities (Conrad and Hayes, 1963; Wang and Bainbridge, 1972), above 200 K. The observed serration at the edges of the twins at 296 K, may possibly be a result of the intersection of the dislocation-induced slip and the deformation twins. The twins do not occur when the initial test temperature is increased to 700 K, as shown in Fig. 8. The black dots inside the grains in Fig. 8 are possibly second-phase precipitates, occurring in the vicinity of dislocation (Edington and Smallman, 1964). They have an effect on the induced slip.

Summarizing the above observations, we note that:

1. at low temperatures and high strain rates, adiabatic shearbands occur after suitably large straining, but no cracks are observed;
Fig. 5. (a) Adiabatic shearbands in as-received vanadium, strained to $\gamma = 0.6$ at 77 K initial temperature and 2500 s$^{-1}$ strain rate. 
(b) Adiabatic shearbands in annealed vanadium, strained to $\gamma = 0.54$ at 77 K initial temperature and 2500 s$^{-1}$ strain rate.

Fig. 6. Microstructure of annealed vanadium, strained to $\gamma = 0.54$ at 77 K initial temperature and 2500 s$^{-1}$ strain rate.
2. extensive twinning occurs at low temperatures and high strain rates, and the twin density decreases with increasing temperature;

3. second-phase precipitates are seen to occur at high temperatures.

4. A physically based model for dynamic response of vanadium

We seek to apply to our vanadium, a model which has been recently proposed and used by Nemat-Nasser and coworkers (Nemat-Nasser and Isaacs, 1997; Nemat-Nasser et al., 1999) to represent the dynamic response of tantalum and molybdenum over a broad temperature range. A modified version of this model has been successfully applied to OFHC copper by Nemat-Nasser and Li (1998). The model uses the basic concept that the resistance to the motion of dislocations, introduced by various microstructural barriers, defines the flow stress of metals (Kocks et al., 1975; Follansbee and Kocks, 1988; Follansbee and Gray III, 1989). Here for vanadium, we assume that the

Fig. 7. (a) Microstructure of annealed vanadium, deformed to $\gamma = 0.54$ at 296 K initial temperature and 2500 s$^{-1}$ strain rate. (b) Microstructure of annealed vanadium, strained to $\gamma = 0.38$ from 77 to 296 K initial temperature and 2500 s$^{-1}$ strain rate.
flow stress, \( \tau \), consists of three parts: one essentially due to the Peierls stress, denoted by \( \tau^* \); another, an athermal component, \( \tau_a \), mainly due to the long-range effects such as the stress field of dislocation forests and grain boundaries; and a remaining viscous-drag component, \( \tau_d \), which is important at high temperatures and high strain rates. Thus, the flow stress is written as

\[
\tau = \tau_a + \tau_d + \tau^*.
\]

In this formulation, \( \tau^* \) depends on the temperature and the strain rate, while \( \tau_a \) is basically temperature- and strain-rate-independent, being the athermal part of the flow stress, and \( \tau_d \) is a function of the strain rate and temperature only.

To apply Eq. (1) to model the high strain-rate response of the considered vanadium, it is first necessary to check experimentally the basic assumptions that:
1. \( \tau^* \) depends on temperature and strain rate only;
2. the observed softening in the flow stress is due to the plastically induced temperature increase (adiabatic heating) of the sample;
3. at sufficiently high temperatures, the flow stress is essentially due to athermal resistance to the dislocation motion and a viscous-drag likely due to either phonon damping or solute-atom drag.

The last assumption implies that, for \( T > T_c \), \( \tau^* \approx 0 \), where \( T_c \) is a critical temperature, depending on the strain rate only. In addition, we need to estimate the amount of the plastic work converted into heat in an adiabatic high strain-rate condition.

Fig. 9 shows the effect of a temperature jump on the flow stress. The light curves are the adiabatic stress–strain relations at indicated temperatures and at a strain rate of 2500 s\(^{-1}\). The dark curves show the flow stress when the sample is first deformed at 77 K to about 18% strain, then it is unloaded and allowed to reach room temperature before it is reloaded at the same strain rate. As is seen, the resulting stress–strain curve essentially follows the light curve of the initial room-temperature test. Fig. 10 shows similar results, but this time there is a jump decrement in the initial temperature. Based on these results, it may thus be assumed that, at a fixed (high) strain rate, \( \tau^* \) is basically a function of temperature, \( T \), only; later on, the strain-rate dependence of \( \tau^* \) is checked experimentally.

As a sample is plastically deformed at high strain rates, its temperature changes by

\[
\Delta T \approx \frac{\eta}{\rho} \int_{\gamma_0}^{\gamma} \frac{\tau}{C_v} \, d\gamma,
\]

Fig. 8. Microstructure of annealed vanadium, strained to \( \gamma = 0.56 \) at 700 K initial temperature and 2500 s\(^{-1}\) strain rate.
where \( \rho \) (6.16 g cm\(^{-3}\)) is the mass density (assumed to remain constant), \( \eta \) the fraction of energy converted into heat, and \( C_v \) is the (temperature-dependent) heat capacity of the material at constant volume. For our application, it appears adequate to use an average constant value of 0.498 J g\(^{-1}\) K\(^{-1}\) for \( C_v \). To estimate \( \eta \), we have used an interrupted test, Fig. 11, where the sample is strained at a 2500 s\(^{-1}\) strain rate to about 18\%, then cooled to its initial room temperature of 23°C, and then subsequently heated to 53°C and reloaded at the same 2500 s\(^{-1}\) strain rate. According to Eq. (2), an increase of 30°C corresponds to \( \eta = 1 \), i.e., when all the plastic work is used to heat the sample. As is evident from Fig. 11, the assumption of \( \eta \approx 1 \) is good within experimental error, and is in accord with other experimental data reported by Kapoor and Nemat-Nasser (1998); see also Nemat-Nasser and Isaacs (1997), Nemat-Nasser et al. (1999) and Nemat-Nasser and Li (1998).

4.1. Athermal stress component, \( \tau_a \)

The athermal resistance to the motion of dislocations, \( \tau_a \), is assumed to be due to the elastic
stress field generated by the dislocations, point defects, grain boundaries, and various other impurities such as those listed in Table 1. Hence, the temperature dependence of \( \tau_a \) is only through the temperature dependence of the elastic modulus, especially the shear moduli \( \mu(T) \). The stress \( \tau_a \) is independent of the strain rate. Based on linear elasticity, \( \tau_a \) would be proportional to \( \mu \). Hence, \( \tau_a \) is independent of both strain rate and temperature, where \( \mu_0 \) is a reference value of the shear modulus. We therefore may write

\[
\tau_a \approx f(\rho, d, \ldots),
\]

where \( \rho \) is the average dislocation density, \( d \) the average grain size, and the dots stand for parameters associated with other impurities. In a general loading, the strain \( \gamma \) represents the “effective” plastic strain which is a monotonically increasing quantity in plastic deformation. In the present case, \( \gamma \) defines the loading path and is also a monotonically increasing quantity, since \( \gamma > 0 \). Hence, it can be used to define the variation of the dislocation density, the average grain size, and other parameters which affect \( \tau_a \), i.e., we may set \( \tau_a = f(\rho(\gamma), d(\gamma), \ldots) = f(\gamma) \). Further, as a first approximation, we may use a simple power-law representation of \( f(\gamma) \), and set

\[
\tau_a \approx a_0 + a_1 \gamma^n + \cdots,
\]

where \( a_0, a_1, \) and \( n \) are free parameters which must be fixed experimentally. In our work, we choose an average value for \( \mu_0/\mu(T) \approx 1 \), so that \( \tau_a \approx \tau_a \). This approximation is used below, but it is not essential to our work, and the actual \( \mu(T) \) can be employed if deemed necessary.

4.2. Viscous-drag component, \( \tau_d \)

It is known that, at high strain rates and temperatures, the strain-rate sensitivity of most metals increases rapidly with increasing strain rate and temperature. This is assumed to be due to the phonon- (mostly at high temperatures) and electron-drag effects on the mobile dislocations (Follansbee and Weertman, 1982; Regazzoni et al., 1987; Chiem, 1992; Zerilli and Armstrong, 1992; Kapoor and Nemat-Nasser, 1999). The viscous-drag stress, \( \tau_d \), is usually taken to be proportional to the average dislocation velocity, \( v \), i.e., \( \tau_d \sim MBv/b \), where \( M \) is the Taylor factor and \( B \) is the drag coefficient. Since \( v \) relates to the strain rate by \( \dot{\gamma} = \rho_m b v \), it follows that we may set \( \tau_d \approx \rho_m B \dot{\gamma}/b \), where \( \rho_m \) is the Taylor factor and \( B \) is the drag coefficient. Since \( v \) relates to the strain rate by \( \dot{\gamma} = \rho_m b v \), it follows that we may set \( \tau_d \approx g(M^2 B/(\rho_m b^2), \dot{\gamma}, T) \). Here, \( M \approx 2.75 \) (Kapoor and Nemat-Nasser, 1998), \( \rho_m = O(10^{13} \text{ m}^{-2}) \) (Follansbee and Weertman, 1982), \( B = O \left( 10^{-3} \text{ Pa s} \right) \), and \( b \) is the magnitude of the Burgers vector. At high temperatures, the flow stress is essentially independent of the temperature \( T \), and we have \( \tau_d \approx g(M^2 B/(\rho_m b^2), \dot{\gamma}) \).

To examine the effects of the viscous-drag on the flow stress, Kapoor and Nemat-Nasser (1999) have performed high-temperature experiments on tantalum, Ta. Their results are replotted in Fig. 12.
From these experimental results, it can be seen that, when the strain rate exceeds about 2000 s$^{-1}$, the flow stress is basically constant. Based on these experimental results and the notion of a drag mechanism, we set

$$
\tau_d = A_0 \left[ 1 - \exp \left( -\frac{c\gamma}{\tau_y} \right) \right], \quad \alpha = \frac{M^2 B}{\rho_m b^2 \mu_0},
$$

where $A_0$ is a material constant which can be measured directly at very high strain rates, and $\alpha$ represents an effective damping coefficient, affecting the dislocation motion. For bcc metals, we take $M^2 = O(10)$, $b = O(10^{-3}$ m), $\rho_m = O(10^{13}$ m$^{-2}$), $B = O(10^{-3}$ Pa s), and a high-temperature yield stress of $\tau_y = 120$ MPa, measured at a 1000 K temperature and 10$^{-3}$ s. Then, for this tantalum, we obtain $\alpha = M^2 B / (\rho_m b^2 \tau_y) \approx 8.74 \times 10^{-4}$. This yields $\tau_a + \tau_d = 120 + 120 [1 - \exp (-8.74 \times 10^{-4} \gamma)]$ which fits the experimental results of Fig. 12.

### 4.3. Athermal- and drag-stress components of vanadium

For the present vanadium, experiments at strain rates of 10$^3$ s$^{-1}$ and 2500 s$^{-1}$ and various temperatures have also been carried out. Fig. 13 shows the resulting relations between the flow stress and temperature, at a 0.1 true strain. From this figure, it is clear that the athermal stress at high temperatures, depends on the strain rate. For a strain rate of 10$^3$ s$^{-1}$ and at temperatures greater than 500 K, the flow stress is about 220 MPa. But at a strain rate of 2500 s$^{-1}$ and temperatures exceeding 800 K, this flow stress is about 300 MPa. The difference, 80 MPa, is possibly due to the dislocation-drag stress. Motivated by the tantalum result in (5), assume for the vanadium that:

$$
\tau_a + \tau_d = a_0 \gamma^n + A \left[ 1 - \exp \left( -\frac{c\gamma}{\tau_y} \right) \right],
$$

where $a_0$, $n$, and $A$ from the corresponding experimental data.

To identify the constitutive parameters for the athermal stress and drag stress of the vanadium, in Eq. (6), the flow stress at a 2500 s$^{-1}$ strain rate is plotted in Fig. 14 as a function of temperature, for indicated values of the strain; these data include the effect of adiabatic heating calculated from (2). From Fig. 14, it is seen that, when the temperature exceeds 700 K, the flow stress of vanadium is independent of temperature. Therefore, we take the stress at 800 K to be the athermal stress for this vanadium at a strain rate of 2500 s$^{-1}$ (this also includes the drag stress component). The limiting values of the flow stress for large values of temperature are then estimated from Fig. 14, plotted in Fig. 15, and fitted by Eq. (6), arriving at the following expression (shown by the solid curve in Fig. 15)
Fig. 13. Flow stress as a function of temperature for indicated strain rates and strain.

Fig. 14. Flow stress as a function of temperature for indicated strains and 2500 s\(^{-1}\) strain rate.

Fig. 15. Limiting values of flow stress as a function of strain.
\[ \tau_a + \tau_d = 305\gamma^{0.2} + 90\left[1 - \exp\left(-8.74 \times 10^{-4}\gamma\right)\right]. \quad (7) \]

4.4. Thermally activated component of flow stress

Once \( \tau_a + \tau_d \) is estimated by (7), we calculate the experimental values of \( \tau^* = \tau - (\tau_a + \tau_d) \), and obtain the results shown in Fig. 16. The data points seem to fall on a single curve, independently of the value of the corresponding strain. This curve is given by

\[ \tau^* = 1050\left[1 - (0.00125T)^{1/2}\right]^{1.5}. \quad (8) \]

The factor 0.00125 combines the activation energy, \( G_0 \), and the reference strain rate, \( \dot{\gamma}_0 \), in the following expression for \( \tau^* \):

\[ \tau^* = \tau^* \left[1 - \left(-\frac{kT}{G_0} \ln \frac{\dot{\gamma}}{\dot{\gamma}_0}\right)^{1/q}\right]^{1/p}, \quad (9) \]

which results if we assume that the energy barrier, \( \Delta G \), that a dislocation must overcome by its thermal activation, is given by

\[ \Delta G = G_0\left[1 - \left(\frac{\tau^*}{\tau^*}\right)^{pq}\right]^q, \quad (10) \]

where \( p \) and \( q \), with \( 0 < p \leq 1 \) and \( 1 \leq q \leq 2 \), are parameters which define the barrier profile, and \( k \) is Boltzmann’s constant (Kocks et al., 1975). Furthermore, the plastic strain rate \( \dot{\gamma} \), relates to \( \Delta G \) and \( T \) by

\[ \dot{\gamma} = \dot{\gamma}_0 \exp\left(-\frac{\Delta G}{kT}\right). \quad (11) \]

with \( \dot{\gamma}_0 = b\rho_mD\omega_0d \), where \( \rho_m \) is the density of the mobile dislocations, \( \omega_0 \) the attempt frequency, and \( d \) is the average barrier spacing (for the Peierls barrier, it is essentially the lattice spacing).

From Fig. 16, we have estimated \( \dot{\gamma}^* \) to be about 1050 MPa, and have taken \( p = 2/3 \) and \( q = 2 \). From (8) and (9), it follows that

\[ -\frac{k}{G_0} \ln \frac{\dot{\gamma}}{\dot{\gamma}_0} = 0.00125. \quad (12) \]

Nemat-Nasser and Isaacs (1997) use \( G_0 \approx 1 \text{ eV atom}^{-1} \) for tantalum. It turns out that, for vanadium, \( G_0 \approx 0.5 \text{ eV atom}^{-1} \) is a good estimate; this leads to \( k/G_0 = 1.72 \times 10^{-4} \) and \( \dot{\gamma}_0 = 3.58 \times 10^6 \text{ s}^{-1} \); see also the results shown in Fig. 17(d). Hence, (9) yields

\[ \tau^* = 1050\left[1 - \left(-1.72 \times 10^{-4}T \ln \left(\frac{\dot{\gamma}}{3.58 \times 10^6}\right)\right)^{1/2}\right]^{3/2}, \quad (13) \]

and the final expression for the flow stress in the temperature range \( T < T_c \) becomes:
Fig. 17. Comparison of model predictions with experimental results for indicated initial temperatures and indicated strain rates.
The critical temperature $T_c$ is given by

$$T_c = \frac{-G_0}{k \ln \left( \dot{\gamma} / \dot{\gamma}_0 \right)}.$$  \hfill (15)

Fig. 17(a) and (b) compare the experimental results with the predictions obtained from Eq. (14). To check the validity of (14) by independent tests, we have implemented a temperature jump in Fig. 17(c), from an initial temperature of 500 K to room temperature. Additional independent validation is discussed below.

4.5. Strain-rate dependency of flow stress

To check the accuracy of the prediction of the strain-rate dependency of the flow stress given by (14), we have performed a series of tests at various initial temperatures, and at a strain rate of 8000 s$^{-1}$. All high-temperature tests are performed in an argon atmosphere. Fig. 17(d) compares the experimental results with the predictions given by Eq. (14). As is seen, good correlation is obtained. Furthermore, using a mini-Hopkinson (3/16 in.) bar system, we have tested this material at strain rates of 13 500, 20 000, and 30 500 s$^{-1}$, starting with room temperature. The results are shown in Fig. 18 together with the predictions of (14). Again, a good correspondence is obtained.

5. Low strain-rate response

In addition to the high strain-rate tests which constitute the main objective of the present work, we have performed a series of low strain-rate tests (at $10^{-3}$ s$^{-1}$ and a few $10^{-2}$ s$^{-1}$ strain rates) on our vanadium, for comparison purposes. The results are given in Figs. 19 and 20 for indicated temperatures and strain rate, respectively. The data display the presence of dynamic strain aging for temperatures above 500 K. It is known that strain aging occurs due to the interaction between mobile dislocations and impurities such as solute atoms which may diffuse toward dislocations and pin them down at suitable temperatures and strain rates. At high enough strain rates, on the other hand, there may not be enough time for the migration of solute atoms to dislocations,
Fig. 19. Flow stress at indicated temperatures and 0.001 s\(^{-1}\) strain rate.

Fig. 20. Flow stress as a function of temperature for indicated strain and 0.001 s\(^{-1}\) strain rate.

Fig. 21. Flow stress as a function for indicated strain rates and 0.1 strain.
and the phenomenon of dynamic strain aging disappears.

Fig. 21 compares the high and low strain-rate data for a true strain of 0.1 and various temperatures. For a strain rate of $10^{-3}$ s$^{-1}$, a peak stress is observed at about 750 K, which even exceeds the flow stress at 2500 s$^{-1}$ at the same temperature. This is most likely due to the presence of impurities listed in Table 1, which also affect the low-temperature response of the material. As is seen in Fig. 21, the flow stress at $10^{-3}$ s$^{-1}$ and 100 K is very close to that at 2500 s$^{-1}$. It is clear that the influence of the impurities listed in Table 1 must be included in modeling the low-strain-rate response of this material. The model presented in Eq. (14) does not include this and hence underestimates the low strain-rate response of the material.

6. Conclusions

1. The initial microstructure (grain sizes and their distribution, the density of the dislocation, etc.) of the considered commercially pure vanadium affects the initial plastic flow (say, for $\gamma < 20\%$), especially at low temperatures.
2. Adiabatic shearbands occur at low temperatures and high strain rates.
3. Deformation twinning occurs at lower temperatures. The density of twins rapidly decreases with increasing temperature.
4. Dynamic strain aging occurs above a 500 K temperature at a strain rate of $10^{-3}$ s$^{-1}$, with the local peak stress at about 750 K.
5. Based on the experimental results at high strain rates, a micromechanical model is developed which accurately predicts the plastic flow stress of this vanadium at high strain rates, but underestimates the low strain-rate stress.

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