



Thermal activation parameters of V–5Cr–5Ti alloy under hot compression

Yu-fei LI¹, Zhen-hong WANG², Yun-hong ZHOU², Chao LUO², Xin-chun LAI¹

1. Science and Technology on Surface Physics and Chemistry Laboratory, Mianyang 621907, China;

2. China Academy of Engineering Physics, Mianyang 621900, China

Received 27 October 2014; accepted 30 January 2015

Abstract: In order to well understand the elementary mechanisms that govern the hot working process of a V–5Cr–5Ti alloy (mass fraction, %), thermal activation parameters under compression were measured in a temperature range of 1373–1673 K by a Gleeble–3800 system. The results show that the stress exponent n is 4.87 and the activation energy Q is 375.89 kJ/mol for the power law equation. The activation energy is determined as 288.34 kJ/mol, which is close to the self-diffusion energy of alloy (270–300 kJ/mol) by introducing a threshold stress (σ_0) variable. The typical values of physical activation volume (V_p) and strain rate sensitivity (m) are measured as $(120–700) b^3$ and 0.075–0.122, respectively, by the repeated stress relaxation tests. These activation parameters indicate that the rate controlling mechanism for V–5Cr–5Ti alloy compressed in ranges of 1373–1673 K and 0.001–1.0 s^{−1} is the dislocation climb by overcoming of forest dislocations.

Key words: V–5Cr–5Ti alloy; hot compression; thermal activation; dislocation climb

1 Introduction

Researches on the high temperature deformation mechanism of V–Cr–Ti alloy, which has been investigated as the potential fusion reactor first-wall blankets over several years, are mainly on the thermal creep [1–6]. The thermal activation parameters, such as activation energy and activation volume, are very important to understand the elementary mechanisms. It is well accepted that the predominant creep mechanism in V–(4–5)Cr–(4–5)Ti alloy between 600–850 °C is climb-assisted glide of dislocations [1–4]. The reported activation energy Q is 180–326 kJ/mol [1–3,5,6] and the stress exponent n is in the range of 2.7–8.0 [1,3,5], depending on the materials and testing methods used. On the other hand, LI et al [7], GUBBI and ROWCLIFFE [8] measured the activation energies of static recrystallization for V–(4–5)Cr–(4–5)Ti alloy are 261.91–289.67 kJ/mol and 576 kJ/mol, respectively.

It is believed that both creep and hot working can share the same dislocation motion mechanism which governs plasticity. The temperature dependent flow stress reflects the interaction of dislocations with energy barriers, which are mostly thermal activated [9]. While,

insufficient data are available in the temperatures of 1373–1573 K, which are the typical temperatures for the hot working of V–Cr–Ti alloy [10]. In order to well understand the elementary mechanisms that govern the hot working process of alloy, it is necessary to determine thermal activation parameters of V–5Cr–5Ti alloy during hot working.

The present paper is intended to determine the activation energy Q and activation volume V of a V–5Cr–5Ti alloy in the range of 1373–1673 K by hot compression experiments, including constant strain rate test and transient test.

2 Experimental

A piece of vertical section disk ($d120 \text{ mm} \times 50 \text{ mm}$) was cut from a hot isostatic pressed (HIP-ed) V–5Cr–5Ti ingot. Its chemical composition is V–5.12Cr–4.87Ti–0.056O (mass fraction, %). A series of cylindrical specimens, $d8 \text{ mm} \times 12 \text{ mm}$, were cut from the sheet by electrical discharge machine and lathe [11]. The mean grain size of specimen is about 860 μm . The hot compression test was performed on a Gleeble–3800 system (Dynamic Systems Inc., USA) in an argon atmosphere (99.999%, $\sim 0.09 \text{ MPa}$).

Two kinds of experiment were carried out. The first kind of test is the constant strain rate compression at three strain rates (0.01, 0.1, and 1 s⁻¹) and five temperatures (1423, 1473, 1523, 1573, 1673 K) to a true strain level of 0.5. The second experiments are transient test, including strain rate jump and stress relaxation at four temperatures (1373, 1473, 1573 and 1673 K), which were performed near the yield point. For the strain rate jump tests, the applied strain rate was changed from 0.001 to 0.01 s⁻¹. For the repeated stress relaxation tests, two applied strain rates (0.001 and 0.01 s⁻¹) were performed for comparison, and the relaxation time of each cycle is 30 s and 5–6 cycles were adopted in each test. The compression axial (CA) direction is along the axial of specimen.

Transmission electron microscopy (TEM, FEI TF 20) was used to identify the Ti-(CNO) phase. The as-received sheet was machined, polished and cut to $d3$ mm foils. The foil specimen was final thinned by electronic two-jet method under DC 20 V, 20 mA. The composition of solution is 15% H₂SO₄ + 85% H₂O (volume fraction).

3 Results and discussion

3.1 Activation energy and stress exponent

3.1.1 Apparent constitutive analysis

For hot working, the phenomenological hyperbolic sine constitutive model combined with Zener–Hollomon parameter (Z), Eq. (1), is well known to predict the flow stress and to evaluate the apparent activation energy [11–14]. In general, Eq. (1) can be simplified to Eqs. (2) and (3) for low stress level (power law) and high stress level (exponential law), respectively.

$$Z = \dot{\epsilon} \exp[Q_a/(RT)] = A \sinh(\alpha\sigma)^n \quad (1)$$

$$\dot{\epsilon} \exp[Q_a/(RT)] = A_1 \sigma^{n_1} \quad \text{for } \alpha\sigma < 0.8 \quad (2)$$

$$\dot{\epsilon} \exp[Q_a/(RT)] = A_2 \exp(\beta\sigma) \quad \text{for } \alpha\sigma > 1.2 \quad (3)$$

where $\dot{\epsilon}$ is the strain rate (s⁻¹), R is the universal gas constant, T is the thermodynamic temperature (K), Q_a is the apparent activation energy for hot deformation (kJ/mol), σ is the flow stress (MPa) for a given strain, A , A_1 , A_2 , α , β , n_1 , n , n_1 and β are the material constants.

It should be noted that the nature of materials constants and equations are dependent on the characteristic stress used to derive them. In general, the peak stress is the most widely accepted one to find the hot working constants [13]. The peak stresses, obtained by constant strain rate compression experiments and revised by friction coefficient [11], were used in the present work (Fig. 1). Following the well established procedures [11–14], the mean values of α and n are determined to be 0.007635 and 6.02, respectively.

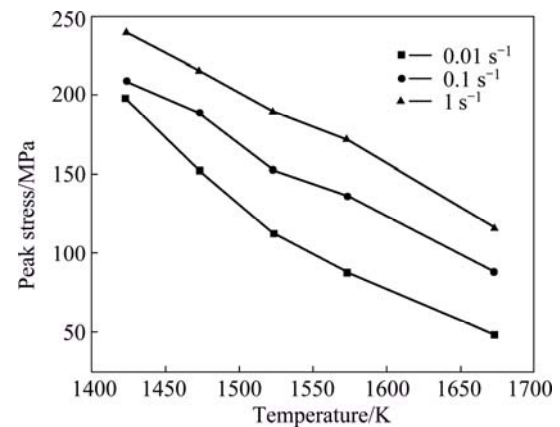


Fig. 1 Peak stress dependent on temperature and strain rate revised by friction coefficient

The slope of the plots of $\ln[\sinh(\alpha\sigma)]$, $\ln\sigma$ and σ against the reciprocal of thermodynamic temperature can be used to obtain the value of Q_a [12]. The values of Q_a are calculated as 637.13, 883.48 and 737.42 kJ/mol based on Eqs. (1)–(3), respectively, by averaging the values under different strain rates. The correlation coefficients (R^2) of these regression values are 0.9894, 0.9756, and 0.9869, for Eqs. (1)–(3), respectively, which reveals that Eq. (1) has a slightly better fit to the experimental data. Therefore, the apparent activation energy Q_a of hot working can be taken as 637.13 kJ/mol.

It can be seen that the Q_a (637.13 kJ/mol) is higher than that for creep (180–326 kJ/mol [1–3,5,6]), because it is without any specific physical meaning (and may be the sum of several thermal activation processes) and is common for many metals [15].

3.1.2 Physical-based approach

As we all know, the n value should be 5 and the Q value should be close to the self-diffusion energy as long as the deformation mechanism is controlled by the glide and climb of dislocations [12]. Moreover, Eq. (2) is also well known for thermal creep equation [1–6]. Usually, Eq. (4) is used to fit the experimental data to get the actual values [12,13]:

$$\ln[\dot{\epsilon}/D(T)] = n \ln[\sigma/E(T)] \quad (4)$$

where $D(T) = D_0 \exp[-Q_{sd}/(RT)]$ is the self-diffusion coefficient and $E(T)$ is the elastic modulus, and D_0 is 3×10^{-5} m²/s, Q_{sd} is 300 kJ/mol and $E(T) = 125.2 \times [1 - 7.69 \times 10^{-5}(T - 20)]$ GPa for V–4Cr–4Ti alloy [1,6,16].

Taking data into Eq. (4), the n value can be evaluated from the slope of plots shown in Fig. 2. It can be seen that the entire n value is 6.47 for all data. While, the n value varying from 23.49 to 4.87 depends on the used stress. Following Eq. (2), the Q value can be determined as 499.89 and 375.89 kJ/mol for n of 6.47 and 4.87, respectively, by linear fitting the plot of $\ln(\sigma/E) - 1/T$.

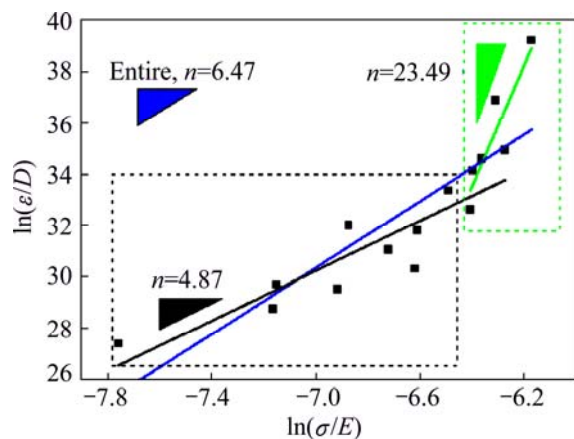


Fig. 2 Plots of $\ln(\dot{\epsilon}/D)$ and $\ln(\sigma/E)$ for peak stress (Data used to determine partial n are denoted as two dashed rectangles)

The compiled data indicate clearly a typical “power-law breakdown” behavior. At high stresses, a high n value of 23.5 is prevailing, while n decreases to around 5.1 at low stresses. A high n value is frequently interpreted as large contribution of dislocation glide to the combined dislocation glide and climb mechanisms [9,17]. Furthermore, the n values are 22.56 and 4.98 at 1423 and 1673 K, respectively, by using the peak stresses directly (not normalized by $D(T)$ and $E(T)$), which are similar to those in Fig. 2. Since, the dislocation climb mechanism is more prevalent with increasing the temperature. Based on this, the values of n and Q are finally determined as 4.87 and 375.89 kJ/mol, respectively.

It can be seen that the values of n and Q (4.87 and 375.89 kJ/mol) are higher than those under creep condition. The fact that Q and n values for hot working are higher than those for creep is common [9,15]. The different impurities contents, different sizes and densities of Ti(CNO) phases, as well as different deformation mechanisms are responsible for this greater Q value.

3.1.3 “Composite materials” consideration

Considering the high stress exponent n (23.49) and Q_a (883.48 kJ/mol) and different Ti(CNO) phases in V–Cr–Ti alloys [7,18], as shown in Fig. 3, these are typical features of a composite materials [19]. Based on this, we try to acquire the “physical” value of n and Q by introducing threshold stress σ_0 .

Firstly, the n values of 2, 3, 5, 8 are used to fit the $\dot{\epsilon}^{1/n}$ and σ_p relationship and we found out that $n = 5$ is the most reasonable one. The fitted σ_0 at 1423, 1473, 1523, 1573 and 1673 K are 167.27, 116.4, 67.2, 39.8 and 9.8 MPa, respectively. By linear fitting the plot of σ_0 – T , the critical temperature T_c above which σ_0 is zero can be evaluated as about 1662 K. The fitted σ_0 at 1673 K is 9.8 MPa, which is reasonable within error.

As shown in Fig. 4, by linear fitting the plots of

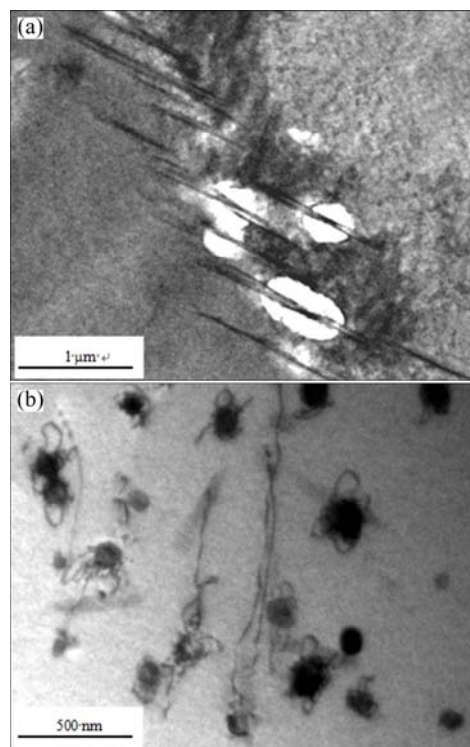


Fig. 3 Two different Ti(CNO) second phases in V–5Cr–5Ti alloy: (a) Plate-shape; (b) Particle-shape

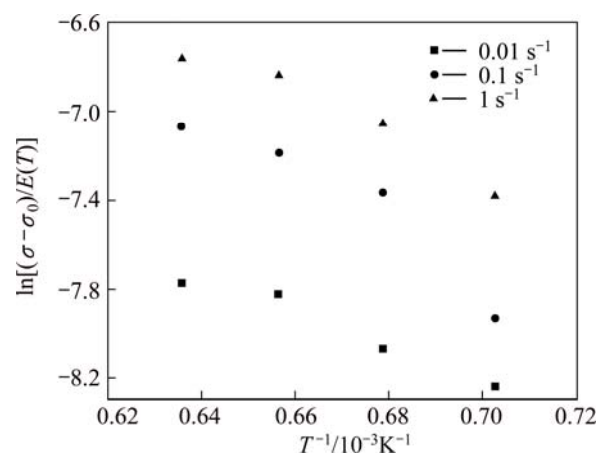


Fig. 4 Plot of $\ln[(\sigma-\sigma_0)/E(T)]-1/T$ to evaluate Q

$\ln[(\sigma-\sigma_0)/E(T)]-1/T$ at three strain rates, the Q value ($Q=R \cdot 5 \cdot |S|$, where S is the slope of fitting line) can be determined. From Fig. 4, the average value of Q is 405.04 kJ/mol. Moreover, if 1423 K is excluded, the average value of Q is 288.34 kJ/mol. This Q value is very close to the typical ones for creep (270 kJ/mol [5,6] and 300 kJ/mol [1,2]) and close to the self-diffusion energy of V–4Cr–4Ti alloy (300 kJ/mol [16] and pure V (270 kJ/mol) [1,3].

It should be noted that the threshold stress is not verified by other experimental results. The σ_0 and Q values are also dependent on the stress used to derive them after testing, and the slope of $\ln[(\sigma-\sigma_0)/E(T)]-1/T$

is negative. Since, it is questionable for this “composite materials” treatment even the result is acceptable apparently and needs further investigation.

3.2 Activation volume and strain rate sensitivity

3.2.1 Transient mechanical tests

Another important parameter to declare the dislocation motion mechanism is the activation volume. The most popular method is the transient mechanical tests, including strain rate jump and stress relaxation experiments, around the yield point [9,17,20–26].

The apparent activation volume V_a can be determined by the formula:

$$V_a = f k T \ln(\dot{\varepsilon}_1 / \dot{\varepsilon}_2) / \Delta \sigma \quad (5)$$

for the strain rate jump experiments, or by fitting the first cycle of a stress relaxation test:

$$V_a = -f k T \ln(1 + t / c_r) / \Delta \sigma \quad (6)$$

where f is the Taylor factor (3.06), k is the Boltzmann constant, t is the time variable, c_r is the time constant, $\Delta \sigma$ is the stress change during the tests.

Using Eq. (7), the physical activation volume V_p can be extracted from data obtained in the consecutive relaxation transients in a repeated stress relaxation test, one gives,

$$V_p = f k T \ln(\dot{\varepsilon}_{i2} / \dot{\varepsilon}_{f1}) / \Delta \sigma \quad (7)$$

$$\dot{\varepsilon}_i = -\dot{\sigma}_i / M = -(d\sigma_i/dt) / M \quad (8)$$

$$\sigma_a = M(\varepsilon_t - \varepsilon_p) \quad (9)$$

where $\dot{\varepsilon}_{i2}$ and $\dot{\varepsilon}_{f1}$ correspond to the strain rates at the onset of stress relaxation 2 and at the finish of stress relaxation 1, respectively, $\dot{\varepsilon}_i$ is the i th run plastic strain rate by measuring the slope of the relaxation curve, $d\sigma_i/dt$, M is the elastic modulus of the specimen-machine assembly, σ_a , ε_t and ε_p are the applied stress, the total and plastic strains, respectively.

On the other hand, the strain rate sensitivity m can be calculated by

$$m = f k T / (V_a \sigma) \quad (10)$$

Repeated stress relaxation cycles, taking the 1473 K, 0.001 s^{-1} as an example, are shown in Fig. 5(a). It shows that the stress decay $\Delta \sigma$ and plastic strain rate during each cycle are almost equal, that is not similar to the common one that $\Delta \sigma_j$ decreases as j (the relaxation number in the series) increases [9,20]. The stress decay curves during the first relaxation run, taking the 1473 and 1673 K as an example, are shown in Fig. 5(b). It can be seen that the stress decay at 0.01 s^{-1} is quicker than that at 0.001 s^{-1} .

Following Eqs. (6)–(10), the V_p , V_a and m of alloy under different conditions can be obtained [9,20,23].

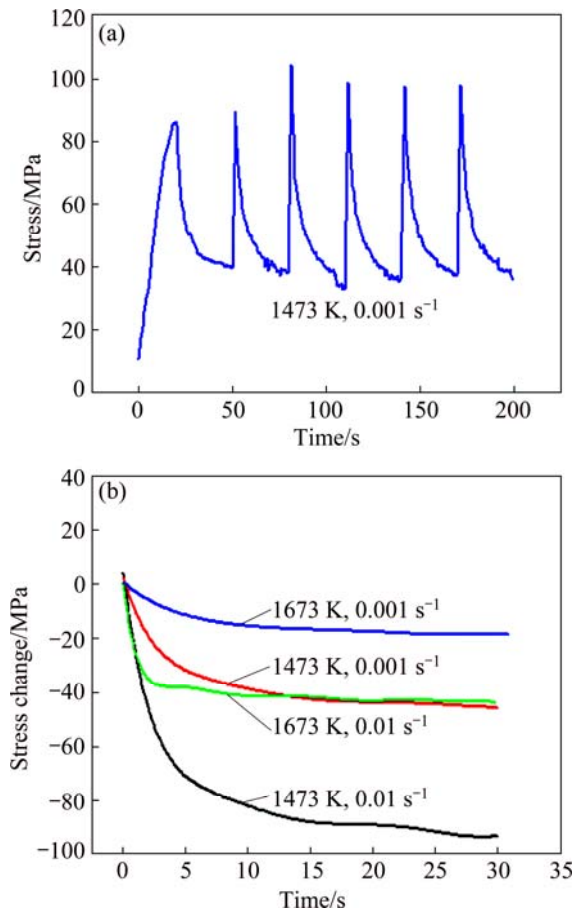


Fig. 5 Repeated relaxation curves (a) and first run curves (b) under different conditions

Figure 6 gives the V_p and V_a values which vary with temperatures and strain rates. The measured V_p and V_a values of V-5Cr-5Ti alloy under the present experimental conditions are about $(120-700) b^3$ and $(150-950) b^3$, respectively. It also shows that both V_p and V_a increase with increasing the temperature and have higher values at 0.001 s^{-1} than at 0.01 s^{-1} . This is because the extent of dynamic recovery (DRV), which can reduce the dislocation density and the distance of forest dislocations, increases with increasing the temperature and decreasing the applied strain rate. Herein, dynamic recrystallization (DRX) can be ignored around the yield point. Furthermore, the values of V_a are in the range of $(150-750) b^3$ evaluated by strain rate jump and single stress relaxation tests, which is consistent with that in Fig. 6.

Following the relation $V_a/V_p = \Omega$ [9,20,24], Ω can be determined and is shown in Fig. 7. The fluctuation of Ω is within 1.0–2.0. Moreover, the value fluctuation of Ω at 0.01 s^{-1} is less than that at 0.001 s^{-1} , and the mean value of Ω is 1.26 at 0.01 s^{-1} .

The dependence of m on temperature and strain rate is given in Fig. 8. Here, m is determined by the following two methods. First, the plots of $\ln(-d\sigma/dt) - \ln \sigma$ for the

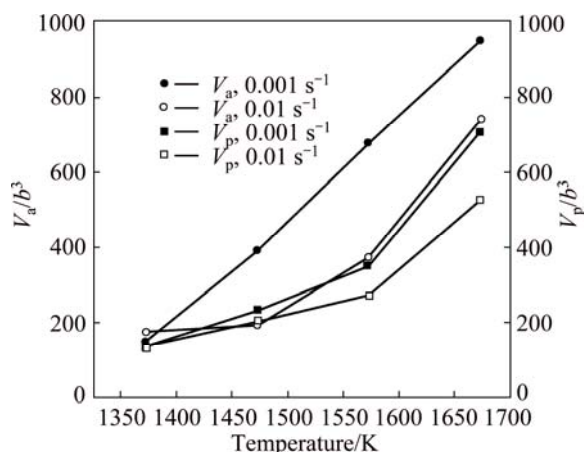


Fig. 6 Evolution of physical and apparent activation volume (V_p and V_a) with temperature and strain rate by transient mechanical tests

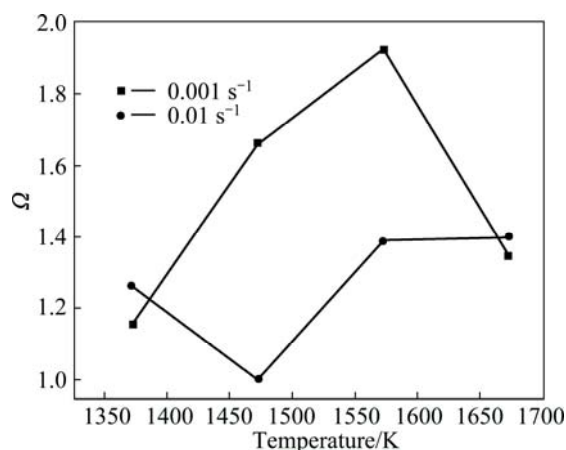


Fig. 7 Evolution of Q with temperature and strain rate

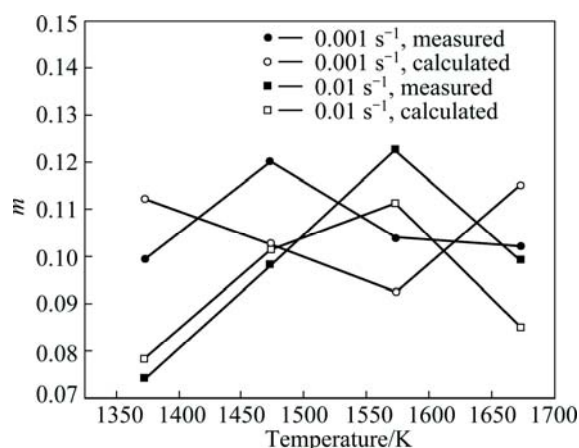


Fig. 8 Evolution of m with temperature and strain rate

first cycle are used to obtain the value of m [23], noted as “measured”. On the other hand, m can also be calculated with Eq. (10) by using the V_a value, noted as “calculated”. Different to activation volume, there is no regular law among m values, while it shows an increase tendency of m with increasing the temperature. A slight

increase of m with increasing the temperature is typical in other materials [9,24]. The typical m values of V–5Cr–5Ti alloy in the present study are 0.075–0.122.

3.2.2 Constant strain rate tests

The transient mechanical tests were performed near the yield point to simplify the deformation mechanism [9,20,23]. While, it is interesting to evaluate the strain dependence, i.e., stress or microstructures dependence, of activation volume and strain rate sensitivity. To achieve this, the data of constant strain rate compression tests are used.

Herein, Eq. (11) [22] and Eq. (10) are adopted to determine m and V_a ,

$$m = \ln(\sigma_1 / \sigma_2) / \ln(\dot{\epsilon}_1 / \dot{\epsilon}_2) \quad (11)$$

where $\dot{\epsilon}_1$ and $\dot{\epsilon}_2$ denote 0.01 s^{-1} and 0.1 s^{-1} , respectively, σ_1 and σ_2 are the stresses at a given strain corresponding to $\dot{\epsilon}_1$ and $\dot{\epsilon}_2$, respectively. Here, all stresses are revised by friction coefficient.

Figure 9 gives the plots of m and V_a against temperature and true strain ($\epsilon=0.05, 0.1, 0.3, 0.5$). As shown in Fig. 9(a), the tendency that m increases with increasing the temperature is more obvious than that in

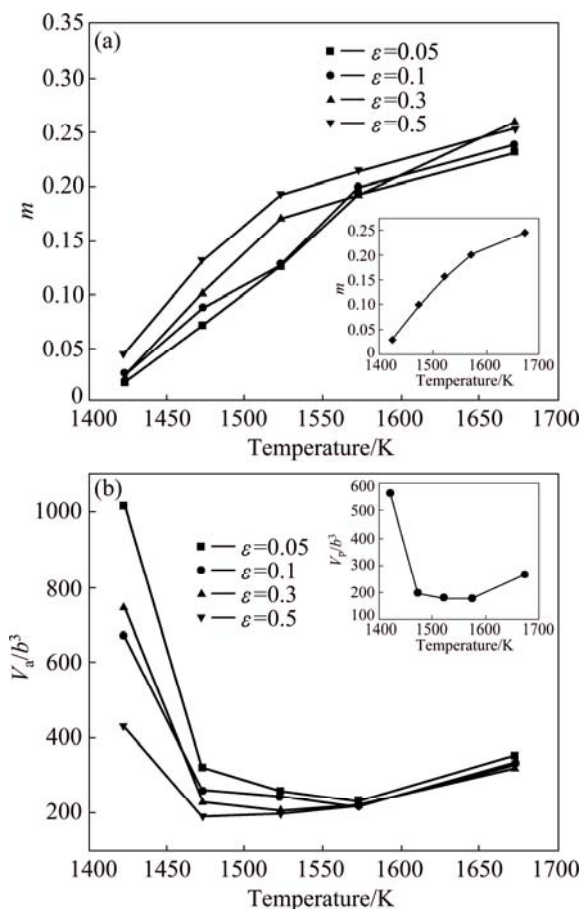


Fig. 9 Dependence of strain rate sensitivity m (a) and apparent activation volume V_a (b) on temperature and strain level during constant strain rate tests (inserts are mean values of m and V_p with temperature)

Fig. 8. The insert in Fig. 9(a) gives the evolution of the average m with temperature. The average values of m increase from 0.027 at 1423 K to 0.245 at 1673 K. Moreover, there is a slight increase of m with increasing the strain at a given temperature.

Figure 9(b) gives the evolution of V_a and V_p (insert) with temperature. Here, the V_p is calculated by using the average V_a at a specific temperature and the mean value of $\dot{\epsilon}$ (1.26 for 0.01 s^{-1}). Except at 1423 K, the increase tendencies of both V_a and V_p (insert) with temperature are similar to that obtained by transient tests (Fig. 6). The calculated V_p values are in the range of $(170\text{--}570) b^3$, which is consist with those of transient tests, $(120\text{--}700) b^3$. There is a decrease tendency of V_a with strain at 1423, 1473 and 1523 K, while the V_a values are almost independent of strain at 1573 and 1673 K. Considering the direct relation of V_a and V_p , it is believed that the V_p values at 1573 and 1673 K are also independent of strain. The DRV progress, which can be accelerated by increasing the temperature, should be responsible for this. As described in Section 3.1, the dislocation climb mechanism is more dominant at high temperature, as well as the quick rate approaching to dynamic balance of dislocation generation and annihilation.

The result inconsistency for transient and constant strain rate tests, as shown in Figs. 6–9, can be ascribed to the accuracy of experiments and the different data processing methods. In summary, the tendency of V_p with plastic strain and temperature and the magnitude of $(120\text{--}700) b^3$ suggest that the rate controlling mechanism may be the overcoming of forest dislocations [25–27].

It should be noted that, LIN et al [25] thought that the Taylor factor f , which converts normal stress into shear stress in polycrystals, should not been introduced in high temperature range because the driving force of dislocation climbing derives from normal stress rather than shear stress. If f is removed from Eqs. (5)–(10), all the V_a and V_p values should be three times lower. That is, the magnitude of V_p will fluctuate within $(40\text{--}200) b^3$.

A complete understanding of the dislocation motion mechanism needs more accurate transient experiments as well as the microstructure investigation in future.

4 Conclusions

1) Following the phenomenological hyperbolic sine constitutive model, the apparent stress exponent n and apparent activation energy Q_a are determined as 6.02 and 637.13 kJ/mol, respectively, for V–5Cr–5Ti alloy (mass fraction, %) compressed at 1423–1673 K using the peak stresses.

2) Following the power law equation, the stress exponent n and activation energy Q are obtained as 4.87 and 375.89 kJ/mol, respectively, by normalizing the

strain rate and stress with the temperature dependence of self-diffusion coefficient $D(T)$ and elastic modulus $E(T)$.

3) The physical activation volume V_p and strain rate sensitivity m increase with increasing the temperature. The typical values of V_p and m are about $(120\text{--}700) b^3$ and 0.075–0.122, respectively. These activation parameters indicate that the rate controlling mechanism may be the dislocation climb by overcoming of forest dislocations.

Acknowledgments

The authors gratefully acknowledge Rong MA and Yong-xin BAO for the help of compression tests.

References

- [1] KURTZ R J, HAMILTON M L. Biaxial thermal creep of V–4Cr–4Ti at 700 °C and 800 °C [J]. *Journal of Nuclear Materials*, 2000, 283–287: 628–632.
- [2] KURTZ R J, ABE K, CHERNOV V M, HOELZER D T, MATSUI H, MUROGA T, ODETTE G R. Recent progress on development of vanadium alloys for fusion [J]. *Journal of Nuclear Materials*, 2004, 329–333: 47–55.
- [3] FUKUMOTO K, YAMAMOTO T, NAKAO S, TAKAHASHI S, MATSUI H. High temperature performance of highly purified V–4Cr–4Ti alloy, NIFS-Heat1 [J]. *Journal of Nuclear Materials*, 2002, 307–311: 610–614.
- [4] FUKUMOTO K, NAGASAKA T, MUROGA T, NITA N, MATSUI H. Creep mechanism of highly purified V–4Cr–4Ti alloys during thermal creep in a vacuum [J]. *Journal of Nuclear Materials*, 2007, 367–370: 834–838.
- [5] NATESAN K, SOPPET W K, PUROHIT A. Uniaxial creep behavior of V–4Cr–4Ti alloy [J]. *Journal of Nuclear Materials*, 2002, 307–311: 585–590.
- [6] GROSSBECK M L. Creep of V–4Cr–4Ti in a lithium environment [J]. *Journal of Nuclear Materials*, 2002, 307–311: 615–619.
- [7] LI Zeng-de, CUI Shun, LIN Chen-guang, LI Ming. Recrystallization and kinetics of V–5Cr–5Ti alloy [J]. *Chinese Journal of Rare Metals*, 2009, 33(6): 785–789. (in Chinese)
- [8] GUBBI A N, ROWCLIFFE A F. Recovery and recrystallization of vanadium alloys [J]. *Journal of Nuclear Materials*, 1996, 233–237: 497–501.
- [9] CAILLARD D, MARTIN J L. Thermally activated mechanisms in crystal plasticity [M]. Oxford: Elsevier, 2003.
- [10] DUQUESNES V, GUILBERT T, MARION L F. French investigation of a new V–4Cr–4Ti grade: CEA-J57–fabrication and microstructure [J]. *Journal of Nuclear Materials*, 2013, 426: 96–111.
- [11] LI Yu-fei, WANG Zhen-hong, ZHANG Lin-ying, LUO Chao, LAI Xin-chun. Arrhenius-type constitutive model and dynamic recrystallization behaviors of V–5Cr–5Ti alloy during hot compression [J]. *Transactions of Nonferrous Metals Society of China*, 2015, 25(6): 1889–1900.
- [12] MIRZADEH H, CABRERA J M, NAJAFIZADEH A. Constitutive relationships for hot deformation of austenite [J]. *Acta Materialia*, 2011, 59(16): 6441–6448.
- [13] MIRZADEH H, CABRERA J M, NAJAFIZADEH A. Modeling and prediction of hot deformation flow curves [J]. *Metallurgical and Materials Transactions A*, 2012, 43(1): 108–123.
- [14] LIN Y C, CHEN X M. A critical review of experimental results and constitutive descriptions for metals and alloys in hot working [J]. *Materials and Design*, 2011, 32: 1733–1759.

- [15] MCQUEEN H J, RYAN N D. Constitutive analysis in hot working [J]. Materials Science and Engineering A, 2002, 322: 43–63.
- [16] MEIMEI LI, ZINKLE S J. Fracture mechanism maps in unirradiated and irradiated metals and alloys [J]. Journal of Nuclear Materials, 2007, 361: 192–205.
- [17] BYSTRZANOWSKI S, BARTELS A, CLEMENS H, GERLING R. Characteristics of the tensile flow behavior of Ti–46Al–9Nb sheet material-analysis of thermally activated processes of plastic deformation [J]. Intermetallics, 2008, 16: 717–726.
- [18] TYUMENTSEV A N, KOROTAEV A D, PINZHIN Y P, DITENBERG I A, LITOVCHENKO S V, HUBA Y V, SHEVCHENKO N V, DROBISHEV V A, POTAPENKO M M, CHERNOV V M. Effect of the modes of thermomechanical treatment on the formation of the multi-phase and grain structure of V–4Ti–4Cr alloys [J]. Journal of Nuclear Materials, 2004, 329–333: 429–433.
- [19] MA Z Y, TJONG S C. Creep deformation characteristics of discontinuously reinforced aluminium-matrix composites [J]. Composites Science and Technology, 2001, 61: 771–786.
- [20] MARTIN J L, KRUML T. Characterizing thermally activated dislocation mobility [J]. Journal of Alloys and Compounds, 2004, 378: 2–12.
- [21] WEI Q, CHENG S, RAMESH K T, MA E. Effect of nanocrystalline and ultrafine grain sizes on the strain rate sensitivity and activation volume: fcc versus bcc metals [J]. Materials Science and Engineering A, 2004, 381: 71–79.
- [22] PICU R C, VINCZE G, OZTURK F, GRACIO J J, BARLAT F, MANIATTY A M. Strain rate sensitivity of the commercial aluminum alloy AA5182-O [J]. Materials Science and Engineering A, 2005, 390: 334–343.
- [23] WANG Y M, HAMZA A V, MA E. Temperature-dependent strain rate sensitivity and activation volume of nano-crystalline Ni [J]. Acta Materialia, 2006, 54: 2715–2726.
- [24] KLAY E, DIOLOGENT F, DURUSSEL A, MORTENSEN A. Thermally activated deformation of two- and three-variant nanotwinned L1₀ Au–Cu–Pt [J]. Intermetallics, 2011, 19: 988–996.
- [25] LIN D L, WANG Y, LAW C C. Thermal activation processes of tensile deformation in γ -TiAl alloy [J]. Materials Science and Engineering A, 1997, 239–240: 369–377.
- [26] LU L, ZHU T, SHEN Y F, DAO M, LU K, SURESH S. Stress relaxation and the structure size-dependence of plastic deformation in nanotwinned copper [J]. Acta Materialia, 2009, 57: 5165–5173.
- [27] CONRAD H. Grain-size dependence of the flow stress of Cu from millimeters to nanometers [J]. Metallurgical and Materials Transactions A, 2004, 35: 2681–2695.

热压缩条件下 V–5Cr–5Ti 合金的热激活参数

李鱼飞¹, 王震宏², 周运洪², 罗超², 赖新春¹

1. 表面物理与化学重点实验室, 绵阳 621907;

2. 中国工程物理研究院, 绵阳 621900

摘 要: 为了了解 V–5Cr–5Ti 合金(质量分数) 热加工过程的基本物理机制, 在 Gleeble-3800 热物理模拟试验机上采用热压缩方法测量合金在 1373~1673 K 范围内的热激活参数。结果表明: 抛物线型本构方程的应力指数 n 和激活能 Q 分别为 4.87 和 375.89 kJ/mol。通过引入临界应力参数(σ_0), 计算得到的激活能 Q 为 288.34 kJ/mol, 这与合金的自扩散激活能(270~300 kJ/mol)相近。重复应力松弛实验得到合金的物理激活体积 V_p 和应变速率敏感指数 m 分别为(120~700) b^3 和 0.075~0.122。上述激活参数表明: V–5Cr–5Ti 合金在 1373~1673 K 和 0.001~1.0 s⁻¹ 范围内的热压缩物理机制为克服林位错钉扎的位错攀移。

关键词: V–5Cr–5Ti 合金; 热压缩; 热激活; 位错攀移

(Edited by Mu-lan QIN)