RAPID COMMUNICATION

A universal function of creep rate*

Li Jing-Tian(李菁田)[a], Rong Xi-Ming(荣曦明)[b], Wang Jian-Lu(王建录)[c], Zhang Bang-Qiang(张邦强)[c], and Ning Xi-Jing(宁西京)[a]†

[a] *Institute of Modern Physics, Department of Nuclear Science and Technology, Fudan University, Shanghai* 200433, *China*
[b] *Department of Optical Science and Engineer, Fudan University, Shanghai* 200433, *China*
[c] *Dongfang Turbine Co. Ltd., Deyang* 618000, *China*

(Received 4 January 2015; revised manuscript received 4 May 2015; published online 3 August 2015)

In this paper, we derive a universal function from a model based on statistical mechanics developed recently, and show that the function is well fitted to all the available experimental data which cannot be described by any function previously established. With the function predicting creep rate, it is unnecessary to consider which creep mechanism dominates the process, but only perform several experiments to determine the three constants in the function. It is expected that the new function would be widely used in industry in the future.

Keywords: creep, statistical mechanics, metal and alloys

PACS: 34.20.Cf, 62.20.Hg, 81.05.Bx **DOI:** 10.1088/1674-1056/24/9/093401

1. Introduction

Creep is the progressive time-dependent plastic deformation of metals under load and elevated temperature conditions,[1] and the rate of creep determines the service lifetime of given materials in engineering. However, theoretically predicting the rate of creep without empirical data is impossible[1–3] and experimental measurement is time consuming because creep resistance is defined by a stress level required to produce a nominal strain, say 0.1%, 0.2%, or 0.5% in periods of about 100000 h.[2,4,5] In fact, creep rate depends not only on the applied stress but also on the ambient temperature, while the stress and temperature may differ greatly in different usages. Thus, to evaluate the creep lifetime of a given material, the creep rate test should cover the combination of various stresses and temperatures entirely. Clearly, this is an impossible task and experimental measurements of creep rate have to be limited to constant stress and constant temperature.[6–12]

If a universal function can be established to describe the dependence of creep rate at stress and temperature, the creep rate of any condition can be obtained by only several measurements performed at limited points of stresses and temperatures. Since the 1920s, much theoretical effort has been made to find such a function,[10,13–16] and several equations are established as follows:

$$\dot{\varepsilon} = A_1 \sigma_n, \tag{1}$$

$$\dot{\varepsilon} = A_2 \exp(B\sigma), \tag{2}$$

$$\dot{\varepsilon} = A_3 (\sinh \alpha\sigma)^n, \tag{3}$$

$$\dot{\varepsilon} = A_4 \exp\left(-\frac{Q_c}{RT}\right), \tag{4}$$

$$\dot{\varepsilon} = A_5 (\sinh \alpha\sigma)^n \exp\left(-\frac{Q_c}{RT}\right), \tag{5}$$

$$\dot{\varepsilon} = A_6 \sigma^n \exp\left(-\frac{Q_c}{RT}\right), \tag{6}$$

where $\dot{\varepsilon}$, σ, Q_c, R, and T are the minimum creep rate, the applied stress, the apparent creep activation energy, the gas constant, and the absolute temperature, respectively, while A_s ($s = 1, 2, \ldots, 6$), B, and n are constant for a given material.

In 1965, Monkman and Grant found that the creep rupture time t_{rup} multiplied by $\dot{p}\varepsilon$ is a constant for many metals:[17]

$$t_{rup}\dot{\varepsilon} = \text{constant}. \tag{7}$$

Based on Eq. (7), the Larson–Miller parameter method[18] was developed from Eq. (4) to predict the creep life and is being used widely to predict long-term creep life from short-term measurement data via

$$\text{LMP} = T\left[\log t_{rup} + C_{LM}\right], \tag{8}$$

where LMP depends on the applied stress.[19] If the units of T and t_{rup} are Kelvin and hour, respectively, the value of C_{LM} was found to be around 20 for many materials and was expected to be the same for different materials.[20] However, recent experimental works showed that the value varies obviously for different materials.[21–26] Even for a given material, it is shown via Eq. (5) that the value of C_{LM} is proportional to Q_c with high correlation coefficient rather than a constant.[27] According to Eq. (6), Wilshire *et al.*[4,5,28] showed that C_{LM} depends on Q_c, and the extrapolation always overestimates long-term

*Project supported by the National Natural Science Foundation of China (Grant Nos. 11274073 and 51071048), the Shanghai Leading Academic Discipline Project, China (Grant No. B107), and the Key Discipline Innovative Training Program of Fudan University, China.

†Corresponding author. E-mail: xjning@fudan.edu.cn

© 2015 Chinese Physical Society and IOP Publishing Ltd

performance with $C_{\rm LM}$ being a constant. For instance, if only the short-term data for creep measurements within 1000 h are applied to determine the constant of Eq. (8), then the prediction of the rupture strength for P92 alloys at 600 °C by extrapolation to 100000 h is 180 MPa. However, if longer measurements data (> 3000 h) are used in Eq. (8), the prediction under the same conditions changes to 128 MPa, obviously smaller than the former prediction.[29]

If a creep rate equation, such as the creep constitute equation [Eq. (6)], is universal for different materials, $C_{\rm LM}$ can be reasonably derived via Eq. (7) and the prediction of rupture time will be reliable. However, previous experiments showed that none of Eqs. (1)–(6) are universal. Taking the creep constitute equation as an example, recent experiments showed that n and $Q_{\rm c}$ depend obviously on stress and temperature in some cases.[5,8,30,31] For example, the value of n drawn from the experiment for K435[30] increases from 3.8 to 11.0 when the stress varies from 140 MPa to 250 MPa, while for polycrystalline copper[10] (or AZ31[31]) the values increase from 1 to 4.5 (or 3 to 5) when the stress varies, and for IN738LC[30] (or TC11 alloy[8]) the value of $Q_{\rm c}$ increases from 420 kJ/mol to 730 kJ/mol (or 67 kJ/mol to 196 kJ/mol) with the temperature increasing. The experimental data for K435 and IN738LC are shown in Fig. 1, where the values of n and $Q_{\rm c}$ fitted by Eq. (6) are also presented. Clearly, a universal and reliable function of creep rate upon the stress and temperature is imperative for predicting creep life.

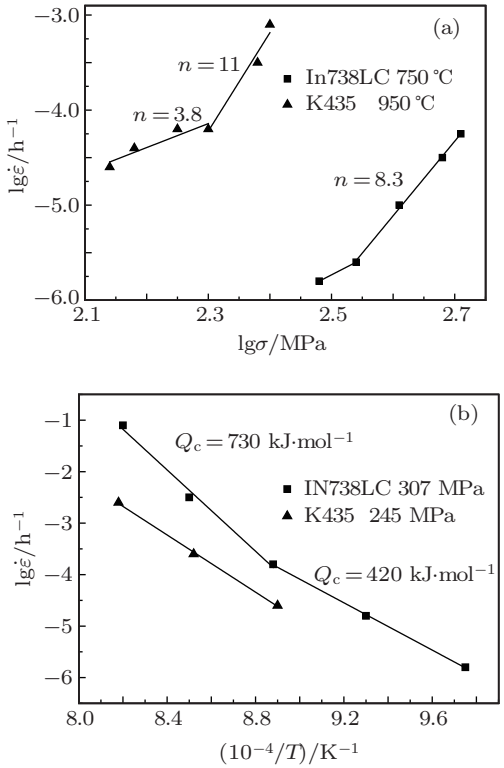


Fig. 1. The experimental $\dot{\varepsilon}$–σ curves (a) and $\dot{\varepsilon}$–T curves (b) for In738LC and K435 alloys.

In our previous works, a model based on statistical mechanics was developed successfully to predict the stability of materials, atom diffusion rates, etc.[32–38] In the present work, it was developed to predict the creep rates and a universal function was derived. In order to test the function, we measured the creep rates of three kinds of materials 10305DA, 10305BA, 10705BU, and collected available experimental data of other materials (IN738LC,[30] K435,[30] NiAl-25Cr,[39] NiAl-28Cr-5Mo,[6] Ti-45Al-2Mn-2Nb,[7] AS31,[9] TC11,[8] Polycrystalline Cu,[10] FGH95,[12] TMW2,[40] 1.25Cr0.5Mo[41]). It was shown that the experimental data are in good agreement with our new function.

2. Model and method

Creep mechanism can be classified as diffusion and dislocation.[42] In any case, when stress is applied on bulk metals, some atoms in a potential valley must cross over a series of barriers of height E_0 to reach another valley along the direction of the applied force [Fig. 2], otherwise creep plastic deformation cannot take place. In the absence of applied stress, thermal motion can also make the atoms cross over the barrier with a small probability, but this motion does not lead to macro-deformation because the motion is spacially isotropic. In what follows, we describe how the isotropic motion is affected by the applied stress to result in creep and give the expression of creep rate.

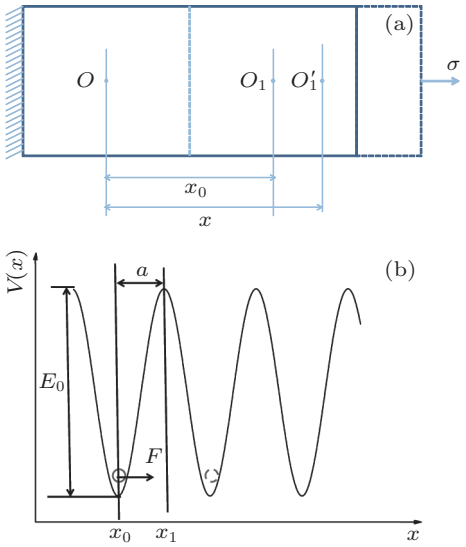


Fig. 2. (color online) Schematic diagram of creep caused by a stress σ applied along the x direction (a) and the potential barriers felt by an atom (b).

According to a statical model of a single atom mentioned in Refs. [32]–[38], if we observe the motion of an atom in a potential valley in one time unit [Fig. 2(b)], the total time for the atom with kinetic energy ε larger than the static barrier E_0 is

$$t = \frac{1}{Z}\int_{E_0}^{\infty} \varepsilon^{1/2} \mathrm{e}^{-\varepsilon/k_{\rm B}T}\,\mathrm{d}\varepsilon, \tag{9}$$

where $Z=\sqrt{\pi}(k_{\rm B}T)^{3/2}/2$ (the partition function) and T is the temperature. Assuming that all the hopping events start when the atom stays in the center of the potential well described by $V(x)$ along the minimum energy path, which should be the most probable, the time taken by the atom to cross over the barrier is

$$\delta t=\sqrt{m}\int_{x_0}^{x_1}{\rm d}x/\sqrt{2(\varepsilon-V(x))}, \tag{10}$$

which results from Newton's law for the atom moving in the potential well. The averaged period can be obtained via

$$\overline{\delta t}=\frac{\int_{E_0}^{\infty}(\delta t)\varepsilon^{1/2}{\rm e}^{-\varepsilon/k_{\rm B}T}{\rm d}\varepsilon}{\int_{E_0}^{\infty}\varepsilon^{1/2}{\rm e}^{-\varepsilon/k_{\rm B}T}{\rm d}\varepsilon}. \tag{11}$$

Hence, the averaged hopping frequency $t/\overline{\delta t}$ is given by

$$\Gamma=\frac{1}{Z}\left(\frac{\int_{E_0}^{\infty}\varepsilon^{1/2}{\rm e}^{-\varepsilon/k_{\rm B}T}{\rm d}\varepsilon}{\int_{E_0}^{\infty}(\delta t)\varepsilon^{1/2}{\rm e}^{-\varepsilon/k_{\rm B}T}{\rm d}\varepsilon}\right)\int_{E_0}^{\infty}\varepsilon^{1/2}{\rm e}^{-\varepsilon/k_{\rm B}T}{\rm d}\varepsilon. \tag{12}$$

When a stress σ is applied along the x direction, the atom feels a force F, and work of the force is Fa if the atom moves from x_0 to x_1 [Fig. 2], leading to the increase of the total energy from ε up to $\varepsilon+Fa$. If the total energy $\varepsilon>E_0-Fa$, the atom is able to cross over the barrier E_0. Thus, in one time unit, the total time for the atom having energy larger than E_0 is

$$t=\frac{1}{Z}\int_{E_0-Fa}^{\infty}\varepsilon^{1/2}{\rm e}^{-\varepsilon/k_{\rm B}T}{\rm d}\varepsilon. \tag{13}$$

Correspondingly, the time taken by the atom to cross over the barrier (from x_0 to x_1) is

$$\delta t=\sqrt{m}\int_{x_0}^{x_1}{\rm d}x/\sqrt{2(\varepsilon-V(x)+Fa)}, \tag{14}$$

and the average period can be obtained via

$$\overline{\delta t}=\frac{\int_{E_0-Fa}^{\infty}(\delta t)\varepsilon^{1/2}{\rm e}^{-\varepsilon/k_{\rm B}T}{\rm d}\varepsilon}{\int_{E_0-Fa}^{\infty}\varepsilon^{1/2}{\rm e}^{-\varepsilon/k_{\rm B}T}{\rm d}\varepsilon}, \tag{15}$$

so the average hopping frequency is given by

$$\Gamma=\frac{1}{Z}\left(\frac{\int_{E_0-Fa}^{\infty}\varepsilon^{1/2}{\rm e}^{-\varepsilon/k_{\rm B}T}{\rm d}\varepsilon}{\int_{E_0-Fa}^{\infty}(\delta t)\varepsilon^{1/2}{\rm e}^{-\varepsilon/k_{\rm B}T}{\rm d}\varepsilon}\right)\times\int_{E_0-Fa}^{\infty}\varepsilon^{1/2}{\rm e}^{-\varepsilon/k_{\rm B}T}{\rm d}\varepsilon, \tag{16}$$

and the creep rate is expressed as

$$\dot{\varepsilon}=ka\Gamma, \tag{17}$$

where k is a constant related to microstructures of the specific materials, such as the density of defects (dislocations or grain boundaries). In principle, if the density of defects and the microstructures are exactly obtained in advance, then the diffusion potential function $V(x)$ can be obtained theoretically, and the creep rate can be exactly determined by Eq. (17). Unfortunately, this method is time-consuming because the potential functions may be very different for different defects with different microstructures.

For applications of the above equations in practice, the creep of a column shaped material of mass M can be regarded as the departure of two mass centers [Fig. 2(a)] with each of mass $M/2$, and an average potential that the mass center feels is described by

$$V(x)=\frac{E_0}{2}\left[1-\cos\frac{\pi}{a}(x-x_0)\right], \tag{18}$$

where E_0 and a are the potential barrier and the half width of the potential valley, respectively. Based on the statics model,[32–38] equations (13)–(17) are still valid for the mass center. Thus, we can determine the three constants, E_0, a, and k by measuring creep rate at several temperatures and stress points, and then apply Eq. (17) to predict creep life for any temperature and stress.

3. Result and discussion

There are two methods to validate the accuracy and the universality of our model. The first method is to check whether E_0, a, and k maintain as constants and are independent of stress and temperature when equation (17) is applied to fit the available experimental data of the creep rate. The second method is to check whether equation (17) with E_0, a, and k is unchanged because a given material can reproduce previous creep rate equations, such as Eqs. (1) and (2), which present different forms for the same material under different loads (or temperature) conditions.

To validate the accuracy of our model sufficiently, we collected a large number of experimental data of creep rates[6,8–10,30,40,41] upon stress σ and temperature T, and the curves of $\dot{\varepsilon}$–σ and $\dot{\varepsilon}$–T are shown in Fig. 3. In addition, we performed creep tests of 10305DA (42CrMo), 10705BA (2Cr12NiW1Mo1V), 10705BU (1Cr12Mo) in air with different loads and temperatures. The alloy ingot was cut into several pieces of cylinder with a diameter of 10 mm and a gauge length of 100 mm. Each test cost several thousand hours until the failure point. The experimental data are listed in Tables 1–3, while the $\dot{\varepsilon}$–σ and $\dot{\varepsilon}$–T curves are shown in Figs. 3(a) and 3(b), respectively. When fitting the experimental data with Eq. (17), the force F applied on a single atom within the material was regarded as the average of the total force applied on the vertical cross section divided by all the atoms in the

cross section. As shown in Fig. 3, our model can reproduce all the experimental data with E_0, a, and k (listed in Table 4) unchanged for given materials. The potential barriers for these materials are $\sim$ eV, which approach the apparent creep activation energy Q_c drawn from previous experiments fitted with Eq. (4).[6,8–10,30,40,41] The half widths of the potential valley obtained from our model are of tens of nanometers which is in agreement with a creep mechanism by Seitz, who proposed that isolated vacancies may condense to form plates or disks[43] in the process of plastic deformation and Dash estimated that the diameter of the disks is about 15 nm according to experimental data.[44] This means that the atoms in the disks would feel a potential valley of tens of nanometers in width when creep occurs.

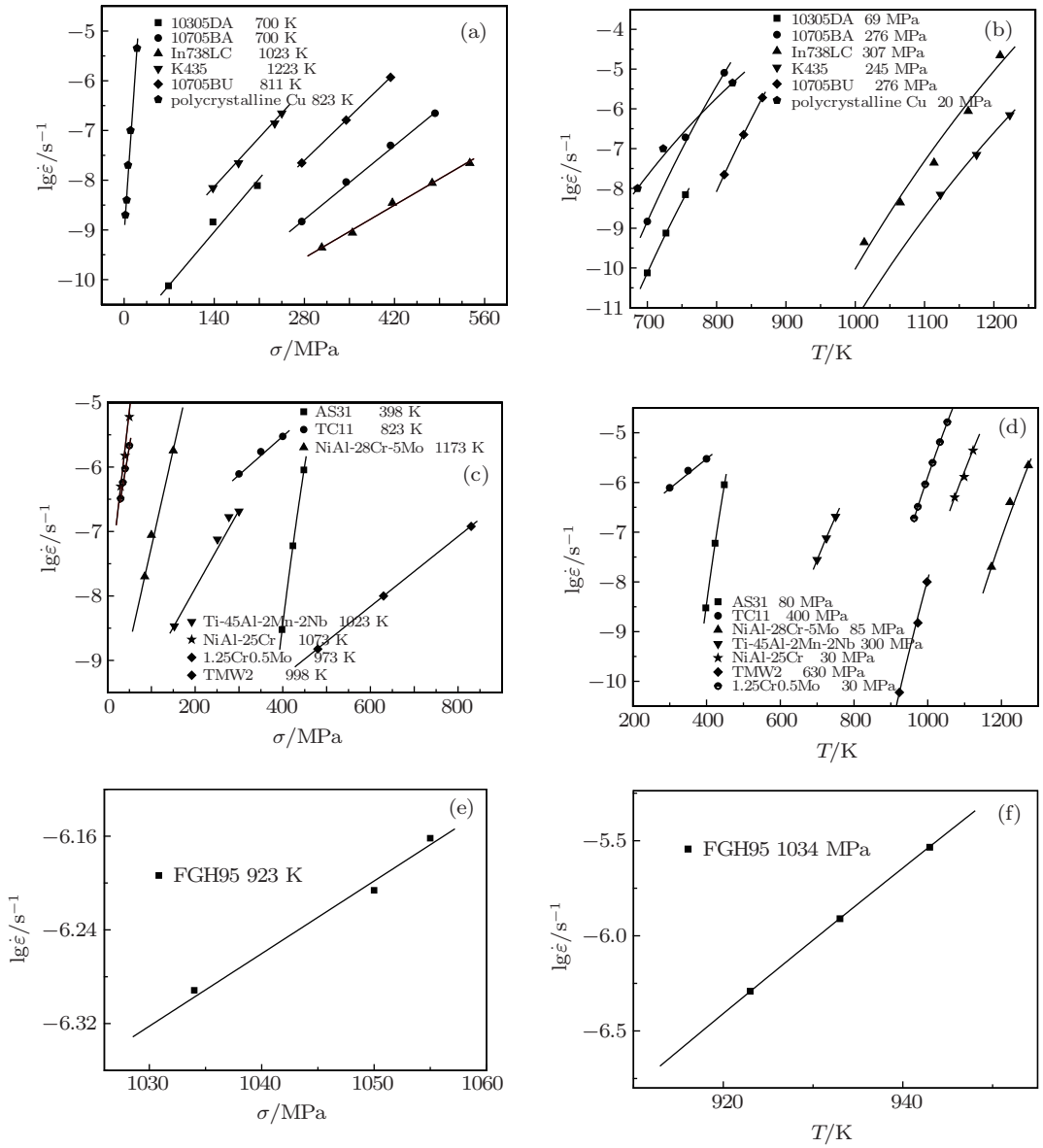


Fig. 3. The experimental data of $\dot{\varepsilon}$–σ ((a), (c), and (e)) and $\dot{\varepsilon} - T$ ((b), (d), and (f)) (black dots, squares, triangles, inverted triangles, rhombi, pentagons) and the curves (black line) obtained by fitting Eq. (17) to 10305DA , 10705BA, 10705BU, IN738LC, K435, NiAl-25Cr, NiAl-28Cr-5Mo, Ti-45Al-2Mn-2Nb, AS31, TC11, Polycrystalline Cu, FGH95, TMW2, 1.25Cr0.5Mo.

Table 1. The experimental creep rates of 10305DA (42CrMoA) alloy measured in the present work.

Stress/Mpa	69	Temperature/°C	427
Temperature/°C	Steady creep rate/(%/h)	Stress/Mpa	Steady creep rate/(%/h)
427	2.70×10^{-7}	69	2.70×10^{-7}
454	2.70×10^{-6}	138	5.20×10^{-6}
482	2.50×10^{-5}	207	2.80×10^{-5}

Table 2. The experimental creep rates of 10705BU (2Cr12NiW1Mo1V) alloy measured in the present work.

Stress/Mpa	276	Temperature/°C	538
Temperature/°C	Steady creep rate/(%/h)	Stress/Mpa	Steady creep rate/(%/h)
538	8.00×10^{-5}	276	8.00×10^{-5}
566	8.10×10^{-4}	345	6.20×10^{-4}
593	6.90×10^{-3}	414	4.30×10^{-3}

Table 3. The experimental creep rates of 10705BA (1Cr12Mo) alloy measured in the present work.

Stress/Mpa	276	Temperature/°C	427
Temperature/°C	Steady creep rate/(%/h)	Stress/Mpa	Steady creep rate/(%/h)
427	5.30×10^{-6}	276	5.30×10^{-6}
482	6.90×10^{-4}	345	3.30×10^{-5}
538	2.90×10^{-2}	414	1.80×10^{-4}
		483	8.00×10^{-4}

Table 4. The average potential barriers of various kinds of alloys.

Alloy types	Potential barrier E_0/eV	Half width a/Å	k
10305DA	4.0	25	1.25×10^{23}
10705BA	4.23	21.5	5.47×10^{22}
10705BU	5.5	29.3	3.20×10^{28}
IN738LC	6.45	22.4	3.02×10^{20}
K435	6.5	50	6.34×10^{18}
NiAl-25Cr	5.0	185	4.17×10^{14}
NiAl-28Cr-5Mo	6.7	103	2.32×10^{17}
Ti-45Al-2Mn-2Nb	4.22	36.2	1.11×10^{9}
FGH95	7.8	16.5	6.71×10^{27}
AS31	2.2	76.5	9.84×10^{12}
TC11	2.0	14.8	1.21×10^{13}
Polycrystalline Cu	2.82	451	9.57×10^{7}
TMW2	6.15	15.7	4.66×10^{17}
1.25Cr0.5Mo	4.6	115	1.37×10^{15}

It is notable that some of the traditional creep rate equations correspond to different creep mechanisms. For example, the stress power law, described by Eq. (1), is considered to correspond to the climb-controlled creep process when the load is small, while equation (2) is applicable to glide-controlled creep under a large load. Obviously, even if the parameters A_1 and n of Eq. (1) are determined by some experiments under certain service conditions, we cannot use the equation to predict the creep rate for other service conditions because the value of n may change obviously. For example, the value of n for polycrystalline copper increases from 1 to 4.5 even if the temperature remains unchanged but the stress increases only from 2 MPa to 20 MPa, which is attributed to the transition of the creep mechanism from diffusion to dislocation.[10] For IN738LC, the value of n increases from 3.3 to 8.3 when the creep mechanism transits from climb-plus-glide to Orawan mechanism.[45]

Using our model, it is unnecessary to distinguish which mechanism dominates the creep and we can reproduce Eqs. (1)–(6) if specific ranges of stress and temperature are given. For example, the experimental $\dot{\varepsilon}$–σ curves for high pure aluminium with stress below 10 MPa can be well described by Eq. (1), but this power law breaks down when the stress surpasses 10 MPa.[46] Thus, when the stress changes from 5 MPa to 20 MPa, the experimental $\dot{\varepsilon}$–σ curves shown in Fig. 4 satisfies neither Eq. (1) nor Eq. (2). Nevertheless, the experimental $\dot{\varepsilon}$–σ curve at 598 K can be well fitted (Fig. 4) by Eq. (17), with the values of E_0, a, and k determined to be 1.6 eV, 38.7 nm, and 1.12×10^7, respectively. Furthermore, using these constants, we can reproduce the experimental $\dot{\varepsilon}$–σ at 613 K for the same stress range (Fig. 4).

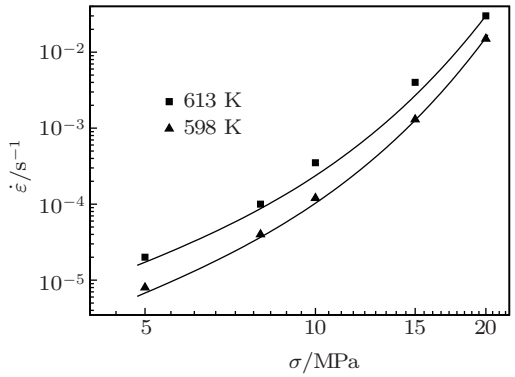


Fig. 4. The experimental $\dot{\varepsilon}$–σ curves for high pure aluminium at 598 K (black triangles) and 613 K (black squares), which can be well fitted (solid line) by Eq. (17) with the same values of E_0, a, and k.

4. Conclusion

In summary, a universal function was established to predict creep rate upon any stress and temperature and the validity was confirmed by available experimental data, especially by the fact that the function can reproduce all the previous creep rate equations obtained empirically for specific service conditions. In the application of the function to predict creep rate, we need not consider which creep mechanism dominates the process, but only perform several experiments to determine the three constants in the function. It is expected that the new function would be widely used in industry in the near future.

References

[1] Babitsky V I and Wittenburg J 2007 *Modeling of Creep for Structural Analysis* (Berlin: Springer)
[2] Penny R K and Mariott D L 1995 *Design for Creep* (London: Chapman & Hall)
[3] Evans R W and Wilshire B 1985 *Creep of Metals and Alloys* (London: Institute of Materials)
[4] Wilshire B and Scharning P J 2007 *Scripta Mater.* **56** 701
[5] Wilshire B and Scharning P J 2008 *Int. Mater. Rev.* **53** 91
[6] Guo J T, Zhang S Z, Yuan C, Zhou W L and Li G S 2003 *J. Aeronautical Mater. Suppl.* **23** 34
[7] Kral P, Dvorak J, Zherebtsov S, Salishchev G, Kvapilova M and Sklenicka V 2013 *J. Mater. Sci.* **48** 4789
[8] Gu Y, Zeng F H, Qi Y L, Xia C Q and Xiong X 2013 *Mater. Sci. Eng. A* **575** 74
[9] Zhu S M, Easton M A, Gibson M A, Dargusch M S and Nie J F 2013 *Mater. Sci. Eng. A* **578** 377
[10] Wilshire B and Battenbough A J 2007 *Mater. Sci. Eng. A* **443** 156
[11] Abdallah Z, Whittaker M T and Bache M R 2013 *Intermetallics* **38** 55
[12] Xie J, Tian S G and Zhou X M 2013 *JMEPEG* **22** 2048
[13] Norton F H 1929 *The Creep of Steel at High Temperature* (London: McGraw-Hill)
[14] McVetty P G 1943 *Trans. ASME* **65** 761
[15] Garofalo F 1965 *Fundamentals of Creep and Creep Rupture in Metals* (New York: MacMillan)
[16] Dorn J E 1955 *J. Mech. Phys. Solids* **3** 85
[17] Grant N J and Mullendore A W 1965 *Deformation and Fracture at Elevated Temperatures* (Boston: MIT Press)
[18] Larson F R and Miller J 1952 *Trans. ASME* **74** 765
[19] Kassner M E and Perez-Prado M T 2004 *Fundamentals of Creep in Metals and Alloys* (Amsterdam: Elsevier)
[20] Maruyama K, Armaki H G and Yoshimi K 2007 *J. Pressure Vessels Piping* **84** 171
[21] Nie M, Zhang J, Huang F, Liu J W, Zhu X K, Chen Z L and Ouyang L Z 2014 *J. Alloys Compd.* **588** 348
[22] Brunnera M, Huttner R, Bolitz M, Volkl R, Mukherji D, Rosler J, Depka T, Somsen C, Eggeler G and Glatzel U 2010 *Mater. Sci. Eng. A* **528** 650
[23] Shrestha T, Basirat M, Charit I, Potirniche G P and Rink K K 2013 *Mater. Sci. Eng. A* **565** 382
[24] Terada Y, Murata Y, Sato T and Morinaga M 2011 *Mater. Chem. Phys.* **128** 32
[25] Terada Y, Murata Y and Sato T 2013 *Mater. Sci. Eng. A* **584** 63
[26] Maruyama K and Yoshimi K 2007 ASME 2007 *Pressure Vessels and Piping Conference* **9** 631
[27] Tamura M, Abe F, Shiba K, Sakasegawa H and Tanigawa H 2013 *Metall. Mater. Trans.* A **44** 2013
[28] Wilshire B, Scharning P J and Hurst R 2009 *Mater. Sci. Eng. A* **510–511** 3
[29] Ennis P J, Shibli I A, Holdsworth S R and Merckling G 2005 *Creep and Fracture in High Temperature Components—Design and Life Assessment Issues* (Lancaster: DEStech Publications)
[30] Guo J T 2008 *Materials Science and Engineering for Superalloys* (Beijing: Science press)
[31] Spingarelli S, Ruano O A, Mehtedi M E and del Valle J A 2013 *Mater. Sci. Eng. A* **570** 135
[32] Lin Z Z, Yu W F, Wang Y and Ning X J 2011 *Europhys. Lett.* **94** 40002
[33] Lin Z Z and Ning X J 2011 *Europhys. Lett.* **95** 47012
[34] Lin Z Z, Zhuang J and Ning X J 2012 *Europhys. Lett.* **97** 27006
[35] Ming C, Lin Z Z, Cao R G, Yu W F and Ning X J 2012 *Carbon* **50** 2651
[36] Lin Z Z, Li W Y and Ning X J 2014 *Chin. Phys. B* **23** 050501
[37] Yu W F, Lin Z Z and Ning X J 2013 *Phys. Rev. E* **87** 062311
[38] Yu W F, Lin Z Z and Ning X J 2013 *Chin. Phys. B* **22** 116802
[39] Zhang G Y, Guo J T and Zhang H 2006 *The Chinese Journal of Nonferrous Metals* **16** 1883
[40] Gu Y F, Cui C, Ping D, Harada H, Fukuda T and Fujioka J 2009 *Mater. Sci. Eng. A* **510–511** 250
[41] Vakili-Tahami F, Sajjadpour M and Attari P 2011 *Strain* **47** 414
[42] Bressers J 1981 *Creep and Fatigue in High Temperature Alloys* (London: Applied Science Publishers LTD)
[43] Seitz F 1950 *Phys. Rev.* **79** 890
[44] Dash W C 1958 *J. Appl. Phys.* **29** 705
[45] Guo J T, Ranucci D, Picco E and Strocchi M 1983 *Metall. Trans. A* **14** 2329
[46] Tobolová Z and Čadek J 1979 *Phil. Mag.* **26** 1419