Physical Meaning of the New Creep Rupture Equation for Heat Resisting Steels

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The authors have proposed a new creep rupture equation assuming a thermally activated process for dislocation motion. The activation energy and the activation volume of the new equation have been formulated on the base of a dislocation model. The residual dislocation density, ρ_r , the gliding distance of movable dislocations, g , and the equivalent obstacle spacing, d/p , where d is the obstacle spacing and p is a characteristic parameter of the obstacles, can be directly estimated using the creep rupture data with some assumptions but without any adjustable parameters. The values of d/p and g are comparable with the reported particle spacing and the width of sub-boundaries, respectively. The estimated dislocation densities are consistent with the observed values in the literature. The values of g decreases and d/p increases with increasing both test temperature and time to rupture. If dynamical aging and/or precipitation occur, d/p should be decreasing.

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1. Introduction

The steady state creep rate of the dislocation creep has been described generally by the product of the diffusion coefficient and the stress to the n^{th} power or exceptionally the exponential stress.¹⁾ The steady state creep rate is inversely proportional to the time to rupture.²⁾ Therefore, the logarithm of the steady state creep rate or the time to rupture has been plotted as a function of the logarithm of the stress without any exception and the long term creep rupture strength is estimated by extrapolation using these diagrams. However, the following problems on the previous creep equations have not yet been solved; (i) the stress exponents for commercially available steels are widely ranging through 3 to 20,³⁾ in spite of some theoretical approaches to explain the stress exponent,⁴⁾ and (ii) though the theoretical activation energy for creep is comparable to that of the self diffusion,¹⁾ the activation energy observed for commercially available steels is generally much larger than that of self diffusion.³⁾

Recently, the authors have proposed the following new creep rupture equation for commercially available heat resisting steels based on a thermally activated process considering the decrease in free energy due to the load drop.⁵⁾ The time to rupture, t_r , can be expressed as

$$t_r = t_{r0} \exp \left(\frac{Q - \sigma V}{RT} \right), \quad (1)$$

where σ , T , R , Q , V and t_{r0} are the applied stress, the testing temperature in K, the gas constant, the activation energy in the absence of the applied stress, the activation volume and the constant, respectively. The equation has been phenomenologically verified by the very long time rupture data of the commercially available heat resisting steels beyond 100000 h. The values of Q and V can be calculated as a partial derivative of $\ln(t_r/t_{r0})$ with respect to $1/T$ and σ , respectively.

In this study, the activation volume and the activation energy of the new creep equation have been formulated assuming a simple dislocation model and the physical meaning of these values have been investigated.

2. Theory

2.1 Dislocation model

The rupture strength of the commercially available heat resisting steels under actual service conditions is controlled mainly by the dislocation creep and the steels are strengthened mainly by precipitates and/or dislocations.³⁾ In the ferritic or martensitic steels, dislocations induced after austenitization usually form the lath boundaries or sub-boundaries during the subsequent heat treatment and/or creep deformation⁶⁾ and the size of these sub-structures are directly related to the creep strength.⁷⁻¹¹⁾ Besides these sub-structures, precipitates in the austenitic or ferritic matrix act as an obstacle against a moving dislocation. If the precipitates are the second phase having interaction with a dislocation, they can pin a moving dislocation. Consequently, the dislocation pinned by the precipitates behaves as a new obstacle against succeeding dislocations. Therefore, the dislocations pinned by the precipitates and/or in the lath or sub-boundaries can be assumed to be the essential obstacles against a moving dislocation when we consider macroscopic deformation.

We usually observed dislocations pinned by precipitates, tangled dislocations and dislocation network under a transmission electron microscope. However, for simplicity, a dislocation model as shown in Fig. 1 is considered in this study.

Imaging straight obstacle dislocations array with spacing of d along the moving direction of a dislocation. The length of the Burgers vector of the obstacle dislocation is assumed as pb , where b is the length of the unit Burgers vector and p is a parameter which can accommodate the difference between the model and the actual structure of the obstacles.

A straight movable dislocation lies the halfway between the obstacles and the dislocation approaches to the obstacle under the applied stress. The interaction energy increases and becomes the maximum. After that the movable dislocation climbs the first obstacle followed by gliding over the subsequent obstacles. The gliding dislocation stops moving at sink sites due to pinning by another obstacles or annihilating by encountering another dislocation with opposite sign. Lath boundaries, sub-boundaries and/or grain boundaries are im-

Fig. 1 Illustration for a dislocation model.

aged as a sink cite of the movable dislocation. The average gliding distance is defined as g .

The principle activation process mentioned above acts through out a specimen.

2.2 Activation volume

When the tensile load, P , is applied to a specimen of the cross section of S_0 and the gage length of L_0 , the specimen is elongated by ΔL_A through the activated process. The potential drop of the loading system per mole should be added to the activation free energy¹²⁾ as shown in eq. (1) and the value is given by

$$P \Delta L_A \frac{1}{S_0 L_0 / v_m} = \sigma \varepsilon_A v_m = \sigma V, \quad (2)$$

where v_m is the molar volume and $\varepsilon_A = \Delta L_A / L_0$ is defined as the activation tensile strain in this paper. Imaging that n dislocations are generating in a unit longitudinal section of the width of d and the height of h and that the dislocations glide a distance, g , the resultant shear strain is equal to $n \cdot (b/h) \cdot (g/d) = (n/hd) \cdot bg \equiv \rho bg$. Therefore, when the conversion factor from shear strain to normal strain is set to 0.5, the following equations are obtained from eq. (2).

$$V = \frac{1}{2} \rho bg v_m, \quad (3)$$

$$\varepsilon_A = \frac{1}{2} \rho bg, \quad (4)$$

and

$$\rho = \frac{n}{dh}. \quad (5)$$

The dislocation density, ρ , in eq. (5) denotes the number of dislocations generating in the cross section of $d \times h$ during the activation process and, thus, ρ means dislocations necessary to operate the activation process independent of time meaning neither the dislocations observed after the rupture test nor the number of dislocations necessary to creep a specimen.

Image a specimen of infinite thickness with the actual slip plane of the length in a direction of the Burger's vector, d_0 , which is comparable to the specimen width. Rupture elongation in a constant load creep test affected many factors, such

as shape of a specimen, the origin of necking and increasing in stress under a constant load test. On the other hand, the Monkman-Grant relation is well established for heat resisting steels, where time to rupture is inversely proportional to creep rate.¹³⁾ This implies that the rupture elongation is strongly affected not only by the origin of necking but also by the creep deformation itself. Plastic deformation can be macroscopically observed only when a lot of dislocations are swept out from a specimen as long as we assume the dislocation creep. Therefore, it can be assumed that only the dislocations in the range of g from the both side surfaces should contribute to rupture elongation. Other movable dislocations than those of the surface area stop moving at the sink cites and contribute to only hardening or softening, not to the macroscopic deformation.

Moreover, assuming the average glide distance in the surface range as $g/2$, the rupture elongation, ε_R , and the total dislocation density necessary to form rupture elongation, ρ_R , are given by the following equations.

$$\varepsilon_R = \frac{1}{2} \rho_R b g / 2 \quad (6)$$

and

$$\rho_R = \frac{2gn}{d_0 dh}. \quad (7)$$

If we define the parameter, α , as the ratio of ε_R to ε_A , following relations are obtained using eqs. (2)–(7).

$$\alpha \equiv \frac{\varepsilon_R}{\varepsilon_A} = \frac{\varepsilon_R}{V/v_m} = \frac{\rho_R}{2\rho} = \frac{g}{d_0}, \quad (8)$$

$$g = d_0 v_m \frac{\varepsilon_R}{V}, \quad (9)$$

$$\rho = \frac{2}{bd_0 v_m^2} \cdot \frac{V^2}{\varepsilon_R}, \quad (10)$$

and

$$\rho_R = \frac{4}{bd_0 v_m} V. \quad (11)$$

2.3 Activation energy

Main factors affecting Q are both the activation energy for climb and the elastic interaction energy between the movable dislocations and the obstacle dislocations. The increase in elastic energy and surface energy during creep is neglected as compared with the potential drop due to the plastic deformation. The activation energy for climb is assumed to be equal to that for self diffusion, Q_d . In case of alloy the definition of self diffusion is unclear. However, since the major constituent element for heat resisting steels is iron, Q_d is assumed as the activation energy for the self diffusion of pure iron. Q_d are dependent on temperature and given by the following equations;

$$Q_d = Q_{d\alpha}(1 + 0.22m^2) \quad (12)$$

and

$$Q_d = Q_{d\gamma}, \quad (13)$$

where $Q_{d\alpha}$ and $Q_{d\gamma}$ are the activation energies for self diffusion of ferro-magnetism and para-magnetism, respectively and are 240.9 and 291.3 kJ/mol, respectively, and $m =$

$I_s(T)/I_s(0)$ and $I_s(T)$ and $I_s(0)$ are the saturation magnetization at a given temperature and 0 K, respectively.¹⁴⁾

The elastic interaction energy between the movable dislocations and the obstacle dislocation is proportional to the shear modulus, μ . Q is calculated from the rupture data at different test temperatures using eq. (1). Therefore, in order to obtain the true elastic interaction energy, Q_{int} , the correction due to the temperature dependence of the shear modulus is necessary.¹⁵⁾ The value, Q_{int} , is calculated as

$$Q = Q_d + Q_{\text{int}} \left(1 - \frac{d \ln \mu}{d \ln T} \right). \quad (14)$$

Among the interactions between the dislocations and the obstacles, the interaction energy between an edge dislocation and an obstacle with the edge character becomes the maximum. When the edge dislocation cites at (r, θ) , the elastic interaction energy is given by the following equation.¹⁶⁾

$$Q_{\text{int}} = \max_{r, \theta} \left\{ \frac{\mu b \cdot pb}{2\pi(1-\nu)} \left\{ \ln \frac{D}{r} - \sin^2 \theta \right\} \frac{g}{d} \rho v_m \right\} \\ = \frac{\mu p b^2}{2\pi(1-\nu)} \ln \left(\frac{d}{2r_0} \right) \rho v_m \frac{g}{d}, \quad (15)$$

where ν is the Poisson's ratio, D is assumed to be equal to $d/2$, r_0 is the length between the nearest neighbor atoms and is equal to b .

2.4 Relations between metallurgical parameters and V , Q_{int} and ε_R

Eliminating ρ from eqs. (3) and (15), d/p is given by

$$\frac{d}{p} = \frac{\mu b}{\pi(1-\nu)} \ln \left(\frac{d}{2r_0} \right) \cdot \frac{V}{Q_{\text{int}}}. \quad (16)$$

Equation (16) shows that the equivalent obstacle spacing for a moving dislocation, d/p , can be directly estimated from creep rupture data with some assumptions.

Orlova and Cadek¹⁷⁾ reported that the total dislocation density including sub-boundaries is about one order magnitude higher than that within the subgrains. We assume that the obstacles are the immobile dislocations. Therefore, if we can neglect the mobile dislocations as compared with immobile dislocations, $(p/d)^2$ is corresponding to the total residual dislocation density, ρ_r . The total residual dislocation density means the total dislocations not annihilated or swept out and is comparable to the observed total dislocation density after the rupture test. Thus,

$$\rho_r = \left(\frac{p}{d} \right)^2 \propto \left(\frac{Q_{\text{int}}}{V} \right)^2. \quad (17)$$

In the above simplified discussion, dislocations with screw character are omitted. Even if the screw dislocations are taken into account, the total residual dislocation density, ρ_r , might increase twice at most.

From eqs. (9) and (16) eliminating d , we obtain the new metallurgical parameter, $s \equiv gp/d$, which may denote the strength of the interaction and/or the difficulty of climb or annihilation.

$$s \equiv \frac{gp}{d} = \frac{\pi d_0 v_m (1-\nu)}{\mu b \ln(d/2r_0)} \cdot \frac{Q_{\text{int}} \varepsilon_R}{V^2}. \quad (18)$$

The value of Q_{int}/V , which is inversely proportional to d/p and is presented by a unit of stress, denotes the resistance stress for the dislocation motion and is named the back stress, σ_b .

$$\sigma_b = Q_{\text{int}}/V \propto p/d. \quad (19)$$

From eqs. (9) and (16)–(19) it is clear that the metallurgical parameters, *i.e.* g , d/p , s , ρ_r and σ_b can be calculated using Q_{int} , V and ε_R . The latter is the macroscopic values entirely calculated from the rupture data without using any adjustable parameters.

Unfortunately, p and d cannot be estimated separately only from the rupture data. However, for example, if the obstacle spacing, d , is given through metallurgical observations, the other parameter, p , can be easily estimated.

3. Materials and Creep Rupture Data

National Research Institute for Metals has systematically performed long term creep rupture tests for the numerous heat resisting alloys and the results have been published as the data sheets. Among these, 0.2% carbon steel¹⁸⁾ (STB410 in JIS, Heat CAN and CAH), 0.5%Cr0.5%Mo steel¹⁹⁾ (STBA20 in JIS, Heat MFA), 2.25%Cr1%Mo steel²⁰⁾ (STBA24 in JIS, Heat MAA), 9%Cr1%MoVNb steel²¹⁾ (ASTM A213-T91, Heat T91) and 18-8 stainless steel²²⁾ (SUS304HTB in JIS, Heat ABD) have been selected for the verification of the present theory. These steels are mainly used for the power boilers.

Q and V for each material can be calculated from eq. (1) as a function of temperature and stress by interpolating the original data except for T91. For this steel the extrapolation method are also adopted, because the data of the same stress level at different temperatures are limited.

4. Analysis of Rupture Data

4.1 Analysis on 0.5%Cr0.5%Mo steel

The values of V , Q_{int} and σ_b of 0.5%Cr0.5%Mo steel (hereinafter, % is omitted) are shown as a function of t_r along with the data of σ and ε_R in Figs. 2 and 3. Q_{int} can be calculated at the average of two adjacent test temperatures. Therefore, though V , t_r and ε_R have been obtained at each test temperature, these values are also re-calculated at the average temperature in order to compare these physical values at the same temperature. The temperature dependence of the shear modulus is calculated referring to the Young's modulus of the typical heat resisting steels found in the literature²³⁾ assuming the Poisson's ratio is 0.3. The plotted data are fitted by a linear line or a spooline.

V stays at a rather low level within a short time and at the low temperatures, but V increases at 575°C with increasing t_r . Q_{int} decreases with increasing both T and t_r within a short time and at rather low temperatures, but Q_{int} tends to increase after a long time and at higher temperatures as compared with the values predicted from the low temperature and short time data. σ_b decreases with increasing both T and t_r .

The metallurgical parameters of 0.5Cr0.5Mo steel, g , d/p , s , and ρ_r , can be calculated using eqs. (9) and (16)–(17) re-

Fig. 2 The applied stress and the rupture elongation of 0.5Cr0.5Mo steel as a function of the average time to rupture.

Fig. 3 The activation volume, the interaction energy and the back stress of 0.5Cr0.5Mo steel as a function of the average time to rupture.

spectively. In the calculation μ is estimated as mentioned above and it is also assumed that $d_0 = 7.5 \text{ mm}$, $v_m = 7.2 \mu\text{m}^3/\text{mol}$, $b = 0.25 \text{ nm}$, $\nu = 0.3$ and d in the logarithm in eqs. (16) and (18) is equal to $1 \mu\text{m}$. The results are plotted as a function of t_r in Figs. 4 and 5. In Fig. 5, ρ and ρ_R are also plotted as a comparison. g and s decrease with increasing both T and t_r . On the other hand, d/p increases with increasing both T and t_r . The temperature and time dependence of ρ_r is similar with that of the back stress as implied by eq. (17). As eqs. (10) and (11) show, the temperature and time dependence of ρ and ρ_R is similar to that of the activation volume.

4.2 Temperature dependence of d/p , g , s , ρ_r , ρ and ρ_R

The temperature dependence of the metallurgical parameters, d/p , g , s , ρ_r , ρ and ρ_R , are more clearly shown in Figs. 6 and 7. The values of Q , V and ϵ_R for 10000 h are obtained by interpolation for each heat resisting steel. The metallurgical parameters are calculated using Q , V and ϵ_R . The

Fig. 4 The microstructural parameters p/d , g and s of 0.5Cr0.5Mo steel as a function of the average time to rupture.

Fig. 5 The dislocation density, ρ , the rupture dislocation density, ρ_R , and the residual dislocation density, ρ_r of 0.5Cr0.5Mo steel as a function of the average time to rupture.

parameters, g and s , decrease with increasing T and tend to approach to each certain value, respectively. On the contrary, d/p ($=g/s$) tends to increase with increasing T , though the behavior of 0.2% carbon steel (Heat N) is exceptional. Accordingly, $\rho_r = (p/d)^2$ decreases with increase in T .

Though the parameters, d and p , cannot be calculated separately with each other, the most possible interpretation for the temperature dependence of the metallurgical parameters is as follows; d increases with increasing T due to increase in the size of lath and subgrain or precipitate and g decreases due to the frequent occurrence of annihilation of moving dislocations. p also decreases with increasing T due to the annihilation or recovery of the obstacle dislocations. As a result s ($=gp/d$) always decreases with increase in T and t_r as shown

Fig. 6 The microstructural parameters p/d , g and s for 10000 h of the typical heat resisting steels as a function of temperature. The observed particle spacing for 2.25Cr1Mo, 1Cr0.5Mo and 0.5Cr0.5Mo steels are also plotted.

in Figs. 4 and 6.

5. Discussion

5.1 Verification of d/p and g

Yukitoshi and Nishida²⁴⁾ observed the extracted replicas of the crept steels, 1Cr0.5Mo and 2.25Cr1Mo, and summarized the number of carbide particles per unit area as a function of the Larson-Miller parameter. Sumida *et al.*²⁵⁾ reported the particle spacing of the aged 2.25Cr1Mo steel and Kimura *et al.*²⁶⁾ reported the transmission electron micrographs of the thin foil of the crept 0.5Cr0.5Mo steel. From these studies, the particle spacings have been calculated and plotted in Fig. 6. Though the observed values for the 2.25Cr1Mo steels scatter largely depending on the reports, it is found that the equivalent obstacle spacing, d/p , coincides with the observed particle spacing within the factor of about 3. The same coincidence is observed between 0.5Cr0.5Mo steel and the similar steel, 1Cr0.5Mo steel, as shown in Fig. 6. This implies that the parameter, p , is nearby unity, though it should depend on material, time and temperature, *i.e.* a single dislocation with edge character is pinned by some particles or in the sub-boundaries and becomes the obstacle to movable dislocations.

Spiradek *et al.*²⁷⁾ reported 9Cr1MoVNB steel that the particle spacing is $0.28\mu\text{m}$ for a ruptured specimen at 600°C for 19464 h. This value closely coincides with the calculated value of d/p in T91 (Fig. 6).

It is also well known that in high Cr heat resisting steels the lath or sub-boundaries are well developed and the width of the boundary for a specimen ruptured at 600°C for about 10000 h is reported as $0.5\text{--}1.4\mu\text{m}$.^{6,11,26)} The values are just smaller than the calculated glide distance, g , of T91, *i.e.* $2\mu\text{m}$. The observed and calculated values are judged to closely coincide with each other, because the aspect ratio of the lath is usually larger than unity.

Fig. 7 The dislocation density, ρ , the rupture dislocation density, ρ_R , and the residual dislocation density, ρ_r for 10000 h of the typical heat resisting steels as a function of temperature.

5.2 Residual dislocation density, ρ_r

ρ_r decreases with increase in T and t_r as shown in Figs. 5 and 7. ρ_r is inversely proportional to $(d/p)^2$ and, therefore, is related to σ through eqs. (1), (16), (17), (19).

$$\begin{aligned}\sigma &= \sigma_b \left(1 - \frac{d \ln \mu}{d \ln T} \right) + \frac{Q_d - RT \ln(t_r/t_{r0})}{V} \\ &= \sqrt{\rho_r} \cdot \frac{\mu b \ln(d/2r_0)}{\pi(1-\nu)} \left(1 - \frac{d \ln \mu}{d \ln T} \right) \\ &\quad + \frac{Q_d - RT \ln(t_r/t_{r0})}{V}.\end{aligned}\quad (20)$$

The 1st term of eq. (20) is related to the back stress. The derivative of σ with respect to $\sqrt{\rho_r}$ is roughly constant for iron base alloy, *i.e.* about 85 TPa/m. The typical value of the 2nd term of eq. (20) is -500 MPa , however, it strongly depends on T , V and t_r .

Usually, the logarithms of dislocation density is plotted against the logarithms of stress,²⁸⁾ but eq. (20) shows that σ in a linear scale should be plotted as a function of $\sqrt{\rho_r}$. The results are summarized in Fig. 8. Investigating the T and t_r dependence on 0.5Cr0.5Mo steel (Δ) in detail it is found that the residual dislocation density decrease with increasing T and t_r and that even the data of the specific alloy should not be classified in one group. However, all of the calculated data in Fig. 8 look to form a single data band independent of the kind of steel. In this sense, the regression analysis was made and the results is shown in the figure. In spite of that, the coefficients obtained should not have any specific physical meaning according to the above discussion. The reason is that the regression coefficient of 17.6 TPa/m is about one fifth of the theoretical value in eq. (20), *i.e.* 85 TPa/m, because the 2nd term of eq. (20) strongly depends on the kind of steel and cannot be neglected as compared with the 1st term.

In the figure, the observed dislocation densities within the lath boundaries of T91 crept at 550 through 650°C ¹⁰⁾ is plot-

Fig. 8 The residual dislocation density, ρ_r , of the several heat resisting steels as a function of the applied stress. The observed dislocation densities for the crept 9Cr1MoVNb steel are plotted as a comparison.

ted. The observed dislocation density within lath boundaries of the similar alloys^{6,10,11,27}) are also plotted. The data are largely scattered depending on the researchers. However, it is surprising that the calculated dislocation density, though it means the total dislocation density, is comparable to the observed values.

5.3 Creep behavior of 0.2% carbon steel, Heat H

The rupture strength for 10000 h at 475°C of Heat H is 94% higher than that of Heat N¹⁸⁾ and Heat H contains 0.019%Mo and 0.0063%N in contrast with 0.005%Mo and 0.0051%N of Heat N. It has been well established that the long term rupture strength of heat resisting carbon steels is strongly affected by such residual elements²⁹⁾ and at a longer time the strength is controlled mainly by Mo not by nitrogen.³⁰⁾ Using the rupture data of Heat H and Heat N it can be estimated that Q_{int} and σ_b of Heat H are much larger and V is much smaller than those of Heat N. As corresponding to this phenomenon, s and ρ_r of Heat H is larger and d/p of Heat H is much smaller than those of Heat N as shown in Figs. 6 and 7. That is, the metallurgical behaviors of Heat H are much different from those of Heat N, but are similar with those of the low alloy steels. This suggests that the rupture strength of Heat H may be controlled by the fine precipitation of Mo_2C as found in the usual low alloy steel.²⁶⁾ This is supported by the computer simulation that the solubility of Mo for Heat H at 450°C is very low and is estimated around 0.005%, and the sufficient precipitation of Mo_2C is expected, though the calculation was made on MoC because the thermodynamic data are available for MoC and not for Mo_2C .³¹⁾

5.4 Abnormal creep behavior of carbon steel, Heat N

As shown in Fig. 6, d/p generally increases with increasing T . This phenomenon is understood that p decreases and d increases with increasing T as mentioned above. However, it is also possible that d decreases and/or p increases with increasing T due to the precipitation during creep and/or the dynamical interaction between dislocations and solute atoms.

Fig. 9 The microstructural parameters p/d , g and s of 0.2% carbon steel (Heat N) as a function of the average time to rupture.

The typical case is found in Fig. 6. d/p of Heat N is not only large as compared with those of Heat H, but also d/p decreases with increasing T .

Figure 9 shows g , s and d/p of Heat N as a function of t_r . d/p of 475°C and 10000 h is about 70% of that of 425°C according to the depression in the d/p curve. The same depression at 425°C is found at a longer time. If the depressions of the d/p vs. $\log t_r$ curves of 425 and 475°C indicated by the arrows in the figure are assumed to have the same physical meaning, an apparent activation energy for the phenomenon is calculated as 91 kJ/mol. This value is much smaller than the activation energy for diffusion of substitutional solute atoms and rather comparable to that of interstitial atoms in α iron.³²⁾ Heat N of 0.2%C steel contains only 0.005% of Mo and, therefore, the fine needle like particles of Mo_2C which are usually observed in the Mo containing steels³⁰⁾ have not been expected. However, the existence of the depression in d/p curves suggests that d/p is considered to decrease due to the interaction between the gliding dislocations and the solute atoms, *i.e.* N, Mn and Si, though the details of a diffusion model is not clear. Yokoi *et al.*³³⁾ have reported the solubility of N in the Al-killed 0.2%C steel determined by the hydrogen extraction method. According to their results about 4 ppm of N can be soluble in the matrix at 450°C, while the steel contains 51 ppm of N. Therefore, the nitrides possibly precipitate in the Heat N of 0.2% carbon steel and may have an interaction with dislocations.

6. Conclusions

Physical meaning of the activation energy and activation volume in the new creep rupture equation which is previously proposed by the authors for the heat resisting steels has been studied on the base of the simple dislocation model. The metallurgical parameters calculated from the long term rupture data are compared with the literature data and following conclusions are obtained.

(1) The residual dislocation density, ρ_r , can be estimated from both the activation volume, V , and the activation energy

in the absence of the applied stress, Q , which are calculated using the new rupture equation. In the estimation, Q is converted to Q_{int} , the interaction energy between the obstacle dislocations and the movable dislocations, $Q_{\text{int}} = Q - Q_d$ where Q_d is the activation energy of the self diffusion.

(2) The estimated dislocation density is comparable to the observed values in the literature.

(3) Back stress to the movable dislocations is estimated by Q_{int}/V .

(4) The equivalent spacing between the obstacles, d/p , and the gliding distance of the movable dislocations, g , can be also estimated from V , Q_{int} and the rupture elongation, where d is the obstacle spacing and p is a characteristic parameter near unity.

(5) The reported particle spacings and the size of the sub-boundaries are comparable with d/p and g , respectively.

(6) The gliding distance, g , decreases and the equivalent obstacle spacing, d/p , increases with increasing temperature and time to rupture. If dynamical aging and/or precipitation occur, d/p should be decreasing.

List of main symbols

b	length of the Burger's vector
d	spacing of the obstacle dislocations
d/p	equivalent obstacle spacing
d_0	length of a slip plane in a direction of the Burger's vector
h	spacing of a slip plane
ΔL_A	activation displacement of a specimen
L_0	gage length
g	gliding distance of movable dislocations
n	dislocation number generating in unit longitudinal section of $h \times d$
p	characteristic parameter for the obstacle dislocations
pb	equivalent length of the Burgers vector of the obstacle dislocations
P	tensile load
Q	activation energy in the absence of the applied stress
Q_d	activation energy for self diffusion
Q_{int}	elastic interaction energy between movable dislocations and the obstacle dislocations
r_0	length between nearest neighbor atoms
R	the gas constant
S_0	cross section of a specimen
s	interaction parameter defined by eq. (18)
t_r	time to rupture
t_{r0}	constant of eq. (1)
T	testing temperature
v_m	molar volume
V	activation volume
α	ratio of $\varepsilon_R/\varepsilon_A$ in eq. (8)
ε_R	rupture elongation
ε_A	activation tensile strain
ρ	number of dislocations generating in the cross section of $d \times h$ during the activation process
ρ_r	residual dislocation density
ρ_R	dislocation density necessary to form rupture elongation

μ	shear modulus
ν	the Poisson's ratio
σ	applied stress
σ_b	back stress

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