**In-situ Observation of Dislocation Motion and Its Mobility in Fe–Mo and Fe–W Solid Solutions at High Temperatures**

Daisuke TERADA, Fuyuki YOSHIDA, Hideharu NAKASHIMA, Hiroshi ABE and Yoshikuni KADOYA

Graduate Student, Kyushu University, Kasugakoen, Kasuga, Fukuoka 816-8580 Japan.
1) Interdisciplinary Graduate School of Engineering Sciences, Kyushu University, Kasugakoen, Kasuga, Fukuoka 816-8580 Japan.
2) Takasago R&D Center, Mitsubishi Heavy Industries, Ltd., Shinhama Arai-cho, Takasago 676-8686 Japan.

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The mobilities of the edge dislocations in Fe–W and Fe–Mo solid solution alloys at high temperatures were investigated with in-situ TEM observation in order to compare the effect of W and Mo on the solid solution hardening of ferrite. The dislocation behaviors were recorded continuously with VTR. TEM observations showed that the dislocations were moving viscously and that the dislocation velocities were constant in both Fe–W and Fe–Mo alloys. These results show that dislocations dragged solute atmosphere. The mobilities were determined to be $5.7 \times 10^{-15} \text{m/(Pa s)}$ at 993 K in the Fe–W alloy and $4.3 \times 10^{-15} \text{m/(Pa s)}$ at 1011 K in the Fe–Mo alloy. It was found that the mobility in the Fe–W alloy is similar to that in the Fe–Mo alloy. Mobilities were estimated by simulation using the interaction between an edge dislocation and solute atoms. The results show that the simulated value was similar to the measured value in the Fe–Mo alloy while the simulated value is different from the measured in the Fe–W alloy. The simulated mobility of the dislocation in Fe–W is one-tenth as large as that in Fe–Mo. The simulated results for Fe–W did not agree with the experimental results. It is considered that the difference between the experimental values and simulated values chiefly came from the shape of the dislocations using for measurement and the diffusion constant which was used in the simulation. Experimental results showed that the mobility in Fe–W is as large as that in Fe–Mo. Therefore, it is suggested that the effect of W on the solid solution hardening of Fe is similar to that of Mo.

KEY WORDS: *in-situ* TEM observation; solid solution hardening; dislocation mobility; Fe–Mo alloy; Fe–W alloy.

1. Introduction

Ultra supercritical (USC) boilers for usage at 650°C in power plants are being developed in order to increase the plant’s thermal efficiency and to decrease the greenhouse gas emission. The steam conditions of the plant’s thermal efficiency and to decrease the greenhouse gas emission. The steam conditions of the current plants, therefore, is needed to develop materials with high creep rupture strength in order to realize the 650°C class USC boilers.

9Cr ferritic heat resistant steels have been developed for USC boilers. Especially Mod.9Cr–1Mo (9Cr–1Mo–V–Nb steel) is used as the materials for boilers of conventional power plants. In recent years, advanced heat resistant steels; such as NF616 (9Cr–0.5Mo–1.8W–V–Nb steel), in which a part of Mo was replaced with W have been developed. It is recognized that the creep rupture strength of advanced heat resistant steels is higher than that of Mod.9Cr–1Mo steel, and that replacement of Mo with W is effective in improving of creep rupture strength.

The strengthening mechanism of replacing Mo with W is considered to be solid solution hardening mechanism, precipitate hardening and/or cluster hardening. Our group studied solid solution hardening of 9Cr ferritic steel with W addition by stress relaxation test, and recognized that W is effective to solid solution hardening.

The ability of solid solution hardening depends on the size misfit between a solution atom and a matrix atom and the diffusion coefficient of a solution atom in the matrix. It is known that the size misfit parameter, $\varepsilon$, of W is 9.97% and that of Mo is 8.43%, where $\varepsilon$ is defined as $\varepsilon = (1/a)(\Delta a/\Delta C_0)$ with a lattice constant, $a$, and an average solution concentration, $C_0$. Diffusion coefficients are calculated by following equation, $D = D_0 \exp(-Q/RT)$, where $D_0$ is the pre exponential term, $Q$ the activation energy, $R$ the gas constant and $T$ the absolute temperature. The values of these parameters of W in ferrite are $D_0 = 2.0 \times 10^{-4} \text{m}^2/\text{s}$ and $Q = 426.2 \text{kJ/mol}$, and that of Mo are $D_0 = 6.63 \times 10^{-5} \text{m}^2/\text{s}$ and $Q = 224.2 \text{kJ/mol}$. W is larger than Mo in misfit parameter and W is smaller than Mo in diffusion coefficient. It is expected from these results that the effect of W addition on solid solution hardening is larger than that of Mo addition. But from experimental result, reverse is true. The creep rupture strength of the Fe–Mo solid solution alloy was larger than that of the Fe–W solid solution alloy. The results reported don’t agree with the above expectation. So it is necessary to compare the effects caused by W with caused by Mo on solid solution hardening.
To estimate solid solution hardening quantitatively, it is necessary to evaluate mobility of edge dislocation. Therefore, it is necessary to evaluate the mobility of edge dislocation in Fe–W and Fe–Mo solid solution alloys in order to compare the effect caused by W with caused by Mo on solid solution hardening. In this study, at first dislocation motions in Fe–W and Fe–Mo alloys were observed in-situ with transmission electron microscope (TEM) and mobilities of dislocation were estimated. Secondly mobilities were also estimated by a simulation using the interaction between an edge dislocation and solute atoms in an imaginary tetragonal lattice.8) The results of TEM observation were compared with the results of simulation. The effect on solid solution hardening caused by W addition and Mo addition is also discussed.

2. Experimental Procedure

2.1. In-situ TEM Observation

Specimens used in this study were ferritic steels with 0.7 at% W and 0.7 at% Mo, respectively. The chemical compositions and the grain sizes of the specimens are listed in Table 1. For TEM observation, thin sheets, approximately 1 mm in thickness, were cut and then disks, 3 mm in diameter, were trepanned from the sheets by spark cutting and were polished to 50 μm in thickness. The disks were polished electrolytically into foil by a twin-jet technique with a solution of perchloric acid and methanol of 1 : 10 volume ratio.

An in-situ TEM observations of dislocation motion in the Fe–W alloy and the Fe–Mo alloy were carried out with JEM-1000 and a heating holder equipped with a two-axes tilt apparatus at HVEM laboratory of Kyushu University. The foil specimens were heated up to about 1 023 K. Dislocations were moved by the thermal stress that is caused by the difference of thermal expansion coefficient between the holder and the specimen. The dislocations motions were recorded with a videotape recorder (VTR), continuously. The stresses acting on dislocations and the velocities were analyzed from movements of dislocations and shape, respectively. The mobilities of dislocations were estimated by the velocities and the stresses.

2.2. Simulation to Estimate Dislocation Mobility

Mobilities of dislocation were estimated by a simulation using the interaction between an edge dislocation and solute atoms in an imaginary lattice.8) An imaginary tetragonal lattice was constructed around a dislocation as shown in Fig. 1 and the lattice constant was taken as the jump distance, s, of solute atoms. Then solute concentration changes at each lattice point were calculated in a short time interval from the jump frequencies of solute atoms between the nearest lattice points, considering the interaction of solute atoms with the dislocation.

The interaction energy is given by the product of a hydrostatic pressure come from an edge dislocation multiplied by a volume expansion caused by replacing matrix atoms with solute atoms.9) If the solute concentration is low, the volume expansion is estimated by the difference of volumes between the solute atom and the solvent atom. In this case, the interaction energy of a lattice point at \((x, y)\) is given by the following equation.

\[
V(x, y) = \frac{4(1 + \nu)}{3(1 - \nu)} G b e_b R_0 \left[ \frac{y}{x^2 + y^2} \right]
\]

Where \(\nu\) is Poisson’s ratio, \(G\) the shear modulus, \(b\) the magnitude of Burgers vector, \(e_b\) the size misfit parameter from the difference between radius of the solute atom and that of the solvent atom and \(R_0\) the radius of the solvent atom. \(e_b\) is defined as \(e_b = (R_s - R_0) / R_s\) with the radius of solute atom, \(R_s\). The relation between \(e_b\) and \(e\) is known as \(e_b = (1 + \nu) / 3(1 - \nu) \cdot e\). If the solute concentration is high, it is necessary to consider the volume expansion produced by a local concentration change from average concentration. As the local concentration changes around dislocation, the hydrostatic pressure produced by an edge dislocation is relaxed. In this case, the interaction energy is given by the following equation.8)

\[
V(x, y) = \frac{4(1 + \nu)}{3(1 - \nu)} G b e_b R_0 \left[ \frac{y}{x^2 + y^2} \right] + \frac{6\pi(1 - \nu)}{1 - 2\nu} \frac{e_b (C - C_0)}{b} \left( 1 - C \right)
\]

Where \(C\) is a local solute concentration. Both Eqs. (1) and (2) were used in this study.

A solute concentration change at a lattice point \((x, y)\) is calculated from the jump frequencies of solute atoms between \((x, y)\) and the nearest lattice points; \((x + s, y)\), \((x, y + s)\) and \((x, y - s)\) as shown in Fig. 1. When the dislocation moves in the \(x\) direction with a velocity, \(v\), the rate of concentration change at the lattice point \((x, y)\) is given by the following equation.

\[
\frac{\partial e_b}{\partial t} = \frac{8\pi(1 - \nu)}{1 - 2\nu} e_b \left( 1 - C \right)
\]
\[
\frac{\partial C(x, y)}{\partial t} = \frac{D}{s^2} \left[ C(x+s, y)[1-C(x, y)] \right] \\
\times \exp \left[ \frac{V(x+s, y)-V(x, y)}{2kT} \right] \\
- C(x, y)[1-C(x+s, y)] \exp \left[ \frac{V(x, y)-V(x+s, y)}{2kT} \right] \\
+ \text{jump}{(x-s, y \leftrightarrow (x, y))} \\
+ \text{jump}{(x, y+s \leftrightarrow (x, y))} \\
+ \text{jump}{(x, y-s \leftrightarrow (x, y))} \\
+ \frac{v}{s}(C(x+s, y)-C(x, y)) \quad \cdots \cdots \cdots \cdots (3)
\]

Where \(C(x, y)\) is the concentration at \((x, y)\), \(D\) the diffusion coefficient of solute atom, \(kT\) usual meaning. In the right hand side in Eq. (3), the first and second terms mean the concentration change caused by the solute atoms jump between lattice points \((x, y)\) and \((x+s, y)\). Similarly, third, fourth and fifth terms mean the solute atom jump between the nearest lattice points, for example “jump\((x-s, y \leftrightarrow (x, y))\)” means the solute atoms jump between \((x, y)\) and \((x-s, y)\). The atmosphere of solute atoms around the dislocation was calculated by simulation based on Eq. (3).

The drag stress \(\tau_d\) caused by solute atoms was evaluated from the solute atmosphere by the follow equation.\(^8\)

\[
\tau_d = \frac{1}{b} \sum \left[ \frac{C(x, y)s^2}{\Omega} \right] \frac{\partial V}{\partial x} \quad \cdots \cdots \cdots \cdots (4)
\]

Where \(\Omega\) is the volume occupied by a solute atom and is expressed by \(a^3/2\) for b.c.c. with the lattice constant \(a\).

The dislocation mobility, \(B\), was estimated by \(B = \frac{\nu}{\tau_d} \).\(^{10,11}\) In the simulation, amount of lattice points is \(200 \times 200\) in the imaginary square lattice, the dislocation velocities were from \(10^{-11}\) to \(10^{-4}\) m/s, the solute concentrations were from 0.5 to 2.0 at% and the temperatures were from 873 to 1023 K. Table 2 shows the value of the parameters used in the simulation. The shear moduli, \(G\), were calculated from the value measured by D. J. Dever\(^{12}\) with geometric average.

### 3. Results

#### 3.1. In-situ TEM Observation

Figure 2 shows micrographs of successive dislocation movements in the Fe–W alloy at 993 K and Fig. 3 shows that in the Fe–Mo alloy at 1011 K from in-situ TEM observation recorded with VTR. Figures 2(a) to 2(d) are pictures that were taken at intervals of 30 s, and Figs. 3(a) to 3(d) are the pictures that were taken at intervals of 15 s. In Fig. 2, the black regions such as a region indicated by circle A are contaminations, not precipitations. Most of dislocations did not move so much and moving distances were small. They probably interacted each other. However, the dislocation indicated by arrow moved viscously from the left to the right.

![Fig. 2. TEM micrographs of Fe–W at 993 K. The arrow indicates the moving dislocation. (a) to (d) are the pictures which were taken at intervals of 30 s. Circle A shows the contamination.](image)

| Parameter | Designation | Value |
|-----------|-------------|-------|
| Lattice constant\(^{13}\) | \(a\) | \(2.8664 \times 10^{-10}\) m |
| Atomic radius | \(R\) | \(1.24 \times 10^{-10}\) m |
| Burgers vector | \(b\) | \(2.48 \times 10^{-10}\) m |
| Shear modulus\(^{12}\) | \(G\) | 46.2 GPa (1023K) |
| Poisson's ratio | \(\nu\) | 0.349 (1023K) |
| Diffusion Parameters\(^5\) | \(D_0\) | 2.0 \(\times 10^4\) m\(^2\) s\(^{-1}\) Fe-W |
| | \(Q\) | 246.2 kJ/mol Fe-W |
| Size misfit Parameter\(^{14}\) | \(\varepsilon\) | 0.0997 Fe-W |

Table 2. Parameters for simulation in the Fe–W alloy and the Fe–Mo alloy.
right in the observation field of VTR and the moving distance was more than 3 μm. Similarly, the dislocation in Fe–Mo alloy moved viscously as is shown in Fig. 3.

An internal stress, $\tau_i$, due to interdislocation interaction is expressed as

$$\tau_i = \alpha G b / \rho = \alpha G b l$$  \hspace{1cm} (5)

where $\alpha$ is a constant depending on dislocation arrangement, $\rho$ the dislocation density and $l$ the average spacing between dislocations. $\tau_i$ were estimated to be 2.4 MPa at 993 K and to be 2.3 MPa at 1011 K by Eq. (5) when it was regarded that $\alpha$ was 0.58 and $l$ was 3 μm.

Figures 4 and 5 show the traces of the dislocations movement indicated by arrows in Figs 2 and 3 at interval 5 s, respectively. The dislocation movements were traced using contaminations in observation fields as reference points because the images drifted. The arrow A and arrow B indicate the direction of dislocation movement in Figs. 4 and 5, respectively. Driving forces moving the dislocation are regarded as the stress acting on the dislocation, and the stress $\tau$ is given by $\tau = G b / 2r$, where $r$ is curvature radius of the dislocation. Circles along the trace lines were drawn and the radii of the circles were measured. The radii were regarded to be curvature radii. The averages of curvature radii estimated from Figs. 4 and 5 were 1.6 μm and 1.1 μm, respectively. From these results, the driving forces which moved dislocation were estimated to be 3.8 MPa in the Fe–W alloy and 5.3 MPa in the Fe–Mo alloy, respectively. Both the driving forces were larger than the internal stresses estimated. Therefore, the internal stress due to interdislocation interaction could be disregarded. Thus the mobilities were estimated using the dislocations.

Figures 6 and 7 show the relationships between the dislocation moving distance and time in Figs. 4 and 5. Here, the moving distance was defined as a distance from the point O to intersections of the tracing lines and the straight lines A and B in Figs. 4 and 5. From Figs. 6 and 7, it is found that moving distance is proportional to time. The slope of Figs. 6 and 7 mean the dislocation velocities. It is clear that the dislocation velocities were constant from the
beginning to the end of TEM observation in both Fe–W and Fe–Mo alloys. The results mean that the dislocations drag the solute atmosphere. The dislocation velocities were estimated to be $2.2 \times 10^{-3}$ m/s in the Fe–W alloy and $2.3 \times 10^{-3}$ m/s in the Fe–Mo alloy from the slopes of Figs. 4 and 5, respectively.

The dislocation mobilities, $B$, were estimated by $B = \frac{v}{\tau_d}$. When dislocations are moving with a constant velocity, it is considered that the driving force which moved the dislocations was equal to the drag stress, $\tau_d$. Therefore, it is possible that the mobility of dislocation can be evaluated by $B = \frac{v}{\tau}$ from in-situ TEM observation. The mobilities of dislocation were calculated to be $5.7 \times 10^{-15}$ m/(Pa·s) in the Fe–W alloy and $4.3 \times 10^{-15}$ m/(Pa·s) in the Fe–Mo alloy, respectively. It is found that the mobility of dislocation in the Fe–W alloy was similar to that in the Fe–Mo alloy.

### 3.2. Estimation of Dislocation Mobility by Simulation

**Figure 8** shows an example of double logarithmic plots of the relationship between the drag stress and dislocation velocity obtained from the simulation with Eq. (1). The solute concentration is 1.0 at% and the temperatures are from 873 to 1023 K in Fig. 8. In this figure, it is found that the slopes were equal to one at low velocities region. Similar results were obtained in other solute concentrations. The results show that the drag stresses are proportional to dislocation velocities at low velocities. As the dislocation velocities increase, the proportional relationships disappear gradually. As the dislocation velocities further increase, the drag stresses decrease.

An example of the relationships between the drag stress and the dislocation velocity with Eq. (2) is shown in **Fig. 9**. The dislocation velocities used in the calculation for in Fig. 9 is the value at low velocity; where the drag stress is proportional to the dislocation velocity. The drag stresses in Fig. 9 are smaller than that in Fig. 8 slightly because the hydrostatic pressure produced by an edge dislocation is relaxed. From Fig. 9, it is found that the drag stresses caused by W are about ten times as large as that caused by Mo at same conditions. For example, the drag stress caused by W is 1.25 MPa and that caused by Mo is 0.135 MPa at 873 K and $1 \times 10^{-11}$ m/s in Fig. 9.

It is possible that the mobilities of dislocation are calculated by $B = \frac{v}{\tau}$ only when the drag stress is proportional to
the dislocation velocity. Figure 10 shows the Arrhenius plot of the mobilities in Fig. 9 including results of other solute concentrations. It is found that the mobilities decrease as the solute concentrations increase and that the mobilities of dislocation in Fe–W are about one-tenth as large as that in Fe–Mo. This result dose not agree with the result of TEM observation that the mobilities in the Fe–W alloy is similar to that in the Fe–Mo alloy. In order to compare the mobilities obtained from simulation and that of TEM observation, the measured values from TEM observation are also plotted in Fig. 10. The measured value in Fe–Mo alloy is similar to the simulated value while the experimental value in Fe–W alloy is different from the simulated value and more than ten times as large as simulated value.

4. Discussion

4.1. Comparison between the Result of TEM Observation and the Result of Simulation

In the in-situ TEM observation, dislocations moved viscously (Figs. 2 and 3) and dislocation velocities during observation were constant (Figs. 6 and 7) in both the Fe–W alloy and the Fe–Mo alloy. These results mean that dislocations dragged solute atmospheres. The solid solution hardening was considered to be effective in experimental conditions of the present TEM observations. The dislocation velocities were 2.2×10−8 m/s and 2.3×10−8 m/s, the stress acting on the dislocations were 3.8 MPa and 5.3 MPa and the mobilities of dislocation were 5.7×10−15 m/(Pa·s) and 4.3×10−15 m/(Pa·s) in Fe–0.7at%W at 993 K and in Fe–0.7at%Mo at 1 011 K, respectively. The mobility in Fe–W alloy is slightly larger than that in Fe–Mo alloy. However, the mobility of Fe–W is about one-tenth as large as the simulated value. The reason of this difference between the measured value and simulated value will be discussed as follows.

One of the reasons is considered to be the effect of impurities in specimens used in TEM observation. W and Mo in ferrite are substitutional atoms. It is known that interstitial atoms and substitutional atoms interact chemically (I–S effect)15,16 and the drag stress of solute atmosphere increases due to the I-S effect. It has been recognized that nitrogen (N), which is interstitial atom, interact with W and Mo and N is effective in the I-S effect in Fe–W and Fe–Mo alloys.17,18

The mobility of dislocation is expressed by the following equation10,11 approximately,

\[ B = \frac{1}{68} \frac{D k T}{C_N g^2 \varepsilon_0^2 R_0 b} \] ........................(6)

where \( D \) is the diffusion coefficient of solute atom, \( N \) amount of atoms in unit volume. It has been reported for Al–Mg alloy that the mobility obtained from Eq. (6) was smaller than the value simulated with Eq. (2).19 However, the mobility is considered to be proportional to the diffusion constant and to be inversely proportional to the solute concentration and the square of the misfit parameter as is shown in Eq. (6). When the I-S effect is effective, the diffusion constant depends on the substitutional atom (W or Mo) and the solute concentration and the misfit parameter depends on interstitial atom (N). The N concentrations in the specimens for TEM observation were 0.12at% in Fe–W alloy and 0.08 at% in Fe–Mo alloy. The N concentration of Fe–W alloy was larger that that of Fe–Mo alloy. If I-S effect is effective the magnitude of the I-S effect in Fe–W alloy is expected to be larger than that in Fe–Mo alloy. Therefore, the mobility measured is expected to be smaller than the simulated mobility especially in Fe–W alloy. The fact is that the mobility measured was larger than the mobility simulated. Thus, the effect on mobilities caused by N is regarded to have been small in our experimental conditions. The difference between the mobility measured and the mobility simulated cannot be explained sufficiently by the I-S effect caused by N.

Another reason is considered to be errors in the in-situ TEM observations. In the TEM observation, a slope of a slip plane against the observation plane is not taken into consideration. If the slip plane is inclined against the observation plane, the evaluated dislocation velocities are smaller than the practical dislocation velocities. Perhaps the estimated mobilities are smaller than the actual mobility of dislocation because the estimation was made by \( B = \frac{u}{\tau} \). However, the slopes of the slip planes are regarded small, because the observed dislocations moved long distance; more than 3 μm in Figs. 2 and 3. Thus, the mobilities obtained by TEM observation can be regarded approximately the same as the practical mobilities.

The dislocations indicated by arrow in Figs. 2 and 3 were not pure edge dislocations because the dislocations were not straight. There is some possibility that the estimated mobility was different from a mobility of pure edge dislo-
cation. On the other hand, the simulation supposed an ideal edge dislocation in an imaginary lattice. Therefore, it is considered that there is some difference between the measured values and simulated values. It is not clear how much the measured values were different from the simulated values. It is necessary to consider the difference hereafter.

Other reason to cause difference between the measured value and the simulated value can be the effect of parameters used in our simulation. The mobility of dislocation is considered to be proportional to the diffusion constant as shown in Eq. (6). A variety of values have been reported with regard to diffusion parameters of Fe–W and Fe–Mo. The parameters which were used in this study for Fe–W were \( D_0 = 2.0 \times 10^{-8} \text{ m}^2/\text{s}, Q = 246.9 \text{ kJ/mol} \). While values \( D_0 = 1.0 \times 10^{-7} \text{ m}^2/\text{s}, Q = 238 \text{ kJ/mol} \) and \( D_0 = 6.9 \times 10^{-3} \text{ m}^2/\text{s}, Q = 265.9 \text{ kJ/mol} \) have been reported in other papers. The mobilities were calculated by Eq. (6) using parameters which were given above and the mobilities at 993 K with 0.7 at% W were \( 6.1 \times 10^{-16} \text{ m/(Pa} \cdot \text{s)} \), \( 9.4 \times 10^{-17} \text{ m/(Pa} \cdot \text{s)} \) and \( 2.6 \times 10^{-15} \text{ m/(Pa} \cdot \text{s)} \), respectively. The values calculated by Eq. (6) cannot be compared with the measured value directly because Eq. (6) is approximate equation. However, the calculated values are useful for consideration of the effect of a diffusion coefficient on dislocation mobility. The calculated values scatter two orders of magnitude. The measured value \( 5.7 \times 10^{-15} \text{ m/(Pa} \cdot \text{s)} \) is close to one of the calculated values \( 2.6 \times 10^{-15} \text{ m/(Pa} \cdot \text{s)} \). It is considered that the results from simulation were affected by the original values which were used in calculations. The parameters used in simulation need to be considered hereafter.

It is suggested that the difference between the experimental values and simulated values chiefly came from the shape of the dislocations using for measurement and the diffusion constant which was used in the simulation.

4.2. The Effect of Addition W and Mo on the Solid Solution Hardening

The results of TEM observation show that the effect of W addition on the solid solution hardening is slightly larger than that of Mo addition. However, the results of simulation don’t agree with the results of TEM observation. Creep test of the same specimens used in this study were carried out at 873 K and 30–120 MPa by Kadoya et al. They evaluated the strengthening factor by \( \bar{\varepsilon}_{\text{m,Fc}}/\bar{\varepsilon}_{\text{ar}} \) where \( \bar{\varepsilon}_{\text{m,Fc}} \) is the minimum strain rate in pure Fe and \( \bar{\varepsilon}_{\text{m}} \) is the minimum strain rate in each alloy. It was noticed that the strengthening factor of Fe–Mo alloy was larger than that of Fe–W alloy. The results mean that the effect of solid solution hardening in Fe–Mo alloy was larger than that in Fe–W alloy indicating that the mobility of dislocation in Fe–W is considered to be larger than that in Fe–Mo. This trend agrees with the result of TEM observation in this study.

Therefore, it is considered that the results of TEM observation are important and the effect of W addition on the solid solution hardening is slightly larger than that of Mo addition. The fact that the Fe–W has higher creep resistance than Fe–Mo can not be explained by solid solution hardening mechanism. It is suggested that some mechanism other than solid solution hardening is the cause of higher creep resistance of Fe with W addition.

5. Conclusion

In-situ TEM observation of dislocation motion in Fe–0.7at%W and Fe–0.7at%Mo solid solution at high temperature and simulation to estimate the mobility of dislocation were carried out. The results are summarized as follows.

(1) TEM observation showed that the dislocations moved viscously and are regarded to drag the solute atmosphere in Fe–W and Fe–Mo solid solution.

(2) From the results of in-situ TEM observation, the dislocation velocities were \( 2.2 \times 10^{-8} \text{ m/s} \) and \( 2.3 \times 10^{-8} \text{ m/s} \), the stress acting on the dislocations were 3.8 MPa and 5.3 MPa and the mobilities of dislocation were \( 5.7 \times 10^{-15} \text{ m/(Pa} \cdot \text{s)} \) and \( 4.3 \times 10^{-15} \text{ m/(Pa} \cdot \text{s)} \) in Fe–W at 993 K and in Fe–Mo at 1 011 K solid solution, respectively.

(3) From the results of in-situ TEM observation, it is considered that there is little difference between the effect of the addition of W and Mo on the solid solution hardening because the mobility of dislocations in Fe–W is similar to that in Fe–Mo.

(4) The mobilities of dislocation in the Fe–W alloys were about one-tenth as large as that in Fe–Mo alloys in the simulation using the interaction between an edge dislocation and solute atoms in an imaginary tetragonal lattice. The measured value in Fe–Mo is almost the same as the simulated value while the measured value in Fe–W is tenth as larger as the simulated value. It is considered that the difference between the experimental and simulated values chiefly came from the shape of the dislocations using for measurement and the diffusion constant which was used in the simulation.

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