Plasticity of Body-Centred Cubic Metals at High Strain Rate*

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Synopsis

Stress—strain curves of metals mentioned in the title were obtained at high strain rate and various temperatures above room temperature by “counter-blow” forging tester. The strain rate can be varied from $3 \times 10^2/\text{sec}$ to $10^3/\text{sec}$ by changing the angle of lifting hammers. The main experimental results are as follows:

(1) Many investigators have reported that the strain-rate dependence of flow stress of metals and alloys increases with temperature. But in the authors' experiment, such a general conclusion can not be adopted to that in the cases of iron, steel and tantalum at strain rate of $10^2/\text{sec}$.

(2) At lower temperature, large strain-rate dependence of flow stress is found in the cases of Ta, C steel and 2%Al iron. The temperature dependence corresponds almost always to the strain-rate dependence. But that of 0.02%C steel, 0.05%C steel containing Ni, 2%Al iron and Ta is small at the lowest part (below 150°C for the iron and steel, and 200°C for Ta) of the temperature range.

(3) In the case of tantalum, a peak of stress—temperature curves is found where the strain-rate dependence is large. It is the different point of flow-stress behaviour from that in the case of iron and steel.

(4) The flow-stress behaviour at lower temperature range mentioned above can be interpreted by the double-kink formation mechanism on moving dislocations.

(5) In the case of 1%Si iron, the flow-stress behaviour below 300°C is of the same nature as that of steel, whereas that above 300°C (small temperature and strain-rate dependence of flow stress) must be due to the local order-hardening effect.

(6) A peak found at 650°C on stress—temperature curves in the case of steel must be due to the interaction between interstitials and dislocations. But the dragging mechanisms fail to explain higher work-hardening rate and higher density of dislocations. Then the other mechanism exists, but it is not yet clear.

(7) At higher strain rate no peak is found on stress—temperature curves in the α - γ coexisting temperature range, whereas at lower strain rate a peak exists. The reason is that strain-rate dependence of flow stress is higher in α -phase temperature than that in γ -phase temperature.

I. Introduction

In 1940, M. Manjoine and A. Nadai noticed in their first paper¹⁾ about high speed tension tests at elevated temperature, “The continuous hot and cold mills built during the last 10 to 15 years in this country stand out as one of the most remarkable developments in the mass production of steel..... Very little is known about the forces required for the pure plastic forming of steel and of other ductile metals at the high straining velocity used for rolling of sheets.....”

Since 1940, many investigators have reported the data about the stress—strain relations at elevated temperatures and high strain rates of common metals as well as steel. Many of their works have been done from the point of view of mechanical engineers. There

is much reliable knowledge for the rolling-mill practice about such stress-strain—strain-rate relation as an experimental equation, i.e.

$$K_f = K_0 \varepsilon^m \dot{\varepsilon}^n e^{A/kT} \dots\dots\dots (1)$$

where K_f is flow stress, ε is strain, $\dot{\varepsilon}$ is strain rate, k is Boltzmann's constant, T is absolute temperature and m , n and A are constants. Below $10^2/\text{sec}$, of strain rate for the flow stress of metals and alloys, it is said that the strain-rate dependence exponent n is small at low temperatures, whereas larger at higher temperatures, as shown in Fig. 1.

T: Testing temperature ($^\circ\text{K}$)
 T_m : Melting temperature ($^\circ\text{K}$)

Fig. 1. The strain-rate dependence exponent “ n ” and testing temperature²⁾

Now in Japan, many investigators²⁾⁻⁴⁾ have studied on the stress-strain—strain-rate relation and ductilities of various sorts of steels at various temperatures, and the data have been utilized in autocontrolling of hot and cold stripmill practices.

The authors can measure the stress—strain curves of metals at the strain rate of magnitude of $10^2/\text{sec}$ at various temperatures above room temperature by the counter-blow-type forging tester. The out-lined feature of tester and the block diagram of the measuring system are shown in Fig. 2.

In this paper, the problems as follows are studied.

The Strain-Rate Dependence of Iron and Steel from Room Temperature to 1000°C at Strain Rate of $10^2/\text{Sec}$.

As shown in Fig. 1, at low temperature the strain-rate dependence indicated by n in the equation (1) is small, but at higher temperature large. However, this is not true for the authors' experiments especially in the case of steel.

The Effect of Solute Atoms of Flow Stress

What effect have solute atoms but for carbon and

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Fig. 2. The counter-blowing forging tests designed by the present authors

nitrogen on the strain-rate and temperature dependence of the flow stress of steel? It is discovered that silicon has the remarkable effect on the strain-rate and temperature dependence to be much smaller at the temperature range between 300° and 650°C.

II. Experimental Procedure and Materials

As mentioned above, the stress—strain curves at various deforming conditions, i.e. from room temperature to 1 000°C at strain rate of 10²/sec, were measured by the counter-blowing type forging tester shown in Fig. 2. The testing samples are cylindrical and of the size of 9–10 mm in diameter and 9–10 mm in length, respectively. The testing conditions are shown in Table 1 and the chemical compositions are shown in Table 2.

In this paper, the main part of the experiments is on the temperature and strain-rate dependence of the flow stress of mild steel. In addition, for the sake of comparing the flow mechanism of iron and steel

at strain rate of 10²/sec with that of Ta which has also body-centred cubic structure, the temperature and strain-rate dependences of the flow stress of Ta were measured, too. The chemical composition is also shown in Table 2.

The specimens were prepared in such ways as shown below.

0.09%C Steel

The cylindrical specimens of 10 mm in length and 10 mm in diameter were made of the 10 mm diameter rimmed steel wire rod. They were annealed for 1/2 hr. at 900°C in cast iron powder.

0.02%C Steel

This is used as one of the soft electro-magnetic materials. The specimens of 10 mm in length were made of wire rod of 9.5 mm in diameter. They were annealed for 1/4 hr. at 950°C in vacuo.

0.04%C Steel

The specimens of 10 mm in length were made of rimmed steel wire rod of 10 mm in diameter, and were annealed for 1 hr. at 950°C in vacuo.

0.05%C Steel-I

The cast ingot was hot-forged to 50 mm in diameter, reduced to 12 mm in diameter by hot rolling and cold drawing, and annealed for 2 hrs. at 900°C in the air. The specimens of 10 mm in length and 10 mm in diameter were made of it.

0.05%C Steel-II and the Eight Sorts of 0.05%C Steel Containing Ni and Cr

Each material was made from electrolytic iron, metallic Ni and Cr. Carbon content was attained by mixing the proper weight of pig iron. 1 200 g of the iron and Ni (in the case of steel containing Ni) was charged in the electronic high frequency melting furnace (15 kW, 30 kc), melted and held for 5 hrs. in H₂ gas for reducing. Then, the gas was pumped out, Ar gas was charged in turn there and 0.015 wt% of

Table 1. Conditions of experiments

Specimen	Testing temp. (°C)	Strain rate (sec ⁻¹)	Lubrication
0.09%C steel	Room temp.	400–1 080	MoS ₂ +grease
	200–860	300–1 080	MoS ₂ +grease
	900	300–1 080	glass
0.02%C steel	Room temp. to 850	300–1 000	MoS ₂ +grease
0.04%C steel	700–1 000	300–1 000	MoS ₂ +grease
0.04%C steel-I	700–1 000	300–1 000	MoS ₂ +grease
0.05%C steel-II and 0.05%C steel containing Ni and Cr	Room temp. to 600	300–1 000	MoS ₂ +grease
2% Al-iron	Room temp. to 1 000	300–1 000	MoS ₂ +grease
1% Si-iron	Room temp. to 1 000	300–1 000	MoS ₂ +grease
Ta	Room temp. to 1 200	300–1 000	MoS ₂ +grease

Table 2. Chemical compositions of specimens (%)

Specimen	C	Si	Mn	P	S	Al	Ni	Cr	N	
0.09%C steel	0.09	0.01	0.30	0.02	0.05	—	—	—	—	
0.02%C steel	0.02	0.01	0.07	0.005	0.011	0.2	—	—	—	
0.04%C steel	0.04	0.01	0.28	0.012	0.016	0.02	—	—	—	
0.05%C steel-I	0.05	(This is rimmed steel.)								
0.05%C steel-II	0.05	(As for Si, Mn, P, S and N these contain them as much as a common electrolytic iron.)								
NC-I	0.05									
NC-II	0.05									
N-III	0.05									
C-I	0.05									
C-II	0.05									
C-III	0.05									
NC-I	0.05									
NC-II	0.05									
2%Al iron	0.002	0.03	trace	0.008	0.008	1.96	0.009	0.005	0.002	
1%Si iron	—	1.2	(This is made from electrolytic iron.)							
Ta	C	O	H	Nb	Mo	Cr	Fe	W	Ti	V
	0.006	0.006	0.002	0.013	<0.001	<0.001	0.0009	<0.005	<0.0005	0.0005

Al was added in the melt. Carbon was alloyed in the form of pig iron together with metallic Cr (only in the case of steel containing Cr), held for 3 min., subsequently solidified in the crucible used for melting.

The cast ingot of 50 mm in diameter and 100 mm in length was hot-forged to 12 mm in diameter and cold drawn to 9 mm in diameter. The specimens of 10 mm in length were made of it, and annealed in vacuo for 1 hr. at 900°C.

2%Al Iron

The zone refined electrolytic iron was melted in vacuo, and Al was alloyed to 2 wt%, so that this alloy is in α -phase at every temperature. It was cooled and solidified in the crucible, and the cast ingot was of the size : 50 mm in diameter and 250 mm in length. This ingot was hot-forged at 700°C in Ar gas and reduced in diameter to 12 mm. Then, by cold drawing the rod of 9 mm in diameter was obtained. The specimens of 10 mm in length were made of it and annealed in vacuo for 1 hr. at 900°C.

1%Si Iron

Electrolytic iron and metallic Si were melted and

alloyed in the air. Si content is about 1%. This alloy was casted in a metallic mould. It was of the size, 50 mm in diameter and 100 mm in length. This ingot was hot-forged at 1 000°C, reduced in diameter to 12 mm and cold drawn to 9 mm in diameter. The specimens of 10 mm in length were made of it and annealed for 1 hr. at 900°C.

Ta

The specimens were made of DF-381 produced Kobe Steel, Ltd. The size is of 9 mm in diameter and length. They were annealed for 1 hr. at 1 300°C.

III. Experimental Results

0.09%C Steel

The stress—strain curves from room temperature to 900°C are shown in Fig. 3. From room temperature to 400°C, the mode of strain-hardening is mild and the strain-rate dependence of flow stress is large. The strain-rate dependence exponent is of the magnitude of 0.15–0.20.

At 600°C, the strain-rate dependence of flow stress is still large, and strain hardening becomes larger

Fig. 3. Stress-strain curves of 0.09%C steel at various temperatures above room temperature to 900°C

than one at the lower temperatures. At strain of 0.10 the flow stress at 600°C is of the same magnitude as that at room temperature.

From 700° to 820°C, the strain-rate dependence of flow stress becomes smaller and the dependence exponent decreases to about 0.12 and strain-hardening is mild again. However, the temperature dependence of flow stress is much larger. At 860°C the structure of 0.09%C steel is composed of 20% α -phase and 80% γ -phase. The strain-rate dependence of flow stress is much smaller than that at the other temperature and dependence exponent decreases to 0.05.

The stress—strain curves at 900°C in the two cases are shown in Fig. 3. The case (a) is that of the specimens held for 5 min. and the case (b) is that of the specimens held for 10 min. The flow stress of specimens held longer is lower than that of ones held shorter, and the strain-rate dependence of that is also smaller. The flow stress at 860°C is less than that at 900°C in the both of two cases.

0.02%C Steel

In Fig. 4 the stress—strain curves of 0.02%C steel at various temperatures are shown. Except for results from 500° to 600°C, at which the strain-rate dependence of flow stress is very small, the others are almost the same as in the case of 0.09%C steel.

The temperature dependence of flow stress is known from Fig. 5 where stress—temperature curves at the strain of 0.10 are shown. It may be said that the curves have a knick point at 150°C.

0.04%C Steel and 0.05%C Steel-I

About 0.09%C steel, it is found that the flow stress at 860°C is lower than that at 900°C. Moreover, a peak of the flow stress—temperature curves at the

Fig. 5. K_f -temperature curves at 0.10 strain of 0.02%C steel

Fig. 4. K_f -strain curves of 0.02%C steel at various temperatures above room temperature to 850°C

temperature range where α -phase transforms to γ -phase, has also been found by some investigators^{6), 7)}. Then, the flow stress measurement about these steels was done between 700° and 1 000°C. In Fig. 6 the flow stress—temperature curves are shown about those at the strain of 0.10. The curves shown in Fig. 6 (a) are of 0.04%C steel and the others shown in (b) are of 0.05%C steel-I. In Fig. 6 (a), the experimental results at lower strain rate measured by Katoh* and

(a) 0.04%C steel

(b) 0.05%C steel-I

Fig. 6. Stress—temperature curves

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Fig. 7. (a)-(j).

Stress-temperature and strain curves of steels containing various contents of Ni and Cr

Kusakabe* with the cam-plasto-meter are shown for comparison.

The peak is not present at such high strain rate, $10^2/\text{sec}$. Especially from Fig. 6(a) the strain-rate dependence of the flow stress is clearly larger at α -rich temperature than at γ -rich temperature. The same tendency can be said about the temperature dependence of flow stress.

0.05%C Steel-II and the Eight Sorts of 0.05%C Steel Containing Small Amount of Ni and Cr

For the sake of detecting the effect of substitutional alloying elements on the flow-stress behaviours such as temperature and strain-rate dependences, the specimens added Ni, Cr, Al and Si were prepared. This experiment about the addition of Ni and Cr has the purpose to clarify the effect of these two elements on the temperature and strain-rate dependence of

the flow stress between room temperature and 600°C .

The stress-temperature curves are shown in Fig. 7 (a)–(i). From Fig. 7, below 500°C the strain-rate dependence of flow stress is as large as those found in the cases of 0.09%C steel and 0.02%C steel. At 100°C , in the case of steels containing Ni, a knick point on the stress-temperature curves is found as in the case of 0.02%C steel. But for the presence of the knick point, the temperature dependence where the strain-rate dependence is large, is also large. The effect on the flow-stress behaviours mentioned above of these alloying elements is not found. In Fig. 7(j), two of the stress-strain curves of 0.05%C steel-II are shown. At 600°C , strain-hardening occurs more rapidly than at room temperature. The strain-rate dependence between 500° and 600°C is as small as that found in the case of 0.02%C steel.

2%Al Iron

C content of this alloy is less than 0.002%. Then,

Fig. 8. Stress-temperature curves of 2%Al iron (strain 0.10)

with this alloy the effect of C and Al on the flow-stress behaviours can be detected. In Fig. 8 the stress—temperature curves at strain of 0.10 are shown. From Fig. 8, both of the strain-rate and temperature dependence of flow stress are large below 500°C, become smaller between 600° and 850°C and again larger to some extent above 850°C. Any of peak is not found on the stress—temperature curves, but a knick is also found at 100°C.

1%Si Iron

The stress—temperature curves are shown in Fig. 9. The flow-stress behaviours below 300°C are of the same feature as the other C steels and alloyed irons. From 300° to 650°C, however, the strain-rate and temperature dependences of flow stress become much smaller and the level of the flow stress 20 kg/mm² higher than the flow stress of the other steels at 400°–500°C. Between 700° and 850°C the flow stress begins to decrease drastically. Above 850°C, the strain-rate and temperature dependences of flow stress becomes much smaller.

Fig. 9. Flow stress—temperature curves of 1%Si iron (strain 0.10)

Tantalum

The stress—temperature curves about Ta are shown in Fig. 10. The flow stress difference between at strain rate 1000/sec and at 300/sec is also shown in the figure. Below 850°C, the strain-rate dependence of flow stress has the same order of magnitude of the stress of iron and steel. A peak on the stress—temperature curve is detected at 500°C and it is due to the blue shortness of Ta at high strain rate 500/sec. At strain rate, 10²/sec, the blue shortness of Ta occurs in the temperature range where the strain-rate dependence of flow stress is large. This is a different point from the flow-stress behaviours of iron and steel. But for the peak below 850°C, the temperature dependence of flow stress is of the same feature as that of iron and steel at the temperature range from room temperature to 500°C.

Fig. 10. Flow stress—temperature curves and strain-rate dependence of Ta (strain 0.10)

IV. Discussions

All experimental results can be summarized as shown below.

(1) From room temperature to 500°C, the strain-rate dependence of flow stress is large in the case of steel and iron alloys except for 1%Si iron. Their temperature dependence is generally large in this temperature range, but in the cases of 0.02%C steel, 0.05%C steel containing Ni and 2%Al iron it is small between room temperature and 100°C.

(2) The flow stress behaviours in the case of 1% Si iron are of the same feature as in the cases of the other steels and iron alloys below 300°C, but are quite different above 300°C. The strain-rate and temperature dependences are very small at the temperature range from 300° to 650°C, and the flow-stress level is 20 kg/mm² higher than those of iron and steel at 500°C.

(3) A peak on the stress—temperature curves of steel is found at the temperature range between 600°C and 700°C. But any peak is not found in the case of 2%Al iron not containing C nor N enough to occur blue shortness.

(4) In the case of steel, the strain-rate dependence of flow stress between 500° and 600°C is small, but above 650°C become as large as below 500°C.

(5) From 500° to 850°C, the strain-rate and temperature dependences of 2%Al iron are small, but above 850°C those become a little larger.

(6) In the case of steel, the strain-rate and temperature dependences of flow stress are smaller above A_3 point. And then no peak are found on the stress—temperature curves at high strain rate, whereas a peak is found at lower strain rate at about A_3 point.

(7) In the case of Si iron the strain rate dependence of flow stress is small, but the temperature dependence is much larger between 700° and 800°C. Above 850°C, both of those become much smaller.

(8) In the case of tantalum the behaviours of flow stress are almost similar to those in the case of steel, but a peak on the stress—temperature curve is found in the temperature range where the strain-rate dependence of flow stress is still large. Above 900°C, temperature does not affect on the flow stress.

As for the experimental results mentioned above, some discussion will be made in the following.

First, the large temperature and strain-rate dependences from room temperature (100°C in the cases of 0.02%C steel, 0.05%C steel containing Ni, 2%Al iron and 1%Si iron) to 500°C (but 300°C in the case of 1%Si iron) have the same feature as those in the case of the low temperature plasticity of iron and steel at much lower strain rate, 10⁻³–10⁻⁶/sec. For instance, if strain rate $\dot{\epsilon}$ can be related to temperature T and flow stress τ effective to move dislocation by equation (2):

$$\dot{\epsilon} = \dot{\epsilon}_0 \exp \{ -U(\tau)/kT \} \dots\dots\dots (2)$$

where $\dot{\epsilon}_0$ is composed of the number of moving dislocations Burger's vector, the average length of moving dislocation segments and the number of free vibrating frequencies of moving dislocations, $U(\tau)$ is the poten-

tial barrier to moving dislocations, and a function of τ decreasing as stress increasing, and k is Boltzmann's constant, $U'(\tau)$ has the same order of magnitude as that in the case of the low temperature and low strain-rate test. With eq.(2), H. Conrad, et al. analysed the low temperature and low strain-rate plasticity of iron and steel, and concluded that the controlling mechanism of the deformation is the process of double kink formation on moving dislocations because of high Peierls-Nabarro force of iron⁽⁸⁾⁻¹⁰⁾. As for what mechanism controls the deformation, however, all other investigators did not agree to H. Conrad, et al. D.F. Stein and Low Jr., et al. studied on the plasticity of iron containing C and N lower than 5×10^{-3} p.p.m. and found the temperature dependence of stress smaller than that of iron containing C and N higher than 1 p.p.m. Then, they concluded that the frictional stress to moving dislocations depends on the presence of C and N.⁽¹¹⁾ On this problem, systematically experiment on C and N dependence of flow stress must be done. Such a study was done by the present authors with the low temperature creep and tensile test on iron containing various amount of C and N, i.e., lower than 1 p.p.m. to 200 p.p.m. They concluded that C and N affect the pre-exponential factor of right side of eq. (2), i.e., moving dislocation density and then τ increases because of need of decreasing $U(\tau)$ ⁽¹²⁾.

In this paper, the strain-rate and temperature dependences of flow stress of Ta are measured for the sake of comparing with those in the case of iron and steel. In the case of Ta, a peak on the stress—temperature curve due to the interstitial impurities is found in the temperature range where the strain-rate dependence of flow stress is large. The temperature dependence of flow stress above 450°C, at which the flow-stress peak exists, is large up to the temperature, 850°C and becomes very small above 900°C. If the strain-rate dependence is caused by the same deformation mechanism both below and above 450°C and the strain-rate dependence is also small above 900°C, it is possible to compare the case of iron with that of Ta.

The strain-rate and temperature dependences of flow stress of iron and steel (especially 2%Al iron) are small above 500°C ($\approx 800^\circ\text{K}$) and those of Ta are also small above 900°C ($\approx 1200^\circ\text{K}$). Above those temperature, the flow-stress component due to the frictional stress determined by the controlling deformation mechanism is zero. Then from eq.(2), the ratio of $U(0)$ in the case of Ta to that in the case of iron is obtained, assuming that $\dot{\epsilon}_0$ of eq. (2) does not differ above 10 fold from each other in the both of two cases. Then,

$$\exp\{-U(0)\text{Fe}/k \cdot 800\} = \exp\{-U(0)\text{Ta}/k \cdot 1200\} \dots (3)$$

where $U(0)\text{Fe}$ is $U(0)$ in the case of iron and $U(0)\text{Ta}$ in the case of Ta. From eq.(3), $U(0)\text{Ta}/U(0)\text{Fe} = 1.5$ is obtained. If the controlling mechanism is that of double kink formation process as proposed by H. Conrad, et al., $U(0)$ must be proportional to μb^3 where μ is the shear modulus. μb^3 of iron is 11 ergs, that of Ta is 16 ergs, and then the ratio is also 1.5. This is favourable result for concluding the deformation mech-

anism to be that of the double kink formation process. In the case of iron, $U'(\tau)$ obtained by $\Delta \ln \dot{\epsilon} / \Delta K_f$ from the experimental data varies from $<10b^3$ to $30b^3$, respectively from room temperature to 500°C, and almost the same result is obtained in the case of Ta. Such a magnitude of $U'(\tau)$ is also favourable to conclude the double kink formation mechanism to control the deformation. But now one problem exists. It is whether the strain-rate history of deformation affects the structure of materials at 0.10 strain. But in other papers the present authors treated this problem, and conclude that neither strain rate nor temperature affect on the structure where flow stress seems to be determined by the double kink formation mechanism.^{(12), (16)}

The flow-stress behaviours of Si iron above 300°C is quite different from those of iron and steel. From 300° to 650°C the flow stress of 1%Si iron does not depend on temperature and strain rate, and is about 20 kg/mm² higher than that at 500°C of iron and steel. This high stress level must be interpreted by the presence of Si atoms. About 10 years ago, Fisher proposed a hardening mechanism by Si atom in iron by which he interpreted the preferential activity of $\{110\}\langle 111 \rangle$ slip system in high Si iron.⁽¹³⁾ The hardening mechanism is called "local or short range order" hardening. Now there is no knowledge about the binding energy of Si-Fe pair enough to calculate the stress to hinder the moving dislocations. It is possible to understand that the strain-rate and temperature dependences are small if the local ordering hinders the dislocation motion, because the effective hindering path of this mechanism can be expected long enough to avoid the effect of thermal fluctuation of dislocations to the contrast to the double kink formation mechanism or jog dragging mechanism. The large temperature dependence above 700°C may correspond to the rapid decrease of the stress by destroying of the local order, if it is true that the flow stress is determined by "local ordering" mechanism between 300° and 650°C.

Now, the peak on the stress—temperature curves is discussed below. The peak is due to the interstitial impurities. The reason why flow stress increases from 500° to 650°C, however, is not yet clear. The mechanisms proposed now are agreeable on a point that the diffusivity of interstitials relates to dislocation velocity. But some investigators proposed the mechanism as that of dragging interstitials of dislocation.⁽¹⁴⁾ At lower temperature the diffusivity of interstitials is too low to move with dislocations and at higher temperature that is high enough to be free from dislocations. Then at optimum temperature this dragging effect on the flow stress is the highest. But the dragging mechanism mentioned above is unable to explain the high work-hardening rate at the peak temperature as shown in Figs. 1, 2 and 7(j). This high work-hardening rate is due to the rapid multiplication of dislocations as shown by transmission electron microscopy⁽¹⁵⁾. Then the high stress level at the peak temperature is interpreted not by the dragging stress but by the back stress due to the

high dislocation density. The reason why the dislocation density varies with temperature and is affected by interstitials, is not yet clear.

The discussion mentioned above about the flow-stress peak is also able to be adopted to explain the large temperature and strain-rate dependences above the peak temperature. The cause of the dependences must be interpreted in the same way by the interaction between dislocations and interstitial impurities.

Then in the temperature range of coexisting of α - and γ -phase, no peak on the stress—temperature curves are found at higher strain rate, whereas at lower strain rate a peak exists there. The strain-rate dependence of flow stress of α -phase is larger than that of γ -phase as shown in Fig. 6(a). This is the reason why a peak exists at lower strain rate and disappears at higher strain rate.

In the case of 2% Al iron, neither a peak due to the interstitial impurities nor a peak due to α - γ transformation exists. However, above 850°C the strain-rate and temperature dependences are a little larger. This seems due to the self diffusion enough to affect the structural change during deformation.

V. Conclusions

(1) Many investigators report that the strain-rate dependence of flow stress of metals and alloys increases with temperature. But in the authors' experiment, such a general conclusion can not be adopted to those in the cases of iron, steel and Ta at strain rate of $10^2/\text{sec}$.

(2) At lower temperature, the large strain-rate dependence of flow stress is found in the case of steel, 2%Al iron and Ta. The temperature dependence corresponds almost always to the strain-rate dependence, but that of 0.02% C steel, 0.05% C steel containing Ni, and Ta is small at the lowest part (below 100°C—150°C for steel and 200°C for Ta) of the temperature range.

(3) In the case of Ta, a peak of the stress—temperature curve is found where the strain-rate dependence of flow stress is large. It is the different point of flow-stress behaviour from those in the case of iron or steel.

(4) The flow stress behaviours at lower temperature mentioned above can be interpreted by the double kink formation mechanism on moving dislocations.

(5) In the case of Si iron, the flow-stress behaviours below 300°C is of the same nature as those of iron or steel. But those above 300°C must be due to the local ordering effect.

(6) A peak found on the stress—temperature curves of steel at 600°–650°C must be due to the interaction between dislocations and interstitials. But the dragging mechanism fails to explain the high work-hardening rate and high density of dislocations. So the other mechanism is necessary to explain, but it is not yet clear.

(7) In the α - γ -coexisting temperature range, at higher strain rate no peak is found on the stress—temperature curves, whereas at lower strain rate a peak exists. The reason is that the strain-rate dependence of flow stress is higher in α -phase temperature than in γ -phase temperature.

(8) In the case of 2%Al iron, both the strain-rate and temperature dependences of flow stress are small from 500° to 850°C, and then become larger above 900°C. It seems that above 850°–900°C the self diffusion is active enough to cause the recovery during deformation.

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