Long-term Creep Rupture Properties and Microstructure of 12 % Cr Heat Resisting Steels*

By Ik-Min PARK** and Toshio FUJITA***

Synopsis

Long-term creep rupture properties and microstructure of 12 % Cr heat resisting steels were investigated. It has been found that solid solution hardening by an addition of Mo is very effective on long-term creep rupture strength, but precipitation or dispersion hardening by an individual addition of V or Nb is little effective. Especially, the presence of excessive amounts of undissolved NbC leads to the coarsening of MX precipitates, and a marked drop in long-term creep rupture strength. The combined addition of V+Nb with Mo and B leads to superior 30 000 h creep rupture strength. Microstructural differences between long-term creep specimens at 600 °C and short-term creep specimens at 650 °C were discussed by considering the parameter, $T(C+\log t)$. Long-term specimens at 600 °C exhibit higher dislocation density substructure caused by intensive precipitation reactions which include both fresh precipitation within lath and agglomeration at boundaries, and exhibit intergranular fracture probably caused by stress concentration around the agglomerates, whereas short-term specimens at 650 °C exhibit intragranular fracture. It is considered that the deviation of long-term creep rupture strength from the extrapolated value is caused by differences in precipitation hardening level and fracture mode.

I. Introduction

The precipitation hardening is widely applied to the current commercial heat resisting steels. In these steels, a complex microstructural change occurs during long-term service at elevated temperature, and ultimately fracture occurs after creep deformation.¹⁾ In order to clarify these problems, some works²⁻⁴⁾ of the effects of microstructure on creep rupture properties have been carried out. However, these previous works were based on the short-term creep and rupture data, and did not give a satisfactory explanation for the time-dependent creep properties, such as the decrease in precipitation hardening and dispersion hardening caused by Ostwald ripening of the precipitates,^{5,6)} or the decrease in solid solution hardening as a result of the precipitation of the solid solutes in long-term service.

In this work, to make further explanations for the relation between creep rupture properties and microstructure, long-term creep rupture properties and microstructure of 12 % Cr steels have been investigated using long-term specimens ruptured at 10 000~60 000 h. And microstructural differences between long-term creep specimens at 600 °C and short-term creep specimens at 650 °C have been studied by adopting the Larson-Miller parameter, $T(C+\log t)$. Furthermore, the effects of alloying elements on long-term creep rupture strength, microstructural changes

during creep¹³⁾ (creep recovery process) were also discussed.

II. Experimental Procedures

Chemical analyses of the steels studied in this work are given in Table 1. S1 is a plain 12 Cr steel, S2 and S3 are Mo and V individual by-added steels, respectively. The S4~S9 steels are complex alloyed steels containing 1.4 % Mo and B (S4~S6; 0.05 %, S7~S9; 0.03 %) with the variations in V and Nb contents.

These steels were cast into 40 kg ingots by air melting and forged at about 950~1 100 °C to bars of 20 mm diameter. These steels were solution treated at 1 150 °C for 30 min and oil quenched, and then tempered at 700 °C for 1 h. Creep rupture specimens with a parallel gauge length of 30 mm were tested by conventional single specimen machines or multiple specimen machines. Determinations of Vickers hardness were made at room temperature using a 20 kg load.

The specimens for thin foils or other microstructural examinations were prepared from the gauge length of the ruptured specimens. Carbides were extracted in a 5 % HCl + 95 % ethanol solution at a current density of 0.02 to 0.03 A/cm². X-ray diffraction studies of extracted residues were made mainly using Cu-K α radiations. Co powder ($a=3.545$ Å) was used as an internal standard for the lattice parameter measurements. The electron microprobe analyses (EPMA) of the residues were made at an accelerating voltage of 20 kV, specimen current of 0.05 μ A, electron beam diameter of 70 μ m, and measuring time of 10 sec.

Precipitation partition ratio is the ratio of the weight percent of an element in the extracted residue to that of the element in the steel. Electrolytic polishing for thin foils was performed in an 80 % H₃PO₄+20 % CrO₃ solution, and for the final stages in a 135 g CH₃COOH+25 g CrO₃+7 g H₂O solution at 0 °C. Thin foils were examined in a 150 kV electron microscope.

Carbon extraction replicas for EDX (Energy Dispersion X-ray analysis) examinations were prepared by etching initially in Villela's reagent and stripped in a 25 % HCl+25 % HNO₃+50 % ethylene-glycol solution.

* Presented to the 97th ISIJ Meeting, April 1979, at The University of Tokyo in Tokyo. Manuscript received July 13, 1981.
© 1982 ISIJ

** Graduate School, The University of Tokyo, Hongo, Bunkyo-ku, Tokyo 113.

*** Department of Materials Science, Faculty of Engineering, The University of Tokyo, Hongo, Bunkyo-ku, Tokyo 113.

Table 1. Chemical composition of specimens. (wt%)

	C	Si	Mn	P	S	Ni	Cr	Mo	V	Nb	B	N
S 1	0.22	0.43	0.83	0.007	0.012	—	10.56	—	—	—	—	0.037
S 2	0.20	0.36	0.53	0.017	0.017	0.09	11.58	1.07	—	—	—	0.015
S 3	0.20	0.44	0.53	0.005	0.005	0.08	11.85	—	0.38	—	—	0.015
S 4	0.21	0.35	0.85	0.007	0.013	0.05	10.51	1.40	—	—	0.051	0.013
S 5	0.20	0.31	0.89	0.007	0.014	0.05	10.40	1.40	0.20	—	0.042	0.015
S 6	0.19	0.36	0.85	0.008	0.012	0.06	10.35	1.41	—	0.19	0.049	0.016
S 7	0.20	0.32	0.84	0.008	0.010	0.05	10.72	1.44	0.25	0.08	0.027	0.016
S 8	0.20	0.38	0.86	0.009	0.010	0.05	10.59	1.43	0.15	0.15	0.026	0.015
S 9	0.20	0.38	0.83	0.009	0.009	0.05	10.73	1.37	0.12	0.20	0.028	0.017

III. Results

1. Long-term Creep Rupture Strength

Results of creep rupture tests are shown in Figs. 1 and 2 as a function of the Larson–Miller parameter. The solid line was constructed based on the short-term data up to 10 000 h, and long-term data beyond 10 000 h were plotted.

In S1~S3 steels, the long-term creep rupture strength is in the order: S2>S3≈S1. The addition of Mo is very effective on long-term creep rupture strength, whereas the addition of V is little effective.

In S4~S9 steels, the long-term creep rupture strength is in the order: S7>S8>S9>S6≈S5>S4. The V-rich complex alloyed steel (S7) exhibits higher long-term (30 000 h) creep rupture strength than the Nb-rich complex alloyed steel (S9). It should be noted that creep rupture strength of the Nb-rich steels decreases markedly at about 10 000 h.

2. Precipitates

The main precipitates of 12 Cr steels in long-term creep tests is $M_{23}C_6$, which is the same as short-term creep tests. In Mo bearing steels, Laves phase (Fe_2Mo) was identified in long-term specimens ruptured beyond 10 000 h. In long-term crept specimens, fine precipitates ($<0.2\ \mu m$) within laths or at lath boundaries, and agglomerates ($>1\ \mu m$) at prior austenite grain boundaries or lath boundaries were observed. The precipitates at lath boundaries or prior austenite grain boundaries were identified to be $M_{23}C_6$, and the precipitates within laths were identified to be MC and $M_{23}C_6$ by electron diffraction and EDX examinations (Photo. 1).

The typical carbide transformation in Cr containing steels,⁷⁾ M_7C_3 – $M_{23}C_6$ – M_6C , was expected during long-term creep test. However, M_6C was detected only in S4, whereas M_7C_3 was detected in long-term crept specimens of S2, S4 and S9 at 600 °C.

M_2X which has been reported^{3,8)} to be the main secondary hardening precipitates of high N bearing ($>0.04\ \%$ N) 12 Cr steels was identified to be Cr_2N only in S1, whose morphology changed from an acicular type to a planar or spherical type during long-term creep.

The precipitation partition ratio of V in S3 is about 60 % after long-term creep test, and V is not contained in M of $M_{23}C_6$. Electron microprobe

Fig. 1. Long-term creep rupture strength of S1~S3 steels. (Solid line was constructed based on short-term data up to 10 000 h, and long-term data beyond 10 000 h were plotted.)

Fig. 2. Long-term creep rupture strength of S4~S9 steels. (Solid line was constructed based on short-term data up to 10 000 h, and long-term data beyond 10 000 h were plotted.)

analyses by EDX showed that V-carbides contain small amounts of Cr (Photo. 2). In the Nb-rich ($>0.1\ \%$ Nb) steels, MX ($a=4.39\sim4.43\ \text{\AA}$) was identified by X-ray diffraction. The lattice parameter of MX is almost the same as that of (Nb, V) (C, N) ($a=4.432\sim4.440\ \text{\AA}$) reported by Hede and Aronsson,³⁾ but the metallic composition was found to be mainly Nb by EDX examinations (Photo. 1(c)). Laves phase (Fe_2Mo ; $a=4.72\sim4.75$, $c=7.67\sim7.75\ \text{\AA}$) identified by X-ray diffraction can not be detected by electron diffraction or EDX examinations.

Amounts of precipitates, precipitation partition ratio of the individual alloying elements, and the lattice parameter of $M_{23}C_6$ are shown in Figs. 3(A) to (C) as a function of rupture time. As increasing the rup-

S7: 600 °C, 20 kg/mm² t_r =26 223 h
(a) Bright field image
(b) EDX spectrum showing the precipitation of $M_{23}C_6$ (A and B type)
(c) EDX spectrum showing the precipitation of MC (C type, within lath)
(d) Dark field image (B type, at lath boundary)
(e) Electron diffraction pattern
(f) Key diagram
Photo. 1. Precipitation characteristics.

S3: 650 °C, 3 kg/mm²
 t_r =13 864 h head region
(a) STEM image (extraction replica)
(b) EDX spectrum of B precipitates ($M_{23}C_6$)
(c) EDX spectrum of A precipitates (VC)
Photo. 2. Precipitation characteristics.

Fig. 3. Precipitation characteristics.

ture time, amounts of precipitates and the lattice parameter of $M_{23}C_6$ increases. Precipitation partition ratio is in the order: $Nb > V > Cr > Mo > Fe$. It will be noted that the precipitation partition ratio of Mo increases markedly in long-term test.

3. Microstructure

1. Optical Microscopy

Types of microstructure can be divided into two by the degree of creep recovery according to optical microscopy. Whilst the specimen with Vickers hardness (Hv) above 180 exhibits the tempered martensitic structure, the specimen with Vickers hardness below 180 no more exhibits the original tempered martensitic structure and prior austenite grain boundaries can not be distinguished (Photo. 3). Furthermore, recrystallized ferrite grains formed in long-term creep test, and decarburized surface regions depleted in Cr (with a width of about $50\ \mu\text{m}$) were observed.

2. Transmission Microscopy

Transmission micrographs (Photo. 4(a)) of the specimens, which no more exhibit the original tempered martensitic structure under an optical microscope, exhibit the creep recovery stage of subgrain growth. Transmission electron micrographs of the specimens with Vickers hardness above 180 exhibit the clear lath martensitic structure which have some polygonized regions. Microstructure of long-term crept specimens of S7 and S9 at 600 and 650 °C are shown in Photo. 5. In the long-term crept specimens at 600 °C, in spite of the agglomeration of $M_{23}C_6$ at boundaries, fine precipitates are dispersed within lath. In the long-term crept specimens at 650 °C, subgrain growth is occurring with the coarsening of $M_{23}C_6$, and the fine precipitates within lath is markedly decreased. It will be noted that the creep recovery in S9 is accelerated by the marked coarsening of MX.

IV. Discussion

1. Long-term Creep Rupture Strength and Microstructure

1. Microstructure and Parabola-like Decrease in Creep Rupture Curve

Stress-rupture time relations at 600 and 650 °C are shown in Figs. 4 and 5. The number in the figure indicates the rupture elongation. Creep rupture strength of the steels which exhibit low creep rupture strength in long-term tests at 600 °C (for example, S9) decrease markedly in intermediate-term tests at 650 °C. This marked decrease in creep rupture strength (parabola-like decrease in the stress-rupture time relation) is an important factor in the error of the extrapolation method. It is considered that this parabola-like decrease in the stress-rupture time relation is greatly dependent on both fracture mode changes and microstructural changes, such as the agglomeration of precipitates, subgrain formation, and recrystallization.

We discuss the above two changes during creep at the testing temperature of 650 °C. The Vickers hardness and lath width change (or the change of mean diameter of subgrain) as a function of rupture

(a) S9: 600 °C, 20 kg/mm²
 $t_r = 10\ 595\ \text{h}$ (DPN=238)
 (b) S2: 650 °C, 3 kg/mm²
 $t_r = 57\ 819\ \text{h}$ (DPN=163)

Photo. 3. Optical micrographs.

(a) S2: 650 °C, 3 kg/mm² $t_r = 57\ 819\ \text{h}$
 (b) S6: 650 °C, 16 kg/mm²
 $t_r = 366\ \text{h}$ (A; $M_{23}C_6$, B; MX)

Photo. 4. Transmission electron micrographs.

time are shown in Figs. 6 and 7. The change of Vickers hardness as a function of rupture time is similar to the relation of stress-rupture time. This indicates that Vickers hardness and lath width, which are greatly concerned with the degree of creep recovery, well correspond to the creep rupture strength at 650 °C. All of the ruptured steels at 650 °C, with

(a) S7: 600 °C, 20 kg/mm² t_r =26 223 h
 (b) S9: 600 °C, 20 kg/mm² t_r =10 595 h
 (c) S7: 650 °C, 12 kg/mm² t_r =32 070 h
 (d) S9: 650 °C, 12 kg/mm² t_r =6 398 h
 Photo. 5.
 Transmission electron micrographs.

Fig. 4. Creep rupture properties at 600 °C.

Fig. 5. Creep rupture properties at 650 °C.

an exception of S4–8 kg/mm², exhibit typical transgranular ductile fracture surfaces (Photo. 7(a)). In the light of that, the parabola-like decrease in creep rupture strength at 650 °C is mainly attributed to the microstructural changes during creep rather than to the fracture mode changes.

In the steel which exhibits a marked drop in long-term creep rupture strength, microstructural changes, such as subgrain formation and subgrain growth markedly occur with the coarsening of the precipitates within laths. Electron micrographs (Photos. 4(b), 5(b), 5(d)) at the deviation point of linear extrapolation in the stress–rupture time relations, exhibit the creep recovery stage of subgrain growth with the coarsening of the precipitates at lath boundaries as a

Fig. 6. The change of hardness at gauge length region as a function of rupture time.

Fig. 7. The change of lath width as a function of rupture time.

result of the intensive lath boundary migration. The marked drop in long-term creep rupture strength of martensitic 12 % Cr steels at 650 °C is considered to be attributed to the subgrain growth (first recrystallization) with marked agglomerations of the precipitates within laths.

2. Effect of Alloying Element on Microstructure

The steels studied in this work have three different B content levels due to different melting process. It is well known that a minute addition of B astonish-

ingly affects creep rupture strength. Boron addition generally leads to high creep rupture strength due to matrix dot precipitation of $M_{23}(C,B)_6$. However, it sometimes leads to low creep rupture strength due to the formation of undissolved $M_{23}(C,B)_6$.¹⁵⁾ Unfortunately, the effect of B on long-term creep rupture strength is not apparent in this work. Here we only discuss the effect of V and Nb on long-term creep rupture strength of the same level of B contents.

In long-term creep tests, Mo added steels (S2) exhibit much higher creep resistance than V added steels (S3). In short-term creep tests, Mo exhibits 1) substitutional solid solution hardening and IS effects^{5,9)} (chemical interaction between interstitial atoms and substitutional atoms) and 2) stabilizing effects of $M_{23}C_6$ by depressing the rate of Ostwald ripening as a result of the low diffusion rate of Mo contained in $M_{23}C_6$, whereas, in long-term tests, the attribution of precipitation hardening markedly decreases with the coarsening of the precipitates, and IS effects decreases as a result of the precipitation of interstitial solutes, and therefore the main strengthening mechanism becomes substitutional solid solution hardening. The precipitation partition ratio of Mo after long-term creep test (Fig. 3(C)) is about 40 %, and therefore, 60 % of Mo is in solution and exhibits solid solution hardening in long-term creep test. Effect of the Laves phase precipitation on long-term creep rupture strength can not be fully explained because Fe_2Mo was identified by X-ray diffraction alone in this work. According to the recent work,¹⁰⁾ the agglomeration rate of Fe_2Mo is much higher than that of $M_{23}C_6$. Considering the decrease in solid solution hardening of Mo as a result of the precipitation of Fe_2Mo together with the rapid agglomeration rate of Fe_2Mo , the precipitation of Fe_2Mo is considered to be little effective on the long-term creep rupture strength. V exhibits dispersion hardening in short-term tests, whereas in long-term tests dispersion hardening markedly decreases with the coarsening of V_6C_6 up to certain diameters.^{6,14)} Transmission electron micrographs (Photo. 6) of S2 and S3 after long-term creep tests show that the creep recovery in S2 is greatly delayed due to the solid solution hardening of Mo, whereas the creep recovery in S3 is markedly accelerated with the coarsening of V_6C_6 . It should be noted that substitutional solid solution hardening is the main strengthening mechanism in long-term creep tests where precipitation or dispersion hardening is no longer effective.

In complex alloyed steels (S4~S6) containing 1.4 % Mo and 0.05 % B with variations in V and Nb contents, creep rupture strength of Nb-added steels (S6) decreases markedly at about 10 000 h. In S7~S9 steels containing 1.4 % Mo and 0.03 % B with variations in V and Nb contents, V-rich steels (S7) exhibit higher creep rupture strength than Nb-rich steels (S9). Alloying elements of V and Nb exhibit almost the same behavior in the point of dispersion hardening by similar types of MX precipitates. On the other hand, the solubility of V and Nb in austenite^{11,12)} greatly differ. Considering the solubility

(a) S2: 650 °C, 3 kg/mm²
 t_r = 57 819 h head region
 (b) S3: 650 °C, 3 kg/mm²
 t_r = 13 864 h head region

Photo. 6. Transmission electron micrographs.

of V and Nb in austenite in the case of 1 150 °C solution treatments, added amount of V about 0.2 % is completely in solution, whereas an addition of Nb above 0.04 % leads to undissolved NbC. It is considered that this undissolved NbC accelerates the MX-MX agglomeration during creep, and this leads to a marked drop in creep rupture strength of Nb-rich steels (S6, S9). V-rich steels (S7) exhibit higher 30 000 h creep rupture strength than Nb-rich steels (S9), which is probably attributed to the more fresh MX precipitation during creep and the lesser MX-MX agglomeration of V-rich steels compared with Nb-rich steels.

2. Microstructure and L.M. Extrapolation Method

In order to extrapolate the long-term creep rupture strength, the Larson-Miller parameter is often used. In this section, we discuss whether there exists any microstructural differences or not between the specimens tested at 600 and 650 °C which have the same value of the Larson-Miller parameter, $T(C + \log t)$.

The comparison of microstructures between long-term crept specimens at 600 °C and short-term crept specimens at 650 °C is shown in Table 2. In constant applied stress condition of 20 kg/mm², there seems to be little deviation in creep rupture strength between the experimental value at 600 °C and the extrapolated value from the data at 650 °C. However, there are marked differences in microstructures between long-term crept specimens at 600 °C and short-term crept specimens at 650 °C.

Long-term crept specimens at 600 °C have larger values, than short-term crept specimens at 650 °C, in the amounts of precipitates, the mean diameter of precipitates, the lattice parameter of $M_{23}C_6$, and

Table 2. Microstructural differences between long-term crept specimens at 600 °C and short-term crept specimens at 650 °C.

20 kg/mm ²	<i>t_r</i> (h)		<i>T</i> (25+log <i>t</i>) × 10 ⁻³		D.P.N.		Lath width (μ)		Precipitates amounts (wt%)		Mean diameter of ppts (μ)		Lattice parameter of M ₂₃ C ₆ (Å)	
	600 °C	650 °C	600 °C	650 °C	600 °C	650 °C	600 °C	650 °C	600 °C	650 °C	600 °C	650 °C	600 °C	650 °C
S 7	26 223	498	25.68	25.56	286	257	0.31	0.37	5.48	4.58	0.147	0.109	10.640	10.595
S 8	11 650	321	25.37	25.39	278	244	0.41	0.57	5.54	4.57	0.167	0.121	10.609	10.522
S 9	10 595	568	25.34	25.62	281	254	0.38	0.49	5.43	4.47	0.188	0.124	10.587	10.569

Vickers hardness at gauge length, and have smaller values in the mean width of martensite laths. That is, long-term crept specimens at 600 °C exhibit the marked delay in creep recovery in spite of agglomerations of precipitates. Volume fraction of precipitates in steel as a function of size is shown in Fig. 8. This figure shows that this marked delay of the creep recovery in long-term specimens at 600 °C is attributed to the fresh precipitation of fine precipitates (<0.2 μm) in long-term creep tests.

Considering the relation between the creep recovery level and creep rupture strength at 650 °C mentioned at Section IV. 1, the creep rupture strength expected by these marked delays of recovery was not obtained at 600 °C. This inconsistency between microstructure and creep rupture strength of long-term specimens at 600 °C can be explained in terms of the differences in fracture mode. The long-term crept specimens of S7~S9 steels at 600 °C-20 kg/mm² exhibit intergranular brittle fracture surface (Photo. 7(b)) which was fractured along the agglomerates at prior austenite grain boundaries. This intergranular fracture is probably caused by stress concentration at the incoherent agglomerates at boundaries. It is considered that this intergranular fracture leads to a marked drop in creep rupture strength.

3. Microstructural Change during Long-term Creep

In our previous work,¹³⁾ the microstructural change during creep (creep recovery process) was modeled as a function of the L.M. parameter. According to that model, microstructure of long-term specimens at 600 °C was expected to be in the subgrain growth stage. However, the long-term specimens at 600 °C did not exhibit the subgrain growth stage. It should be noted that it is inadvisable to use the L.M. parameter for the estimation of microstructural changes during creep, if there is a possibility of a change in precipitation reactions with creep testing temperature.

Microstructural changes during long-term creep was established based on the long-term crept microstructure (Fig. 9). Transformation product of the oil quenched specimen is shown in Fig. 9(A). The substructure is a duplex structure of lath martensite and lower bainite. The prior austenite grain consisted of bundle and co-variant packet of martensite in which lath was observed. Small amounts of cementite and undissolved NbC were also observed within lath.

Microstructure of long-term crept specimens at 600 °C is shown in Fig. 9(B). In spite of the agglomeration at grain boundaries, the original lath martensitic structure remained due to continuing fresh pre-

Fig. 8. Relation between precipitates size and volume fraction of precipitates for S7 steel.

(a) S8: 650 °C, 20 kg/mm² *t_r* = 321 h
(b) S8: 600 °C, 20 kg/mm² *t_r* = 11 650 h

Photo. 7. Scanning electron micrographs.

cipitation of fine precipitates. Note that the marked agglomeration of M₂₃C₆ at packet boundaries occurs together with agglomeration at prior austenite grain boundaries.

Microstructure of long-term crept specimens at 650 °C shows little differences from the model of previous work. As creep recovery continues, a secondary recrystallization nucleus which has a very low dislocation density is observed, and the original tempered martensitic substructure can not be distinguished even through an optical microscope.

Fig. 9. Microstructure of the 12CrMoVNb steel after long-term creep test.

V. Conclusions

Long-term creep rupture strength and microstructure of 12 Cr steels have been investigated with the specimens ruptured at 10 000~60 000 h. The main results are as follows.

(1) Solid solution hardening by an addition of Mo is very effective on long-term creep rupture strength, but precipitation or dispersion hardening by an individual addition of V or Nb is little effective. Especially, the presence of excessive amounts of undissolved NbC leads to the coarsening of MX precipitates, and a marked drop in long-term creep rupture strength. The combined addition of V+Nb with 1.5 % Mo and 0.03 % B leads to superior 30 000 h creep rupture strength. The V-rich 12Cr-MoVNb steels have higher 30 000 h creep rupture strengths than the Nb-rich steels, which is probably attributed to the more fresh MX precipitation during creep and the lesser MX-MX agglomeration of the V-rich steels.

(2) The marked parabola-like decrease in the creep rupture strength of 12CrMoVNb steels at 650 °C is well concerned with the microstructural changes, in particular, the subgrain growth stage (first recrystallization).

(3) Microstructural differences between long-term crept specimens at 650 °C have been studied by adopting the Larson-Miller parameter, $T(C+\log t)$. Long-term crept specimens at 600 °C exhibit higher dislocation density substructure caused by intensive precipitation reactions which include both fresh precipitation within martensite lath and agglomeration at grain boundaries. On the other hand, long-term specimens at 600 °C exhibit the intergranular brittle fracture, whereas short-term specimens at 650 °C exhibit the transgranular ductile fracture. It is considered that the deviation of long-term creep rupture

strength from the extrapolated value is caused by differences in precipitation reactions and fracture mode.

Acknowledgements

The authors are grateful to Dr. Norio Takahashi, Mr. K. Asakura, Mr. T. Tsuchiyama, Mr. K. Nagai, Mr. M. Murakami for many helpful discussions and their encouragement throughout the work.

REFERENCES

- 1) J. Watanabe: Proceedings of the 23th Japan National Symposium on Strength, Fracture and Fatigue, Tokyo, (1978), 79.
- 2) E. A. Little, D. R. Harries and F. B. Pickering: Ferritic Steels for Fast Reactor Steam Generators, BNES, London, (1978), 136.
- 3) A. Hede and B. Aronsson: *JISI*, **207** (1969), 1241.
- 4) J. Koutsky: *JISI*, **205** (1967), 763.
- 5) B. B. Argent, M. N. van Niekerk and G. A. Redfern: *JISI*, **208** (1970), 830.
- 6) K. Suzuki, T. Nishi, T. Muta and T. Hirota: *Trans. ISIJ*, **11** (1971), 1.
- 7) M. Tanino: *Bull. Japan Inst. Metals*, **6** (1967), 23.
- 8) K. J. Irvine, D. J. Crowe and F. B. Pickering: *JISI*, **195** (1960), 386.
- 9) K. Monma, H. Sato and H. Kokita: *J. Japan Inst. Metals*, **29** (1965), 941.
- 10) S. Komatsu, M. Nakahashi, S. Nakamura and Y. Nakamura: Report of the 123rd Committee (Heat Resisting Metals and Alloys), Japan Society for the Promotion of Science (JSPS), **21** (1980), 1.
- 11) H. Nordberg and B. Aronsson: *JISI*, **206** (1968), 1263.
- 12) K. Narita: *Bull. Japan Inst. Metals*, **8** (1969), 49.
- 13) I. M. Park, T. Fujita and K. Asakura: *Trans. ISIJ*, **20** (1980), 99.
- 14) J. D. Baird, A. Jamieson, R. R. Preston and R. C. Cochran: Creep Strength in Steel and High Temperature Alloys, The Metals Society, London, (1974), 207.
- 15) I. M. Park and T. Fujita: *Trans. ISIJ*, **20** (1980), B129.