Effect of Matrix Substructures on Precipitation of the Laves Phase in Fe–Cr–Nb–Ni System

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In order to make clear the effect of matrix substructures on the Laves phase precipitation in Fe–Cr–Nb–Ni system, age hardening behavior was examined during isothermal aging up to 1 000 h at 873 and 973 K and the precipitation morphology of the Laves phase was investigated by transmission electron microscopy (TEM). The microstructures of Fe–10Cr–1Nb and Fe–15Cr–1Nb (mol%) alloys with ferrite matrix contain high density of rod-shaped Laves phase which precipitates within the ferrite grains during aging. On the other hand, in the Fe–10Cr–1Nb–1Ni alloy with massive ferrite matrix, most of the Laves phase precipitates in subgrain boundaries. Additionally the deformed ferrite matrix has been prepared for examining the effect of dislocation density on morphology of the Laves phase. It is found that the difference of matrix substructures between the ferrite and deformed ferrite provides the different number and distribution of nucleation sites for precipitation of the Laves phase. Mechanical properties were evaluated by tensile tests conducted at room temperature, 873 and 973 K.

KEY WORDS: precipitation; microscopy.

1. Introduction

High Cr ferritic heat resistant steels are widely used in power generating plants because of their high creep strength and superior oxidation resistance. Conventional ferritic steels are usually based on tempered martensite and their creep strength is sustained by precipitation hardening of MC carbides and MX carbonitrides as well as by solid solution hardening of Mo, V, Nb and Ti. The precipitation sequence in high Cr ferritic steels has been revealed to be quite complicated as that it involves precipitation of some meta-stable and stable intermetallic phases other than MC carbides and MX carbonitrides. Recently, C14 type Laves phases have drawn a growing interest as strengthener to improve the creep resistance of the steels.¹⁾

Laves phases are A₂B-type intermetallic compounds with TCP (Topologically Close Packed) structures; hexagonal (C14 and C36) and cubic (C15). These structures are expected to have high strength at elevated temperatures. Uchida *et al.*²⁾ found that the creep rupture strength was improved by the additions of Mo, W and Ti to 2.25 wt% Cr ferritic steels due to the precipitation of Laves phases (Fe₂Mo, Fe₂W and Fe₂Ti) during tempering. Also it is suggested that fine precipitation of the Laves phase Fe₂W under the service condition improves the creep resistance of recently developed ferritic steels containing W.³⁾ It has been pointed out, however, that the precipitation of the Laves phase reduces toughness of the steels particularly at ambient temperature because of its poor deformability due to the

complicated crystal structure.⁴⁾

The difference of matrix substructures could affect the morphology of precipitates. Number of nucleation sites for the precipitation increases with increasing excess energy arising from defects in matrix substructures. It is necessary to understand the effect of matrix substructures on the precipitation behavior for the present alloy design based on precipitation hardening of the Laves phase. Suzuki *et al.*⁵⁾ systematically investigated the age-hardening behavior of maraging steels and ferrite matrix steels and showed that the activation energy of precipitation in the martensite matrix is generally lower than that in the ferrite. Maruyama *et al.*⁶⁾ have investigated the precipitation and phase transformation of copper particles in low alloy ferritic and martensitic steels to show that the precipitation of copper particles and the corresponding hardening behavior in the martensite matrix are different from that in the ferrite. The main reason for this difference is considered to be responsible for the much higher dislocation density in martensite than in ferrite.

The present authors have made every effort to establish the basis of the development of carbon-free ferritic heat resistant steels in which the Laves phase is solely the strengthener in the ferrite matrix.⁷⁾ In the ternary Fe–Cr–Nb system, it is possible to design the two-phase alloys consisting of α -Fe and C14 Fe₂Nb which are stable up to high temperatures at around 10 at% Cr. With the addition of an austenite former Ni, the γ/α transformation can be involved in the quaternary Fe–Cr–Nb–Ni system. Utilizing the γ/α

transformation through proper adjustment in alloy chemistry and heat treatment provides various types of α -Fe matrix substructure, and thereby we believe it is possible to modify the morphology of the Laves phase precipitates in α -Fe matrix by selecting its substructure. Consequently, we could improve elevated temperature strength as well as toughness of the ferritic steels. The objective of the present study is to investigate the effects of α -Fe matrix substructures on the morphology and thermal stability of the Laves phase, Fe_2Nb , precipitation in Fe–Cr–Nb–Ni system. Variations of the α -Fe matrix substructure widely controlled in this work are ferrite, massive ferrite. Additionally, the deformed ferrite matrix has been prepared for examining the effect of dislocation density on the Laves phase precipitates. Moreover mechanical properties were investigated to evaluate high temperature strength and ambient temperature ductility.

2. Experimental Procedure

Nominal compositions of the alloys used in the present study are Fe–10Cr–1Nb, Fe–10Cr–1Nb–1Ni and Fe–15Cr–1Nb in mol%. The corresponding alloy compositions in wt% are Fe–9.3Cr–1.7Nb, Fe–9.3Cr–1.7Nb–1Ni and Fe–14Cr–1.7Nb respectively. They were prepared by induction melting in an alumina crucible and casting into a mold in an argon atmosphere. The ingots of about 300 g (30 mm in diameter and 50 mm in height) were then hot rolled at 1373 K to a reduction ratio of 80% in thickness. Temperature of the ingots at the final rolling path was monitored at least higher than 1073 K by a radiation thermometer.

The γ/α transformation temperature was determined by a dilatometer. The cylindrical specimens with the size of 3 mm in diameter and 10 mm in height were machined from the as-hot-rolled plates. Specimens were heated to 1273 K at 8 K min^{-1} , held for 5 min, and cooled to room temperature at 50 K min^{-1} .

Tensile specimens with the dimension of 15 mm in gage length, 3 mm in width and 1 mm in thickness were prepared by electrodischarge machining from the rolled plate. The tensile tests were carried out on as-rolled specimens and those annealed at 1023 K for 1 h. The tests were carried out at the initial strain rate of $1.1 \times 10^{-4}\text{ s}^{-1}$ at room temperature, 873 and 973 K in a vacuum of about 10^{-4} torr. Loading direction was chosen as to be parallel to the rolling direction.

Age hardening behavior was examined on as-hot-rolled materials by measuring the hardness change during isothermal aging for 0.1 to 1000 h at 873 and 973 K. Micro-Vickers tests were conducted for this purpose under 1 kg load.

Microstructures of the alloys were observed by optical microscope (OM), scanning electron microscope (SEM), and transmission electron microscope (TEM) at an accelerating voltage of 200 kV. The specimens for observations by OM and SEM were mechanically polished and then etched in a picric acid solution. The SEM observations were done by using back-scattered electron images of compositional mode. Thin foils for TEM observation were prepared by twin-jet electropolishing using a solution of 10% perchloric

acid and 90% methanol at below 223 K. The TEM observations were carried out by using bright field images and selected area diffraction (SAD) patterns for structure analysis.

3. Results and Discussion

3.1. Control of the Matrix Substructures

It is possible to modify the morphology of the Laves phase precipitates by controlling substructure of the α -Fe matrix by selecting alloy compositions and heat treatments. **Table 1** summarizes the matrix substructures and notations of the alloys. The alloys are hereafter expressed by chromium and nickel concentrations in mol% and the characteristic of matrix substructure, for example, the Fe–10Cr–1Nb–0Ni (mol%) alloy with ferrite matrix is denoted as 10–0–F. Two types of matrix substructures, massive ferrite (10–1–MF alloy) and ferrite (10–0–F, 15–0–F alloy), are provided by utilizing the γ/α transformation. The γ/α transformation temperatures are listed in **Table 2**. Since 10–0–F and 15–0–F alloys have no γ/α transformation, the matrix of these alloys is ferrite. With the addition of 1 mol% Ni, the Ar_3 (γ/α transformation temperature during cooling) point appears at about 823 K in 10–1–MF alloy. This Ar_3 temperature is far above the M_s temperature (starting martensitic transformation temperature) which is estimated to be 603 K by the empirical calculation using concentrations of ferrite forming and austenite forming elements.⁸⁾ Therefore the massive transformation from γ -Fe to α -Fe takes place during cooling in the 10–1–MF alloy instead of the martensitic transformation. In low-carbon and ultra low-carbon steels, it is known that the transformation temperature of massive ferrite formation is higher than the M_s temperature.⁹⁾

3.1.1. Microstructures of the As-hot-rolled Alloys

The back-scattered electron images of the as-hot-rolled 10–0–F and 10–1–MF alloys on their transverse sections are shown in **Fig. 1**. The bright phase is Fe_2Nb , and the dark phase is α -Fe matrix. The microstructure of the 10–0–F alloy shown in **Fig. 1(a)** consists of ferrite matrix and a small amount of Fe_2Nb particles. In the 10–1–MF alloy, γ -Fe and Fe_2Nb should coexist with each other at the temperature that hot-rolling was employed, and the γ -Fe matrix transforms to massive ferrite upon subsequent cooling. As shown in **Fig. 1(b)**, the matrix is massive ferrite with irregular boundaries, in which the fine spherical Fe_2Nb precipitates are dispersed uniformly.

Table 1. Characteristics of the matrix microstructure of the experimental alloys.

Alloys (mol%)	Matrix (symbol)	Notation
Fe–10Cr–1Nb–0Ni	Ferrite (F)	10–0–F
Fe–15Cr–1Nb–0Ni	Ferrite (F)	15–0–F
Fe–10Cr–1Nb–1Ni	Massive ferrite (MF)	10–1–MF

Table 2. γ - α transformation temperatures of the experimental alloys.

Alloys	γ - α transformation temperature (K)			
	$\alpha \rightarrow \alpha + \gamma$	$\alpha + \gamma \rightarrow \gamma$	$\gamma \rightarrow \alpha + \gamma$	$\alpha + \gamma \rightarrow \alpha$
10–0–F, 15–0–F	*	*	*	*
10–1–MF	1103	1133	823	733

*=No γ - α transformation

Fig. 1. Back scattered electron images of as-rolled alloys, (a) 10-0-F, (b) 10-1-MF.

Fig. 2. Bright field images of as-rolled alloys, (a) 10-0-F, (b) 10-1-MF.

The bright field images of the as-hot-rolled 10-0-F and 10-1-MF alloys are shown in Fig. 2. As shown in Fig. 2(a), the dislocation density in the ferrite matrix is low in 10-0-F alloy. Fine Fe_2Nb particles of about a few hundreds nm in size which presumably form during hot rolling are located along the ferrite grain boundaries. On the other hand, the cellular substructure with high dislocation density is the microstructural feature of the matrix in 10-1-MF alloy, and spherical Fe_2Nb precipitates with less than $1\text{ }\mu\text{m}$ in diameter are distributed in the matrix as shown in Fig. 2(b). It is noted that Fe_2Nb exists in the as-hot-rolled microstructure of the 10-0-F and 10-1-MF alloys because both the alloys are consisting of the two-phase $\alpha\text{-Fe}/\text{Fe}_2\text{Nb}$ microstructure at hot-rolled temperature.

3.1.2. Microstructures of the Annealed Alloys

Figure 3 shows a bright field image and the corresponding selected area diffraction (SAD) of 10-0-F alloy annealed at 1023 K for 1 h. The microstructure consists of numbers of rod-shaped precipitates and the ferrite matrix. These fine precipitates are identified as Fe_2Nb Laves phase with the C14 crystal structure, which must have secondarily precipitated upon annealing.

The crystallographic orientation relationship between the C14 Fe_2Nb phase and bcc matrix is analyzed using SAD patterns as $(0001)_{\text{Fe}_2\text{Nb}}// (211)_{\alpha\text{-Fe}}$ and $[4\bar{5}10]_{\text{Fe}_2\text{Nb}}/[31\bar{7}]_{\alpha\text{-Fe}}$ (Figs. 3(b), 3(c)). Figure 4 gives the stereographic projection of the orientation relationship between $\alpha\text{-Fe}$ and Fe_2Nb Laves phase. This orientation relationship is equivalent to the following expression by lower mirror indices; $(0001)_{\text{Fe}_2\text{Nb}}// (211)_{\alpha\text{-Fe}}$, $[\bar{2}110]_{\text{Fe}_2\text{Nb}}/[0\bar{1}1]_{\alpha\text{-Fe}}$, and $[01\bar{1}0]_{\text{Fe}_2\text{Nb}}/[\bar{1}11]_{\alpha\text{-Fe}}$. Note that the orientation relationship obtained

Fig. 3. (a) Bright field image of 10-0-F alloy after aging at 1023 K for 1 h, (b) SAD pattern, and (c) indexing of the SAD pattern.

here is the same as that of $\alpha\text{-Fe}$ and C14 Fe_2W in the Cr-W steels reported by Abe *et al.* and Murata *et al.*^{10,11)}

The microstructure of the 10-1-MF alloy annealed at 1023 K for 1 h is shown in Fig. 5. Recovery of the massive

Fig. 4. Stereographic projection of the orientation relationship between α -Fe and Fe_2Nb , C14 Laves phase.

Fig. 5. Bright field image of 10-1-MF alloy after aging at 1023 K for 1 h.

ferrite and subgrain growth are the features of the annealed matrix compared with as-hot-rolled microstructure shown in Fig. 2(b). The Fe_2Nb phase precipitates preferentially along subgrain boundaries, and they are thought to exert pinning force on subgrain boundaries. Comparing Fig. 3(a) with Fig. 5, it is found that extensively large volume fraction of Fe_2Nb phase precipitates within grains of the 10-0-F alloy (Fig. 3(a)) but not of 10-1-MF alloy (Fig. 5) annealed at 1023 K for 1 h. In 9Cr-W steels, the Fe_2W Laves phase is known to precipitate extensively in ferrite rather than in martensite.¹⁰⁾

3.1.3. Relationship between Microstructures and Tensile Properties

To evaluate the effect of matrix substructure on static strength and elongation to failure, tensile tests at room temperature and at 873, 973 K were conducted on as-hot-rolled and annealed 10-0-F and 10-1-MF alloys. The present annealing condition at 1023 K for 1 h is nearly the same heat treatment applied to commercial ferritic heat resistant steels with tempered martensite matrix. Figure 6 shows the temperature dependence of the tensile 0.2% flow stress and

Fig. 6. Temperature dependence of the tensile 0.2% flow stress and elongation to failure of 10-0-F and 10-1-MF alloys for as-rolled and annealed at 1023 K for 1 h. Data available for a commercial ferritic heat resistant, ASME SA-213/SA-213M Grade T91 steel, are shown together by broken lines in Fig. 6.¹²⁾

elongation to failure of the as-hot-rolled and annealed 10-0-F, 10-1-MF alloys. For comparison, those of the commercial ferritic heat resistant steel, ASME SA-213/SA-213M Grade T91, based on tempered martensite matrix are shown together by broken lines in Fig. 6.¹²⁾

By 1 mol% Ni addition, the tensile 0.2% flow stress of as-hot-rolled 10-1-MF alloy clearly increases and the elongation to failure decreases at room temperature as compared to the as-hot-rolled 10-0-F alloy. It is attributed to the higher dislocation density in the matrix of 10-1-MF alloy than that of 10-0-F alloy as shown in Fig. 2. The strength of 10-1-MF alloy is also much higher than that of 10-0-F alloy at 873 K, however it is almost comparable at 973 K between both alloys. By annealing at 1023 K for 1 h, the strength of 10-1-MF alloy considerably reduced while the elongation to failure is improved at room temperature. The annealed 10-1-MF alloy exhibits much lower strength and much higher elongation to failure than those of as-hot-rolled alloy at 873 K.

The strength of the as-hot-rolled 10-0-F alloys is almost temperature independent up to 973 K. The reason would be that precipitation of Fe_2Nb within grains occurs during holding about one hour prior to and upon the tensile tests at high temperatures and that it suppresses the decrease of strength. Fujita *et al.* indicated that precipitation of Fe_2Nb in ferritic stainless steels during the tensile tests at 1223 K might contribute to high strength at elevated temperatures.¹³⁾

3.2. Thermal Stability of Microstructure during Prolonged Aging

3.2.1. Changes in Microstructure and Hardness during Isothermal Aging

The thermal stability of the microstructure during long term exposure at high temperatures is very important for evaluating the durability of heat resistant steels since mechanical properties such as creep strength and toughness are strongly affected by the microstructures. **Figure 7** shows the changes in Vickers hardness of 10-0-F, 10-1-MF and 15-0-F alloys during isothermal aging at 873 and 973 K from 0.1 to 1000 h. The hardness of 15-0-F alloy is higher than that of 10-0-F alloy even though they have the ferrite matrix in common and comparable amount of the Laves phase precipitation. Solid solution hardening by chromium is presumably higher in the 15-0-F alloy than in the 10-0-F alloy. The as-hot-rolled 10-1-MF alloy with massive ferrite matrix has considerably high hardness as compared to the ferrite matrix alloys. This difference in hardness due to matrix substructures becomes smaller as the aging temperature and/or aging time increases, because the recovery of massive ferrite matrix proceeds in the 10-1-MF alloy during aging. Finally, the hardness of 10-1-MF alloy becomes lower than that of the as-rolled state, and comparable to that of 10-0-F alloy, after aging at 873 K for 1000 h and at 973 K as well.

The microstructures of 10-0-F, 10-1-MF and 15-0-F alloys aged at 873 and 973 K for 1000 h are shown in **Fig. 8**. The microstructure of 10-0-F alloy aged at 873 K contains high-density rod-shaped precipitates, less than 200 nm in length, existing in the ferrite matrix and on the subgrain boundaries. The size of the precipitates in the same alloy aged at 973 K is about 500 nm in length as shown in Fig. 8(b). Spherical Laves phase precipitates are observed in the subgrain boundaries in the 10-1-MF alloy aged at 873 K as shown in Fig. 8(c). Comparing aging temperatures, the recovery of the matrix is more extensive at 973 K than at 873 K as clearly seen in Fig. 8(d); the subgrains are larger and the dislocation density is much lower in the alloy aged

at 973 K, 8(d), than at 873 K, 8(c). Figures 8(e) and 8(f) show bright field images of 15-0-F alloy aged at 873 and 973 K for 1000 h, respectively. The microstructures contain the rod-shaped precipitates within the ferrite matrix and in the subgrain boundaries, which is similar to the microstructures of the 10-0-F alloy shown in Figs. 8(a) and 8(b). By tilting experiment, it is confirmed that the precipitates are rod-shaped within ferrite matrix. It is noted that the shape of Fe_2W Laves phase within martensite lath is plate as reported by Murata *et al.*¹⁴⁾

Fig. 7. Changes in Vickers hardness of the as-rolled 10-0-F, 10-1-MF and 15-0-F alloys as a function of isothermal aging at 873 and 973 K.

Fig. 8. Bright field images of 10-0-F, 10-1-MF and 15-0-F alloys aged at 873 and 973 K for 1000 h.

3.2.2. Consideration for Thermal Stability of Microstructures

The thermal stability of microstructures during aging of long duration more than 1 000 h could be discussed in terms of the morphology and dispersion of the Laves phase precipitates. The precipitation of the Laves phase in the ferrite matrix has relatively uniform distribution and rod-shaped morphology. These alloys maintain a high hardness level though slightly less than peak hardness even after aging for 1 000 h at 873 K. On the other hand in the massive ferrite matrix, the Laves phase precipitates in the subgrain boundaries with spherical morphology. Aging at 873 K for 1 000 h is obviously the over-aged condition for this alloy. The hardness of the ferrite matrix alloys is almost equal to that of the massive ferrite matrix alloy after aging for 1 000 h at both 873 and 973 K, however it is speculated through the above discussion that the microstructures of ferrite matrix alloys would be more stable due to less coarsening of precipitates which are more uniformly distributed and less recovery of the matrix than those of the massive ferrite matrix alloys by comparing under the same aging condition at 873 and 973 K for over 1 000 h.

3.3. Effect of Matrix Substructures on Precipitation of Laves Phase in the Fe-10Cr-1Nb-1Ni Alloy

3.3.1. Variation of Matrix Substructures

As described in the Sec. 3.2, it is found that the age hardening behavior and precipitation morphology of the Laves phase in the Ni-free ferrite matrix alloys are considerably different from those in the massive ferrite matrix alloy with Ni addition. The number of available nucleation sites is thought to be provided depending on the dislocation density in the matrix substructure. To elucidate this speculation, the isothermal age hardening and precipitation morphology are investigated in the Fe-10Cr-1Nb-1Ni alloy for the three types of matrix substructures prepared by various heat treatments including compressive deformation. The matrix substructures and heat treatment conditions are listed in **Table 3**. As mentioned earlier, as-hot-rolled 10-1-MF alloy has the massive ferrite matrix as shown in the optical micrograph in **Fig. 9(a)**. The alloy having ferrite matrix with large grain size as shown in **Fig. 9(b)** is prepared by the heat treatment of heating to 1 573 K in the δ -ferrite region for 1 h and subsequently quenching into water. By cold-rolling with the reduction ratio of about 50% in thickness, the matrix becomes deformed ferrite which is characterized by the elongated ferrite grains with deformation bands as shown in **Fig. 9(c)**.

3.3.2. Precipitation of Laves Phase and Hardness Change of the Alloys

The changes in hardness of the alloys with ferrite, massive ferrite and deformed ferrite matrix substructures during isothermal aging at 873 K are shown in **Fig. 10**. Particularly in the ferrite matrix alloy, the peak of hardness shifts to longer aging time in comparison with those in the massive ferrite matrix and deformed ferrite matrix alloys. The age-hardening behavior of deformed ferrite matrix alloy is almost the same as that of massive ferrite matrix alloy, although hardness of deformed ferrite matrix alloy is higher than that of massive ferrite matrix alloy.

Bright field images of the alloys with ferrite matrix and

Table 3. Characteristics of the matrix substructures in the Fe-10Cr-1Nb-1Ni alloy.

Matrix	Processing
Massive ferrite	Hot rolling W.Q.
Ferrite	Hot rolling W.Q.+1573K1h W.Q.
Deformed ferrite	Hot rolling W.Q.+1573K1h W.Q.+Cold rolling

Fig. 9. Optical micrographs in the three types of matrix substructures in Fe-10Cr-1Nb-1Ni alloy, (a) massive ferrite, (b) ferrite and (c) deformed ferrite matrix.

Fig. 10. Changes in Vickers hardness of massive ferrite, ferrite and deformed ferrite matrixes in Fe-10Cr-1Nb-1Ni alloy during isothermal aging at 873 K.

Fig. 11. Bright field images of Fe-10Cr-1Nb-1Ni alloy aged at 873 K for 1000 h, (a) ferrite and (b) deformed ferrite matrix.

deformed ferrite matrix aged at 873 K for 1000 h are shown in Fig. 11. A large number of precipitates in the ferrite matrix are observed as shown in Fig. 11(a). The dislocation density in the ferrite matrix is low and the shape of the precipitates is rod-shaped. In the deformed ferrite matrix, the Laves phase preferably precipitates in the subgrain boundaries as shown in Fig. 11(b). It is assumed that the substructure of deformed ferrite matrix changes from dislocation cell structures to subgrain structures by recovery during aging. It is found that the difference of matrix substructures between the ferrite and the deformed ferrite provides the different number and distribution of nucleation sites for precipitation of the Laves phase. In the deformed ferrite matrix, the coarse Laves phase preferably precipitates in the subgrain boundaries, which is similar to the case in the alloy with massive ferrite matrix. Therefore there seems to be no obvious difference between the massive ferrite and the deformed ferrite in respect of the number and distribution of nucleation sites at subgrain boundaries. While in the alloy with the ferrite matrix, rather small Laves phase precipitates uniformly in the matrix and the microstructure is more stable than the other two alloys.

4. Conclusions

Effect of the matrix substructures; ferrite, massive ferrite and deformed ferrite, on precipitation of C14 Laves phase was studied in Fe-Cr-Nb-Ni system.

The following conclusions are drawn from the present work.

(1) The microstructures of as-hot-rolled 10-0-F and 15-0-F alloys consist of the ferrite matrix and a small amount of Fe_2Nb particles, since those alloys have no γ/α transformation. The microstructure of as-hot-rolled 10-1-MF alloy consists of the massive ferrite matrix with high dislocation density and spherical Fe_2Nb precipitates.

(2) The microstructure of the alloy with the ferrite matrix contains high density of rod-shaped Laves phase which precipitates within the ferrite grains during annealing at 1023 K for 1 h. On the other hand in the massive ferrite matrix, most of the Laves phase precipitates in the subgrain boundaries.

(3) 0.2% flow stress in tensile test of 10-1-MF alloy with high dislocation density is much higher than that of 10-0-F alloy up to 873 K, however it is almost comparable at 973 K between both alloys.

(4) The hardness of 10-1-MF alloy decreases because of the recovery of massive ferrite matrix as the aging temperature and/or aging time increases. Finally, the hardness of this alloy after aging at 873 and 973 K for 1000 h is almost equal to those of 10-0-F and 15-0-F alloys. It is suggested that high temperature strength can be maintained for long duration in the ferrite matrix alloys because of relatively uniform dispersion of the precipitations.

(5) In comparison of aging behavior in the massive ferrite, the ferrite and the deformed ferrite matrix in the Fe-10Cr-1Nb-1Ni alloy, the age-hardening behavior and morphology of precipitates in the deformed ferrite matrix is almost the same as that of the massive ferrite. The difference between the ferrite and the deformed ferrite matrix is the number and distribution of nucleation sites for the Laves phase precipitates.

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