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Mechanical properties of Fe–36 Ni base alloys with chromium addition up to 5 mass%

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Mechanical properties, such as yield strength ($\sigma_{0.2}$), tensile strength (σ_B), elongation (Δl), strain-hardening coefficient (n) and modulus of elasticity (E), for chromium added Fe–36 mass% Ni with chromium addition of up to 5 mass% were studied as a function of annealing temperature from 1053 to 1423 K in a dry hydrogen atmosphere. When chromium was added to Fe–36 Ni, $\sigma_{0.2}$ decreased, but n and E increased. σ_B and Δl scarcely changed with the chromium addition. $\sigma_{0.2}$, σ_B, Δl and E decreased with the increase in annealing temperature. The relation between $\sigma_{0.2}$ and grain size was in agreement with Petch's equation.

1. Introduction

Fe–36 mass % Ni (Fe–36 Ni), called invar, is known as an alloy with very low thermal expansion coefficient. It is used as a shadow mask [1, 2] for cathode ray tubes for colour TV, and for example as the structural material for liquid nitrogen gas tanks in industrial fields [3]. Some investigations on Fe–36 Ni sheet for shadow mask were carried out to obtain good formability [1, 4, 5], etching [1] and oxidizability [6]. Chromium-added Fe–36 Ni, developed by the author and co-workers for use in a shadow mask, is also a useful material as it is more formable than Fe–36 Ni [4]. In this paper, mechanical properties such as 0.2% yield strength, tensile strength, modulus of elasticity and elongation are examined as function of the annealing temperature, and the softening mechanism due to chromium addition, which has already been investigated by the author using Mössbauer spectroscopy [7, 8], is considered and reported on.

2. Experimental procedure

Three samples, whose chromium concentration was 0, 2.0 and 5.0 mass %, respectively, were prepared as 10-kg ingots of iron, nickel and chromium for 99.9 mass % or better purity by vacuum melting at 10^{-4} torr in a zirconia crucible. The ingot (10 kg, C < 0.01, Si < 0.02, P < 0.005, S < 0.001, Mn < 0.3 mass %) was hot-rolled into a 5-mm-thick sheet. It was further reduced to 0.2-mm thickness by repeated cold rolling and annealing. Under a final reduction of 80% and final annealing at 823 K, the grain size was adjusted to about 20000 mm^{-2}. Tension test specimens were cut from these sheets and annealed for 1800 s in a dry hydrogen atmosphere in the range from 1053 to 1423 K. The tensile tests were carried out using an Instron 1153 testing machine. The yield strength ($\sigma_{0.2}$) was determined as 0.2% proof stress. The tensile strength (σ_B), equal to breaking point in this experiment, was the stress on breaking. The elongation (Δl) was measured with the fractured specimen. The strain-hardening coefficient (n) was determined at a strain range from 4 to 10%. The grain size (d) was determined by the number of grains in 25 mm^2, in a micrograph of × 100 magnification. The modulus of elasticity (E) was measured by a conventional tester using the vibration method.

3. Results

Fig. 1 shows the change in $\sigma_{0.2}$ as a function of annealing temperature. The $\sigma_{0.2}$ values of three types of invar alloys are nearly equal before annealing. At each annealing temperature, the chromium addition decreased $\sigma_{0.2}$ for Fe–36 Ni. $\sigma_{0.2}$ showed a straight decrease with increasing annealing temperature. This result is ascribed to an increase in grain size associated with recrystallization. The d value for the alloys becomes 0.016, 0.032, 0.061 and 0.125 mm at the annealing temperatures of 1057, 1177, 1277 and 1423 K, respectively. This tendency is similar in the three alloys. The relation between $\sigma_{0.2}$ and d is commonly represented as the following equation (Petch's equation)

$$\sigma_{0.2} = k_f d^{-1/2} + \sigma_0 \quad (1)$$

where k_f is a constant indicating the stopping power of dislocations, and σ_0 is the friction stress for dislocations. As shown in Fig. 2, the $\sigma_{0.2}$ value is proportional to $d^{-1/2}$, which suggests that Petch's equation can be adopted for this experiment. From the slope of the straight line, k_f is calculated as 1.2 kg $mm^{-3/2}$ both for Fe–36 Ni and Fe–36 Ni–Cr. This result suggests that the stopping power is not influenced by the chromium addition in Fe–36 Ni. The σ_0 values are 17.4, 14.4 and 8.4 kg mm^{-2} in the Fe–36 Ni, Fe–36 Ni–2Cr and Fe–36 Ni–5Cr samples, respectively. The friction stress of dislocations decreases with the increase of chromium addition.

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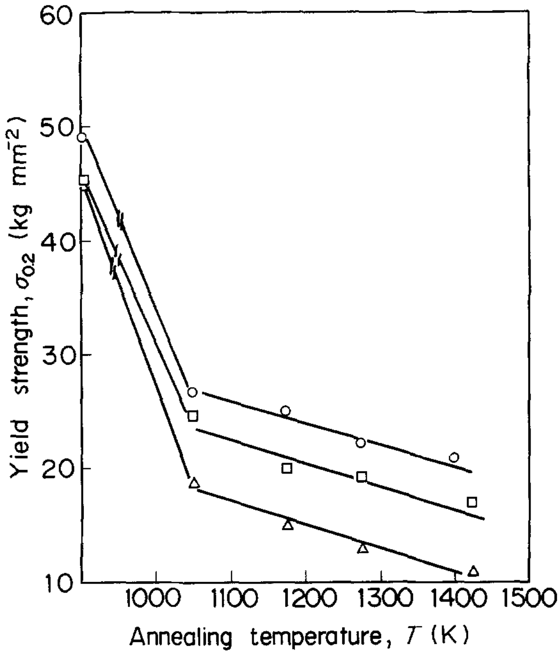

Figure 1 Change in $\sigma_{0.2}$ as a function of annealing temperature. ○, Fe–36 Ni; □, Fe–36 Ni–2Cr; △, Fe–36 Ni–5Cr.

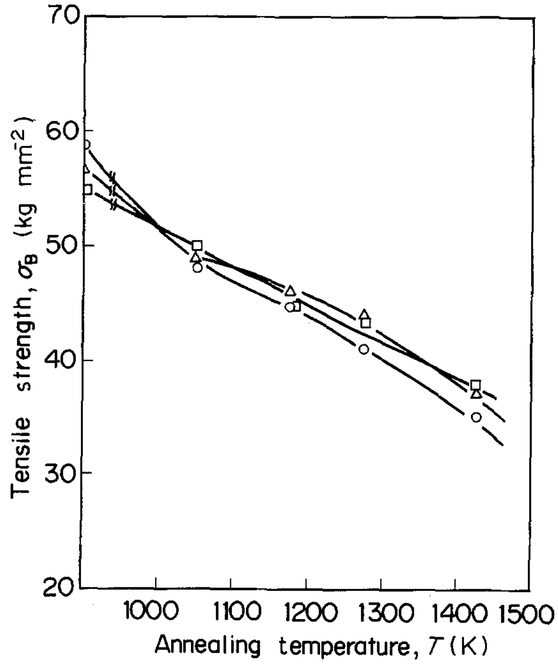

Figure 3 Changes in σ_B as a function of annealing temperature. ○, Fe–36 Ni; □, Fe–36 Ni–2Cr; △, Fe–36 Ni–5Cr.

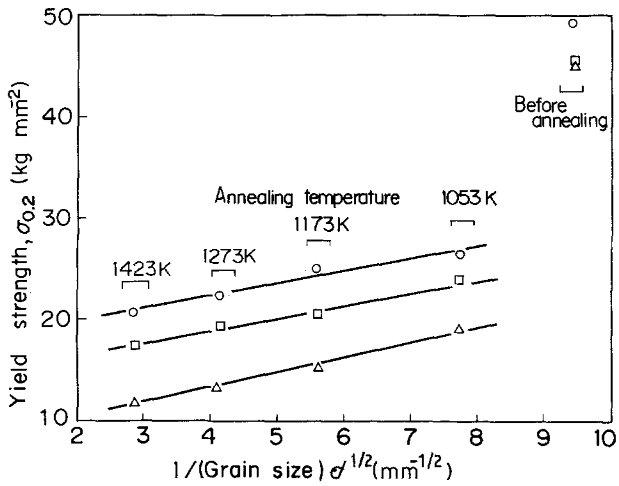

Figure 2 Relations between $\sigma_{0.2}$ and $d^{-1/2}$. ○, Fe–36 Ni; □, Fe–36 Ni–2Cr; △, Fe–36 Ni–5Cr.

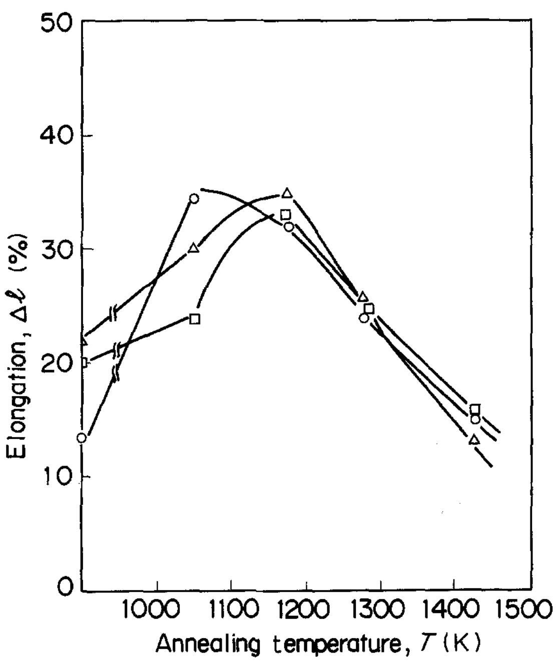

Figure 4 Changes in Δl on breaking point as a function of annealing temperature. ○, Fe–36 Ni; □, Fe–36 Ni–2Cr; △, Fe–36 Ni–5Cr.

Fig. 3 shows the change in σ_B as a function of annealing temperature. As shown in the figure, σ_B is independent of chromium concentration at each annealing temperature and decreases linearly with increasing annealing temperature. This tendency suggests that the deformation mode for Fe–36 Ni–Cr became similar to that for Fe–36 Ni on the high strain.

Fig. 4 shows the change in Δl at breaking point as a function of annealing temperature. Δl for each sample decreased linearly with increasing annealing temperature above 1173 K. The Δl values were nearly equal for three samples with annealing temperatures larger than 1273 K. As the grain growth occurred above 1273 K, this result suggests that the breaking mechanism for Fe–36 Ni–Cr was similar to that for Fe–36 Ni.

Fig. 5 shows the change in n as a function of annealing temperature. The n values are almost the same for the three alloys before annealing. Though the n value for Fe–36 Ni indicated a constant value independent of annealing temperature, n for Fe–36 Ni–Cr is higher than that for Fe–36 Ni and increases with increasing annealing temperature. As reported by the author and co-workers [4] this result suggests that the slip in Fe–36 Ni was easily propagated once it had begun, but that in Fe–36 Ni–Cr it was disturbed. This tendency was contrary to the result that $\sigma_{0.2}$ for Fe–36 Ni–Cr was smaller than that of Fe–36 Ni. As the n value is the average value at the strain range from 4 to 8%, while $\sigma_{0.2}$ is calculated on the strain of 0.2%, the deformation mechanisms may be different among these samples. Another approach to this problem has

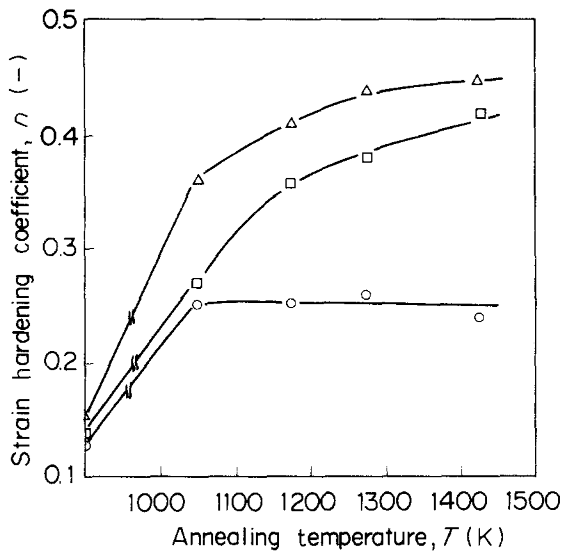

Figure 5 Changes in n as a function of annealing temperature. ○, Fe–36 Ni; □, Fe–36 Ni-2Cr; △, Fe–36 Ni-5Cr.

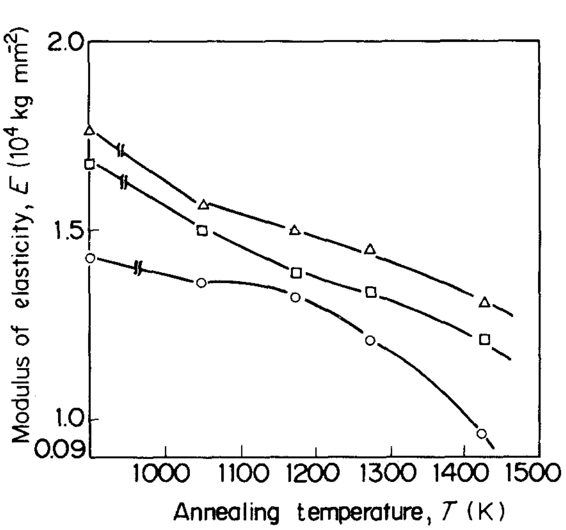

Figure 6 Changes in E as a function of annealing temperature. ○, Fe–36 Ni; □, Fe–36 Ni–2Cr; △, Fe–36 Ni-5Cr.

been accomplished with Mössbauer spectrometry [7, 8] by the author and co-workers.

Fig. 6 shows the change in E as a function of annealing temperature. Although the E values for three kinds of alloys decrease with increasing annealing temperature, the E value for Fe–36 Ni–Cr is larger than that for Fe–36 Ni at each annealing temperature.

4. Discussion

The difficulty in slip propagation indicated a difficulty in dislocation movement. The Peierls force (τ_c) can be expressed by the following equation

$$\tau_c = \frac{2G}{1 - \nu} \exp[-2\pi/(1 - \nu)] \tag{2}$$

where G is the modulus of rigidity and ν is Poisson's ratio. This equation is based on the following three hypotheses: (i) the potential energy on the dislocation movement is expressed as a sine curve; (ii) the influence of kinks is neglected; and (iii) the effect of thermal lattice vibrations is neglected. In this experiment, though the τ_c value is calculated to be larger than the measured value for an actual sample, the τ_c value is considered as a criterion for the dislocation movement.

The following relation between G and E

$$G = \frac{E}{2(1 - \nu)} \tag{3}$$

gives τ_c from the observed E and ν.

The values of ν for Fe–36 Ni, Fe–36 Ni–2Cr and Fe–36 Ni–5Cr were 0.28, 0.30 and 0.31, respectively, on measuring with the tension tester. When these ν values were substituted for Equations 2 and 3, τ_c values for Fe–36 Ni, Fe–36 Ni–2Cr and Fe–36 Ni–5Cr were calculated as 2.66, 2.08 and 1.96 kg mm^{-2}, respectively. However, these τ_c values were smaller than $\sigma_{0.2}$ values obtained from Fig. 2. Referring with the $\sigma_{0.2}$ for Fe–36 Ni, Fe–36 Ni–2Cr and Fe–36 Ni–5Cr mentioned above, the excessive force of 14.74, 12.32 and 6.44 kg mm^{-2} to move dislocation was needed in Fe–36 Ni, Fe–36 Ni–2Cr and Fe–36 Ni–5Cr, respectively.

The dominating cause for $\sigma_{0.2}$ value was not only τ_c but also the cross-slip. The microstructures of Fe–36 Ni and Fe–36 Ni–Cr strained to about 0.2% at room temperature with transmission electron microscopy (TEM) have been already published [4]. These results showed that Fe–36 Ni–Cr has a low stacking-fault energy, whereas Fe–36 Ni has a high stacking-fault energy. Once the dislocations in Fe–36 Ni–Cr start moving (strain range $\sim$ 0–0.2%), the dislocations extend because of their low stacking-fault energy, and cross-slip is difficult, generating an accumulation of dislocations. This phenomenon is different from that in Fe–36 Ni. As the stacking-fault energy for Fe–36 Ni is high, the cross-slip easily occurs to generate many jogs, and $\sigma_{0.2}$ consequently becomes high. These jogs result in excessive force to move dislocations, especially in Fe–36 Ni. In the second region (strain range $\sim$ 0.2–10%), the multiple of dislocation density in Fe–36 Ni–Cr increased as n began to increase. In Fe–Ni–Cr alloy, the stress of deformation starting is low because the accumulation of dislocations with increase of strain is obvious. In a previous paper by the author and co-workers [8], the accumulation of dislocations was accelerated in Fe–Ni–Cr alloy as the internal magnetic field, influenced by the jog density, decreased with hardening. This tendency is in good agreement with that noted above. Therefore the strain-hardening mechanism gradually approaches that for Fe–36 Ni. At the fracture, the breaking mechanism for Fe–36 Ni and Fe–36 Ni–Cr probably resemble each other fairly well, because σ_B and Δl are similar among the three samples (strain range > 10%).

5. Conclusion

The changes in yield strength ($\sigma_{0.2}$), tensile strength (σ_B), elongation (Δl), strain-hardening coefficient (n) and modulus of elasticity (E) in chromium-added Fe–36 Ni, were studied as a function of annealing temperature. The results are as follows.

1. When chromium was added to Fe–36 Ni, $\sigma_{0.2}$ decreased but n and E increased. σ_B and Δl scarcely changed with the chromium addition.

2. $\sigma_{0.2}$, σ_B, Δl and E for Fe–36 Ni–Cr decreased with the increase in annealing temperature. The relation between $\sigma_{0.2}$ and grain size was found to obey Petch's equation.

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References
1. M. INABA, K. TESHIMA, E. HIGASHINAKAGAWA and Y. OHTAKE, *IEEE Trans. Electron Devices* **35** (1988) 1721.
2. H. YAMAZAKI and Y. OHTAKE, *Toshiba Review* **156** (1986) 29.
3. T. KANAMARU and S. HIROTSU, *Nisshin Steel Tech. Rep.* **33** (1975) 18.
4. M. INABA, M. SATO, E. HIGASHINAKAGAWA and Y. OHTAKE, *Displays* **9** (1988) 17.
5. M. INABA and K. TESHIMA, *J. Mater. Sci. Lett.* **6** (1987) 727.
6. M. INABA, Y. HONMA, T. HATANAKA and Y. OHTAKE, *Appl. Surf. Sci.* **27** (1986) 164.
7. M. INABA, K. NOMURA and Y. UJIHIRA, *J. Jpn Inst. Metals* **52** (1988) 1121.
8. M. INABA, Y. NOMURA and Y. UJIHIRA, *ibid.* **53** (1989) 356.

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