Effect of Ni Doping on Strength and Damping Capacity of Fe–Al Alloys

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Strength and damping capacity of Fe–Al alloys doped with different amounts of Ni were examined. Fe–Al alloys with an appropriate amount of Ni doping exhibited yield stress higher than 1 GPa and internal friction of above 0.032. In these alloys, the NiAl phase with the B2 structure densely precipitated in the Fe–Al matrix, satisfying the cube-on-cube orientation relationship. The precipitation of NiAl resulted in high strength, which was associated with the difference in primary slip systems between the NiAl precipitates and the Fe–Al matrix. Moreover, the internal friction measured under an external magnetic field showed low values, suggesting that Fe–Al alloys with the NiAl precipitates demonstrated magneto-mechanical damping caused by the irreversible motion of magnetic domain walls. The Al concentration in the Fe–Al matrix became appropriate for damping properties by the precipitation of NiAl, resulting in an increase in the internal friction of the alloy.

KEY WORDS: internal friction; damping materials; particle strengthening; magneto-mechanical damping.

1. Introduction

The development of damping materials with high strength and damping capacity has attracted much attention, especially for automobile and seismic applications.^{1–4)} For instance, high strength above 1 GPa and internal friction (Q^{-1}) above 0.032 are required for practical application to a vibration source of automobiles, such as engines and gears.²⁾ Recently, Fe–Al alloys with 16–18 at%Al have been expected as damping materials to absorb the vibration and noise due to their high damping capacity.^{5,6)} These Fe–Al alloys have been successfully developed and are commercially available in Japan now. At low strain amplitudes and ambient temperature, the high damping capacity of Fe–Al alloys is usually attributed to the magneto-mechanical damping associated with the irreversible motion of magnetic domain walls.^{5–12)} However, the ultimate tensile strength of binary Fe–Al alloys was approximately 500 MPa, not enough to use the alloys to the automobile vibration sources.

In general, in conventional damping materials, there is a trade-off between the strength and damping capacity and therefore, it is difficult to obtain materials that exhibit both high strength and high damping capacity. In the present study, however, we found that small amount of Ni doping to Fe–Al alloys increased not only the yield stress (σ_y) but also the Q^{-1} and consequently, $\sigma_y > 1$ GPa and $Q^{-1} > 0.032$ could be achieved by optimizing the chemical composition. In this paper, we report the effect of Ni doping on the

strength and damping capacity of Fe–Al alloys. In addition, the strengthening and damping mechanisms in Ni-doped Fe–Al alloys will also be discussed.

2. Experimental Procedure

Master ingots of Fe–16.0–25.0Al (at%) doped with 0–10.0Ni were prepared by melting high purity elements in a plasma arc furnace in a purified Ar atmosphere. The chemical compositions selected in the present study are plotted in the isothermal section of Fe–Al–Ni ternary phase diagram at 750°C as shown in **Fig. 1**.¹³⁾ Note that most of selected chemical compositions were located near the α – β' two-phase field. The α and β' phases are the Fe–Al solid solution and NiAl with the bcc and B2 structures, respectively. After homogenization at 1100°C for 48 h, the specimens were

Fig. 1. Fe–Al–Ni ternary phase diagram at 750°C¹³⁾ and chemical compositions of the alloys prepared in the present study.

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Fig. 2. $[110]$ SAED patterns ((a)–(d)) and DF images ((e)–(h)) of Fe–23.0Al– x Ni alloys. DF images were taken with 001 reflection. (a), (e) $x = 2.0$, (b), (f) $x = 4.0$, (c), (g) $x = 6.0$ and (d), (h) $x = 8.0$.

slowly cooled to room temperature at 80°C/h. The microstructure of Fe–Al–Ni alloys was observed using a transmission electron microscope (TEM) operated at 300 kV. The plane and direction of TEM observation were indexed with respect to the bcc structure.

Rectangular specimens of which dimension is $2 \times 2 \times 5 \text{ mm}^3$ were cut from Fe–Al–Ni ingots by spark machining. These specimens were mechanically and electrolytically polished to remove surface damage. Compression tests were performed in air at room temperature at an initial strain rate of $1.7 \times 10^{-4} \text{ s}^{-1}$.

The Q^{-1} values of Fe–Al–Ni alloys were measured with a dynamical mechanical analyzer (DMA) equipped with a single cantilever module at strain amplitudes (ϵ) ranging between 10^{-4} and 5×10^{-3} . The DMA cyclically bent the cantilever plates of Fe–Al–Ni alloys and could directly measure the phase shift (ϕ) between the force and displacement signals. The Q^{-1} value is given by the following equation,

$$Q^{-1} = \tan \phi \dots \dots \dots (1)$$

Measurement of Q^{-1} was conducted at room temperature at a frequency (f) of 1 Hz. In order to examine the damping mechanism of Ni-doped specimens, the Q^{-1} was also measured by a free decay method under an external magnetic field of 0.25 T; the details of the method are described in a previous paper.¹²⁾ The dimension and shape of the cantilever specimens for the free-decay method were the same as those for the DMA measurement. The specimens were bent at a resonance frequency of about 1050 Hz and were released. The decay of free vibration was recorded by a laser displacement meter.

It is well known that magnetostriction is closely related to magneto-mechanical damping.¹⁴⁾ The magnetostriction (λ_m) of Fe–Al–Ni alloys with 0, 2.0 and 6.0 at%Ni was measured using a strain gauge method. A non-magnetoresistive strain gauge was put on the alloys. It is also noted that a dummy gauge was put on a non-magnetic sample to cancel the magnetoresistance effect. Magnetic field of 0.8 MA/m was applied to Fe–Al–Ni alloys at room temperature and the magnetostriction parallel to the field was measured.

3. Results

3.1. Microstructure of Fe–Al–Ni Alloys

Figure 2 shows the selected area electron diffraction (SAED) patterns and dark-field (DF) images of Fe–23.0Al–

Fig. 3. Phase constituents of Fe–Al–Ni alloys slowly cooled to room temperature at 80°C/h.

2.0–8.0Ni alloys, taken by the TEM. In Fe–23.0Al–2.0Ni alloys, both $\frac{1}{2} 111$ and 001 superlattice spots can be seen and a typical $D0_3$ ordered domain structure develops in the DF image taken with 001 reflection (Figs. 2(a), 2(e)). In contrast, at and above 4.0 at%Ni, $\frac{1}{2} 111$ reflection becomes diffuse while 001 reflection is clearly visible as shown in Figs. 2(b)–2(d). Moreover, in the DF images with 001 reflection, precipitation of the β' -NiAl phase with the B2 structure can be observed in the α -Fe–Al matrix (Figs. 2(f)–2(h)). Moreover, the amount of the β' precipitates increases with increasing Ni content. From the SAED patterns, the α and β' phases satisfy the cube-on-cube orientation relationship. In addition, there is no misfit dislocation between the α and β' phases, suggesting that these phases are coherent. Moreover, the diffuse $\frac{1}{2} 111$ reflection at and above 4.0 at%Ni suggests that a short-range order develops in the α matrix.⁴⁾ **Figure 3** shows the phase constituent of Fe–Al–Ni alloys prepared in the present study. The α – β' two-phase field extends over wide chemical compositions, which is fairly consistent with Fe–Al–Ni phase diagram shown in Fig. 1. It should be noted that Fe–Al–Ni alloys prepared in the present study were slowly cooled to room temperature after homogenization at 1100°C. Since the ternary Fe–Al–Ni phase diagram shown in Fig. 1 was determined at 750°C, the α – β' two-phase field in the slowly cooled alloys becomes wider than that at 750°C.

3.2. Strength and Damping Capacity of Fe–Al–Ni Alloys

Figure 4 shows the stress-strain curves of Fe–23.0Al–2.0–10.0Ni alloys compressed at room temperature. Fe–

Fig. 4. Stress-strain curves of Fe-23.0Al-2.0–10.0Ni alloys compressed at room temperature.

Fig. 5. ε dependence of Q^{-1} in Fe-23.0Al-0–10.0Ni alloys at room temperature; $f = 1$ Hz.

23.0Al binary alloys exhibit a yield stress (σ_y) of about 570 MPa. The σ_y increases with increasing amount of Ni doping and exceeds 1 GPa at and above 4.0 at%Ni. As shown in Fig. 2, the β' -NiAl phase also precipitated at and above 4.0 at%Ni. This means that strengthening by Ni doping in Fe–Al alloys is closely related to the β' -NiAl precipitation. Variation in Q^{-1} with ε in Fe-23.0Al-0–10.0Ni alloys is shown in Fig. 5. Binary Fe-23.0Al and ternary Fe-23.0Al-10.0Ni alloys demonstrate low Q^{-1} values in the wide ε range. However, the Q^{-1} in Fe-23.0Al-2.0–8.0Ni alloys increases rapidly with increasing ε up to about 1×10^{-4} . After showing a peak, further increase in ε results in a decrease in Q^{-1} followed by the saturation. The ε dependence of Q^{-1} in the alloys is in good agreement with magneto-mechanical damping.^{5–11)} Moreover, the peak value of Q^{-1} ($Q^{-1}_{\max}$) at Ni contents between 4.0 at% and 8.0 at% exceeds 0.032.

Figure 6 shows change in σ_y and $Q^{-1}_{\max}$ with amount of Ni in alloys containing 23.0 at%Al. Both σ_y and $Q^{-1}_{\max}$ rise abruptly at around 4.0 at%Ni and remain almost constant up to 8.0 at%Ni. The precipitation of β' -NiAl started around 4.0 at%Ni, suggesting that the strength and damping capacity of Fe–Al–Ni alloys are closely related to the β' precipitation. However, further increase in Ni content results in a decrease in $Q^{-1}_{\max}$ though the σ_y remains almost constant at and above 4.0 at%Ni. Hence, appropriate amounts of Ni doping were found to be effective in the increase in both σ_y and Q^{-1} .

Fig. 6. Variation in σ_y and $Q^{-1}_{\max}$ with amount of Ni in alloys containing 23.0 at%Al; at room temperature, $f = 1$ Hz.

Fig. 7. Variation in Q^{-1} with ε in Fe-23.0Al-6.0Ni alloys at room temperature with or without magnetic field; free decay method, at a resonance frequency of about 1 050 Hz.

Fig. 8. Saturation magnetostriction of Fe-23.0Al- x Ni alloys ($x = 0, 2.0$ and 6.0), parallel to a magnetic field of 0.8 MA/m at room temperature.

3.3. Behavior of Fe–Al–Ni Alloys under Magnetic Field

In order to examine the damping mechanism, the Q^{-1} was measured under a magnetic field of 0.25 T. Figure 7 shows variation of Q^{-1} with ε in Fe-23.0Al-6.0Ni alloys, measured by a free decay method with or without magnetic field. At $\varepsilon = 2 \times 10^{-5}$, the Q^{-1} at zero magnetic field is much higher than that at 0.25 T, suggesting that high damping capacity of Fe–Al–Ni alloys is caused by magneto-mechanical damping. It is well known that the magneto-mechanical damping depends on λ_m .¹⁴⁾ Thus, the λ_m measurement parallel to a magnetic field of 0.8 MA/m was also conducted as shown in Fig. 8. The λ_m of Fe-23.0Al-6.0Ni alloys is higher than

Fig. 9. Chemical compositions of Fe–Al–Ni alloys exhibiting $\sigma_y > 1$ GPa and $Q^{-1}_{\max} > 0.032$; at room temperature, $f = 1$ Hz and $\varepsilon \approx 1 \times 10^{-4}$.

that of Fe–23.0Al binary alloys.

4. Discussion

In the present study, the effect of Ni addition on the strength and damping capacity of Fe–Al–Ni alloys was examined. Appropriate amounts of Ni doping were found to be effective in the increase in both σ_y and Q^{-1} . In particular, $\sigma_y > 1$ GPa and $Q^{-1} > 0.032$ (specific damping index $> 20\%$) which are requirements for the automobile application under strong vibration could be achieved by optimizing the amount of Ni doping, as shown in Fig. 9. However, it should be noted that the high Q^{-1} above 0.032 could be obtained only at limited strain amplitude (Fig. 5). Thus, further study is needed for the automobile application. In addition, the compositions of the alloys which showed high σ_y and Q^{-1} are present in the α – β' two-phase field. This also means that the β' -NiAl phase plays an important role in enhancing the strength and damping capacity of Fe–Al–Ni alloys. Therefore, the role of the β' -NiAl on the strength and damping capacity were discussed. In addition, the deformation and damping mechanisms of Fe–Al–Ni alloys were also discussed.

4.1. Strengthening Mechanism in Fe–Al–Ni Alloys

It is well known that the primary slip system of B2-type NiAl is $\{110\}\langle 001\rangle$ while $\langle 111\rangle$ slip was only activated at loading axis in the vicinity of $\langle 001\rangle$ orientation.¹⁵⁾ It should also be noted that the critical resolved shear stress (CRSS) for $\langle 111\rangle$ slip is about six times higher than that of $\{110\}\langle 001\rangle$ slip. In Fe–Al–Ni alloys, the β' -NiAl precipitates coherent with the α matrix were inevitably sheared by $\langle 111\rangle$ slip of the α -Fe–Al matrix. In fact, the β' precipitates were observed to be cut by dislocations of the α matrix, the results are published in elsewhere. If the β' precipitates are sheared by $\langle 111\rangle$ direction in Fe–Al–Ni alloys, the frictional stress of dislocations inside the precipitates abruptly increases due to the difference in CRSS between $\langle 001\rangle$ and $\langle 111\rangle$ slips. Hence, the formation of the β' precipitates results in extreme hardening of Fe–Al–Ni alloys. It is also noted that precipitation-hardened stainless steels also contained the NiAl precipitates in the matrix composed of the martensite phase. In the stainless steels, the NiAl precipitates were also sheared by dislocations of the matrix phase, which would also lead to the hardening.

4.2. High Damping Capacity in Fe–Al–Ni Alloys

It is well known that the damping mechanism of Fe–Al binary alloys is based on the irreversible motion of magnetic domain walls during vibration.^{5–11)} Since the Q^{-1} of Fe–23.0Al–6.0Ni alloys shows low values under magnetic field, the magneto-mechanical damping seems to appear. It is also noted that the λ_m of the alloys showed higher values compared with Fe–23.0Al binary alloys as shown in Fig. 8. The Q^{-1} based on magneto-mechanical damping was proposed by Kornetzki *et al.*¹⁵⁾ and is given by,

$$Q^{-1} = 3 \times 10^{15} \frac{\nu |\lambda_m|^3 E^2}{\pi I_s^3} \varepsilon \quad (2)$$

where ν is the magnetic Rayleigh constant, λ_m is the average saturation magnetostriction, E is the Young's modulus, I_s is the saturation magnetization and ε is the applied strain. From Eq. (2), the Q^{-1} is strongly dependent on λ_m . Thus, high λ_m in Fe–Al–Ni alloys with appropriate amount of Ni resulted in higher Q^{-1} .

It is also interesting to note that high Q^{-1} above 0.032 could also be obtained at optimum chemical compositions, as shown in Fig. 9. The chemical composition of the α -Fe–Al matrix varied accompanied by the precipitation of the β' -NiAl phase. In other words, Al concentration in the α matrix decreased associated with the β' precipitation. For instance, if the β' phase precipitates in Fe–23.0Al–6.0Ni alloys, Al content of the α matrix would be 17 at%. It is well known that superior damping capacity with high Q^{-1} can be obtained at around 16–17 at%Al in Fe–Al binary alloys.^{5,6)} In fact, Fe–17.0Al binary and Fe–23.0Al–6.0Ni ternary alloys showed similar damping properties. Thus, it is reasonable that the α matrix and β' precipitates acted as a high damping and high strength phases, respectively. Therefore, in association with the β' precipitation, the Al concentration in the α matrix became favorable for damping properties, resulting in high Q^{-1} in Fe–Al–Ni alloys. Moreover, the β' precipitates may not interfere the motion of magnetic domain walls leading to magneto-mechanical damping if the amount of the β' phase is small. This is reasonable assumption because the misfit strain between the α and β' phases was small. However, in Fe–23.0Al–10.0Ni alloys, the volume fraction of the β' precipitates which shows a paramagnetism was so high that the magneto-mechanical damping was hindered by the precipitates (Fig. 6). Thus, Fe–Al–Ni alloys with the appropriate amount of the β' -NiAl phase could exhibit both high σ_y and high Q^{-1} .

5. Conclusions

The deformation and damping behavior of Ni-doped Fe–Al alloys were examined focusing on the β' -NiAl precipitates. The following conclusions were reached.

- (1) Fe–Al–Ni alloys with appropriate chemical compositions exhibited $\sigma_y > 1$ GPa and $Q^{-1} > 0.032$.
- (2) The β' -NiAl precipitates significantly strengthened Fe–Al–Ni alloys owing to the difference in primary slip systems between the β' precipitates and the α matrix.
- (3) The vibration attenuation in Fe–Al–Ni alloys was considered to be based on magneto-mechanical damping.
- (4) Associated with the precipitation of the β' -NiAl phase, Al concentration of the α matrix became favorable

for magneto-mechanical damping, which resulted in high Q^{-1} .

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REFERENCES

- 1) I. G. Ritchie and Z. L. Pan: *Metall. Trans. A*, **22** (1991), 607.
- 2) T. Sakaguchi: *Kinzoku*, **74** (2004), 277 (in Japanese).
- 3) M. Dolce and D. Cardone: *Int. J. Mech. Sci.*, **43** (2001), 2631.
- 4) M. Dolce and D. Cardone: *Int. J. Mech. Sci.*, **43** (2001), 2657.
- 5) I. S. Golovin, H. Neuhäuser, A. Riviere and A. Strahl: *Intermetallics*, **12** (2004), 125.
- 6) K. Mizuuchi, Y. Okanda and S. Motogi: *Sci. Industry*, **64** (1990), 417 (in Japanese).
- 7) T. Yamada, T. Takamura, S. Hashizume, T. Odake, T. Omori and K. Hahori: *NKK Tech. Rep.*, **137** (1991), 18.
- 8) D. P. Sagraadi, M. Sagraadi, A. Karimi and J. L. Martin: *Scr. Mater.*, **39** (1998), 131.
- 9) Z. C. Zhou, J. N. Wei and F. S. Han: *Phys. Status Solidi (a)*, **191** (2002), 89.
- 10) Y. I. Seo, B. H. Lee, Y. D. Kim and K. H. Lee: *Mater. Sci. Eng. A*, **431** (2006), 80.
- 11) X. Y. Gang, L. Ning, S. B. Luo and H. H. Xin: *Mater. Sci. Eng. A*, **447** (2007), 163.
- 12) H. Y. Yasuda, K. Fukushima, Y. Koizumi, Y. Minamino and Y. Umakoshi: *ISIJ Int.*, **49** (2009), 1630.
- 13) V. G. Rivlin and G. V. Raynor: *Int. Metals Rev.*, **25** (1980), 79.
- 14) V. M. Kometzki: *Z. Phys.*, **121** (1943), 560.
- 15) D. B. Miracle: *Acta Metall. Mater.*, **41** (1993), 649.