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Change in stacking-fault energy with Mn content and its influence on the damping capacity of the austenitic phase in Fe-high Mn alloys

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The effect of Mn content on the damping capacity of γ austenitic phase for Fe-(17–28)%Mn alloys has been studied in relation to the stacking-fault energy (SFE), on the basis of the results of microstructural observations, thermodynamic calculations and vibration tests in a torsional mode. The damping capacity of γ austenite decreases with the increase in the Mn content. The decrease in density of stacking-faults in γ austenite and mobility of the stacking-fault boundary, on which the damping capacity of γ phase depends, is considered to be responsible for the deterioration of damping capacity.

1. Introduction

Recently, Fe-high Mn alloys possessing the unique combination of a pronounced damping capacity and good mechanical properties, have been paid increasing attention as promising materials for frame of structures and components of precise machines [1–4]. The damping mechanism of the Fe-Mn alloys is thought to be an internal friction due to the movement of the various boundaries such as stacking-fault boundaries inside ε martensite and γ austenite, ε martensite variant boundaries, and γ/ε interfaces during vibration in an elastic region [5–9]. Accordingly, the total damping capacity of the Fe-Mn alloys (δ_{total}) can be given by:

$$\delta_{\text{total}} = \Delta\delta_{\text{v}} + \Delta\delta_{\text{sf}} + \Delta\delta_{\gamma} + \Delta\delta_{\gamma/\varepsilon} \qquad (1)$$

where $\Delta\delta_{\text{v}}$ is the damping capacity due to the movement of ε martensite variant boundaries, $\Delta\delta_{\text{sf}}$ and $\Delta\delta_{\gamma}$ are the damping capacity by the migration of stacking-fault boundaries inside ε martensite and γ austenite, respectively, and $\Delta\delta_{\gamma/\varepsilon}$ is the damping capacity associated with the movement of γ/ε interfaces.

The damping capacity shows a maximum value around 17%Mn, and decreases in higher Mn content [9]. By a systematic examination of the microstructural evolution and damping capacity with respect to Mn content, it has been found that the deterioration of the damping capacity above 17%Mn is principally ascribed to the decrease in both ε martensite content and mobility of γ/ε interface, which eventually leads to the decrease in contributions of $\Delta\delta_{\text{v}}$, $\Delta\delta_{\text{sf}}$ and $\delta_{\gamma/\varepsilon}$ to δ_{total} [9]. However, the influence of Mn content on the damping capacity of γ austenite ($\Delta\delta_{\gamma}$) remains unclear until now. Therefore, this study is intended to report the damping behavior of γ austenite as a function of the Mn content for Fe-(17–28)%Mn alloys, and to discuss the results in relation to the stacking-fault energy (SFE).

2. Experimental

Fe-17%Mn, Fe-20%Mn, Fe-23%Mn and Fe-28%Mn alloys (in weight), were prepared by high frequency induction melting in vacuum. The ingots were homogenized by keeping at 1473 K for 24 hours and hot-rolled into bars with a diameter of 13 mm. From these bars, the specimens for damping measurement, dilatation test and microstructural observation were prepared by subsequent rolling and machining. Solution treatment was performed at 1323 K for 1 hour followed by a water quench.

Martensitic transformation temperatures (M_{s}, A_{s} and A_{f}) were measured by dilatometry. The chemical compositions and transformation temperatures of the experimental alloys are listed in Table I. Samples for optical microstructure were electro-polished in a solution of 90% CH_3COOH + 10% $HClO_4$, and etched with a 1.2% $K_2S_2O_5$ aqueous solution. Foils for transmission electron microscopy (TEM) were jet-polished in the 90% CH_3COOH + 10% $HClO_4$ solution, and were examined in a JEOL JEM-200CX operating at 200 kV.

In order to investigate the damping capacity of γ austenite single phase, the measurement of damping

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TABLE I Chemical compositions (wt %) and martensitic transformation temperatures (K) of experimental alloys

Alloy	Mn	C	Fe	M_s	A_s	A_f
Fe-17%Mn	17.2	0.019	Bal.	423	474	508
Fe-20%Mn	20.2	0.020	Bal.	409	455	484
Fe-23%Mn	23.3	0.018	Bal.	390	436	470
Fe-28%Mn	27.8	0.021	Bal.	—	—	—

capacity was conducted at 438 K using a Föpple-Pertz type torsional pendulum apparatus [10], after heating the specimens to 533 K above A_f temperature for the complete $\varepsilon \rightarrow \gamma$ reverse transformation, and subsequent cooling to 438 K above M_s temperature to retain the reversed γ single phase. The damping test assembly and dimensions of the specimen are illustrated in Fig. 1. In this study, the damping capacity was measured in a range of 4×10^{-4} to 6×10^{-4} strain amplitude, and is evaluated in a logarithmic decrement (δ) defined as:

$$\delta = 1\mathrm{n}(A_n / A_{n+1}) \tag{2}$$

where A_n and A_{n+1} represent the strain amplitudes of nth and $(n+1)$th cycles in free decay.

3. Results and discussion

Fig. 2 represents a typical example of the optical microstructure at room temperature for Fe-(17–28)%Mn alloys in as-quenched state. As is known, the microstructures are distinctively characterized by γ austenite (the dark part) and thin plate ε martensite (the bright part). With the increase of Mn content, the amount of ε martensite appears to decrease and only γ single phase is observed in Fe-28%Mn alloy. Obviously, this is ascribed to the fact that Mn, an austenite stabilizing element, has a role to lower the M_s temperature of $\gamma \rightarrow \varepsilon$ martensitic transformation by decreasing the thermodynamic equilibrium temperature (T_0) between fcc and hcp phases (Table I) [11]. The change in damping capacity of γ single phase at 438 K with Mn content is given in Fig. 3. It is evident that the damping capacity becomes lower at high Mn contents in Fe-(17–28)%Mn alloys. The reasons for this behavior are as follows: (i) *the decrease in density of stacking-faults in γ austenite*: In the Fe-(17–28)%Mn alloys, the change in SFE with Mn content at 438 K can be calculated thermodynamically. In a regular solution model, the difference in chemical Gibbs energy between γ austenite and ε martensite ($\Delta G^{\gamma \rightarrow \varepsilon}$) for Fe-Mn binary alloy is given by:

$$\Delta G^{\gamma \rightarrow \varepsilon} = X_{\mathrm{Fe}} \Delta G_{\mathrm{Fe}}^{\gamma \rightarrow \varepsilon} + X_{\mathrm{Mn}} \Delta G_{\mathrm{Mn}}^{\gamma \rightarrow \varepsilon} + X_{\mathrm{Fe}} X_{\mathrm{Mn}} \Delta \Omega_{\mathrm{FeMn}}^{\gamma \rightarrow \varepsilon} \tag{3}$$

where X_{Fe} are X_{Mn} are the atomic fractions of Fe and Mn, respectively, $\Delta G_{\mathrm{Fe}}^{\gamma \rightarrow \varepsilon}$ is the change in molar free

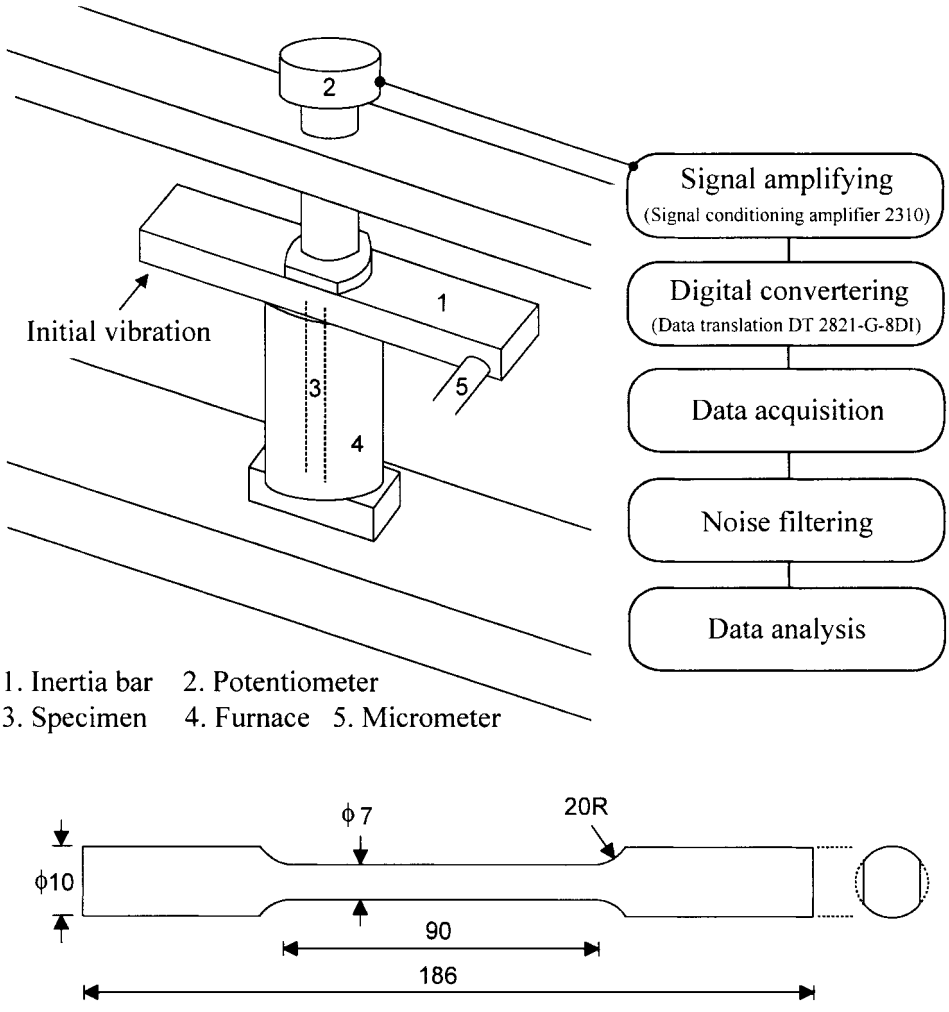

Figure 1 Schematic illustration showing damping test assembly and dimension of its specimen (mm).

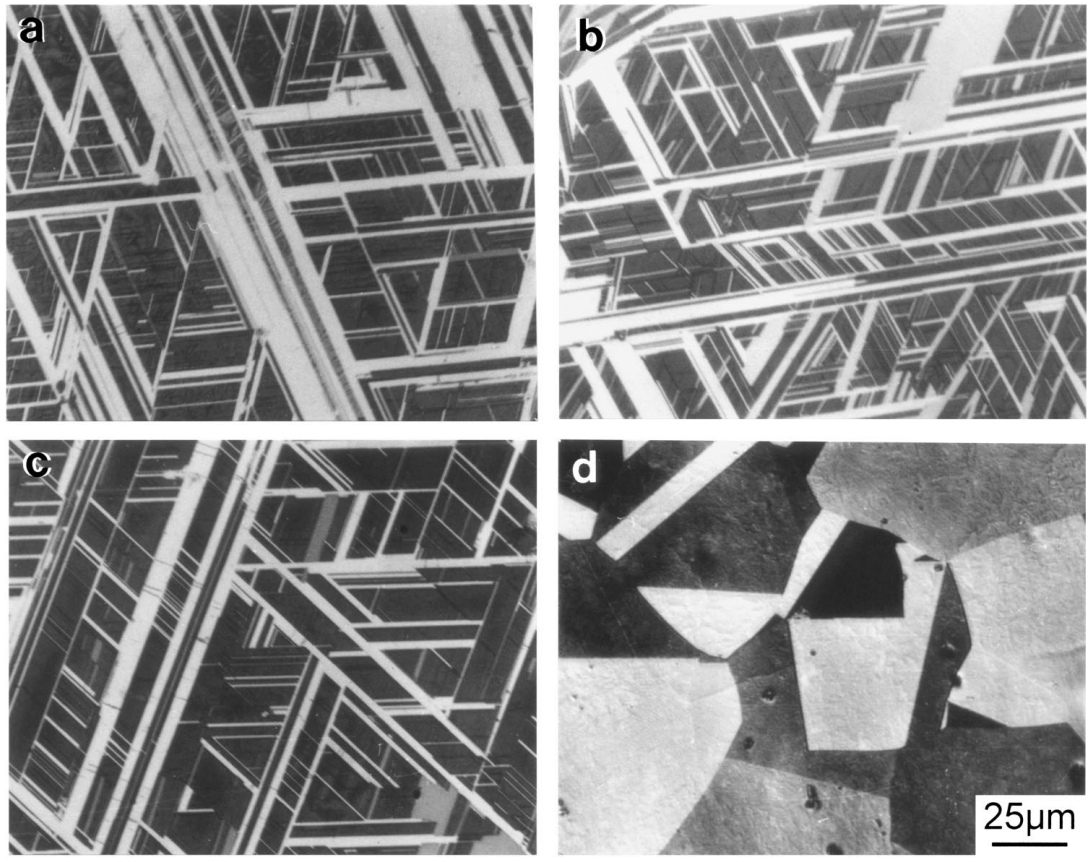

Figure 2 Change in optical microstructure with Mn content in Fe-(17–28)%Mn alloys. (a) Fe-17%Mn alloy, (b) Fe-20%Mn alloy, (c) Fe-23%Mn alloy, and (d) Fe-28%Mn alloy.

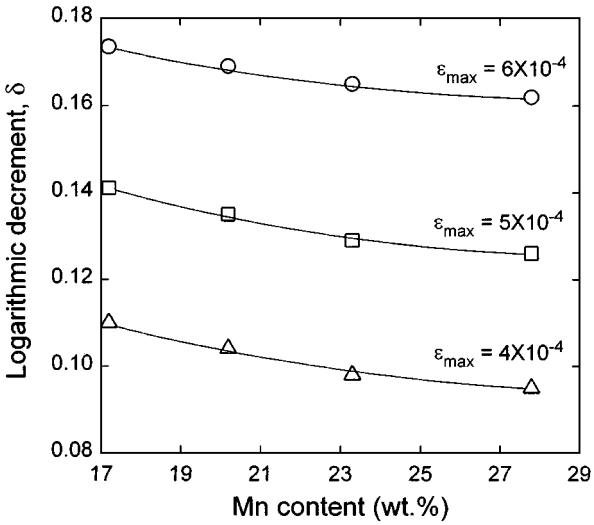

Figure 3 Variation of damping capacity of γ austenite at 438 K with Mn content in Fe-(17–28)%Mn alloys.

energy between γ and ε phases in Fe and Mn, respectively and $\Delta\Omega_{\mathrm{FeMn}}^{\gamma\rightarrow\varepsilon}$ is the interaction energy parameter for Fe and Mn.

In this study, $\Delta G_{\mathrm{Fe}}^{\gamma\rightarrow\varepsilon}$ and $\Delta G_{\mathrm{Mn}}^{\gamma\rightarrow\varepsilon}$ were formulated as a function of temperature by fitting the data reported by Blackburn *et al*. [12] and by Breedis *et al*. [13], respectively, whereas for $\Delta\Omega_{\mathrm{FeMn}}^{\gamma\rightarrow\varepsilon}$, the equation previously reported by Ishida *et al*. [14] was adopted.

$$\Delta G_{\mathrm{Fe}}^{\gamma\rightarrow\varepsilon}(\mathrm{J/mol}) = -820.7 + 1.683T + 0.00226T^2 \quad (4)$$

$$\Delta G_{\mathrm{Mn}}^{\gamma\rightarrow\varepsilon}(\mathrm{J/mol}) = 3972.4 - 2.97T + 0.0051T^2 \quad (5)$$

$$\Delta\Omega_{\mathrm{FeMn}}^{\gamma\rightarrow\varepsilon}(\mathrm{J/mol}) = -10836.6 + 22886.5X_{\mathrm{Mn}} \quad (6)$$

In accordance with a thermodynamic model of fault energy suggested by Olson and Cohen [15], the intrinsic SFE (γ_{i}) is expressed as:

$$\gamma_{\mathrm{i}} = 2\rho_{\mathrm{A}}\Delta G^{\gamma\rightarrow\varepsilon} + 2\sigma^{\gamma-\varepsilon} \quad (7)$$

where ρ_{A} is the planar packing density of a close-packed plane (in mol/area) and $\sigma^{\gamma-\varepsilon}$ is the coherent fcc-hcp interfacial energy. The values of ρ_{A} and $\sigma^{\gamma-\varepsilon}$ have been determined as 2.987×10^{-5} mol/m^2 for planar packing density of $(111)_{\mathrm{fcc}}$ plane [16] and 15 mJ/m^2 for transition metals [15], respectively. The calculated SFE at 438 K is plotted as a function of Mn content in Fig. 4. It is observed that the SFE increases from 27 to 54 mJ/m^2 with increasing from 17%Mn to 28%Mn.

Vassamillet [17] has reported that the stacking-fault probability (SFP, α) in fcc alloys is described as:

$$\alpha = (\mathrm{K}\cdot\rho)/(24\pi\cdot\gamma) \quad (8)$$

where ρ is the density of dislocation, γ is the SFE and K is the constant related to the elastic constant and lattice parameter. Assuming that the ρ is a constant after recrystallization, Equation 8 suggests that the lower the SFE, the higher the number of stacking-faults per unit area for Fe-(17–28)%Mn alloys. In order to test this, TEM examinations were made on the Fe-17%Mn and

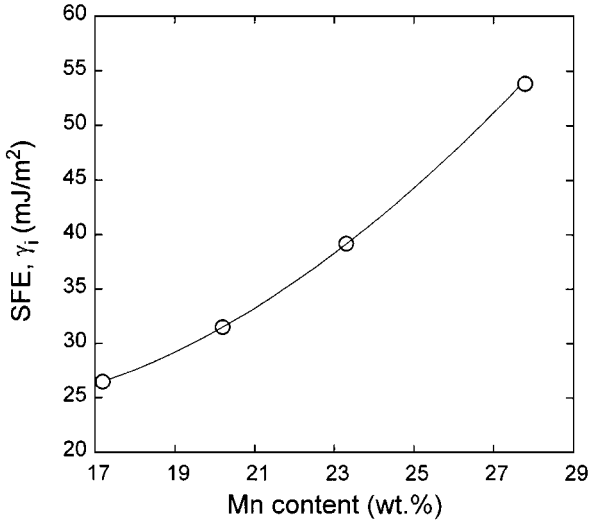

Figure 4 Variation of SFE at 438 K with Mn content in Fe-(17–28)%Mn alloys.

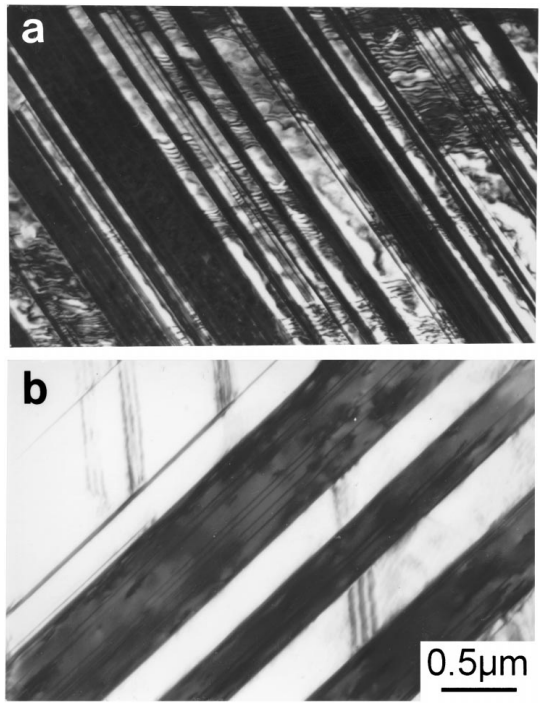

Figure 5 TEM images of (a) Fe-17%Mn and (b) Fe-23%Mn alloys in as-quenched state.

Fe-23%Mn alloys in as-quenched state, respectively, and the result is given in Fig. 5. It is seen that the density of stacking-faults in γ matrix is higher in Fe-17%Mn alloy than in Fe-23%Mn alloy. Fig. 6 shows the correlation between inverse of SFE and damping capacity of γ austenite at 438 K. The damping capacity is low in high inverse of SFE (low SFP), indicating that the increase in Mn content decreases the damping value of γ austenite by decreasing the number of stacking-faults per unit area.

(ii) *The deterioration of mobility of stacking-fault boundaries*: In Fig. 6, it is observed that the damping capacity of γ phase does not depend linearly on the SFP, suggesting that the variation of the SFE with respect to Mn content may affect the mobility of stacking-fault boundary. Therefore, the relation between SFE and

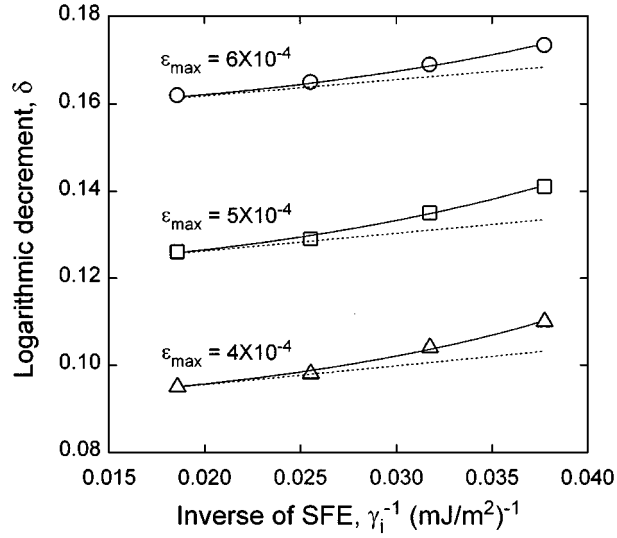

Figure 6 Variation of damping capacity of γ austenite at 438 K with inverse of SFE in Fe-(17–28)%Mn alloys.

mobility of stacking-fault boundary will be theoretically discussed.

In the Fe-Mn based alloys with low SFE, a perfect dislocation in fcc matrix can be easily split into two Shockley partial dislocations, forming a stacking-fault. The equilibrium width between two partial dislocations constituting stacking-fault boundary (d_0) is given by [18]:

$$d_0 = (Gb^2/4\pi) \cdot (1/\gamma) \tag{9}$$

where G is the shear modulus, b is the Burgers vector and γ is the SFE. Under an applied stress of τ on the (111) plane, the forces on the two partial dislocations with Burgers vectors b_1 and b_2, respectively, are [18]:

$$\tau_1 = \tau b_1 \cos(\theta - \pi/6) \tag{10}$$

$$\tau_2 = \tau b_2 \cos(\theta + \pi/6) \tag{11}$$

where θ is the angle between b and τ. Equations 10 and 11 indicate that the values of τ_1 and τ_2 change significantly with respect to θ. When $\theta > 0$, τ_1 becomes higher than τ_2 and thus the stacking-fault would be extended wider, resulting in a new equilibrium width between two partial dislocations (d_0^*). In this case, Equation 9 can be modified as follows:

$$\left(Gb_1^2/4\pi\right) \cdot \left(1/d_0^*\right) = \gamma - (\tau_1 - \tau_2) \tag{12}$$

Substituting for $Gb_1^2/4\pi$ from Equation 9 into Equation 12, we obtain

$$(d_0^* - d_0)/(d_0) = (\tau_1 - \tau_2)/(\gamma - (\tau_1 - \tau_2)) \tag{13}$$

Equation 13 indicates that the stacking-fault is easily extended when the SFE is low. Thus, the decrease in damping capacity of γ austenite by increasing Mn content, is attributed to the decrease in the mobility of the stacking-fault boundaries as well as the decrease in the number of stacking-faults per unit area.

4. Conclusions

The damping capacity of γ austenite single phase decreases with increasing Mn content in Fe-(17–28)%Mn alloys. The microstructural examinations and thermodynamic calculations presented in this paper have shown that the simultaneous decrease in the number of stacking-fault per unit area and the mobility of Shockley partial dislocations constituting stacking-fault boundaries with the increase of Mn content, is responsible for the deterioration of damping capacity.

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