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Shape-memory effect in Co-Ni single crystal ①

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Abstract: The thermal shape-memory effect at room temperature for Co-32% Ni(mass fraction) magnetic shape memory alloy of single crystal was presented. When compressing the sample along the [001] direction at room temperature, strain can be recovered to some extent during later heating and the recovery rate varies with the pre-strain. But no obvious recoverable strain can be obtained along other crystal directions. For the thermal-mechanical training of the sample along [001], the recovery strain decreases obviously during the second round of compress and nearly no recovery happens after the third round of compress. A possible mechanism based on reversible motions of Shockley partial dislocations was proposed.

Key words: Co-based alloys; shape memory effect; Shockley partial dislocation

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1 INTRODUCTION

Ferromagnetic Co-based alloys which exhibit fcc (γ) $\leftrightarrow$ hcp(ε) martensitic transformation are potential magneto-shape-memory (MSM) materials[1-4]. Recently Lee[5] studied the shape-memory effect(SME) in Co-xNi alloys. However, the study on SME of single crystals in Co-Ni alloys has not been yet reported. Our group has found that about 3% reversible strain can be induced by an applied pulse magnetic field of 2 T along the [001] direction in the temperature range between M_d and M_s in Co-33% Ni single crystal[1,3]. However, for this single crystal, the M_s temperature is too low so that the magnetic shape-memory effect can only appear at temperature below 180 K. Therefore a new single crystal with lower nickel content, Co-32% Ni, was prepared in order to induce the thermal and magnetic shape-memory effect at room temperature. In this paper the thermal shape-memory effect along different crystal directions at room temperature in the Co-Ni single crystal was studied.

2 EXPERIMENTAL

A Co-Ni alloy with a nominal composition of Co-32% Ni(mass fraction) was prepared by using metal elements Co and Ni with the purity of more than 99.5%. A single crystal was grown from an ingot with an optical floating zone furnace in purified argon gas. It was oriented by the X-ray back-reflection Laue camera and checked by the X-ray crystal orientation instrument. The samples were cut into dimensions about 3 mm × 3 mm × 5 mm. The dimensions along the [001], [110], [111] and [112] directions were 5mm respectively. The samples were heated at 1 073 K for 30 min and then quenched in icy water. The magnetization under a constant field as a function of temperature (M-T measurement) was measured by Oxford VSM 3001 vibrating sample magnetometer. The differential scanning calorimetry (DSC) tests were done on a TA2910 DSC. The mechanical compress was performed on a Shimazu AG-100kNA mechanical machine. Then the samples were heated to 573 K at a rate of 10 K/min on a LK-02 dilatometer and the length of samples as a function of temperature was recorded. The degree of shape recovery η and the recovery strain ε_R were determined from the change in the length of the specimen at room temperature by the equation:

$$\eta = (l_1 - l_2)/(l_0 - l_1) \times 100\% \quad (1)$$

$$\varepsilon_r = (l_1 - l_2)/l_0 \times 100\% \quad (2)$$

where l_0, l_1 and l_2 are the length before and after deformation, as well as after recovery. The pre-strain was calculated as:

$$\varepsilon_0 = (l_0 - l_1)/l_0 \times 100\% \quad (3)$$

3 RESULTS AND DISCUSSION

X-ray analysis showed that the as-annealed samples exhibit only austenitic phase with fcc structure at room temperature. Based on magnetization curve as a

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function of temperature under 0.4 T magnetic field, the martensitic transformation temperature M_s was determined to be 238 K. The reverse temperatures A_s and A_f were measured to be 429 K and 473 K by DSC.

The mechanical compress behavior of Co-Ni single crystal was studied. Fig. 1 shows the stress—strain curve of single crystal along the [001], [110], [111] and [112] directions respectively. The stress—strain curve of the polycrystalline sample with same composition is also shown for the comparison. Obviously the nominal yield strength of the sample along the [001] direction is much lower than those along other directions, which means that the critical shear stress for stress-induced martensitic transformation is the lowest. For the polycrystalline sample, the nominal yield strength is a medial one.

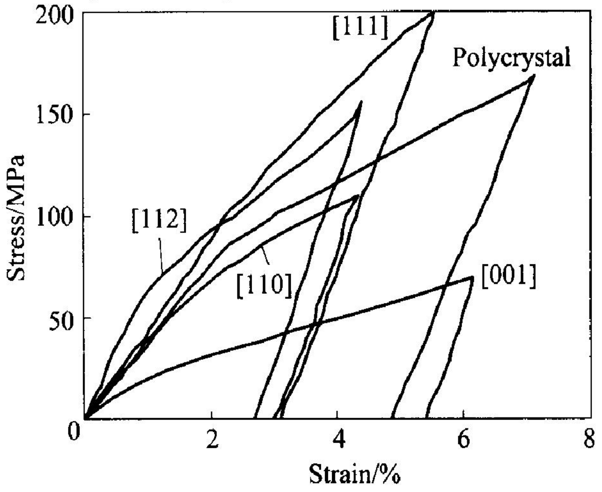

Fig. 1 Strain—stress curves of single crystal along [001], [110], [111], [112] directions and polycrystal

Afterwards the samples were heated to 573 K for recovery. Fig. 2 shows the length change of the samples versus temperature. Only the sample along the [001] direction shows the recovery of strain, other samples expand linearly as the temperature rises and they have no SME. It can be seen that the sample along the [001] begins to recover around 450 K and ends at 500 K. The temperature range is just that when the reverse transformation takes place in the sample.

Fig. 3 shows the degree of shape recovery and the recoverable strain against different pre-strain along the [001] direction, which shows that both the recovery strain and the degree of shape recovery show the tendency to increase as the pre-strain rises, but the degree of shape recovery seems to have a maximum value at certain pre-strain.

For the sample along the [001] direction, the thermal-mechanical training of compression was done three times, as shown in Fig. 4. It is found that the recovery strain decreases obviously during the second round of compress and nearly no recovery happens after the third round of compress.

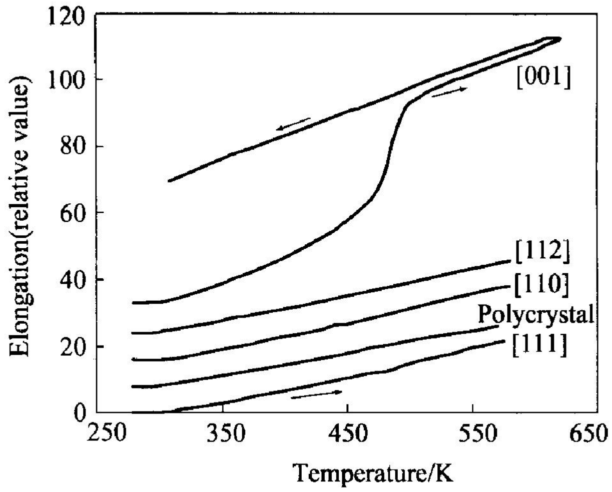

Fig. 2 Length change after compression for single crystal along [001], [110], [111], [112] directions and polycrystal

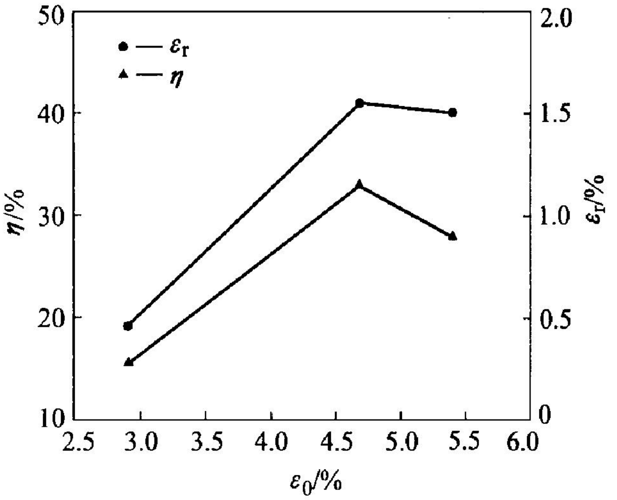

Fig. 3 Variation of recovery strain ε_r and degree of shape recovery η with pre-strain along [001] direction

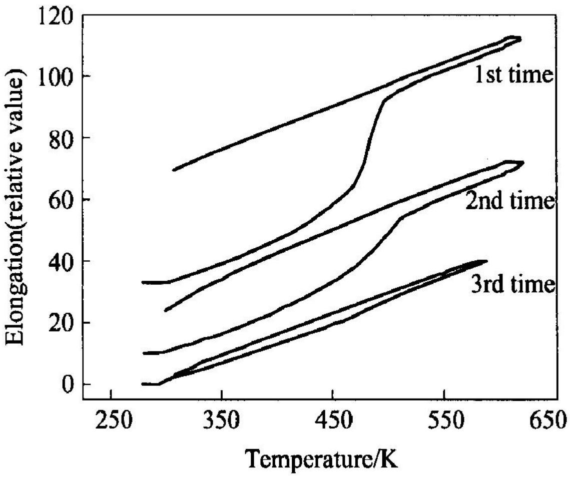

Fig. 4 Elongation along [001] direction during thermal-mechanical training

It is interesting to find that only the sample with the [001] direction shows thermo-SME and also only that direction shows recoverable strain under magnetic field[1], so there may be some similarity in the mechanism of magnetic and thermo SME.

In Co-Ni alloys a martensitic transformation takes place between the face centered cubic(fcc) high temperature phase and the hexagonal close packed (hcp) phase. For the alloys which undergo fcc(γ) $\leftrightarrow$ hcp(ε) the reversible movement of Shockley partial dislocations plays an important role in thermo- and stress-induced martensitic transformation as well as shape memory effect[6-11]. The analysis of the TEM results[11-14] in Co-Ni shows that the slip of Shockley partial dislocations on alternative {111} planes leads to the stacking faults, which act as the nucleus of ε martensite. In our previous paper[15] we have found stacking fault by TEM in Co-Ni alloys and also measured stacking fault probability by X-ray diffraction method. If the sample is deformed, only the strain associated with the formation of stress-induced martensitic transformation by the movement of Shockley partial dislocations will be recovered during later heating. If the perfect dislocations start to move, the permanent plastic deformation will happen and the strain will not be recovered during heating. The movement of $a/6\langle 112\rangle$ partial dislocations on {111} planes also obeys the Schmid law in a similar manner as that of $a/2\langle 110\rangle$ perfect dislocations[16,17]. In fcc parent phase there are 3 directions for $\langle 112\rangle$ or $\langle 110\rangle$ on every {111} plane, so that totally there are 12 motion systems for $a/6\langle 112\rangle$ or $a/2\langle 110\rangle$ dislocations. When the external force is applied along different directions, the Schmid factors may be different and the shear stress will be varied, so the most favorable shear direction may be diverted, respectively. In Table 1 the maximum value of Schmid factors S_s for Shockley partial dislocations and S for perfect dislocations, the number of corresponding moving system, and the average values of $\overline{S}_s$, $\overline{S}$ are listed when the applied force is along [001], [110], [111] and [112] directions, respectively. According to Table 1 the maximum values of S_s and S along [001] are the largest and S_s is larger than S. Although the Schmid factors S_s and S along [110] are the same as that along [001], the $\overline{S}_s$, however, which represents the probability of the movement of partial dislocations, is much smaller in the [110] direction than that in the [001] direction. That means the Shockley partial dislocations are easier to move in the [001] direction than in other directions.

It is noted from Fig. 3 that the recovery strain decreases noticeably during the second compress and no recovery happens after the third compress. This means that a lot of perfect dislocations have formed inside the sample during the repeated compress, which hinder the movement of Shockley partial dislocations.

As mentioned before, the mechanism of thermo-SME and magnetic-induced SME may be similar in some way. However, there is also some diversity between thermo-SME and magnetic-SME. Whereas the magnetic field-induced strain is recoverable simultaneously when the applied magnetic field is unloaded[1], mechanically induced strain is not recoverable during unloading. The mechanically induced strain can only be recovered during subsequent heating. Such phenomena is the same as that in Fe-Mn-Si based alloys, in which the stress-induced martensites can only be recovered during later heating[8,9]. The detailed works is still in progress.

4 CONCLUSIONS

1) While compressing specimens of Co-32% Ni single crystal along the [001], [110], [111] and [112] directions respectively, some part of strain can be recovered along the [001] during later heating. However, no obvious SME can be detected in other crystal directions.

2) For the thermal-mechanical training of the sample along the [001], the recovery strain decreases obviously during the second round of compress and nearly no recovery happens after the third round of compress.

Table 1 Maximum value of Schmid factors S_s for partial dislocations or S for perfect dislocations, number of corresponding system, and average values of $\overline{S}_s$, $\overline{S}$

Direction of applied force	Partial dislocations			Perfect dislocations		
	S_s	Number of corresponding systems	$\overline{S}_s$	S	Number of corresponding slip systems	$\overline{S}$
[001]	0.471	4	0.314	0.408	8	0.272
[110]	0.471	2	0.157	0.408	4	0.136
[112]	0.393	2	0.183	0.408	2	0.181
[111]	0.314	3	0.157	0.272	6	0.136

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