Martensitic Transformation and Shape Memory Effect in Ausaged Fe–Ni–Si Alloys

Yoshiyuki HIMURO, Ryosuke KAINUMA and Kiyohito ISHIDA

Department of Materials Science, Graduate School of Engineering, Tohoku University, Aramaki Aoba-yama 02, Sendai 980-8579 Japan.

(Received on August 20, 2001; accepted in final form on November 12, 2001)

Martensitic transformations in the Fe–(24–30)Ni–(5–8)Si (mass%) alloys have been investigated by means of optical and transmission electron microscopy, differential scanning calorimetry and hardness-testing. The Ms temperature is decreased by Si addition and the morphology of martensite is mainly lenticular in the un-aged specimens. However, a pronounced decrease in the Ms temperature and a change in the martensite morphology from the lenticular to the thin plate type are observed on ausaging at 400°C. The increase in austenite hardness, the decrease in the Ms temperature and the increase in the tetragonality of martensite after ausaging at 400°C are clarified as due to the formation of nanoscale particles of γ’-[(Ni,Fe)3Si with the L12 structure during ausaging at 400°C. The Fe–Ni–Si alloys that form thin plate martensite show the shape memory effect, which arises from the reverse transformation of stress-induced martensite to austenite. Precipitation hardening of the austenite phase by fine γ’ particles during ausaging improves the degree of shape recovery.

KEY WORDS: Fe–Ni–Si alloy; martensitic transformation; ausaging; precipitation; thin-plate martensite; shape memory effect.

1. Introduction

It is well known that some types of ferrous alloys exhibit a shape memory (SM) effect.1,2) From a crystallographic point of view, the martensite phases in such ferrous SM alloys can be classified under three categories: as the BCT martensites, i.e. (1) the Fe–Pt3) and Fe–Ni-based alloys4–8) the HCP martensites, i.e. (2) the Fe–Mn–Si-based alloys 9) and the FCT martensites, i.e. (3) the Fe–Pt10) and Fe–Pd11) alloys, respectively. Among them, the Fe–Ni- and Fe–Mn–Si-based SM alloys are the preferred candidates for technological applications, because they are likely to be much cheaper than Fe–Pt or Fe–Pd based alloys. Between the two of systems of SM alloys based on Fe–Mn–Si and Fe–Ni, the Fe–Ni-based alloys are potentially more attractive, because they have good cold-workability and the BCT martensite is ferromagnetic. This opens up possibilities for development of new types of smart materials that can show, not only a thermally induced SM effect, but also a magnetically induced change accompanying the martensitic transformation. Some of the Fe–Ni-based SM alloys, such as Fe–Ni–C,4) Fe–Ni–Co–Ti,5,6) Fe–Ni–Al–Co–C,7) Fe–Ni–Nb,8) etc. exhibit the SM effect. In these alloy systems, the SM effect is caused by the stress-induced martensitic transformation and its reversion, while the well-ausaged Fe–Ni–Co–Ti alloy exhibits the thermoelastic martensite. The following conditions are indispensable to achieve a good SM effect: (a) the yield stress of the parent austenite should be as high as possible to prevent slip movement in the parent phase during deformation; (b) the martensite should have a thin-plate type morphology with a mobile interface so that the reverse transformation on heating is accomplished by the movement of interface between martensite and austenite; and (c) the tetragonality of the BCT martensite must be large so that the twin boundary energy in martensite can be lowered. It is well known that the coherent precipitates of γ’-(Ni,Fe,Co)3Ti with the L12 structure, Ni3AlC with the perovskite structure and γ0-(Ni,Fe)3Nb with the D022 structure induces not only the hardening of parent phase, but also the formation of thin-plate martensite in the case of Fe–Ni–Co–Ti,5,6) Fe–Ni–Al–Co–C7) and Fe–Ni–Nb8) SM alloys, respectively. It is also known that the strengthening of austenite by ausforming improves the SM property of Fe–Ni–C alloy.5)

Recently, the present authors reported that the thin-plate martensite forms in the as-annealed Fe–Ni–Si alloy.12) Figure 1 shows the vertical section of the phase diagram in the Fe–28%Ni–Si system13) and the morphologies of martensite formed in the as-quenched specimens. The Ms temperature decreases and the morphology of martensite changes from lath to lenticular with increasing Si content. Thin-plate martensite appears in the high Si alloys with very low Ms temperature below −180°C. Moreover, the γ parent phase, which transforms to the thin-plate martensite, shows short-range ordering in the L12 structure. In the BCC...
phase region, on the other hands, since the atomic ordering transitions \( \alpha(A2) \rightarrow \gamma_1(B2) \) and \( \alpha_1 \rightarrow \gamma_2(D0_3) \) occur in a wide composition range as shown in Fig. 1, it is seen that there are very strong attractive interactions between the Fe or Ni and Si atoms. In view of these facts, ausaging at an appropriate temperature could be expected to lead the formation of the metastable \( L1_2 \) phase in the FCC austenite phase with high Si content. In this paper, the effects of ausaging on the microstructure of austenite, the martensitic transformation, and the SM properties of the Fe–Ni–Si alloys are reported.

2. Experimental Procedure

Fe–(24–30)Ni–(5–8)Si (mass%) alloys were prepared from pure Fe (99.9%), Ni (99.9%) and Si (99.999%) by induction melting in an \( Al\textsubscript{2}O\textsubscript{3} \) crucible under an argon atmosphere. After hot rolling to a sheet of thickness about 3 mm, the specimens were austenitized at 1 100°C for 3 h in an evacuated quartz tube, and then quenched into ice water. Some of them were further cold rolled to 0.2 mm in thickness and disk shaped (\( \phi \times 30 \times 40 \) mm) specimens were cut out for transmission electron microscopy (TEM). Several \( 0.2 \times 5 \times 40 \) mm specimens were also cut out for shape memory tests. Aging treatments were conducted at 400°C for various periods up to 3 d after austenitizing. The morphology of martensite was examined by optical microscopy after subzero-cooling down to liquid nitrogen temperature. The Ms temperatures were determined by differential scanning calorimetry (DSC) at a cooling rate of 10°C/min. Vickers micro hardness tester was used for measuring the austenite hardness. Microstructure and crystal structure analyses of the austenite and martensite phases were carried out by TEM using a JEM2000EX microscope. Thin foils were prepared by jet-polishing at room temperature in a solution of one part perchloric acid and four parts methanol. The SM effect was evaluated by the conventional bending tests as described in our previous paper. Sheet specimens were bent at various temperatures to reach up to 1 or 2% surface strain and the curvature of the specimen was measured after releasing the applied strain. The deformed specimen was heated up to 1 100°C and the curvature was measured again at room temperature. The degree of shape recovery was determined by the change in the curvature of the specimens before and after heating.

3. Results and Discussion

3.1. Microstructure of Austenite

Very fine and uniformly dispersed particles were observed in the austenite matrix in some specimens ausaged at 400°C. Figure 2(a) shows the electron diffraction (ED) pattern taken from the Fe–28%Ni–6%Si alloy specimen ausaged at 400°C for 3 d. Some extra spots corresponding to the \( L1_2 \) ordered phase (Cu3Au type) are faint as shown in Fig. 3(a), and no precipitates are observed. This specimen seems to show relatively high degree of short-range order. In the case of ausaged Fe–25%Ni–5%Si alloy, as superlattice reflections are hardly observed in the ED pattern as shown in Fig. 3(b), it is inferred that the austenite phase has the \( A1 \) disordered structure with low degree of short-range order. From the results of TEM examinations carried out on several specimens ausaged at 400°C for 3 d in a similar manner, a phase boundary between the disordered \( A1 \) single phase region and \( A1+L1_2 \) two phases region has been...
determined and plotted in the isothermal section at 1 100°C as shown in Fig. 4. Taking into account the facts that the ordered B2 ($\alpha_1$) and D0$_3$ ($\alpha_2$) BCC phases are formed in a wide composition range in this alloy system as shown in Fig. 1 and a stable L1$_2$ phase of Ni$_3$Si is formed below 1 035°C in the Ni–Si binary system, it is reasonable to understand that the $\gamma'-(Ni,Fe)_3Si$ phase would have a tendency to precipitate during ausaging in the austenite matrix of Fe–Ni–Si alloys with increasing Si contents.

3.2. Mechanical Properties of Austenite

The variation in the hardness with aging time of the austenite phase of the Fe–25%Ni–7.5%Si alloy is shown in Fig. 5. The hardness gradually increases with increasing ausaging time. Figures 6(a) and 6(b) show the iso-hardness contours for the as-quenched and ausaged austenite phases, respectively. Ausaging increases the hardness of austenite, a high hardness value of over 200HV being obtained in a specimen with high Si content over about 7 mass% Si.
change in the hardness value of austenite, $\Delta\text{HV}$, due to aging at 400°C for 3 d, is shown in Fig. 6(c), in which the $\gamma/\gamma'$ phase boundary is shown as a dotted line. It can be seen that the iso-$\Delta\text{HV}$ lines are roughly parallel to the dotted line and that $\Delta\text{HV}$ drastically increases with increasing Si content in the $\gamma/\gamma'$ two phase region. These results suggest that the increase in the hardness of austenite is brought about by the precipitation of $\gamma'$ particles in the austenite matrix during ausaging at 400°C. It is seen also in the $\gamma$ single phase region that the hardness of austenite slightly increases by aging. The origin of this behavior is not clear at present, but slight increase of the degree of short-range order in the austenite phase, which may be enhanced by aging, may be one of the reasons.

3.3. Aging Effect on the Ms Temperature
The Ms temperature of the Fe–25%Ni–7.5%Si alloy sharply decreases due to ausaging, as shown in Fig. 5. The iso-Ms lines estimated from the Ms data of all as-quenched and ausaged specimens are shown in Figs. 7(a) and 7(b), respectively. The decrement ($\Delta\text{Ms}$) in the Ms temperature due to ausaging, $\Delta\text{Ms} = \text{Ms}^{\text{as}} - \text{Ms}^{\text{aus}}$, is shown in (c).

Fig. 7. Ms temperature of Fe–Ni–Si alloys (a) as-quenched and (b) ausaged at 400°C for 3 d. The morphologies of martensite, formed by subzero cooling to $-196^\circ\text{C}$, are also shown. The change in Ms temperature due to ausaging, $\Delta\text{Ms} = \text{Ms}^{\text{as}} - \text{Ms}^{\text{aus}}$, is shown in (c).

3.4. Morphological Change in Martensite
Figure 8 shows the morphologies of the martensites formed in the Fe–25%Ni–7.5%Si alloy. The martensite morphology of this alloy is lenticular in the as-quenched state as shown in Fig. 8(a). It changes to a mixture of lenticular and thin-plate martensites in the early stage of aging as shown in Fig. 8(b). Mainly thin-plate martensites are observed in the specimen aged for longer periods, as shown in Fig. 8(c). The types of martensite morphologies observed in the as-quenched and ausaged specimens are given in Figs. 6 and 7 along with the iso-\text{HV} and iso-Ms lines. The martensite morphology changes in several alloys during ausaging. Especially, it is interesting to note that in the ausaged Fe–24%Ni–7%Si alloy, even though the Ms temperature is relatively high (Ms $= -62^\circ\text{C}$), formation of thin-plate martensites becomes predominant by ausaging.

3.5. Substructure of Thin-plate Martensite
Figure 9 shows the TEM micrograph and the associated selected area diffraction pattern (SAPD) of the thin-plate martensite which forms in the Fe–24%Ni–7%Si alloy.
ausaged at 400°C for 3 d. It is seen that the martensite plate contains a high density of the \( \{112\}_a \) transformation twins that traverse completely from one plate edge to the other. The microstructural features of this martensite plate in the ausaged specimen are similar to those of the thin-plate martensites observed in the Fe–Ni–Co–Ti,\(^6\) the Fe–Ni–C\(^16\) and the as-quenched Fe–28%Ni–7.5%Si alloys.\(^12\) It can be seen that there are some weak superlattice spots corresponding to the \( \{110\}_L \) reflection between the \( \alpha \) martensite fundamental spots as indicated by white arrows in Fig. 9(b). This result suggests that the \( \gamma' \) particles are martensitically transformed to an ordered BCT structure by the martensitic transformation of the matrix to sustain the coherency with the matrix, as reported in our previous paper.\(^17\) This might be the reason for the decrease in Ms temperature due to ausaging as mentioned above, since the additional driving force is needed for stress-induced transformation of the \( \gamma' \) particles, as discussed in the cases of Fe–Ni–Ti\(^18\) and Fe–Ni–Al\(^19\) alloys.

The tetragonality of the \( \alpha' \) martensite and the transformation volume change at Ms temperature in the Fe–24%Ni–7%Si alloy were measured by XRD analysis, and the results obtained are listed in Table 1. It was confirmed that the \( c/a \) ratio of \( \alpha' \) martensite changed from 1.00 (BCC) to 1.01 (BCT) due to ausaging at 400°C for 3 d. This change can be explained as being the result of coherency precipitation of \( \gamma' \) with the L1\(_2\) structure.\(^18\)

The occurrence of thin plate martensites in these alloys seems to be influenced by several parameters. It is well known that a large value of tetragonality of the martensite, which corresponds to a small twinning shear, a low shape strain and a low twin boundary energy\(^20,21\) facilitates the formation of the thin-plate martensite in ferrous alloys.\(^1,2\) However, this change alone may not necessarily be the only factor leading to the formation of the thin plate martensites in all cases. The increase in the tetragonality of the martensite in the Fe–24%Ni–7%Si alloy due to ausaging is one of the factors which changes the morphology of the martensite from the lenticular to the thin-plate type. As already pointed out, the austenite hardness of this alloy also increased and the Ms temperature decreased due to ausaging. These changes as well as the large tetragonality of the martensite are likely to be responsible for the formation of the thin-plate martensite in the ausaged Fe–24%Ni–7%Si alloy. On the other hand, even though the tetragonality of the martensite in the unaged Fe–28%Ni–7.5%Si alloy was 1.00, thin-plate martensites were observed. In this case, the low Ms temperature and the small transformation volume change, \( \Delta V \), at Ms temperature might have been the effective factors influencing the formation of the thin-plate martensite, as listed in Table 1. In contrast, the microstructure of the Fe–25%Ni–7.5%Si alloy ausaged at 600°C for 4 h, which showed a tetragonality value of \( c/a = 1.02 \), was a mixture of thin-plate martensites and lenticular martensites. It is suggested that this might be due to the high Ms temperature (Ms = 41°C) for this case. Therefore, it seems that several factors (such as Ms temperature, \( c/a \), \( \Delta V \) at Ms temperature, and so on) affect the morphology of martensite.

### Table 1.

| Specimen       | Heat treatment | Ms temp, \(^\circ\)C | Hardness of \( \gamma' \) /HV | Morphology | \( \Delta V \) at Ms | \( c/a \) | *Shape recovery / % |
|----------------|----------------|---------------------|-----------------------------|------------|---------------------|--------|-------------------|
| Fe-28%Ni-7.5%Si| As-quenched    | >180                | 154                         | thin-plate | +1.58%              | 1.00   | 19 (25)           |
|                | As-quenched    | -8                  | 140                         | lenticular | +1.77%              | 1.00   | -                 |
|                | 400°C × 3 days | -62                 | 179                         | thin-plate | +2.44%              | 1.01   | 38 (56)           |

* : Specimens were deformed at -196°C under the applied surface strain of 2% (1%) and heated to 1100°C.
martensite, the degree of shape recovery was only 19% under the applied surface strain of 2% at −196°C. Compared with the as-quenched specimen, relatively improved SM effect could be obtained in the ausaged specimen. Figure 10 shows a relation between the degree of shape recovery and the deformation (bending) temperature of an ausaged Fe–24%Ni–7%Si alloy where the applied surface strain was 2%. The degree of shape recovery gradually increases with decreasing the deformation temperature and the maximum shape recovery percentage of 38% could be obtained at a deformation temperature of −196°C.

Figures 11(a)–11(c) are the optical micrographs taken from the same area of the sheet specimen before deformation, after deformation at −196°C and subsequent heating up to 1 100°C, respectively. In this case, the applied surface strain was 1% and the shape recovery percentage was 56. Although the Ms temperature of this ausaged specimen was −62°C, the specimen contained a large amount of retained austenite even after the specimen was cooled down to −196°C as shown in Fig. 11(a). It appears that thermally transformed martensite plates which existed before deformation grew up on deformation by the movement of the flat interface between α′ and γ, as indicated by arrows Nos. 1 and 2, in Fig. 11(a). In addition, a large amount of stress-induced martensites was also formed in the retained austenite on deformation at −196°C. On subsequent heating, the very thin martensite plates such as those indicated by the arrow No. 3 in Fig. 11(b) completely disappeared, while the thick martensite plates such as those indicated by arrow No. 4 in Fig. 11(c) shrunk by the movement of interface, but did not completely disappear.

It has been pointed out that a complete SM effect should result, if a stress-induced thin plate martensite can be easily formed just above the Ms temperature without any accompanying slip deformation in the austenite phase. In the present ausaged Fe–24%Ni–7%Si alloy, however, many slip lines that traverse austenite grains were observed after bending the sheet specimen at −50°C (just above Ms temperature), as shown in Fig. 12(b). As a result, the SM effect...
was hardly obtained by subsequent heating the specimen. On the other hand, the relatively good SM effect is obtained when the sheet specimen is deformed at −196°C, as mentioned above. One could possibly attribute this difference to the presence of a large amount of retained austenite at −196°C, as shown in Fig. 11(a). It could be expected that the irreversible slip deformation in austenite is suppressed more effectively in deformation at much lower temperature. Also the precipitation hardening may be one of the reasons for improvement of SM effect due to aging. However, some of the stress-induced martensites still do not completely vanish on heating as seen in Fig. 11(c). This may be the reason for the incompleteness of shape recovery in this ausaged alloy.

It is reported that the good SM effect in ferrous alloys can be achieved by the large tetragonality of the martensite and the high hardness of the austenite.\(^1\)\(^2\) For example, the tetragonality of the well-ausaged Fe–33%Ni–10%Co–4%Ti alloy is reported to be 1.14\(^5\) and the austenite hardness of the ausaged Fe–31%Ni–10%Co–3%Ti alloy and the ausformed Fe–31%Ni–0.4%C alloy are reported to be 293HV\(^6\) and 316HV\(^4\), respectively. On the basis of this observation, the following explanation is put forward for the incomplete SM effect in the Fe–24%Ni–7%Si alloys. Even though the tetragonality of the ausaged Fe–24%Ni–7%Si alloy is larger than that of the as-quenched Fe–28%Ni–7.5%Si alloy as shown in Table 1, compared with other ferrous SM alloys mentioned above, the tetragonality of the ausaged Fe–24%Ni–7%Si alloy is much smaller (c/a = 1.01). The hardness of the austenite phase (HV170–200) is also much lower, as shown in Fig. 6. It is therefore surmised that, because of the smaller magnitude of the tetragonality and the lower hardness values of the austenite in the ausaged Fe–24%Ni–7%Si alloy, the reverse transformation of the stress-induced martensite is not complete.

Judging from these results, it is clear that ways need to be found for facilitating the formation of thin plate martensite and strengthening of the austenite in order to improve the shape memory effect of this alloy system. Early indications are that the addition of Co would improve the strength of the austenite matrix. The irreversible slip deformation in austenite is suppressed more effectively in deformation at much lower temperature. Also the precipitation hardening may be one of the reasons for improvement of SM effect due to aging. However, some of the stress-induced martensites still do not completely vanish on heating as seen in Fig. 11(c). This may be the reason for the incompleteness of shape recovery in this ausaged alloy.

It is reported that the good SM effect in ferrous alloys can be achieved by the large tetragonality of the martensite and the high hardness of the austenite.\(^1\)\(^2\) For example, the tetragonality of the well-ausaged Fe–33%Ni–10%Co–4%Ti alloy is reported to be 1.14\(^5\) and the austenite hardness of the ausaged Fe–31%Ni–10%Co–3%Ti alloy and the ausformed Fe–31%Ni–0.4%C alloy are reported to be 293HV\(^6\) and 316HV\(^4\), respectively. On the basis of this observation, the following explanation is put forward for the incomplete SM effect in the Fe–24%Ni–7%Si alloys. Even though the tetragonality of the ausaged Fe–24%Ni–7%Si alloy is larger than that of the as-quenched Fe–28%Ni–7.5%Si alloy as shown in Table 1, compared with other ferrous SM alloys mentioned above, the tetragonality of the ausaged Fe–24%Ni–7%Si alloy is much smaller (c/a = 1.01). The hardness of the austenite phase (HV170–200) is also much lower, as shown in Fig. 6. It is therefore surmised that, because of the smaller magnitude of the tetragonality and the lower hardness values of the austenite in the ausaged Fe–24%Ni–7%Si alloy, the reverse transformation of the stress-induced martensite is not complete.

Judging from these results, it is clear that ways need to be found for facilitating the formation of thin plate martensite and strengthening of the austenite in order to improve the shape memory effect of this alloy system. Early indications are that the addition of Co would improve the strength of the austenite matrix. The irreversible slip deformation in austenite is suppressed more effectively in deformation at much lower temperature. Also the precipitation hardening may be one of the reasons for improvement of SM effect due to aging. However, some of the stress-induced martensites still do not completely vanish on heating as seen in Fig. 11(c). This may be the reason for the incompleteness of shape recovery in this ausaged alloy.

4. Conclusions

Results of investigations on the effect of ausaging on the martensitic transformation and shape memory properties of the Fe–Ni–Si alloys lead us to the following conclusions:

1. Ausaging at 400°C leads to the formation of coherent \(\gamma’-(\text{Ni,Fe})_3\text{Si}\) precipitates with the L1\(_2\) structure in the austenite matrix.

2. The formation of coherent \(\gamma’-(\text{Ni,Fe})_3\text{Si}\) precipitates results in an increase in the austenite phase hardness, a sharp decrease in the Ms temperature and the formation of thin-plate martensites below Ms.

3. The formation of thin-plate martensites results in a partial shape memory effect. The degree of shape recovery gradually increases with decreasing the deformation temperature.

4. The improvements in shape memory property due to ausaging can be attributed to the increase in tetragonality of the martensite and the increase in the hardness of the austenite phase by the precipitation hardening of fine \(\gamma’\) particles.

Acknowledgements

The authors wish to thank Dr. L. Chandrasekaran of QinetiQ, UK for help in preparation of this manuscript. One of the authors (R. K.) would like to thank the Steel Research Foundation of the Iron and Steel Institute of Japan (ISIJ) for their support.

REFERENCE

1) T. Maki: Shape Memory Materials, ed. by K. Otsuka and C. M. Wayman, Cambridge University Press, (1998), 117.
2) S. Kajiwara: Mater. Sci. Eng., A273–275 (1999), 67.
3) C. M. Wayman: Scr. Metall., 5 (1971), 489.
4) S. Kajiwara: Trans. JIM, 26 (1985), 595.
5) T. Maki, K. Kobayashi, M. Minato and I. Tamura: Scr. Metall., 18 (1984), 1105.
6) T. Maki, S. Furutani and I. Tamura: ISIJ Int., 29 (1989), 438.
7) S. Kajiwara, T. Kikuchi and N. Sakuma: Proc. of Int. Conf. on Martensitic Transformations (ICOMAT-86), Japan Inst. Met., Sendai, (1986), 991.
8) Yu. N. Koval and G. E. Monastyrsky: Scr. Metall., 28 (1993), 41.
9) A. Sato, E. Chishima, K. Soma and T. Mori: Acta Metall., 30 (1982), 1177.
10) R. Oshima, S. Sugimoto, M. Sugiyama, T. Hanada and F. E. Fujita: Trans. JIM, 26 (1985), 523.
11) T. Sohmura, R. Oshima and F. E. Fujita: Scr Metall., 14 (1980), 855.
12) O. Ikeda, Y. Himuro, I. Ohnuma, R. Kainuma and K. Ishida: Proc. of the 3rd Pacific Rim Int. Conf. on Advanced Materials and Processing (PRICOM 3), Vol. 1, ed. by M. A. Iman et al., TMS, Warrendale, PA (1998), 1503.
13) O. Ikeda, Y. Himuro, I. Ohnuma, R. Kainuma and K. Ishida: J. Alloys & Comp., 268 (1998), 130.
14) R. Kainuma, S. Takahashi and K. Ishida: Metall. Mater. Trans. A, 27A (1996), 2187.
15) P. Nash and A. Nash: Binary Alloy Phase Diagrams, 2nd ed., Vol. 3, ed. by T. B. Massalski et al., ASM Int., Materials Park, OH, (1990), 2859.
16) T. Maki, S. Shimooka, M. Umemoto and I. Tamura: Trans. JIM, 13 (1972), 400.
17) Y. Himuro, R. Kainuma and K. Ishida: “Effect of Ausaging on the Morphology of Martensite in an Fe-25%Ni-7.5%Si Alloy”, J. de Phys. IV, in press.
18) J. K. Abraham and J. S. Pascoe: Trans. AIME, 245 (1969), 759.
19) E. Hornbogen and W. M. Meyer: Acta Metall., 15 (1967), 584.
20) S. Kajiwara and W. S. Owen: Scr. Metall., 11 (1977), 137.
21) M. Umemoto and C. M. Wayman: Metall. Trans. A, 9A (1978), 891.