# **Fe–Pd Ferromagnetic Shape Memory Alloy Thin Films Made By Dual Source DC Magnetron Sputtering**<sup>\*1</sup>

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Fe–Pd films have been deposited onto fused quartz substrates using a dual source dc magnetron sputtering apparatus, equipped with two independent targets of pure Fe and Pd. The Pd content of deposited films can be controlled with an accuracy of 1 at%Pd by varying the power for the Pd target at constant power for the Fe target. Fe–Pd films containing about 29 at%Pd show a fct structure after annealing at 900°C followed by quenching into iced water. These films underwent a thermoelastic fcc-to-fct martensitic transformation and the fct-phase region, where the fct phase is present at room temperature in the annealed films, has a tendency of shifting toward higher Pd content with increasing film thickness. This thickness effect is attributed to the difference in internal stress created during annealing. The martensite start temperature ( $M_s$ ) of films containing 28.5–30.0 at%Pd was higher than room temperature, and it became lower with increasing Pd content. When Fe–30 at%Pd films were separated from the quartz substrate, they showed shape memory behavior upon heating after deformation.

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### 1. Introduction

Shape memory (SM) alloys such as Ti-Ni are widely used as various actuators, in which no rapid actuation response is required. Among actuators made of SM alloys, thin films of Ni-Ti SM alloys have recently received much attention because of potential for rapid actuation response possible by improved heat dissipation. Such rapid actuation response can be expected to be in the shape of thin films at least two orders of magnitude fast, as compared with that in bulk shapes. That is, the actuation response of conventional SM alloys films is as fast as  $\sim 100 \,\text{Hz}^{11}$  Improvement of actuation response is significant in fabricating high performance and power density actuators, especially, for micromachines.<sup>1-3)</sup> More importantly, further improvement of actuation response is possible if ferromagnetic SM alloys can be used. This improvement is unique and can be done by changing driving (or loading) mode necessary for phase transformation to take place from temperature to magnetic field.

There are several ferromagnetic SM alloy systems available at the present moment. They are Ni<sub>2</sub>MnGa,<sup>4)</sup> Fe–Pd,<sup>5)</sup> Fe–Pt,<sup>6)</sup> Fe–Ni–Co–Ti,<sup>7)</sup> Fe–Pd–Pt<sup>8)</sup> and so on. In the present study, ductile Fe–Pd system has been chosen. Fe–Pd alloy system is known to show a thermoelastic fcc-to-fct martensitic transformation in alloys near Fe–30 at%Pd and this transformation accompanies with SM effects in bulk forms.<sup>5,9)</sup> Because the martensite phase is ferromagnetic, magnetic-field-induced SM effects can be expected, suggesting fast actuation response, compared with conventional SM alloys. Recently Ni<sub>2</sub>MnGa single crystal showed 6% strain when a magnetic field of about 400 kA/m is applied.<sup>10)</sup> Fukuda *et al.*<sup>11)</sup> also reported 3% magnetic field induced strain for Fe–31.2 at%Pd single crystal. However, magnetic induced strain of these polycrystalline alloys is not so large at this stage.<sup>12)</sup> Although magnetic field induced strain is small, Fe–Pd alloy system is very attractive to study because of its excellent ductile nature. It should be pointed out that small field-induced-strain may not matter if thin film forms are used. Compared with Ni<sub>2</sub>MnGa, the basic knowledge of physical, mechanical and magnetic properties of Fe–30 at%Pd bulk materials is very poor.<sup>13–17)</sup> The fabrication and characterization of Fe–Pd ferromagnetic SM alloy thin films has scarcely been reported.<sup>18–20)</sup>

Among few studies of Fe–Pd alloy thin films,<sup>19,20)</sup> we have already reported the SM behavior of Fe-28-32 at%Pd films deposited from a composite target consisting of an Fe-Pd alloy disk and fine Pd wires. In our previous studies, however, films thicker than 2 µm were not deposited because of the configuration of the target arrangement, in which alloy composition was adjusted by using Pd wires, which became thinner rather rapidly. Hence, controlling film composition was not easy, especially, when thick films were to be produced. To overcome this problem, we have developed a sputter deposition apparatus with dual dc-magnetron cathodes, which allows us to deposit thicker films under independently controlling incident Fe and Pd atoms onto substrates. The primary purposes of the present study are to investigate some basic knowledge of the fabrication of Fe-Pd ferromagnetic SM alloy thin films by the developed system and to characterize their SM behavior.

### 2. Experimental Procedure

Fe–Pd alloy films were deposited onto fused quartz substrates by a dual source dc magnetron sputtering apparatus. Figure 1 shows a schematic illustration of the apparatus used. This apparatus has two magnetron cathodes (AJA International, A315 and A320), which can independently be powered by two dc power supplies (Advanced Energy, MDX-500) to sputter pure iron ( $\phi$ 50 × 3 mm, 99.99%) and

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Fig. 1 Schematic illustration of dual source sputtering apparatus used.

pure Pd ( $\phi$ 38 × 3 mm, 99.9%) disk targets. Each magnetron inclines at an angle of  $22.5^{\circ}$  with respect to the rotation axis of the substrate holder. The substrate was rotated at a speed of 6 rpm during deposition to try to keep composition and thickness uniform throughout the film plane. The thickness variation of deposited films was found to be less than  $\pm 5\%$ and no compositional variation could be detected for asdeposited films over the entire substrate surface area  $(25 \text{ mm} \times 25 \text{ mm})$ . The distance between the center of the substrate holder and each target was fixed to be 95 mm. Prior to deposition, the apparatus was evacuated to less than  $3.0 \times 10^{-5}$  Pa using a turbomolecular pump. Pure Ar (99.999%) gas was used as working gas and its working pressure was kept to be 0.3 Pa throughout every experiment. No intentional heating of substrates was performed during deposition.

The micro- and crystal-structures, morphology and composition of the films deposited were studied by X-ray diffractometry (XRD: Philips, X'pert MPD, Cu K $\alpha$  radiation), field emission scanning electron microscopy (FE-SEM: Hitachi, S-900) equipped with an energy dispersive X-ray analyzer (EDX: Horiba, EMAX-2770) and atomic force microscopy (AFM: SEIKO Instruments, Nanopics 100).

# 3. Results and Discussion

### 3.1 Control of alloy composition for as-deposited films

Fe–Pd thin films were deposited using a dual source dc sputtering apparatus by adjusting a dc power applied to a target of Pd in the range between 27 and 37 W while keeping a power for an Fe target constant at 300 W so as to control the alloy composition. The deposition rate of such films increases linearly from 65 to 77 nm/min with increasing the applied power for the Pd target in the power range employed. The deposited films were chemically analyzed using a conventional standardless ZAF method for EDX.<sup>19</sup> Results obtained



Fig. 2 Pd contents in as-deposited films as a function of the dc power applied for Pd target. The power applied for Fe target was fixed at 300 W. Circles, rectangles and triangles show data obtained for films with thickness of 300 nm, 2  $\mu$ m and 4  $\mu$ m, respectively.

are depicted in Fig. 2, showing the composition of Pd for asdeposited films as a function of the applied power for the Pd target. As seen, the Pd content in the films changes linearly with the applied power for Pd target. That is, the Pd content can increase by about 0.6 at% per applied power of 1 W for Pd target in the present deposition conditions, indicating that the Pd content in deposited films can be controlled by an order of 1 at%. It was found that the reproducibility of EDX analysis was fairly good, and the measured values fell within an accuracy of  $\pm 0.5$  at%. In this figure, the composition of the specimens with three kinds of thickness is shown, and no thickness effect on the Pd content in as-deposited films can be seen.

Figure 3 shows the typical X-ray diffraction patterns of asdeposited 2-µm-thick films with several Pd compositions that created in the study. It was found that the diffraction patterns of films containing Pd between 28.1 at% and 30.2 at% showed a bcc structure with the strong (110) preferred orientation. On the other hand, films containing higher Pd between 30.7 at% and 31.2 at% showed an fcc structure with the (111) preferred orientation. Microstructural observations made by SEM and AFM clearly showed that as-deposited Fe–Pd films consisted of very fine grains as small as ~0.1 µm and their grains became slightly smaller with increasing Pd content. The surface of films also became smoother with increasing Pd content. Similar tendencies were also observed in a previous study, where Fe–Pd films were deposited using a Fe–Pd alloy target.<sup>19)</sup>

# 3.2 Structure and morphology of heat-treated films

It is in general accepted that as-sputtered alloy films may have phases different from those for thermodynamically stable phases. Thus, we encapsulated all as-deposited Fe–Pd films in evacuated quartz tubes and annealed them at 900°C for 60 min followed by quenching into iced water. Chemical analysis by an EDX method revealed that there is no compositional change after annealing.

Annealed Fe-Pd films were all subjected to X-ray diffractometry to determine their crystal structures and



Fig. 3 X-ray diffraction patterns of as-deposited Fe–Pd films with various Pd contents. The vertical axis is shown in log scale.

features. Some results obtained are shown in Fig. 4. As seen, all diffraction peaks of annealed films are sharp and strong, as compared with those for as-deposited films (Fig. 3). It should be noted that the diffraction patterns of annealed films all show somewhat complicated and more peaks, as compared with those for as-deposited films. The detailed analysis of the diffraction patterns revealed that the annealed Fe-28.1 at%Pd film shows essentially two phases with fct and bcc structures. Based on these fct and bcc crystal structures, the diffraction peaks in Fig. 4 could all be indexed. On the other hand, the films containing 29.1 and 30.2 at%Pd show quite different diffraction pattern. That is, these films show an fct structure with a strong 111<sub>fct</sub> peak, indicating that grains are preferentially orientated along the 111 direction. No bcc phase was retained after annealing, although their structure of as-deposited state was bcc. As Pd content increased further to 30.7 at% and greater, annealed films only show a fcc structure, indicating that this fcc phase formed upon sputter deposition is stable and unchanged under the heat treatment employed in the study. Similar structural alterations were also found to take place after annealing when Fe-Pd films were dc-magnetron sputter-deposited using an alloy target.<sup>19)</sup>

The effect of heat treatment on the crystal structure of the annealed films was investigated as a function of Pd content and film thickness, and some results obtained are depicted in Fig. 5. Here, circles, triangles and rectangles, respectively, denote fcc, fct and bcc phases, and these three phases were observed at room temperature after annealing. The bound-



Fig. 4 X-ray diffraction patterns of annealed Fe–Pd films with various Pd content. Annealing was carried out in an evacuated quartz tube at 900°C for 60 min followed by iced water quenching.



Fig. 5 The crystal structure of annealed films with various thickness as a function of Pd content. Circle, triangle and rectangle plots indicate that the film has fcc, fct and bcc structure at RT, respectively.

aries between bcc and fct phases and between fct and fcc phases shift towards higher Pd content for a given film thickness. The width of Pd content for the fct phase region seems constant regardless film thickness and the width is about 2 at%. It was found that this width is larger than that for bulk material.<sup>16</sup>



27.6at%Pd, fct 29.2at%Pd, fct 30.3at%Pd, fcc 300 nm

Fig. 6 Surface morphology of annealed films with various thickness and Pd contents.

The morphology and size of grains were investigated for all annealed films with various Pd contents and thickness and the results of AFM images are shown in Fig. 6. As seen, the grain size of annealed films becomes larger with increasing thickness and the grains appear to be somewhat smaller for films with higher Pd content. The grain morphology is essentially the same regardless Pd content and crystal structures, especially, when their grains are small. In this case, grains do not show any substructure. When grains are larger, on the other hand, plate-form markings appear. No detailed observations have been pursued in the present study to identify what these markings are. However, they are considered to be martensite plates with different crystal structures. Further detailed observations are underway using, for example, the electron back reflection channeling method to identify grain orientations and habits planes of the plateform markings.

# 3.3 Martensitic transformation behavior in annealed films

Thermoelastic martensite transformation is known to occur in Fe–30 at%Pd alloys in the form of bulk specimens when the crystal structure changes from fcc to fct,<sup>5)</sup> which accompanies SM effects.<sup>9)</sup> In an attempt to investigate whether such thermoelastic martensitic transformation accompanying SM effects also occurs in the present annealed Fe–Pd films, *in-situ* observation of crystal structures was performed during thermal cycling between RT and 90°C using an X-ray diffraction technique. Here, the XRD

apparatus used was equipped with a temperature-control specimen holder, and the temperature of specimen surface was monitored by a copper-constantan thermocouple and controlled within  $\pm 1^{\circ}$ C during measurement. X-ray intensities were measured in an angle between 40° and 53° while keeping the specimen temperature constant. *In-situ* X-ray diffraction experiments were carried out for Fe–Pd alloy films with various Pd contents and some results obtained for Fe–28.5 at%Pd are presented below.

X-ray intensity profiles measured from Fe-28.5 at%Pd films (4 µm thick) are depicted for various test temperatures in Fig. 7. In order to clearly show the behavior of reflection peaks, the X-ray intensity profiles were limited within an angle of  $2\theta$  between  $46^{\circ}$  and  $53^{\circ}$ . As clearly seen in Fig. 7(a), two reflection peaks are present at room temperature at around  $2\theta = 47^{\circ}$  and  $51^{\circ}$ , and these peaks can be indexed as 200<sub>fct</sub> and 002<sub>fct</sub>, respectively. As temperature increases from RT, these two fct peaks become weaker in intensity and move toward around  $2\theta = 49^\circ$ , becoming a single reflection peak at a temperature of 90°C. This single reflection peak can be indexed as the 200<sub>fcc</sub>. Although no intensity profile is presented below  $2\theta = 46^{\circ}$  here, there is a reflection peak at around  $2\theta = 42^{\circ}$  and this peak does not move in this temperature range up to 90°C.15) The peaks at around  $2\theta=49^\circ$  and  $42^\circ$  can be indexed as  $200_{fcc}$  and  $111_{fcc},$ respectively. The reverse behavior of reflection peaks at around  $2\theta = 47^{\circ}$  and  $51^{\circ}$  is also observed during cooling, as seen in Fig. 7(b). That is, the fcc structure goes back to the fct structure during cooling from 90°C to RT.

For understanding the behavior of the reflection peaks



Fig. 7  $200_{fct}$  and  $002_{fct}$  reflection peaks of Fe–28.5 at%Pd film (4  $\mu$ m thick) measured at various temperatures during (a) heating and (b) cooling cycle.



Fig. 8 Change in lattice parameters of the Fe–28.5 at%Pd film (4  $\mu m$  thick) during cooling.

described above, the lattice spacings of the (100) and (001)planes of fct and that of the (100) plane of fcc were calculated for Fe-28.5 at%Pd films at various temperatures (Fig. 7(b)). The calculated values during cooling cycle are plotted as a function of temperature, as shown in Fig. 8. As seen, an fcc structure is present at temperatures between 90°C and 70°C and the (100) lattice spacing of the fcc phase is found to be unchanged at around 0.371 nm. When temperature is lowered to around 70°C, on the other hand, an fct phase occurs and coexists with the fcc phase. The lattice spacings of (001)<sub>fct</sub> and  $(100)_{fct}$  are found to be 0.384 and 0.362 nm, respectively. The lattice spacing of  $(001)_{fct}$  decreases with lowering temperature, while the lattice spacing of  $(100)_{fct}$  increases slightly. The change in lattice parameters as a function of temperature is an indication of a second-order phase transformation taking place in the Fe-Pd films investigated in the present study, although the jump of the lattice spacing accompanied with transformation is found to be larger than bulk material.<sup>17)</sup> Furthermore, the volume of the unit cell for the fct structure is calculated to be larger by  $\sim 4.0\%$  than that for the fcc structure in the present work. It is pointed out that the present annealed Fe-Pd films show a typical thermoelastic martensite transformation, despite the fact that the volume change is not less than one percent, which is the case in most thermoelastic martensite transformation.<sup>21)</sup> This unique property may be related to the morphology of the thin films made in the present study. The reason is not known yet at this moment.

To get more insight into the transformation characteristics, the martensite transformation start and finish temperatures,  $M_s$  and  $M_f$ , of the fcc to fct structure were determined from the relationship between lattice spacing and temperature. Since there are practically a few definitions of the transformation temperatures, here we define  $M_s$  as the temperature where the fct reflection peak was first detected during cooling cycle. Similarly  $M_f$  is defined as the temperature where the high-temperature fcc reflection peak disappears. From Fig. 8,  $M_s$  and  $M_f$  are estimated to be 70 and 56°C, respectively. The fct-to-fcc Austenite start and finish temperatures,  $A_s$  and  $A_f$ , can also be estimated in a similar manner using the relationship between lattice spacing and temperature during heating cycle. It was found that  $A_s$  and  $A_f$  are very close to  $M_f$ 



Fig. 9 Martensitic transformation temperature  $(M_s)$  of Fe–Pd alloy films with various thickness as a function of Pd content.

and  $M_s$ , respectively. The temperature difference between  $M_s$  and  $A_f$  seemed to be as small as 4–6°C, indicating that the temperature hysteresis of the martensite and austenite transformation is very small.

For further understanding of the martensite transformation in the present Fe–Pd film, the relationship between  $M_s$  and Pd content was investigated for annealed films with various thickness. The  $M_s$  points of annealed Fe–Pd alloy films measured in a way described above are plotted in Fig. 9 as a function of Pd content. As seen in Fig. 9,  $M_s$  decreases with increasing Pd content. Similar tendencies were also observed in bulk alloy specimens.<sup>14,16)</sup> It should be emphasized that the 300 nm-thick Fe-27.0 at%Pd film shows slightly lower  $M_{\rm s}$  than that obtained by extrapolation from thicker films. In addition,  $M_s$  of the Fe–Pd annealed films is much higher than that of bulk material.<sup>14,16)</sup> Furthermore, the fct phase is present at room temperature in Fe-Pd films at a wider range of Pd content than that of Fe-Pd bulk material. These differences in  $M_s$  characteristics as compared with those for bulk alloy specimens may be caused by the internal stress created from the difference in thermal expansion coefficient between films and quartz substrates. Since such internal stress is dependent on the film thickness, the transformation temperatures such as  $M_s$  are considered to change with thickness. That is, low  $M_s$  for 300 nm-thick Fe–27.0 at%Pd film compared with that for thicker films may be related to small internal stress created in thin films. This could also be related to that the fct phase was present at higher Pd content when film thickness increased (see Fig. 5). As described above, the  $M_s$  point was defined as the temperature where the X-ray reflection peaks of the fct phase was firstly detected. This may cause some discrepancy in determining  $M_s$  points when films are not thick enough. It should be pointed out that the  $M_{\rm s}$  points measured here from Fe–Pd films with various thickness do not show obvious the thickness effect on  $M_s$ under the present experimental condition. However, systematic experiments may reveal the thickness effect on  $M_s$  if tested in an improved fashion.



Fig. 10 Shape memory behavior occurring for  $4 \mu$ m-thick Fe–30 at%Pd film annealed at 900°C for 60 min followed by iced water quenching. (a) after deformation at liquid nitrogen temperature, and (b) to (d) shape recovery during heating process up to R.T.

### 3.4 Shape memory behavior in annealed films

Our previous study has clearly demonstrated that SM effects occur when an fct-fcc reverse martensite transformation takes place upon heating Fe-29 at%Pd films. These films were annealed and then peeled off from a quartz substrate after sputter deposition using a composite target consisting of an Fe-Pd alloy target and Pd fine wires.<sup>20)</sup> In the present study, an Fe-30 at%Pd film (4 µm thick) was also peeled off from a quartz substrate after annealing and deformed at 20°C prior to heating. Although the As point of the Fe-Pd film bound to the substrate was estimated about 40°C based on Xray diffraction measurements, no shape recovery was observed. When the same film specimen was deformed under liquid nitrogen temperature, on the other hand, SM behavior took place during heating up to room temperature. SM behavior observed in the present study is shown in Fig. 10. As seen in Fig. 10(a), the film specimen is bent immediately after deformation in liquid nitrogen and the bent specimen clearly shows shape recovery during heating to room temperature (i.e., 20°C), Fig. 10(b,c,d). Although no photo is presented here to show the shape prior to deformation, the specimen almost completely went back to the original flat shape at 20°C, Fig. 10(d). Similar perfect shape recovery was also observed when films were formed by sputter deposition using a composite target.<sup>20)</sup> In the present study as described above, no shape recovery took place in the case when peeled-off films were deformed at above room temperature. On the other hand, the martensite transformation behavior has been observed in films when they were not removed from the quartz substrate. That is, internal stresses induced in annealed films raises the  $M_s$  temperature and stabilized the fct phase at room temperature. This is supported by the X-ray diffraction observations, in that the fcc parent phase was present at room temperature in a peeledoff Fe-30 at%Pd film, while the fct martensite phase was present when the film was not removed from the quartz substrate.

In the present study, no experiments have been performed to observe SM effects under magnetic field. Since ferromagnetic Ni<sub>2</sub>MnGa-based SM alloys clearly show magnetic induced SM behavior, similar behavior will be highly anticipated to take place. Because Fe–Pd alloys are ductile and have high saturation magnetization compared with Ni<sub>2</sub>MnGa-based alloys, it is quite interesting to demonstrate magnetic field induced SM behavior of Fe–Pd films and their fatigue behavior under magnetic field loading for various applications. As a series of investigation, we have initiated to study magnetic field induced shape memory behavior using Fe–Pd thin films.

### 4. Conclusions

We have deposited Fe–Pd ferromagnetic shape memory alloy thin films by means of a custom-made dual dcmagnetron sputtering apparatus. The crystal structure, morphology, martensitic transformation and shape memory behavior of such thin films fabricated were investigated. The results obtained are summarized as follows.

- (1) Fe–Pd films can be deposited in a controlling manner to tailor Pd content within an accuracy of 1 at% by changing a sputtering power applied for a Pd target, while keeping power for an Fe target constant. The Pd content affects the stability of phases present in and the morphology of as-deposited films.
- (2) The crystal structure of Fe–Pd films containing 27.5– 30 at%Pd becomes an fct structure after annealing at 900°C for 60 min followed by quenching to iced water. The fct phase region where the fct phase is present at room temperature in the annealed films tends to shift higher Pd content with increasing the film thickness.

- (3) The fct phase in annealed Fe–Pd films transforms into an fcc phase upon heating up to 90°C from room temperature. This fct-to-fcc transformation is thermoelastic in martensite nature, and the martensite transformation start temperature,  $M_s$ , becomes lower with increasing Pd content in the films.
- (4) Annealed Fe–30 at%Pd films deformed at liquid nitrogen temperature showed shape recovery during heating up to room temperature. The shape recovery is associated with the fct-to-fcc reverse martensite transformation. There is some discrepancy between the reverse transformation temperature observed by XRD measurements and shape recovery temperature, which is attributed to the difference in internal stresses created in the film.

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## REFERENCES

- K. D. Skrobanek, M. Kohl and S. Miyazaki: Proc. IEEE 10th Ann. Int. Workshop on Micro Electro Mechanical Systems, (Nagoya, 1997) pp. 256–261.
- W. Van Moorleghem, C. Chandrasekaran and S. Miyazaki: Proc. 5th Intnl. Conf. on New Actuator, (Bremen, 1996) pp. 355–361.
- W. L. Benard, H. Kahn, A. H. Heuer and M. A. Huff: J. Microelectromech. Sys. 7 (1998) 245–251.

- P. J. Webster, K. R. A. Ziebeck, S. L. Town and M. S. Peak: Philos. Mag. B 49 (1984) 295–310.
- M. Sugiyama, R. Oshima and F. E. Fujita: Trans. JIM 27 (1986) 719– 730.
- 6) T. Tadaki: Trans. JIM 18 (1977) 864-870.
- T. Maki, K. Kobayashi, M. Minato and I. Tamura: Scr. Metal. 18 (1984) 1105–1109.
- T. Kakeshita, T. Takeushi, T. Fukuda, T. Saburi, R. Oshima, S. Muto and K. Kishio: Mater. Trans., JIM 8 (2000) 882–887.
- T. Sohmura, R. Oshima and F. E. Fujita: Scr. Metal. 14 (1980) 855– 856.
- 10) S. J. Murray, M. A. Marioni, S. M. Allen, R. C. O'Handley and T. A. Lograsso: Appl. Phys. Lett. 77 (2000) 886–888.
- 11) T. Fukuda, T. Takeuchi and T. Kakeshita: Materia Japan 40 (2001) 544–547. [in Japanese]
- K. Tsuchiya, A. Ohashi, D. Ohtoyo, H. Nakayama, M. Umemoto and P. G. McCormick: Mater. Trans., JIM 41 (2000) 938–942.
- 13) R. D. James and M. Wutting: Philos. Mag. A 77 (1998) 1273.
- 14) M. Sugiyama, R. Oshima and F. E. Fujita: Trans. JIM 25 (1984) 585– 592.
- M. Sugiyama, R. Oshima and F. E. Fujita: J. Japan Inst. Metals 48 (1984) 881–889. [in Japanese]
- R. Oshima: Proc. Intnl. Conf. on Martensitic Transformations, (Nara, 1986) pp. 971–978.
- 17) R. Oshima and M. Sugiyama: J. Phys. 43 (1982) C4-383-388.
- 18) Z. Wang, T. Iijima, G. He, K. Oikawa, L. Wulff, N. Sanada and Y. Furuya: Mater. Trans., JIM 41 (2000) 1139–1141.
- S. Inoue, K. Inoue, K. Koterazawa, K. Mizuuchi and T. Tsurui: Proc. THERMEC '2000, (Las Vegas, 2000), E08-07.pdf, (J. Mater. Processing Technol. 117 (2001)).
- 20) S. Inoue, K. Inoue, K. Koterazawa and K. Mizuuchi: Mater. Sci. Eng. A 339 (2003) 29–34.
- 21) C. M. Wayman: Proc. 1st. JIM Int. Symp. on New Aspects of Martensitic Transformation, Supplements to Trans. JIM, Vol. 17 (1976) pp. 159–170.