Deformation and Recrystallization Behaviour of the TiAl Phase Constituting the TiAl/Ti₃Al Lamellar Structure of Ti-rich TiAl Compounds

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We have grown TiAl crystals of a nearly stoichiometric composition composed of a single lamellar grain as well as those composed of two lamellar grains and studied their deformation behaviour at room temperature. The deformation behaviour of lamellar grains depends strongly on their lamellar orientation. Yield stress is high for uniaxial loading perpendicular or parallel to the lamellar boundaries (the hard mode of deformation), while it is low for uniaxial loading at an intermediate angle (the easy mode of deformation). The tensile elongation for the easy mode of deformation can be as large as 20% at room temperature. By utilizing such easy mode of deformation, TiAl can be rolled up to 50% reduction in thickness at room temperature. Microstructures resulted from annealing of cold-rolled TiAl have been found to depend on the amount of reduction in thickness. This paper summarizes the results of a systematic study on the deformation and recrystallization of TiAl crystals with the lamellar structure.

KEY WORDS: intermetallic compound; TiAl; microstructure; deformation; twinning; rolling, texture; recrystallization.

1. Introduction

Much attention has recently been paid to intermetallic compounds because of their potentially valuable properties, in particular for high-temperature structural applications. In the Ti–Al binary system, there are three intermetallic compounds Al₃Ti, TiAl and Ti₃Al, which are all attractive materials because of their properties such as high melting point, good oxidation resistance and high strength-to-weight ratio. Recently, we have succeeded in growing TiAl crystals of a nearly stoichiometric composition composed of a single lamellar grain as well as those composed of two lamellar grains. The lamellar structure in TiAl of a nearly stoichiometric composition is composed of lamellae of the major constituent phase, TiAl and those of the secondary phase, Ti₃Al. Since the TiAl lamellae contains numbers of thin twins, we call TiAl crystals with a structure composed of lamellae of single lamellar orientation only “polysynthetically twinned (PST) crystals”.¹⁾

Using such PST crystals and PST bicrystals of TiAl we have carried out a study on how the lamellar structure affects the deformation behaviour of as-grown TiAl/Ti₃Al two-phase compounds. Then, we made a study of cold-rolling and subsequent annealing of PST crystals of TiAl. This paper summarizes the results of such a systematic study on the deformation and recrystallization of TiAl PST crystals.^{1,8–10)}

2. Growing of PST Crystals and Preparation of Specimens for Tension, Compression and Rolling Experiments

The master ingot used throughout the present investigation was produced by first melting high-purity titanium and aluminium in a plasma arc-furnace and then remelting it in a vacuum arc-furnace. The composition of the master ingot was Ti–49.3at%Al. Rods, 10 mm in diameter and 100 mm long, were cut from the master ingot. From these rods, polysynthetically twinned crystals were grown using an ASGAL FZ-SS35W optical floating zone furnace at a growth rate of 5 mm h⁻¹ under an Ar gas flow. Orientations of as-grown crystals were determined using both X-ray and electron diffraction methods. When X-rays are directed perpendicular to the lamellar plane, <110> directions common to both the matrix and the twin in TiAl lamellae can be easily determined. Then, oriented compression specimens, approximately 3 mm × 3 mm in cross section and 6 mm long, and also oriented tensile specimens, approximately 2 mm × 0.5 mm in cross-section and 5 mm in gauge length were cut from as-grown crystals and electrolytically polished in 5:35:60 perchloric acid:butanol:methanol solution to remove surface damage. Compression and tensile tests were conducted on an Instron-type testing machine at a strain rate of $2.0 \times 10^{-4} \text{ sec}^{-1}$ at room temperature. Some of compression specimens were deformed to a few % of compression strain and deformation structures on their surface were examined with an optical microscope using

Fig. 1. Orientation of TiAl lamellae with respect to the rolling plane and direction in specimens of (a) the parallel and (b) the perpendicular type for rolling.

Nomarski interference contrast.

For rolling specimens with two different types of geometry shown in **Fig. 1**, approximately 3 mm $\times$ 1.5 mm in cross-section and 7 mm long, were cut from as-grown crystals. The lamellae and the rolling plane intersect in the direction parallel to the rolling direction in **Fig. 1(a)**, while they intersect in the direction perpendicular to the rolling direction in **Fig. 1(b)**. Thus we named specimens with the geometry of **Fig. 1(a)** the parallel type and those with the geometry of **Fig. 1(b)** the perpendicular type. In specimens of the parallel type, a $\langle 110 \rangle$ direction common to both the matrix and the twin in TiAl lamellae lies on the lateral planes perpendicular to the rolling direction, while in specimens of the perpendicular type, it lies on the lateral planes parallel to the rolling direction. The $\langle 110 \rangle$ direction and, therefore, the lamellae lie at angle θ from the rolling plane. We made rolling experiments on specimens of both the parallel and perpendicular types with $\theta = 0^\circ, 45^\circ, 60^\circ$ and 90° .

Some of cold-rolled specimens were put in evacuated quartz capsule and then annealed at 900 and 1000°C. The recovery and recrystallization behaviour of cold-rolled specimens was studied by means of microhardness measurements and both optical and transmission electron microscopy as a function of reduction in thickness, annealing temperature and time. The change in orientation upon rolling and subsequent annealing was investigated by examining (111) pole figures showing the texture developed on the rolled surface of specimens.

Thin foil specimens for electron microscopy were cut from deformed specimens as well as those deformed and subsequently annealed by spark machining and electropolished in a solution of 1 : 2 : 7 perchloric acid : glycerin : methanol at -40°C . The thin foils were examined in a Hitachi H-800 electron microscope.

3. Lamellar Structure and Ordered Domains

Recent high temperature X-ray diffraction and solid-

Fig. 2. Lamellar structure in a two-phase TiAl with a nearly stoichiometric composition.

Fig. 3. (a) $\langle 11\bar{2}0 \rangle$ directions and arrangement of atoms on (0001) in the $D0_{19}$ structure and (b) $\langle 110 \rangle$ directions and arrangement of atoms on (111) in the $L1_0$ structure.

ification experiments for the Ti–Al system have shown that the phase in equilibrium with the liquid phase with a nearly stoichiometric composition is the hexagonal closed packed solid solution based on the structure of αTi .²⁾ This indicates that the $L1_0$ phase (TiAl) is produced from the h.c.p. solid solution which is eventually ordered to the $D0_{19}$ phase (Ti_3Al). The microstructural changes during this phase transformation result in a lamellar structure as shown in **Fig. 2**, where TiAl and Ti_3Al phases have the orientation relationship of

$$\{111\}_{\text{TiAl}} // (0001)_{\text{Ti}_3\text{Al}} \quad \text{and} \quad \langle 110 \rangle_{\text{TiAl}} // \langle 11\bar{2}0 \rangle_{\text{Ti}_3\text{Al}}$$

$\langle 11\bar{2}0 \rangle$ directions on the basal plane in Ti_3Al are all equivalent (**Fig. 3(a)**), while $\langle 110 \rangle^*$ and $\langle 101 \rangle$ on (111) in TiAl with the tetragonal $L1_0$ structure are not equivalent to each other (**Fig. 3(b)**). $\langle 11\bar{2}0 \rangle$ directions on the basal plane in the disordered h.c.p. solid solution are also equivalent. Therefore, at the TiAl/TiAl interface parallel to the lamellar boundaries which are formed during the growth of TiAl lamellae in the direction perpendicular to the basal plane of the h.c.p. disordered solid solution, the following six different orientation relationships may occur between the two neighbouring TiAl lamellae⁴⁾:

$$\begin{aligned} &A\bar{A}'A', \quad A\bar{A}'B', \quad A\bar{A}'C' \\ &A\bar{A}'A', \quad A\bar{A}'B', \quad A\bar{A}'C' \end{aligned} \quad \dots\dots\dots(1)$$

* Notations $\langle hkl \rangle$ indicates a set of equivalent crystallographic directions generated by permutations of $\pm h$ and $\pm k$.

where A , B and C are, for example, $[\bar{1}10]$, $[10\bar{1}]$ and $[0\bar{1}1]$ directions on a (111) plane, which is parallel to the lamellar boundaries, in one of the two neighbouring TiAl lamellae and A' , B' and C' are the corresponding three directions on (111) in the other TiAl lamella. $\parallel$ and $\nparallel$ mean parallel and antiparallel.

When A and A' are parallel to each other, a translational order-fault interface may be formed between the two neighbouring lamellae, otherwise no interface is produced. For the other two parallel cases, which are equivalent to each other, a 120° -rotational order-fault interface is formed.⁴⁻⁶ When A and A' are antiparallel, the two neighbouring lamellae have the $(111)[11\bar{2}]$ -type true twinning relation. For the other two antiparallel cases, which are equivalent to each other, pseudo twins are formed.⁴⁻⁶ Even if a Ti_3Al lamella is sandwiched between the two TiAl lamellae, the orientation relationships between these two TiAl lamellae should be the same as those described above. Thus, we look at a TiAl lamella, other TiAl lamellae have the true or the pseudo twinning relation with respect to the TiAl lamella, or they are related to the TiAl lamella by 120° -rotation about the normal to the lamellar boundaries.

4. Uniaxial Deformation at Room Temperature

Yield stress of PST crystals of TiAl depends strikingly on the angle between the lamellar boundaries and loading axis (Fig. 4). It was high when the lamellae were parallel or perpendicular to the loading axis and it was very low for intermediate orientations. When the lamellae were parallel or perpendicular to the loading axis, shear deformation proceeded across the lamellar boundaries, while shear deformation occurred parallel to the lamellar boundaries for intermediate orientations. Then, the former mode of deformation is seen to be hard and the latter mode of deformation is easy. Both modes of deformation occur by twinning of the type $(111)[11\bar{2}]$ which does not disturb the $L1_0$ symmetry of the lattice or slip of the type $(111)[110]$. Figures 5(a) and 5(b) show structures observed in an undeformed specimen and a specimen deformed in compression along the direction inclined at 31° to the

Fig. 4. Definition of the angle ϕ and yield stress of specimens with various crystal orientations of TiAl lamellae as a function of ϕ .^{1,8)}

lamellar boundaries.⁸⁾ Therefore, the specimen was deformed in the easy mode. Compression strain was about 1%. In comparison with the structure of undeformed specimen, the number of lamellae in the deformed specimen is large and their thickness is small at regions indicated by T . This is because shear deformation occurs parallel to the lamellar boundaries by twinning in these lamellae. By analyzing diffraction patterns obtained from regions indicated by T , these twins are confirmed to be true ones of the $(111)[11\bar{2}]$ -type which does not disturb the $L1_0$ symmetry of the lattice. In the lamellae other than those indicated by T , no deformation twins were observed. The occurrence of the deformation twinning in a TiAl lamella in compression depends on whether the twinning leads to compression. When the deformation twinning whose twin plane is parallel to the lamellar boundaries can occur in a lamella in compression, it can not occur, for example, in other lamellae which have the true twinning relation (corresponding to $A \nparallel A'$) with respect to the lamella because it leads to extension for compression test. This is the reason why twinning occurs in some lamellae but not in other lamellae in Fig. 5(b). In the lamellae where no deformation twinning occurs, dislocations are observed. To establish Burgers vectors of these dislocations, contrast analysis was made on these dislocations and their Burgers vector was found to be of the $1/2\langle 110 \rangle$ type. Thus, true twinning and slip along $\langle 110 \rangle$ are complementary to one another.

Deformation structures for the hard mode of deformation were also examined by electron microscope observations. Deformation twins intersecting the lamellar boundaries and dislocations are observed.^{8,11)} From diffraction experiments, it was revealed that these twins and dislocations were also of the $\{111\}\langle 11\bar{2} \rangle$ true-type and of the $1/2\langle 110 \rangle$ -type, respectively. Thus, we may conclude that deformation of the two-phase TiAl compounds occurs by true twinning of the $\{111\}\langle 11\bar{2} \rangle$ -type and slip along $\langle 110 \rangle$ for both and easy modes of deforma-

Fig. 5. (a) Microstructure observed in an undeformed PST crystal of TiAl and (b) that in a PST crystal deformed in compression along the direction inclined at 31° to the lamellar boundaries.⁸⁾

Fig. 6. (a) Lamellar orientations in a PST bicrystal (schematic) and (b) microstructure of the boundary region in the PST bicrystal.

Fig. 7. Tensile stress-strain curve for the PST bicrystal of Fig. 6 at room temperature.

tion. This is in good agreement with the results of deformation experiments on a $\text{Ti}_{52}\text{Al}_{48}$ compound by Hall and Huang.⁷⁾ There is no difference in the deformation mechanism between the hard and the easy modes of deformation. This indicates that the hard mode is hard solely because the lamellar boundaries constitute an effective barrier to the propagation of the twinning and slip. Recently, we have found that PST specimens with intermediate orientations can be deformed even to about 20% elongation. In view of these findings, it may be concluded that the TiAl phase itself in equilibrium with the Ti_3Al phase is soft and easily deformable. Comparison of yield stresses in compression and in tension in Fig. 4 indicates that no tension-compression asymmetry in yield stress exists in the two-phase TiAl PST crystals. This is because the lamellae which deform by twinning in compression deform by slip in tension and vice versa.

However, polycrystalline TiAl compounds composed of grains with the lamellar structure, in particular, of coarse lamellar grains, show poor ductility. Their tensile ductility is usually 0~3% at room temperature. Since the yield stress and deformation behaviour of lamellar grains is highly anisotropic with respect to the lamellar orientation, a large strain discontinuity may arise at grain boundaries between coarse lamellar grains depending on how the lamellar structure of one of the two adjacent grains is misoriented with respect to that of the other grain. Such a strain discontinuity would lead to crack initiation in the grain boundary region.

Figure 6(a) shows schematically the lamellar orientation

Fig. 8. A crack observed in the PST bicrystal (Fig. 6) tested in tension at room temperature.

of two lamellar grains constituting of a PST bicrystal and Fig. 6(b) shows the microstructure of the boundary region of the PST bicrystal. The upper and the lower grains are expected to deform respectively in the easy mode and mostly in the hard mode. Figure 7 shows the tensile stress-strain curve for the PST bicrystal of Fig. 6. The bicrystal is seen to be deformed to about 4% plastic strain. However, the plastic deformation occurs mostly in the upper grain which can deform in the easy mode. Therefore as the plastic deformation in the upper grain proceeds, strain discontinuity between the upper and the lower grains becomes larger. Finally crack initiation occurs at a grain boundary region in the lower grain as shown in Fig. 8. Plastic deformation occurs also in the region near the grain boundary in the lower grain. However, the amount of plastic deformation in the upper grain is far larger than that in the lower grain. The crack seems to be initiated by failure of a lamellar boundary, probably an interphase boundary TiAl and Ti_3Al and to propagate first along the interphase boundary and then along the boundary between the two lamellar grains. Such crack initiation at lamellar boundaries has been observed also on rolling of PST crystals of TiAl .⁶⁾ Therefore, if lamellar grains are refined, the strain discontinuity problem would become less acute and then a better ductility would be obtained in TiAl compounds with a structure composed of lamellar grains.

5. Rolling at Room Temperature

As shown in the previous section, PST crystals of TiAl can be deformed very easily when the loading axis is inclined at an intermediate angle to the lamellar boundaries, that is deformation occurs in the easy mode. Then, we made an attempt to roll PST crystals of TiAl at room temperature by fully utilizing the easy mode of deformation. Figure 9 shows the maximum total reduction in thickness which was attainable without fracture for specimens of both the parallel and the perpendicular types with four different values of θ .⁹⁾ The values of the maximum total reduction are low for both types of specimens with $\theta=0^\circ$ and $\theta=90^\circ$. The values of the maximum total reduction for specimens of the parallel type with $\theta=45^\circ$ and $\theta=60^\circ$ are also low, while specimens

of the perpendicular type with $\theta=45^\circ$ and $\theta=60^\circ$ are found to be able to be rolled to as much as 50% reduction in thickness without fracture. **Figure 10** shows the shapes of a specimen before rolling (a), a specimen of the parallel type rolled to 20% (b), and a specimen of the perpendicular type rolled to 45% (c). Length strain is seen to predominate for the perpendicular type specimen, whereas width strain predominates for the parallel type specimen. The parallel type specimen of Fig. 10(b) shows about 5% extending and 30% widening at 20% reduction, whereas the perpendicular type specimen of Fig. 10(c) shows about 80% extending and 10% widening at 45% reduction. This indicates that the plastic deformation of these specimens during rolling occurs mainly by shear deformation along the $\langle 110 \rangle$ direction

Fig. 9. The maximum total reduction in thickness for specimens of the parallel and the perpendicular types as a function of the angle θ which is defined in Fig. 1.⁹⁾

Fig. 10. Shapes of specimens before and after rolling: (a) a specimen before rolling; (b) a specimen of the parallel type rolled to 20%; (c) a specimen of the perpendicular type rolled to 45%.⁹⁾

on the planes parallel to the lamellar boundaries, that is by the easy mode of deformation. Upon this type of shear deformation, specimens of the perpendicular type are deformed as shown in **Fig. 11**. This view is supported by the examination of the textures developed on the surface of specimens after rolling to several different amounts of reduction.

Figure 12 shows (111) pole figures obtained for specimens of the perpendicular type, before rolling (a), rolled to 20% (b), and rolled to 40% (c). The starting value of θ was 45° . The pole marked by an arrow in Fig. 12(a) indicates the (111) pole normal to the lamellar boundaries and the other $\{111\}$ poles are consistently located. The specimens rolled to 20% still shows high-intensity areas corresponding to the $\{111\}$ poles (Fig. 12(b)). On comparing Figs. 12(a) and 12(b), however, it is seen that the (111) pole normal to the lamellar boundaries in the specimen rolled to 20% rotates around the direction along which the lamellar boundaries and the rolling plane intersect in such a manner that the value of θ decreases. This observation is consistent with the view that the predominant mode of deformation occurring in specimens of the perpendicular type during rolling is shear deformation such as that in Fig. 11.

The orientation spread of the surface texture in the specimen rolled to 40% is much larger than the corresponding spread in the specimen rolled to 20%. Such spreading of the texture on rolling suggests that not only shear deformation along the $\langle 110 \rangle$ direction on (111) parallel to the lamellar boundaries but also other deformation modes play a role in plastic deformation during rolling and their contribution to the deformation of specimens becomes larger as the amount of rolling reduction increases.

Fig. 11. Plastic deformation of specimens of the perpendicular type by rolling (schematic): (a) before rolling; (b) after rolling.⁹⁾

Fig. 12. (111) pole figures for specimens of the perpendicular type: (a) before rolling; (b) rolled to 20%; (c) rolled to 45%.⁹⁾

6. Recovery and Recrystallization of Cold-rolled PST Crystals

The recovery in microhardness is found to occur in two stages.¹⁰⁾ From electron microscope observations of microstructures of specimens annealed for various periods of time at 900°C, the first stage was found to be associated with the decrease in dislocation density and the second stage with the annealing-out of deformation induced twins.

Figures 13(a) and 13(b) show microstructures of specimens rolled to 12% and annealed at 900°C for 10.5 hr and rolled to 50% and annealed at 900°C for 10.5 hr, respectively. Hardening due to rolling at room temperature fully disappears in both annealed specimens and their hardness is about the same as that of as-grown crystals. However, they show quite different microstructures. This indicates that the recrystallization mode strongly depends on the amount of rolling reduction. Figure 13(b) shows a structure composed mostly of equiaxed grains of TiAl, but the structure of Fig. 13(a) seems to preserve the characteristics of the lamellar structure. If the lamellar structure is preserved, high intensity areas corresponding to $\{111\}$ poles should be

Fig. 13. Microstructures of PST crystals of TiAl: (a) 20% rolled and annealed at 900°C for 10.5 h; (b) 50% rolled and annealed at 900°C for 10.5 h.

Fig. 14. (111) pole figure for a PST crystal of TiAl which was rolled to 15% and annealed at 900°C for 10.5 h.¹⁰⁾

observed on the (111) pole figure obtained from the specimen of Fig. 13(a) since in the lamellar structure of the two-phase TiAl compounds, a $\{111\}$ plane in TiAl lamellae is always parallel to the lamellar boundaries. On comparing Figs. 12(a), 12(b) and 14 which is the (111) pole figure obtained for the specimen of Fig. 13(a), it is obvious that the structure shown in Fig. 13(a) is a lamellar structure with the same crystallographic characteristics as those of the lamellar structure of as-grown and as-rolled PST crystals. Thus, we may conclude that when the PST crystals of TiAl are rolled to more than 40% reduction at room temperature, a structure consisting of equiaxed grains is obtained after recrystallization, while when the rolling reduction is smaller than 20%, subsequent annealing results in a lamellar structure.

Since six different types of ordered domains corresponding to Eq. (1) exist in the TiAl lamellae and these types of domains have different crystal orientations with respect to the loading axis, the operative deformation mechanism may vary with types of domains. The major mechanisms of deformation of the two-phase TiAl compounds are the true twinning of the (111) $[11\bar{2}]$ -type which does not disturb the $L1_0$ symmetry of the lattice and slip of the (111) $[110]$ -type.^{7,8)} However, twinning and slip have been found to be complementary to one another for the easy mode of deformation, in particular, when plastic strain is small.⁸⁾ This means that when PST crystals of TiAl deform by the easy mode of deformation, twinning is operative in some domains and slip occurs in other domains where twinning can not be operative.

When PST crystals with such a deformation structure are annealed, the difference in the stored energy between domains deformed by twinning and those deformed by slip would increase during annealing since the rearrangement and annihilation of dislocations occur much faster than annealing-out of deformation twins. Thus, the growth of domains mostly free of dislocations occurs at the expense of domains deformed by twinning without destroying the lamellar structure. This would be a reason why the lamellar structure is preserved in specimens rolled to less than 20% reduction in thickness even after the full recovery in hardness. When a crystal such as uranium containing deformation twin lamellae is annealed, each of the lamellae, initially lens-shaped is known to become blunted and often contract and eventually disappear.¹²⁾ Any observation of similar effects have not been made so far in cold-rolled PST crystals of TiAl. When the rolling reduction exceeds 40%, TiAl lamellae are deformed in not only the easy mode but also the hard mode. This results in a homogeneous deformation structure comprising both twins and dislocations and then recrystallization results in a structure composed of equiaxed grains.

7. Conclusions

We have grown TiAl crystals of a nearly stoichiometric composition with a lamellar structure composed of TiAl and Ti_3Al lamellae of a single lamellar orientation only. Since the lamellae of the TiAl phase, that is the major

constituent phase of the lamellar structure, contain large numbers of thin twins, we call our crystals polysynthetically twinned (PST) crystals. Then, a systematic study of the microstructure, deformation and recrystallization of PST crystals and PST bicrystals of TiAl has been made. It was found that:

(1) Six different types of ordered domains described by Eq. (1) exist in the TiAl lamellae.

(2) Yield stress depends strikingly on the angle between the lamellar boundaries and loading axis. It is high when the lamellar boundaries are parallel or perpendicular to the loading axis and it is very low for intermediate orientations. No tension-compression asymmetry in yield stress was observed.

(3) The result of (2) indicates that two modes of deformation, *i.e.* hard and easy modes, can occur in PST crystals of TiAl. The hard and the easy modes of deformation were found to correspond to shear deformation across the lamellar boundaries and shear deformation parallel to the lamellar boundaries, respectively.

(4) The tensile elongation at room temperature can be as large as 20% for specimens which deform in the easy mode.

(5) The results of deformation of PST bicrystals suggest that crack initiation often occurs along a lamellar boundary in the boundary region of bicrystal.

(6) If PST crystals of TiAl are oriented such that the easy mode of deformation is operative and produces length strain during rolling, they can be rolled to as much as 50% reduction in thickness at room temperature.

(7) When PST crystals are rolled at room temperature and subsequently annealed at temperatures in the range 900°–1 000°, recovery in microhardness occurs in two stages. The first and the second stages correspond to the annealing-out of dislocations and that of deformation twins, respectively.

(8) Recrystallization mode of cold-rolled PST crystals depends on rolling reduction. When the rolling reduction is larger than 40%, a recrystallized structure composed of equiaxed grains is obtained, while when it is smaller than 20%, the lamellar structure is preserved even after the complete recovery in hardness.

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REFERENCES

- 1) T. Fujiwara, A. Nakamura, M. Hosomi, S. R. Nishitani, Y. Shirai and M. Yamaguchi: *Philos. Mag. A*, **61** (1990), 591.
- 2) C. McCullough, J. J. Valencia, C. G. Levi and R. Mehrabian: *Acta Metall.*, **37** (1989), 1321.
- 3) S. M. L. Sastry and H. A. Lipsitt: *Metall. Trans. A*, **8** (1977), 299.
- 4) M. Yamaguchi and Y. Umakoshi: *Prog. Mater. Sci.*, **34** (1990), 1.
- 5) C. R. Feng, D. J. Michel and C. R. Crowe: *Scr. Metall.*, **22** (1988), 1481; **23** (1989), 241, 1135.
- 6) M. Yamaguchi, S. R. Nishitani and Y. Shirai: High Temperature Aluminides and Intermetallics, ed. by S. H. Whang, C. T. Liu, D. P. Pope and J. O. Stiegler, TMS, Warrendale, (1990), 63.
- 7) E. L. Hall and S. C. Huang: *J. Mater. Res.*, **4** (1989), 595.
- 8) H. Inui, A. Nakamura and M. Yamaguchi: High Temperature Ordered Intermetallic Alloys IV, MRS, Pittsburgh, (1991), 569.
- 9) S. R. Nishitani, M. H. Oh, A. Nakamura, T. Fujiwara and M. Yamaguchi: *J. Mater. Res.*, **5** (1990), 484.
- 10) M. H. Oh, H. Inui, S. R. Nishitani and M. Yamaguchi: High Temperature Ordered Intermetallic Alloys IV, MRS, Pittsburgh, (1991), 871.
- 11) T. Hanamura, R. Uemori and T. Tanino: *J. Mater. Res.*, **3** (1988), 656.
- 12) R. W. Cahn: Deformation Twinning, Metallurgical Society Conferences Vol. 25, ed by R. E. Reed-Hill, J. P. Hirth and H. C. Rogers, Gordon and Breach Science Publishers, New York, (1964), 1.