Effects of Particle-Dispersion on the Strength of an Alumina Fiber-Reinforced Aluminum Alloy Matrix Composite

Kazunori Asano and Hiroyuki Yoneda

Department of Mechanical Engineering, Faculty of Science and Engineering, Kinki University, Higashiosaka 577-8502, Japan

Aluminum alloy matrix hybrid composites, in which alumina particles were dispersed among continuous alumina fibers, were fabricated by squeeze casting, and the influence of temperature and the effects of the particle-dispersion on the strength of the composites were then investigated. The particle-dispersion among the fibers minimized preform contraction and fiber-to-fiber contact due to the melt infiltration during the squeeze casting. The tensile strength, 0.2% proof stress and elastic modulus of the composites in the longitudinal direction increased with increasing the fiber volume fraction, retaining nearly the same values up to 623 K. These values of the hybrid composite were larger than those of the particle-free composite at every temperature. This is because the fibers were distributed uniformly owing to the particles that prevented the fiber-to-fiber contact, leading to the reduction of stress concentration at the points of direct fiber contact, and stress transmission between the fiber and the matrix becomes easy. At every temperature, the transverse tensile strength and proof stress of the particle-free composite was lower than that of the unreinforced alloy, because the fracture was initiated at the fiber-to-fiber contact point and the cracks propagated mainly along the fiber-matrix interface. In contrast, the strength of the hybrid composite was close to that of the unreinforced alloy even at high-temperature, because the cracks propagated mainly throughout the matrix owing to the uniform distribution of the fibers and the strong fiber-matrix bond.

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1. Introduction

There are many reports on the properties of the continuous fiber-reinforced aluminum alloy matrix composites because the reinforcement with the fiber considerably improves the strength of the aluminum alloy not only at room temperature but also at high temperature.¹⁻⁹⁾ These composites are mainly fabricated by squeeze casting¹⁻⁷⁾ and hot pressing.^{8,9)} Squeeze casting is an attractive process because it can be used even poor wettability between the melt and reinforcement¹⁰⁾ and is suitable for strengthening selected regions of the casting and processing complex shapes.¹¹⁾ Also, it shows high cost performance.¹²⁾ However, because of the fiber contact, attributed to the contraction of the preform during melt infiltration, the transverse mechanical properties obtained are not sufficient. To prevent the fiber contact, dispersion of fine particles or whiskers between the fibers in the composite using a preform of the fibers having particles or whiskers attached to their surfaces has been proposed. The composite is called a hybrid composite. Aluminum alloy matrix hybrid composites have been fabricated using the following reinforcements: continuous SiC fibers having SiC particles attached to their surfaces,¹³⁾ continuous carbon fibers having SiC particles and whiskers attached to their surfaces,^{14,15)} and continuous Tyranno fibers having β -SiC particles attached to their surfaces.¹⁶⁾ As Al_2O_3 (alumina) has high rigidity and strength, and outstanding resistance to heat and wear, the alumina can improve the strength of aluminum alloy not only at room temperature but also at high temperature. With this improvement, the alloy could be used as structural heat-resistant parts. However, no reports on high-temperature properties of the hybrid composites using fine alumina particles and continuous alumina fibers could be found.

In this study, Al-Cu-Mg alloy matrix hybrid composites using continuous alumina fibers having fine alumina particles

attached to their surfaces were produced by squeeze casting. The relationship among microstructure, strength characteristics in the range from room temperature to 623 K, and the volume fraction of the reinforcements examined. Additionally, the effects of the alumina particle-dispersion on the high-temperature strength and the strengthening mechanism were investigated based on fractography.

2. Experimental Procedures

JIS AC1B aluminum alloy (Al-4.3Cu-0.2Mg alloy) was used as the matrix metal, and continuous γ -alumina fibers (Sumitomo Chemical Inc., SV-10-1K) and α -alumina particles (Showa Denko, Inc., CB-A01) were used as the reinforcement. The chemical composition of AC1B aluminum alloy and the characteristics of the reinforcements¹⁷⁾ are listed in Tables 1 and 2, respectively. It has been reported that the tensile strength of the fibers decreases little after a long-

Table 1 Chemical composition of AC1B (mass%).

Cu	Mg	Fe	Ti	Al
4.31	0.20	0.15	0.14	Bal.

Table 2 Properties of alumina fiber and alumina particle.

	alumina fiber	alumina particle
Diameter (μm)	15	1.0
Density (Mg/m^3)	3.3	4.0
Elastic modulus (GPa)	210	—
Tensile strength (MPa)	1800	—
Crystal phase	γ	α
Composition (mass%)	Al_2O_3 85 SiO_2 15	Al_2O_3 99.9

Fig. 1 Appearance of preform.

time exposure at 1373 K in the air.¹⁸⁾ The preform was fabricated as follows: (1) alumina particles were dispersed in water containing a small amount of polyvinyl alcohol as a binder, (2) the particles were attached to the fiber surfaces by soaking the fibers in the solution, and (3) the fibers were packed unidirectionally in a steel square pipe with open ends, dried at 373 K, and then heated at 973 K for 1 hour to remove the binder in the preform. A schematic illustration and a SEM image of the preform are shown in Fig. 1. The particles can be seen attached to the fiber surfaces thus preventing fiber-to-fiber contact.

The composite was fabricated by squeeze casting. The preform, still in the steel pipe, was placed horizontally in the mold, and the molten aluminum alloy was poured into the mold at 1073 K. Pressure was quickly applied using a plunger and maintained until solidification was complete. The steel square pipe with open ends makes it easy to infiltrate the melt parallel to the fiber axis from both ends simultaneously. In this way, preform contraction is limited. The preform temperature was 973 K, the mold temperature was 573 K, the applied pressure was 40 MPa, and the pressurization retention time was 60 s. The volume fraction of the fibers (hereinafter V_f) was made 40, 50 and 60%, and the volume fraction of the particles (hereinafter V_p) was 0 and 10%. T6 heat treatment (solutionized at 788 K for 36 ks and water-quenched, followed by aging at 433 K for 14.4 ks) was applied to the composites.

Their microstructures and the fiber-matrix interfaces were observed. Tensile tests in the longitudinal and transverse direction were conducted with machined plate specimens 3 mm in thickness, 8 mm in width and 10 mm in gage length to obtain tensile strength, 0.2% proof stress, elastic modulus and elongation in the range from room temperature (293 K) to 623 K. In the tensile test, an Instron testing system was used and the crosshead speed was 1.67×10^{-5} m/s. Specimen numbers were determined by the combination of V_f and V_p . For example, composite 40-0 contains 40% fibers and 0% particles, and composite 40-10 contains 40% fibers and 10% particles.

3. Results and Discussion

3.1 Microstructure

Figure 2 shows the microstructures of the transverse section of particle-free composites (40-0 and 60-0) and

Fig. 2 Microstructures and fiber distribution in composites [(a) (e) 40-0, (b) (f) 40-10, (c) (g) 60-0, (d) (h) 60-10].

Fig. 3 TEM photograph of a region near a fiber-matrix interface in the composite (50-10).

hybrid composites (40-10 and 60-10). The frequency distributions of the distance between adjacent fibers are shown in Figs. 2(e) through (h). The distance between adjacent fibers was measured with an image analyzer using microstructure photographs. No defects or damage in the particles or fibers were observed in the microstructure. When the particles were not introduced, the fibers were distributed non-uniformly having fiber-to-fiber contact (Figs. 2(a) and (c)). The frequency of fibers touching was between 50% and 60% (Figs. 2(e) and (g)). In contrast, in the hybrid composites, the particles on the fiber surfaces prevented the fiber contact, and some particles were dispersed in the matrix (Figs. 2(b) and (d)). The frequency of fibers touching was almost nil, while that 1–10 μm apart was about 70% (Figs. 2(f) and (h)). These results indicate that the fibers are distributed more uniformly in the hybrid composite.

A TEM image near the fiber-matrix interface in a hybrid composite (50-10) is shown in Fig. 3. There was no reaction product near the interface between the fiber and the matrix, indicating that the fibers bonded with the matrix without causing any excessive reaction or degradation. The grain size of the γ -alumina in the fiber is about 10 nm, which is almost equal to the grain size in the as-received fiber.¹⁹⁾ This indicates that degradation of the fiber strength due to the grain growth during the de-binder process, the infiltration of molten aluminum alloy, and the solution treatment did not occur.

3.2 Mechanical properties

Figure 4 shows stress-strain curves of 50% fiber-reinforced composites tested at room temperature (293 K) and 623 K. In the longitudinal direction, stress increased almost linearly as the strain increased at every temperature. Tensile resistance to deformation of the hybrid composite (50-10) was larger than that of the particle-free composite (50-0). Tensile resistance to deformation in the transverse direction was smaller than that in the longitudinal direction and decreased as the temperature increased. The resistance increased with the introduction of the particles as well as in the longitudinal direction.

Fig. 4 Stress-strain curves of composites ($V_f = 50\%$) tested at (a) 293 K and (b) 623 K.

Figure 5 shows the relationship between the tensile strength and the temperature for the unreinforced alloy and the composites. The tensile strength of the unreinforced alloy decreased little up to about 423 K, above which it decreased considerably (indicated by the open circles in Fig. 5). For the composites, the longitudinal tensile strength increased as V_f increased and decreased little even at high-temperature. This is probably because the fibers are subject to longitudinal tensile stress. The longitudinal tensile strength of the hybrid composite was larger than that of the particle-free composite with the same V_f . The transverse tensile strength of the particle-free composite were lower than those of the unreinforced alloy and decreased as V_f increased. They were also improved by the introduction of the particles; the transverse tensile strength of the hybrid composite were almost the same as those of the unreinforced alloy (Fig. 5(b)). The increase in the strength by the introduction of particles was particularly large up to about 423 K, above which the increase was small owing to the low strength of the matrix. Nevertheless, the strength of the hybrid composite was slightly larger than that of the particle-free composite at 623 K.

Generally, the longitudinal tensile strength of the continuous fiber-reinforced composite can be estimated by the rule of mixture expressed as eq. (1).²⁰⁾

$$\sigma_{\text{rom}} = \sigma_f V_f + \sigma'_m (1 - V_f) \quad (1)$$

where σ_{rom} is the tensile strength of the composite, σ_f is the tensile strength of the fiber, and σ'_m is the stress on the matrix when the fibers reach their ultimate tensile stress in the composite. Table 3 shows σ_{rom} , experimental values of the

(a) Longitudinal direction

(b) Transverse direction

Fig. 5 Tensile strength of unreinforced alloy and composites at various temperatures.

Table 3 Comparison between experimental and calculated tensile strengths in the longitudinal direction.

Specimen	293 K			623 K		
	σ_{cL}	σ_{rom}	σ_{cL}/σ_{rom}	σ_{cL}	σ_{rom}	σ_{cL}/σ_{rom}
40-0	510	887	0.57	480	782	0.61
40-10	600	890	0.67	570	782	0.73
60-0	645	1186	0.54	615	1120	0.55
60-10	730	1190	0.61	710	1120	0.63

 σ_{cL} : Experimental tensile strength (MPa) σ_{rom} : Tensile strength calculated by rule of mixture (MPa)

tensile strength (σ_{cL}), and the ratios of σ_{cL} to σ_{rom} . From Table 2, the value of σ_f was set at 1800 MPa. Although σ_{cL} was smaller than σ_{rom} for every composite, the tensile strength and the ratios σ_{cL}/σ_{rom} of the hybrid composites (40-10 and 60-10) were larger than those of the particle-free composites. This suggests that the strengthening effect of the fibers with the particles is large.

Figure 6 shows the relationship between 0.2% proof stress and temperature for the unreinforced alloy and the composites. The variation in the 0.2% proof stress as a function of temperature was similar to that in the tensile strength in the longitudinal and transverse directions.

Figure 7 shows the relationship between the elastic modulus and the temperature for the unreinforced alloy and

(a) Longitudinal direction

(b) Transverse direction

Fig. 6 0.2% proof stress of unreinforced alloy and composites at various temperatures.

Fig. 7 Elastic modulus of unreinforced alloy and composites at various temperatures.

the composites. The longitudinal and transverse elastic modulus increased as V_f increased and further increased with the introduction of the particles. Although the longitudinal elastic modulus was almost constant over the temperature range measured, the transverse elastic modulus decreased as the temperature increased. The decrease in the transverse elastic modulus is probably because the elastic modulus of the matrix decreased as the temperature increased.

As well as the tensile strength, the longitudinal elastic modulus can be estimated by the rule of mixture expressed as eq. (2).²⁰⁾

$$E_{rom} = E_f V_f + E_m (1 - V_f) \quad (2)$$

where E_{rom} , E_f and E_m are the elastic modulus of the composite, the fiber and the matrix, respectively. From Table 2, E_f was set at 210 GPa. Table 4 shows E_{rom} , the experimental elastic modulus (E_{cL}), and the ratios of E_{cL} to E_{rom} . For the particle-free composite, E_{cL} was smaller than E_{rom} . The introduction of the particles increased E_{cL} , resulting in almost the same values as E_{rom} at room temperature and slightly larger than E_{rom} at 623 K. This is probably because the introduction of the particles reduces the fiber-to-fiber contact and the stress transmission between the fiber and the matrix becomes easy, and increases the elastic

modulus of the matrix (E_m).

Figure 8 shows the relationship between the elongation and the temperature for the unreinforced alloy and the composites. Although the elongation of the unreinforced alloy increased as the temperature increased, that of the particle-free composite in the longitudinal (Fig. 8(a)) and transverse (Fig. 8(b)) directions changed little even as the temperature increased. In contrast, the elongation of the hybrid composite increased as the temperature increased. This tendency was pronounced when the volume fraction of the matrix was large (V_f was small).

3.3 Relationship between strength and fracture morphology

3.3.1 Longitudinal fracture morphology

Figure 9 shows the fracture surfaces of the composites after a longitudinal tensile load was applied at room temperature. There is almost no fiber pullout on either fracture surface, suggesting that the interfacial bond between the fibers and the matrix was relatively strong. Under more specific observation, the surface of the particle-free composite was flat at the fiber-to-fiber contact points (arrows in the Fig. 9(a-1)). Fig. 9(a-2) is a high magnification micrograph near the contact point, showing the chevron patterns on the fracture surfaces of the fibers initiated at the fiber-to-fiber contact points indicated by arrows. In contrast, the surface of the hybrid composite was irregular and a flat region was not observed (Fig. 9(b-1)). The chevron patterns were observed on the fracture surfaces of the fibers, but the initial points of the fracture varied: the surface of the fiber, the interior of the fiber (Fig. 9(b-2)).

Figure 10 shows the fracture surfaces of the composites at 623 K. The variation in the morphology of the fracture surface by the particle-dispersion was similar to that at room temperature.

Based on these fractographs, the fracture morphology when the tensile load was applied in the longitudinal

Table 4 Comparison between experimental and calculated elastic modulus in the longitudinal direction.

Specimen	293 K			623 K		
	E_{cL}	E_{rom}	E_{cL}/E_{rom}	E_{cL}	E_{rom}	E_{cL}/E_{rom}
40-0	102	126	0.81	96	104	0.92
40-10	124	126	0.98	122	104	1.17
60-0	131	154	0.85	126	139	0.91
60-10	152	154	0.99	150	139	1.08

E_{cL} : Experimental elastic modulus (GPa)

E_{rom} : Elastic modulus calculated by rule of mixture (GPa)

Fig. 8 Elongation of unreinforced alloy and composites at various temperatures.

Fig. 9 SEM photographs of longitudinal fracture surfaces of composites after tensile test at 293 K.

Fig. 10 SEM photographs of longitudinal fracture surfaces of composites after tensile test at 623 K.

direction was considered. The model for the crack propagation is shown in Fig. 11. When a fiber breaks, a neighboring fiber directly contacted by the broken fiber breaks easily because stress concentration occurs in the neighboring fiber. Close-packed fibers were frequently observed on the fracture surfaces of the particle-free composite. Thus, progressive fracture of touching fibers was considered to occur easily, as shown in Fig. 11(a). When the touching fibers break, the fracture surface must be flat. This would be why the surface of the particle-free composite became flat. Under this fracture morphology, the strengthening effect of the fiber cannot be fully displayed. In contrast, for the hybrid composite, in which the fibers were distributed uniformly, the matrix around a fiber relieves the stress concentration after the fiber breaks. Thus the stress concentration does not easily occur,

and each fiber is subject to the stress. Consequently, irregular surfaces are frequently formed as shown in Fig. 11(b). That is, because the introduction of the particles reduces the fiber-to-fiber contact and the stress transmission between the fiber and the matrix becomes easy, the longitudinal strength of the hybrid composite was higher than that of the particle-free composite. At higher temperature, stress relaxation by the matrix would be more pronounced because the ductility of the matrix increased. This idea is supported by the fact that the ratios σ_{CL}/σ_{rom} at 623 K are larger than those at room temperature with the same V_f , as shown in Table 3.

3.3.2 Transverse fracture morphology

Figure 12 shows the fracture surfaces of the composites after a transverse tensile load was applied at room temperature. In this investigation, both sides of the fracture surfaces

Fig. 11 Schematic illustrations of the longitudinal fracture in composites.

Fig. 12 SEM photographs of transverse fracture surfaces of composites after tensile test at 293 K (f-f: Fracture at fiber, f-m: Fracture at fiber-matrix interface, m-m: Fracture at matrix).

were observed to determine where the fracture occurred. On the fracture surface of the particle-free composite, many fibers and grooves remaining in the matrix can clearly be observed (Fig. 12(a)). In contrast, the matrix is dominant on the fracture surface of the hybrid composite (Fig. 12(b)). This suggests that the cracks propagated mainly throughout the matrix.

Figure 13 shows the fracture surfaces of the composites at 623 K. The variation in the morphology of fracture surface by the particle-dispersion was similar to that at room temperature.

Based on these fractographs, the fracture morphology

Fig. 13 SEM photographs of transverse fracture surfaces of composites after tensile test at 623 K (f-f: Fracture at fiber, f-m: Fracture at fiber-matrix interface, m-m: Fracture at matrix).

Fig. 14 Schematic illustrations of the transverse fracture in composites.

when the tensile load was applied in the transverse direction was considered. The model for crack propagation is shown in Fig. 14. The transverse tensile strength and elongation of the particle-free composite were significantly smaller than those of the unreinforced alloy. The fiber surfaces and grooves remaining in the matrix shown in Figs. 12 and 13 correspond to the points of direct fiber contact where the bond between the fibers and the matrix was insufficient. Because the stress transmission does not completely occur at the point of direct fiber contact, the cracks are considered to propagate selectively near the fiber-matrix interface, as shown in Fig. 14(a). The transverse tensile strength of the hybrid composite was close to that of the unreinforced alloy. This is probably

Table 5 Transverse tensile strength of composites and each area fraction on the fracture surfaces.

Specimen	293 K				623 K			
	A_{ff}	A_{fm}	A_{mm}	σ_{cT}	A_{ff}	A_{fm}	A_{mm}	σ_{cT}
40-0	13	47	40	150	4	51	45	92
40-10	12	9	79	290	4	23	73	116
50-0	14	65	21	115	6	70	24	78
50-10	10	17	73	288	5	33	62	110
60-0	15	77	8	100	7	76	17	65
60-10	10	31	59	260	5	46	49	101

A_{ff} : Area fraction of fiber (%)

A_{fm} : Area fraction of fiber-matrix interface (%)

A_{mm} : Area fraction of matrix (%)

σ_{cT} : Transverse tensile strength (MPa)

due to the particles that distribute the fibers uniformly and prevent the fiber contact. Additionally, the interfacial bond between the fiber and the matrix was sufficient and strong. Consequently, the cracks are considered to propagate mainly throughout the matrix, as shown in Fig. 14(b).

In order to quantitatively analyze these fracture morphologies, the fracture surface was classified into three parts: the interface between fiber and matrix (f-m), matrix (m-m), and fiber (f-f). Each area fraction on the fracture surface was calculated, and the relationship between the area fraction and the transverse tensile strength was investigated. The results are summarized in Table 5. The area fraction of the interface, matrix, and fiber are expressed as A_{fm} , A_{mm} , and A_{ff} , respectively. From Table 5, the following results were obtained: (1) A_{ff} is nearly constant at the same temperature, (2) A_{fm} increased as V_f increased, (3) A_{fm} was larger than A_{mm} in the particle-free composites, but smaller in the hybrid composites, and (4) A_{mm} of the hybrid composite was larger than that of the particle-free composite with the same V_f . These results indicate that A_{fm} and A_{mm} have an appreciable influence on the transverse strength. The relationship between transverse tensile strength σ_{cT} (MPa), A_{fm} (%), A_{mm} (%), and A_{ff} (%) was calculated by multiple regression analysis. As a result, the following eq. (3) was obtained at room temperature:

$$\sigma_{cT} = 4.41A_{mm} + 1.85A_{fm} - 6.28A_{ff} \quad (3)$$

The multiple correlation coefficient (R) of eq. (3) is 0.978, signifying substantial correlation. The relationship between each fraction can be expressed as

$$A_{ff} = 100 - (A_{mm} + A_{fm}) \quad (4)$$

When eq. (4) is substituted in eq. (3), the following eq. (5) is obtained:

$$\sigma_{cT} = 10.7A_{mm} + 8.13A_{fm} - 628 \quad (5)$$

σ_{cT} of the composite 50-10 calculated by eq. (5) is 291 MPa, and it is in good agreement with the experimental tensile strength of 288 MPa.

Likewise, the following eq. (6) was obtained at 623 K:

$$\sigma_{cT} = 1.43A_{mm} + 0.54A_{fm} - 0.53A_{ff} \quad (6)$$

R of eq. (6) is 0.978, signifying substantial correlation. When eq. (4) is substituted in eq. (6), the following equation

is obtained:

$$\sigma_{cT} = 1.96A_{mm} + 1.07A_{fm} - 53 \quad (7)$$

σ_{cT} of the composite 50-10 calculated by eq. (5) is 104 MPa and is in good agreement with the experimental tensile strength of 110 MPa.

These results indicate that the transverse tensile strength increases as the area fraction of the matrix occupied on the fracture surface increases. That is, fiber-distribution which makes the cracks propagate throughout the matrix is effective in improving the transverse tensile strength.

4. Conclusions

The preform of continuous alumina fibers with fine alumina particles attached to the fiber surfaces was infiltrated with molten AC1B aluminum alloy by squeeze casting, and hybrid composites were fabricated. Their microstructure and tensile properties in the range from room temperature to high temperature were compared with those of particle-free composites, and the effects of the particle-dispersion were investigated. The following results were obtained.

- (1) In the particle-free composite, many fibers contacted each other, and the fibers were distributed non-uniformly. In the hybrid composite, in which the particles were introduced, the fibers were distributed uniformly because the particles prevented a direct contact between neighboring fibers.
- (2) The tensile strength and 0.2% proof stress, and elastic modulus in the longitudinal direction increased as V_f increased, and these values decreased little up to 623 K. The tensile strength of the hybrid composite was larger than that of the particle-free composite at every temperature. This is because the uniform distribution of the fibers and the reduction in the fiber-to-fiber contact reduce the stress concentration at contact points, and stress transmission between the fiber and the matrix becomes easy.
- (3) The tensile strength and the 0.2% proof stress in the transverse direction decreased as V_f increased. These were improved by the introduction of the particles, leading to the fact that the strength of the hybrid composite was close to that of the unreinforced alloy even at high-temperature. This is because the uniform distribution of the fibers reduces the fracture initiated at the points of fiber contact, and the strong fiber-matrix bond promotes the fracture throughout the matrix.

REFERENCES

- 1) H. R. Shetty and T. W. Chou: Metall. Trans. A **16A** (1985) 853–864.
- 2) H. Mabuchi and Y. Nakayama: J. JILM **38** (1988) 608–613.
- 3) Y. Kagawa and H. C. Bae: J. Japan Inst. Metals **53** (1989) 339–348.
- 4) K. Matsuura, C. Meikon and N. Matsuda: J. Japan Inst. Metals **54** (1990) 671–678.
- 5) Y. Ikeuchi, K. Iga and T. Hanabusa: J. Soc. Mater. Sci. JP. **43** (1994) 792–798.
- 6) K. Yamada, S. Sekiguchi, T. Hanamura and T. Inoue: J. Japan Inst. Metals **59** (1995) 1108–1117.
- 7) H. N. Liu and K. Ogi: Mater. Sci. Technol. **15** (1999) 821–826.
- 8) T. Sato, J. Furukawa and K. Shimakage: J. JILM **40** (1990) 822–827.
- 9) K. T. V. Rao, S. C. Siu and R. O. Ritchie: Metall. Mater. Trans. A **24A**

- (1993) 721–734.
- 10) G. Chadwick: *ECCM* **4** (1990) 3–13.
- 11) E. Nakata: in *New Materials Handbook*, (Maruzen, Tokyo, 1988) pp. 129–140.
- 12) Y. Nishida: *Materia Japan* **36** (1997) 40–46.
- 13) S. De Bondt, L. Froyen and A. Deruyttere: *Mater. Sci. Eng.* **A135** (1991) 29–32.
- 14) S. Towata, H. Ikuno and S. Yamada: *Trans. JIM* **29** (1988) 314–321.
- 15) H. Ikuno, S. Towata, S. Yamada and Y. Awano: *J. JILM* **44** (1994) 379–384.
- 16) K. Matsunaga, S. Ochiai, K. Osamura, Y. Waku and T. Yamamura: *J. JILM* **43** (1993) 219–224.
- 17) Sumitomo Chemical Co.: Data sheet for Altex (1995).
- 18) Y. Abe and M. Nakatani: *J. JILM* **36** (1986) 372–380.
- 19) M. Morita: in *New Materials Handbook*, (Maruzen, Tokyo, 1988) pp. 99–113.
- 20) A. Kelly and N. H. Macmillan: *Strong Solids*, 3rd (Oxford Univ. Press, 1986) pp. 240–325.