Cyclic Deformation and Development of Dislocation Structures in a Centrifugally Cast Al-Al₃Ti of Functionally Graded Material

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An Al-Al₃Ti composite has been prepared to contain a large volume fraction of D₀₂₂ ordered Al₃Ti phase on one side and a smaller volume fraction on the other side by a functional grading process. The cyclic stress-strain curves obtained by strain control tests are compared with the available data on Al to delineate the effects of the Al₃Ti plates distributed with concentration gradient. Examination of the crack nucleation and growth by a replica method has revealed that introduction of the non-deformable Al₃Ti accelerates crack nucleation but concurrently hinders the spontaneous propagation in the cyclic tests. Observation of the final structure by HVEM has further revealed that presence of the non-deformable Al₃Ti induces non-uniform distribution of dislocations near the Al₃Ti plate, causing build-up of stress inhomogeneity, especially on the Al₃Ti rich side. On the basis of these observations, utility of the FGM specimens is discussed with a particular interest in the lifetime prediction under cyclic loading.

(Received May 10, 2000; Accepted September 4, 2000)

Keywords: functionally graded material (FGM), fatigue, Al-Al₃Ti interface, high voltage electron microscope, compatibility, crack nucleation, crack propagation, D₀₂₂

1. Introduction

It was found in the previous studies¹,²) that the intermetallic compounds of Al₃Ti and Al₃Ni embedded in Al matrix deformed by dislocation glide at room temperature under high stress level. Such a condition was attained either by compressing a thin foil specimen with the thickness less than the size of Al₃Ti and Al₃Ni or by hammering a bulk specimen in case of Al-Al₃Ti. In the usual tensile or compression tests at room temperature, however, these compounds are not deformable when they are embedded in Al.

Meanwhile, Fukui and Watanabe³) have proposed a centrifugal method to produce a functionally graded material (FGM), by which an Al-Al₃Ti composite can be fabricated with concentration gradient of Al₃Ti. Since Al₃Ti has higher melting point⁴) and the larger specific weight than Al, the centrifugal force acting on the mixture of molten Al and Al₃Ti pushes the Al₃Ti towards the outer side of a cylinder in the centrifugal casting.⁵⁻⁷) The densities are 3.17 Mg/m³ for Al₃Ti and 2.37 Mg/m³ for molten Al, respectively. Watanabe et al.⁸) have studied the effect of dispersion hardening on wearing with particular emphasis on the orientation asymmetry of Al₃Ti in Al matrix and shown that wear resistance increases notably at outer surfaces with strong dependence on the platelet orientation.

For application of the newly developed material, it is important to testify the resistance against fracture by cyclic deformation before the practical use. The crack initiation and propagation, which determine the fatigue life, are affected strongly by introduction of the second phase particles. Crack propagation through the matrix is resisted by strong particles, thus lifetime to fracture can be extended by introduction of the hard entities.⁹⁻¹¹) But as mentioned above, the second phase particles Al₃Ti are not sufficiently ductile. Then, it is likely that the incompatibility between the matrix and the particles may accelerate crack nucleation at the Al-Al₃Ti interface. Contribution of the second phase particles on the fatigue behavior is not simple and can cause counter effects in the crack nucleation and propagation.

Furthermore, the FGM specimen contain the second phase particles distributed non-uniformly but with gradient across the specimen thickness. In this study, cyclic stress-strain behaviors will be examined with particular interest in the development of microscopic structures in the Al-Al₃Ti FGM. The major subjects focussed in this study are to clarify how the cyclic deformation proceeds in the immediate vicinity of the Al₃Ti plates and how the fracture occurs through the Al₃Ti distributed uniformly but with concentration gradient across the specimen.

2. Experimental

The specimens used in the present study were prepared by a centrifugal casting method described in the previous reports.⁵⁻⁷) Specimens for cyclic deformation tests were cut by a spark machine from the as-cast Al-Al₃Ti tube of 9 mm in thickness, 90 mm in outside diameter and 90 mm in length to have the final shape as shown in Fig. 1. The Al₃Ti plates were distributed homogeneously with the flat planes of [001]₀₀₂ respectively parallel to the cylinder face as determined in the previous studies⁵⁻⁷) and shown later by an optical micrograph taken after fracture. Since the Al matrix is considered to solidify around the Al₃Ti plates of a high melting point of 1615 K⁴) in the repeated centrifugal casting, the Al matrix grains tend to have a [001] axis nearly perpendicular to the surface. Volume fraction of the Al₃Ti plates varied from 0.4 to 0.2 across 6 mm

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towards the inner portion of the heavy wall cylinder.

The cyclic deformation tests were conducted in air at room temperature by plastic strain control for three kinds of strain amplitudes, $1 \times 10^{-4}$, $1 \times 10^{-3}$ and $5 \times 10^{-3}$ at 0.1 hertz. The stress-strain diagram was measured on a $x$-$y$ recorder by using a strain gage pasted on the specimen surface. In the course of cyclic deformation, stress-strain hysteresis loops were observed on an oscilloscope to stop the deformation just before the specimen was about to fracture. Some of the specimens were tested without using a strain gage in order to take surface replica at several points judged critical in the behavior of cyclic stress-strain curves. In this case, the plastic strain amplitude was adjusted on the hysteresis curve by measuring the gap in the total strain between push and pull motion at zero stress level.

Replica films were pasted on both sides of a specimen, $\text{Al}_3\text{Ti}$ rich on one side and Al matrix rich on the other side, at the applied tensile load equal to 20% of the maximum value by interrupting the cyclic loading. Specimens for the electron microscopic observations were prepared from the portions near the inner and outer surfaces of the deformed crystals. The thin foils sliced and mechanically thinned down to 150 $\mu$m were, further, jet polished down to several $\mu$m in the thinnest portion and finally polished to perforation. In the course of electrolytic polishing special care was exercised to control the polishing speed as same as possible in the Al matrix and $\text{Al}_3\text{Ti}$. The best condition was 60 V at 218 K for jet polishing with a mixture of perchloric acid : methanol = 1 : 9 and 25 V at 203 K for the final polishing with the same solution. However, the polishing rate was always somewhat faster for the Al matrix. Thus, the $\text{Al}_3\text{Ti}$ plates remained thicker than Al matrix and some of thick Al–$\text{Al}_3\text{Ti}$ interfaces were just sufficient for transmission by HVEM. The electron microscopic observations were made by using a Hitachi-1250S HVEM equipped with a double tilt specimen holder at the accelerating voltage of 1000 keV.

3. Experimental Results

3.1 Cyclic hardening and fracture

Stress amplitudes obtained in the plastic strain control tests are plotted against the cumulative strain defined by $\varepsilon_{cum} = 4N \varepsilon_p$ in Fig. 2, where $N$ is the number of cycles and $\varepsilon_p$ is the plastic strain amplitude. Results for 3003 Al are also shown in Fig. 2 for comparison. Hereafter the 3003 Al will be called simply as Al. The final cycle is denoted by a $\times$ mark at which clear asymmetry was noted on the cyclic stress-strain curve. The specimens, thus, cycled to the $\times$ marks were actually just about to fracture. The circles marked on the curves denote the specimens examined by HVEM after the cyclic deformation.

The stress amplitude increased rather quickly in the early stage of cyclic deformation for the large strain amplitude tests of $\varepsilon_p = 1 \times 10^{-3}$ and $5 \times 10^{-3}$, and the stress level as well as the cyclic hardening was quite similar to Al shown by broken lines. This fact is noteworthy in the presence of substantial volume fraction of $\text{Al}_3\text{Ti}$ on one side of the FGM specimen. On the other hand, the cyclic hardening saturates at a low stress level in the small strain amplitude tests of FGM specimen. The negative effect of $\text{Al}_3\text{Ti}$ is, thus, more pronounced in the smaller strain amplitude tests. It is found in Fig. 2 that the FGM specimen fractures without exhibiting the second stage hardening such as found in Al. Origin of the different behaviors found in the Al–$\text{Al}_3\text{Ti}$ FGM specimen may be attributed to the incompatibility induced by the non-deformable $\text{Al}_3\text{Ti}$ particles and the resultant local hardening in the cyclic tests. This point will be discussed later in terms of microstructure examined by a replica method and electron microscopy.

The cumulative strain for fracture was also affected by the presence of non-deformable AlTi plates distributed with gradient across the specimen as shown in Fig. 2. The shortening effect is pronounced in the small strain amplitude tests.
as mentioned above. The Al tested at $\varepsilon_{pl} = 1 \times 10^{-4}$ and $1 \times 10^{-3}$ did not fracture to $\varepsilon_{cum} = 80$ below which the cyclic deformation was stopped because of fracture for all the tests in the present study. In contrast, it is known from the previous study\cite{13} that Al exhibits secondary hardening to the strain beyond $\varepsilon_{cum} = 100$ without fracture at the strain amplitudes of $\varepsilon_{pl} = 1 \times 10^{-4}$ and $1 \times 10^{-3}$. At a larger strain amplitude of $\varepsilon_{pl} = 5 \times 10^{-3}$, the present FGM specimen exhibited comparable or even better ductility than Al as shown in Fig. 2. Ductility of the present specimen was also superior to some other Al based alloys such as Al–Si–Mg. The latter alloy is known to fracture in the range $\varepsilon_{cum} = 1–8$ at the strain amplitude of $\varepsilon_{pl} = 1 \times 10^{-4}$,\cite{13}

3.2 Surface morphologies

Figure 3 shows variation of the surface morphology taken by using replica at various stages of cyclic loading. The replica films were pasted on both sides of the specimen; the Al$_3$Ti being rich on one side and less on the other side as shown in the two series of figures (a) to (c) and (d) to (f). In these tests, the strain amplitude was controlled at $\varepsilon_{pl} = 1 \times 10^{-3}$ on an oscilloscope as described in Section 2. A tensile load of 20% of the maximum value was applied upon pasting replica films so that crack opening could be detected easier. The viewing areas shown in Fig. 3 were chosen on the replica films after finishing the fatigue test. Thus, the crack propagation could be traced back on the replica films taken successively at both sides of the surfaces in the course

Fig. 3 Variation of surface morphology recorded by a replica method. The cyclic tests were made at a strain amplitude of $\varepsilon_{pl} = 1 \times 10^{-3}$. (a)–(c) and (d)–(f) show the Al$_3$Ti rich and Al matrix rich sides, respectively. Number in the picture indicates the cycle at which replica was taken by interrupting the cyclic loading.
of cyclic tests. Comparing the Al<sub>3</sub>Ti plates in (a)–(c) and (d)–(f), it is apparent that the centrifugal casting is effective in increasing the volume fraction of Al<sub>3</sub>Ti on the outer side of the FGM tube as reported in the earlier studies<sup>5,6,8</sup> and shown schematically in Fig. 1.

Cracks were noted in the early stage of deformation at the cycle of N = 25 on the Al<sub>3</sub>Ti rich surface as shown by arrows in (a). Then, the small cracks induced in the vicinity or inside of the Al<sub>3</sub>Ti plates were connected and expanded gradually to fracture with increase in the loading cycles as shown in (b) and (c). It is also a typical structure that cracks were introduced preferentially in the region where Al<sub>3</sub>Ti are clustered as shown by arrows in (a)–(c). On the Al matrix rich side, crack nucleation was less frequent and a large crack was introduced suddenly in the later stage of cyclic deformation, penetrating through the Al matrix as shown in (f).

Nucleation of a crack must be accelerated around the non-deformable Al<sub>3</sub>Ti and the origin may be ascribed to the inhomogeneous deformation near the Al<sub>3</sub>Ti as shown later by electron microscopy. Since generation of a crack reduces the effective cross section, the resultant increase in the stress will be balanced by the work hardening in the Al matrix. Such an increase in the stress without fracture is readily guaranteed in the cyclic hardening curve of Al shown in Fig. 2. Thus, the crack is imagined to develop gradually from the Al<sub>3</sub>Ti rich side towards the Al matrix rich side, leading the specimen to fracture entirely. This is the consequence of good coherency between the Al<sub>3</sub>Ti plate and Al matrix as demonstrated in the previous work, examining the orientation relationship between the Al<sub>3</sub>Ti and Al and deformability of Al<sub>3</sub>Ti.<sup>1,7</sup>

### 3.3 Electron microscopy

In the present specimen Al<sub>3</sub>Ti plates embedded in Al matrix are distributed with their flat surfaces nearly parallel to the surface of the centrifugally cast cylinder as shown in Fig. 3. Thin foil specimens for electron microscopy were prepared by slicing the fractured specimen along the flat surfaces from both Al<sub>3</sub>Ti rich and Al matrix rich sides, corresponding to the outside and inside of the cylinder as shown Fig. 1. A thin foil specimen prepared from the outer portion of a specimen after deformation to fracture at a plastic strain amplitude of ε<sub>p</sub> = 1 × 10<sup>−4</sup> is shown in Fig. 4. Dislocation structures are examined in the vicinity of an Al<sub>3</sub>Ti plate, near the straight and curved edges of the plate in (a) and (b). In contrast to the Al matrix, no dislocations can be seen in the Al<sub>3</sub>Ti plate. Thus, Al<sub>3</sub>Ti plates are non-deformable to fracture in the cyclic tests of the present strain amplitude. This observation differs from the plastic deformation observed in the Al<sub>3</sub>Ti after the simple compression test of a thin FGM specimen.<sup>13</sup>

It is also interesting to note that the dislocation structure differs considerably in (a) and (b). Since Al matrix has solidified coherent to the Al<sub>3</sub>Ti plate in Fig. 4, orientation of the Al is naturally same for (a) and (b). The structural difference in (a) and (b) is, then, attributed to termination of the Al<sub>3</sub>Ti plate at bottom in (b). Non-uniform deformation in Al will, thus, accommodate the incompatibility caused by the presence of non-deformable Al<sub>3</sub>Ti. But at the same time, accumulation of the non-uniformly distributed dislocations in Al matrix will accelerate crack nucleation in the vicinity of the Al<sub>3</sub>Ti plate. The crack will eventually grow into Al matrix. However, presence of the Al<sub>3</sub>Ti plates prevents spontaneous propagation as evidenced in Figs. 3(a)–(c) where evolution of small cracks was noted without substantial growth on the Al<sub>3</sub>Ti rich side.

Another example of dislocation structure after cyclic deformation to fracture is shown in Fig. 5. This figure shows the Al<sub>3</sub>Ti rich side of a specimen deformed to fracture at a plastic strain amplitude of ε<sub>p</sub> = 1 × 10<sup>−3</sup>. The dislocations seen in Fig. 5(a) are of Al and those piled-up against the Al–Al<sub>3</sub>Ti interface, which runs vertical in the figure and inclined with respect to the foil of ~1 μm in thickness. When g vector is adjusted to operate in Al<sub>3</sub>Ti only, the dislocation contrasts have disappeared completely as shown in Fig. 5(b). Dislocations introduced in the Al matrix are invisible but strain contrasts of straight lines generated in Al<sub>3</sub>Ti by the piled-up dislocations are visible at the interface. The line contrasts are identified to be parallel to the intersection between the slip plane in the Al matrix and the Al–Al<sub>3</sub>Ti interface. Thus, glide dislocations are stopped by the non-deformable Al<sub>3</sub>Ti and distributed homogeneously in fine scale at the interface. The Al<sub>3</sub>Ti phase without overlapped Al matrix can be seen at the bottom left corner in Fig. 5(b), being completely free from dislocations.

Figure 6 shows the Al matrix rich side of a specimen de-
formed to fracture at plastic strain amplitude of $1 \times 10^{-3}$. The cell structure varied from one grain to another as shown in (a) and (b). The cells were elongated along a (001) direction in (a). When the similar cells are viewed along different type of (001) direction in the other grain, they are found to have a similar size of round shape as shown in (b). These structures are seemingly the same as that of Al deformed at the same strain amplitude of $\varepsilon_{pl} = 1 \times 10^{-3}$ to the cumulative strain of $\varepsilon_{cum} = 1.0$. However, the Al specimen fractures after notable increase in the stress level following the plateau on the $\sigma$-$\varepsilon_{cum}$ curve as shown in Fig. 2. Presence of Al-Ti apparently accelerates fracture in the tests at $\varepsilon_{pl} = 10^{-4}$ and $10^{-3}$ and the origin must be attributed to the easier nucleation of cracks in the Al-Ti rich side as observed in Fig. 3.

In contrast, the present FGM specimen containing Al-Ti showed quite different structure after the cyclic deformation to fracture at $\varepsilon_{pl} = 5 \times 10^{-3}$ as shown in Fig. 7. Rapid change in the Bragg condition near the Al-Ti plate indicates large elastic distortion sustained by the presence of non-deformable Al-Ti. On the other hand, microstructure at the Al rich side was quite similar to that reported for Al in the previous study, consisting of well-developed patches. Accordingly, at a large strain amplitude of $\varepsilon_{pl} = 5 \times 10^{-3}$, the $\sigma$-$\varepsilon_{cum}$ curve of the FGM specimen exhibited a similar behavior as Al and even the better ductility than Al as shown in Fig. 2.

4. Discussion

4.1 On the deformability of Al-Ti in the Al–Al-Ti FGM

As it is well known, the Al-Ti takes a D0$_{22}$ structure with tetragonality along one of the cube axes and usual slip hardly takes place at an ambient temperature. Following the notation used by Yamaguchi et al., a particular (110) direction in the Al matrix is parallel to a (110)$_{D0_{22}}$ direction in the Al-Ti as reported previously. Furthermore, one of the (111) plane is nearly parallel to the (111)$_{D0_{22}}$ plane of the Al-Ti, while the other (101)$_{D0_{22}}$ directions and (111/2)$_{D0_{22}}$ planes are inclined by several degrees with respect to (101) and (111) in the Al matrix. Therefore good orientation relationship between the Al-Ti and Al matrix was found for one type of the shear system only as given below:

$\langle 1\bar{1}1 \rangle_{\text{fcc}}//\langle 1\bar{1}1/2 \rangle_{D0_{22}} ; [1\bar{1}0]_{\text{fcc}}//[1\bar{1}0]_{D0_{22}} \quad (1)$

Because of the tetragonality of the D0$_{22}$, the Al-Ti takes a plate shape as seen in Fig. 4 by optical microscopy. The c axis of the D0$_{22}$ structure was determined to be nearly perpendic-
cular to the wide plane of the platelet.\(^7\)

Despite of the crystallographic continuity, in the usual compression and in the present cyclic deformation tests, glide dislocations were not found in the Al\(_x\)Ti. On the other hand, in other mechanical tests made by hammering or by compression of a foil specimen thinner than the Al\(_x\)Ti size, both twinning partial and superpartial dislocations were found in the Al\(_x\)Ti. The shear system, satisfying the orientation relationship given in eq. (1) was shown to have induced the observed dislocations in the Al\(_x\)Ti plates.\(^1\) Absence of mobile dislocations in the Al\(_x\)Ti after the large cumulative strain in the present cyclic tests implies that stress level at the Al–Al\(_x\)Ti interface was not sufficiently high as that in the hammering or in the thin foil compression tests.\(^1,7\)

In general, dislocations accumulated at the Al–Al\(_x\)Ti interfaces were distributed homogeneously in fine scale after the cyclic tests. Nevertheless, cracks were found to nucleate at the interfaces as shown in Fig. 3. Judging from the inhomogeneous dislocation structure developed near the corner of an Al\(_x\)Ti plate shown in Fig. 4(b), a large stress inhomogeneity is expected to develop at the curved portion of an Al\(_x\)Ti plate, resulting in acceleration of crack nucleation at the interface. It means that the cracks nucleated at special portion of the interface will grow along the interface and continue to propagate into the neighboring Al matrix. But the growth is stopped by the presence of other Al\(_x\)Ti plates.

Another possibility is the initiation of some shear systems in the Al\(_x\)Ti plate, possibly due to development of the internal stress at the Al–Al\(_x\)Ti interface. As pointed out above, limited number of shear systems are operative in the Al\(_x\)Ti due to the poor symmetry in the D\(_{030}\) structure. The limitation must be strict when a particular shear system is activated by the local stress developed in the surrounding Al which has a specific orientation relationship to the Al\(_x\)Ti as given by eq. (1). Then, once a particular shear system starts to operate inside the plate, the tensile stress acting on the shear plane, possibly amplified at the interface will lead to either local deformation or spontaneous fracture in the Al\(_x\)Ti. This may be the reason why the Al\(_x\)Ti plates have never contained twinning dislocations nor superpartial dislocations in the electron microscopic observation, despite that some of the Al\(_x\)Ti plates were cracked as shown in Fig. 3.

### 4.2 Effects of Al\(_x\)Ti on the cyclic deformation of the Al–Al\(_x\)Ti FGM

As pointed out above, crack nucleation is expected to take place preferentially along or inside the Al\(_x\)Ti in absence of usual slip in the Al\(_x\)Ti. However, presence of the Al\(_x\)Ti concurrently suppresses spontaneous propagation of the cracks as it is manifested in the large number of cycles required for a crack to propagate through the FGM specimen. A similar role of intermetallic compounds to retard crack growth has been suggested in the earlier reports of cyclic deformation.\(^9,11,16,17\) At this point it must be recalled that Al exhibits good ductility with large work hardening as shown in Fig. 2. Growth of some of the small cracks was actually terminated inside the Al matrix as shown in Fig. 3. Therefore, the origin of the good lifetime observed under the condition of plateau stress, say up to \(\varepsilon_{\text{cum}} = 80\) in the \(\varepsilon_p = 1 \times 10^{-4}\) tests in Fig. 2 must be attributed to the combined effects of the Al\(_x\)Ti with high strength and the Al matrix with good workability to prevent spontaneous propagation of cracks.

By comparing the dislocation structures in Figs. 4(a), 5(a) and 7, it is noted that development of the cell structure near the Al\(_x\)Ti plate is less distinct in the larger strain amplitude tests. The increase in the strain amplitude caused decrease in the cumulative strain to fracture but the change was more drastic in Al, thus overwhelming the apparent brittleness of the FGM specimen as shown in Fig. 2. These observations are consistent with the fact that the cracks nucleated near the Al\(_x\)Ti rich region hardly propagate through the FGM specimen as shown in Fig. 3. Such a property that the Al\(_x\)Ti plates embedded in Al can sustain a large internal stress at the interface must be important in practical use of the functionally graded material.

Finally, it is gratifying that the present FGM specimen has exhibited good ductility, despite that cracks were formed frequently in the immediate vicinity of the Al\(_x\)Ti plates in the early stage of deformation. Thus, the cyclic deformation increased number of small cracks on the Al\(_x\)Ti rich side without leading the FGM specimen to fracture, which is essentially controlled by the development of a cell structure in the Al rich environment.
region in the present FGM specimen. With this knowledge of characteristic behaviors in mind, it is suggested that careful examination of crack nucleation on the Al$_3$Ti rich face may lend a tool to predict life-time of the FGM specimen having a good wear resistance on the Al$_3$Ti rich side.

5. Conclusions

FGM specimens of Al–Al$_3$Ti prepared by a centrifugal casting method have been examined by cyclic deformation in a strain control mode. Surface observations were made by a replica method at several stages of cyclic deformation. The specimens deformed to fracture were examined at both Al$_3$Ti rich and Al matrix rich sides by HVEM. From the results of mechanical tests coupled with the microscopic observations of the dislocation structures, following conclusions were obtained in describing the effects of Al$_3$Ti distributed with gradient on the cyclic deformation behaviors.

(1) The secondary hardening, which characterizes the cyclic deformation of Al at a small strain amplitude, does not appear in the FGM specimen for the tests made at the strain amplitudes of $1 \times 10^{-4}$ and $1 \times 10^{-3}$.

(2) Crack nucleation occurs preferentially along or through the Al$_3$Ti on the Al$_3$Ti rich side but they hardly propagate through the matrix during cyclic tests.

(3) The acceleration of crack nucleation at the Al$_3$Ti is attributed to the development of inhomogeneous stress accumulated near the non-deformable Al$_3$Ti.

(4) Life-time of the FGM specimen was shorter than that of 3003 Al at smaller strain amplitudes of $1 \times 10^{-4}$ and $1 \times 10^{-3}$ but was similar or even longer than Al at $5 \times 10^{-3}$, suggesting that the life-time is closely related to the development of deformation structure in the Al matrix.

(5) It is learned from the above results that lifetime of the FGM specimens with high wear resistance should be predicted by examining the cracks nucleated on the Al$_3$Ti rich surface.

Acknowledgement

Part of this study, which was conducted at the Shinshu University, was supported by the Grant-in-Aid for COE Research (IOCE2003) by the Ministry of Education, Science, Sports and Culture of Japan.

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