Dynamic Recrystallization Based on Twinning in Coarse-Grained Mg

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Abstract. Mechanisms of dynamic recrystallization (DRX) associated with twinning are proposed and discussed on the results of microstructure evolution during warm and hot deformation of a coarse-grained Mg. "Twin" DRX is composed of following elementary processes: i.e. formation of twin boundaries and their transformation into random boundaries followed by their local boundary migration taking place during deformation. The formation of DRX nuclei can result from mutual intersection of deformation twins belonging to primary twinning systems and subdividing of primary twin lamellas by secondary twins or subboundaries. Twin boundaries eventually convert into random boundaries in high strain due to their interaction with mobile dislocations. The new grains formed in twin lamellas are in non-equilibrium states and exhibit almost rectangular shape. A limited boundary migration takes place locally, resulting in development of equiaxed grains in high strain. Under high temperature deformation, when mechanical twinning is suppressed, DRX can also develop through growth of twins presented in an initial structure.

Introduction

Twinning is known [1-4] to play an important role in plastic deformation of Mg and Mg alloys in a wide temperature range. In hexagonal closed packed (hcp) materials, twinning is especially significant because it occurs in grains resisting basal slip and reorients the lattice to facilitate such slip [1,4]. There has been numerous works on the role of twinning in structure evolution in Mg alloys to date [2,3]. Recent investigations have shown that twinning can initiate dynamic recrystallization (DRX) in early stages of plastic deformation [2-4]. While many experimental evidences of dynamic nucleation caused by twinning have been presented and discussed for magnesiums [3-4], there exists, however, a lack of knowledge on the mechanisms, by which recrystallized grains were evolved.

The microstructure evolution during plastic deformation of a pure Mg was investigated in the previous works [2]. Remarkable fractions of twinned grains were developed in recrystallized regions and their contribution to microstructural development was studied. However "twin DRX" was considered in combination with other deformation mechanisms taking place in Mg. Accordingly, structural changes associated by twinning have been poorly examined under such complex considerations. In this paper, recrystallization behavior of pure magnesium causing by twinning [2] is summarized and mechanisms of nucleation and growth of recrystallized grains are reexamined in more detail. We pay special attention to the effects of deformation temperature on this process.

Experimental

A commercial as-cast Mg with an average grain size of 2 mm was used for this investigation. Cylindrical samples, 10 mm in diameter and 12 mm in height, were machined from the ingot that was preliminarily homogenized at 450°C for 6 h. An uniaxial compression test was carried out on Instron type universal dynamometer at temperatures of 150, 300 and 450°C, and at a strain rate of 10^{-3} s^{-1}. Metallographic analysis of the deformed samples was performed on a Neophot-32 light microscope by using ordinary and polarized lights. For qualitative determination of misorientations, color etching was used [5]. Deformation relief and dislocation substructures were observed using a JSM-840 SEM and a JEOL-2000EX TEM.

Results and discussion

Temperature dependence of twinning and DRX. Twins belonging to one system were developed in original grain interiors of the annealed Mg due to any mechanical damage introduced by the preparation of samples (Fig. 1a). The average thickness of twin lamellas was about 8 μm and their volume fraction did not exceed 15%. Under early deformation at 150 and 300°C, various kinds of deformation twins are evolved in the Mg, as shown in Figs. 1b and c. At ε=5-15%, the volume fraction of twins increased up to around 60%. The number of mutually nonparallel twins was 5-6 in some grains; this suggests the operation of two kinds of twinning systems: the main one is $\{10\bar{1}2\}\langle\bar{1}011\rangle$ and the secondary one is $\{10\bar{1}1\}\langle10\bar{1}2\rangle$ [6]. As a result, any difference between initial and deformation twins at low strains could not be discerned.

A main peculiarity of twinning at 150°C is the frequent formation of very fine twins with a transverse size of around 1.5 – 2 μm. Such twins are developed in twin colonies belonging to various systems (Fig.1b). Narrow twin bundles within such bands were often developed laterally on each other. As temperature increases, the twin bands and twin themselves become thicker. Twins develop to form thick lenses with curved boundaries scarcely at 150°C and frequently at 300°C (Fig.1c).

Further deformation to ε=40-70% at 150 and 300°C results in colonies of recrystallized grains formed in the regions of twins (Fig.2). The new grain size evolved near twins can be determinated by the width of twin lamellas. This suggests that grain formation in Mg can be closely related to the operation of specific structural mechanisms caused by twinning. It is interesting to note in Figs.2b and c that similar grain structures are formed in the place of twins at 150 and 300 °C. This suggests that the evolution of recrystallized grains in twin lamellas can occur by similar ways.

During high-temperature deformation, in contrast, deformation twinning scarcely takes place. A main structural change in early stages of deformation at 450°C, as shown in Fig. 3a, is only an extensive growth of initial twin lamellas in Fig. 1a. The latter leads to the appearance of large twin formation with stepped boundaries. At ε=5-10%, the transverse size of twins was sometimes around 50 μm and their volume fraction reached 30-40%. At moderate to high strains (ε=40-70%), coarse layered grains with an irregular shape are frequently evolved in the places of initial twins (Fig. 3b). These grains are surrounded by corrugated high angle boundaries and subdivided by low angle boundaries in transverse direction. Let us provide a detail consideration of the mechanisms of new grain formation caused by twinning in coarse-grained Mg.

Development of DRX associated with mechanical twinning. The "twin" DRX can be composed of following three elementary processes: i.e. formation of twin boundaries and their transformation into random boundaries, followed by their local boundary migration. During deformation, twin-twin and/or dislocation-twin interaction may cause dynamic nucleation of new grains. New grains can be nucleated by three kinds of mechanisms, which will be discussed below in detail.

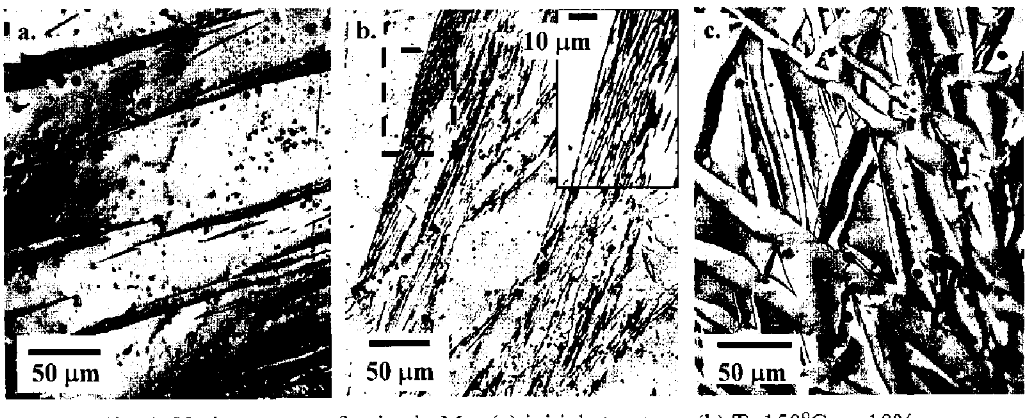

Fig. 1 Various types of twins in Mg: (a) initial structure; (b) T=150°C, ε=10%; (c) T=300°C, ε=7%.

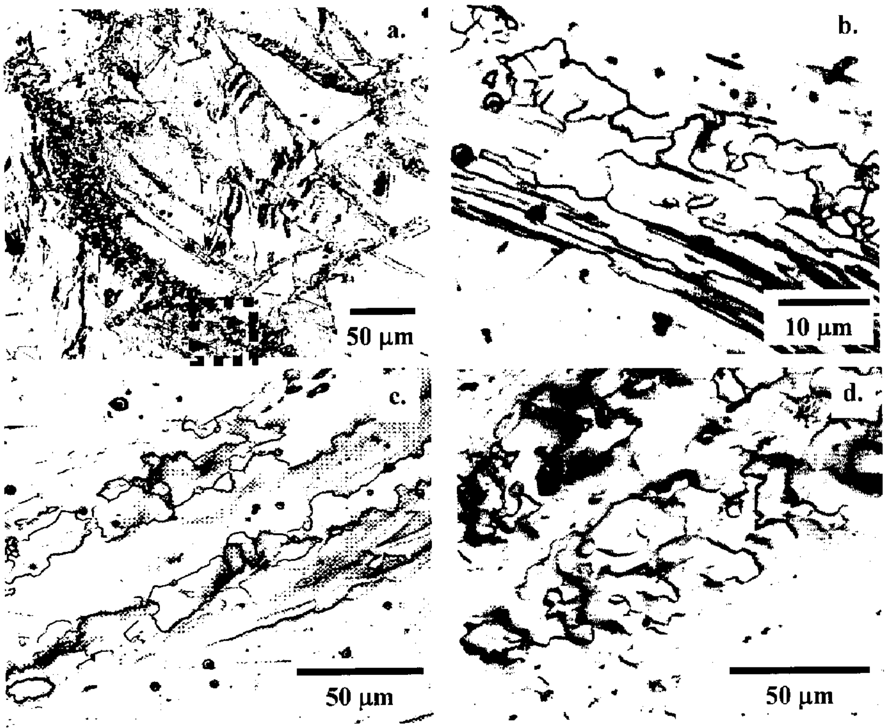

Fig.2 Microstructure formed in the place of twins in Mg: (a), (b) T=150°C, ε=40%, (c) T=300°C, ε=40%, (d) T=300°C, ε=70%.

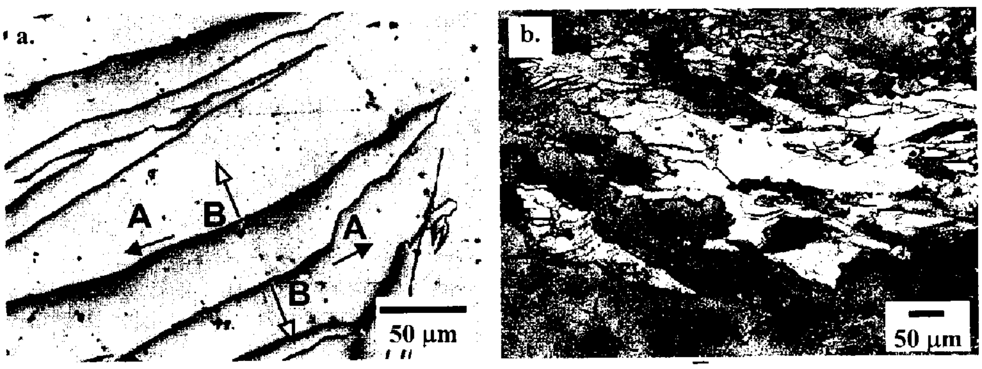

Fig.3 Microstructure formed in the place of initial twins in Mg at T=450°C: (a) ε=10%; (b) ε=50%; an explanation of the markers in the Fig. 3a is presented in the text.

(I) The first step of nucleation can be a mutual crossing of deformation twins belonging to the $\{10\bar{1}2\}\langle\bar{1}011\rangle$ system and/or intersection of initial twins by deformation twins (Fig.4). At 150°C, mutual intersection of twins belonging to various twin bands frequently takes place (Figs.4a, b). Such twin-twin interactions result in the formation of tetragonal regions surrounded by two pairs of twin boundaries (Figs. 4a, c). The misorientation between the matrix and nuclei may be caused by specific features of lattice rotation under development of multiple twinning [7]. This is confirmed by the microstructure in polarized light after color etching (Fig. 4d). This type of nucleation occurs frequently at 150 and 300°C. The development of DRX grains in the places of twin-twin intersection at low to intermediate temperatures has been also reported for an Mg-Al-Zn alloy [4].

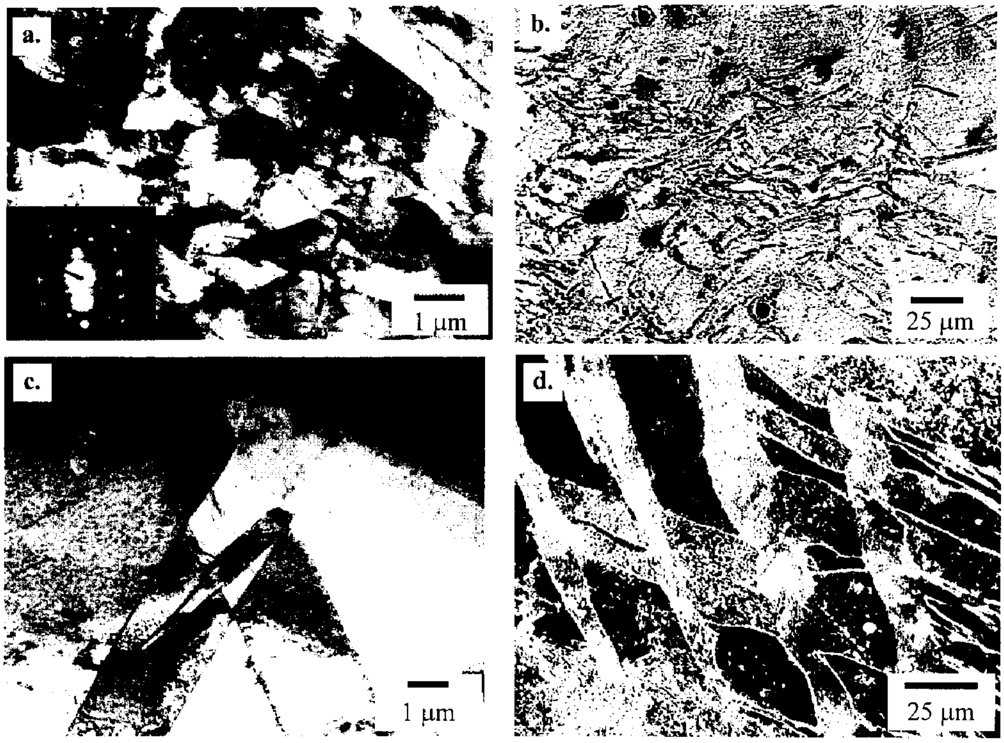

Fig.4 Typical TEM (a), (c) and optical microstructures (b), (d) showing the twin-twin intersection in coarse-grained Mg: (a), (b) T=150°C, ε=15%; (c), (d) T=300°C, ε=15%.

(II) The second step is related to the secondary twinning on $\{10\bar{1}1\}$, which is resulted by the specific features of twinning in Mg [6]. Secondary twins are formed as thin lamellas inside primary coarse twins (Fig. 5). The nuclei of recrystallized grains are bounded by $\{10\bar{1}2\}$ primary and $\{10\bar{1}1\}$ secondary twin boundaries [6,7]. It is important to note that, twinning on the $\{10\bar{1}1\}$ system is more difficult than that on the $\{10\bar{1}2\}$ system and requires of a high elastic stress concentration in Mg [8]. Accordingly, secondary twinning can be developed predominantly in the vicinities of the initial grain boundaries, where the highest stress concentrations can be reached [9].

(III) The third step is caused by formation of low-angle boundaries across twin lamellas (Fig. 6). Such subboundaries are developed sometimes at right angles to twin boundaries. Nuclei evolved are separated by a pair of low-angle boundaries from the twin body and by pair of twin boundaries from the matrix. This nucleation mechanism was found to be predominantly operative at strains of 15-30% at T=300°C [2]. The present microstructural results in Fig. 6 exhibit the good correspondence to those reported for an Mg alloy at elevated temperatures [4].

At the second stage of "twin" DRX, various types of nuclei transform into recrystallized grains with irregular shape. The main process at this stage is a transformation of special type boundaries into random HABs due to deviation of misorientation angle from coincident site lattice relationship on a value over than the standard Brandon criteria [10]. Such transformation can be caused by interaction between moving lattice dislocations and twin boundaries, followed by accumulation of some misfit dislocations at the twin boundaries. As it is seen in Fig. 7 change in dislocation slip direction takes place when dislocations intersect a twin boundary. As a result, orientation misfit dislocations [11] are formed, as presented schematically in Fig. 7. The Burgers vector b_3 of misfit dislocations compensates the change in the Burgers vector b_1 of the mobile lattice dislocations when they intersect the twin boundary ($b_1 \rightarrow b_2$).

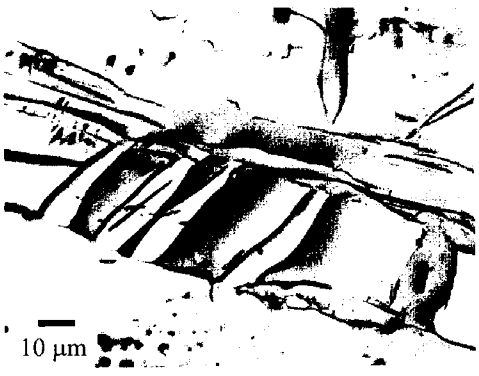

Fig. 5 Optical microstructure of Mg showing the development of secondary twinning: T=300°C, ε=15%.

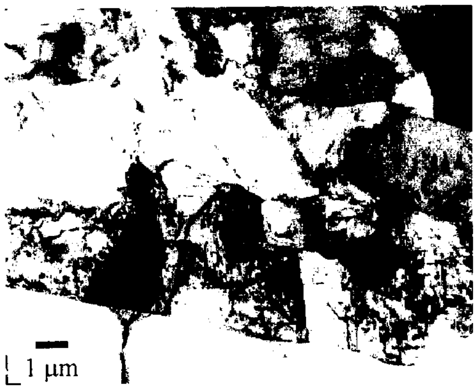

Fig.6 TEM structure showing subdivision of twin by transverse low-angle boundaries: T=300°C, ε=15%.

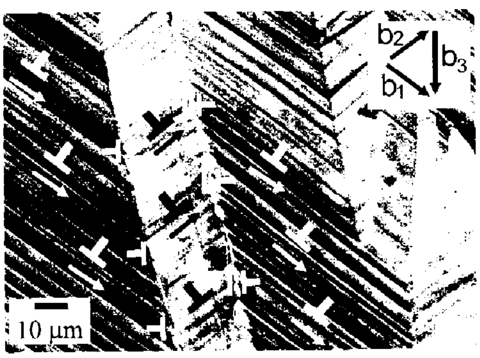

Fig.7 Deformation relief of Mg at T=300°C, ε=15% and scheme showing formation of orientation misfit dislocations in twin boundaries.

Accumulation of such misfit dislocations, their rearrangement and/or their absorption in twin boundaries at elevated temperature can lead to an additional misorientation change and deviation of twin boundary from a special misorientation. This makes the original twin boundaries "non-transparent" for mobile lattice dislocations. The latters are accumulated in near-boundary regions. This can provide the elastic stress concentration in such regions and activation of non-basal slip systems [2,9]. The dislocations of non-basal systems can be also generated in near-boundary regions and absorbed by former twin boundaries [2,4]. This in turn leads to additional changes in the misorientation angles of twin boundaries and accelerates processes of transformation of these boundaries into random grain boundaries. The transverse low-angle boundaries evolved by the mechanism III are also gradually transformed into HABs by accumulation of lattice dislocations during deformation [2,3,9].

Thus at the first and second stages of "twin" DRX, the formation of grains of an irregular tetragonal shape takes place (Figs. 2b, c). It is known [10] that the tetragonal shape of grains is unstable because of the presence of triple and quadruple junctions with angles close to 90°. A limited grain boundary migration takes place during following deformation and results in a structure with a more stable configuration of boundaries and the triple junction angles close to 120°. It is clearly seen in Figs. 2b and c that former twin boundaries start to migrate locally at moderate strains. As a result, equiaxed recrystallized grains with an average size of 25-30% larger than that of rectangular nuclei are formed in place of the former twins (Fig.2d). It is worth noting that the intense migration of the former twin boundaries starts only after their transformation into random grain boundaries. It is known that in comparison with twin boundaries, random HABs have a high ability of migration due to a larger grain boundary energy [10]. This explains the fact that the formation of an equilibrium grain structure requires achievement of relatively high strains (Fig.2d).

DRX mechanism caused by evolution of initial twins in high-temperature region. The initial twins may have a great importance for new grain evolution extending the operation of twin DRX up to high temperatures. Such twins are most likely introduced into the initial coarse-grained Mg during previous machining operations. However, from point of view of the structural mechanisms operating during subsequent deformation, the origin of these twins seems to be not important. Their effect on microstructural evolution may be similar to that of some annealing twins formed in cubic metals with low stacking fault energy [12].

The mechanism of extensive growth of twin lamellas during hot plastic deformation was proposed and discussed in [12]. According to [12], high-temperature plastic deformation is accompanied by dissociation of twin boundaries in Mg into longitudinal coherent (**B**) and transverse incoherent (**A**) segments (Fig.3a). Growth of twins can occur as a result of deformation-induced migration of incoherent segments **A** in longitudinal direction [12]. This leads to a simultaneous migration of low-mobile coherent segments **B** in transverse direction. The driving force for this process can be the difference between the energy stored in the deformed matrix and that of twinned regions with lower dislocation density. Under such twin boundary migration, lattice dislocations are swept and trapped by moving incoherent segments. Accordingly the energy of these segments increases and this leads to additional acceleration of twin growth. Note here, that increasing deformation temperature provides a high mobility of incoherent boundary segments [10]. Due to that, operation of such mechanism becomes remarkable only in high-temperature region [2].

In initial twins, the nucleation of recrystallized grains is possible only in accordance with the mechanism III. The formation of low-angle grain boundaries across twin lamellas restrains their growth (Fig. 3b). In this case, coarse grains with corrugated boundaries can be formed in place of twins due to concurrent transformation of twin and low-angle boundaries into HABs. Note that this "twin" DRX mechanism can be operated with significantly lower rate than that in the case of deformation twins. As a result, the appearance of recrystallized grains at 450°C is observed at higher strains than that at 150 or 300°C [2].

Summary

Analysis of resent experimental data shows that the DRX mechanism related to twinning includes the three sequential stages. At the first stage, nucleation occurs by either intersection of various systems of twins or rearrangement of lattice dislocations within the twin lamellas. At the second stage, twin boundaries are changed into random high-angle boundaries due to formation of orientation misfit dislocations. As a result, the nuclei are transformed into recrystallized grains, which have non-equilibrium shape close to tetragonal one. At the third stage, migration of their boundaries begins to occur. Note that such multistage character of the microstructure evolution is a common feature of the "twin" DRX operating in a wide range of temperatures.

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