Deformation twin formed by self-thickening, cross-slip mechanism in nanocrystalline Ni

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We report the observation of a deformation twin formed by a recently proposed self-thickening, cross-slip twinning mechanism. This observation verifies one more twinning mechanism, in addition to those reported before, in nanocrystalline face-centered-cubic metals. In this mechanism, once the first Shockley partial is emitted from a grain boundary, and cross slips onto another slip plane, a deformation twin could nucleate and grow in both the primary and cross-slip planes without requiring the nucleation of additional Shockley partials from the grain boundary. © *2008 American Institute of Physics*. [DOI: 10.1063/1.2949685]

In face-centered-cubic (fcc) metals, twinning is usually not observed in coarse-grained structures with medium to high stacking-fault (SF) energy at normal rates of deformation and room temperature.[1–6] However, twinning has been observed in fcc metals under the following conditions: low-temperature deformation,[6] high-strain-rate deformation,[6,7] severe plastic (large strain) deformation, and deformation of nanocrystalline materials.[2,3,8–12] Contrary to coarse-grained metals, which become more difficult to deform by twinning with decreasing grain size, nanocrystalline materials may deform via partial dislocations originated from grain boundaries as elucidated by molecular dynamics simulation studies[4,5,13,14] and experimental observations.[1,3,8,15,16]

Deformation twins usually form in nanocrystalline fcc metals via mechanisms different from those proposed for coarse-grained materials.[17] Several twinning mechanisms have been proposed for and observed in nanocrystalline fcc metals, including the coincidental overlapping of wide stacking-fault ribbons inside nanosized grains,[15] partial emission from grain boundaries,[1,3,4] grain boundary splitting and migration,[11,18] and random activation of partials from grain boundaries.[17] Recently, a self-thickening, cross-slip mechanism was proposed and observed in coarse-grained Cu–Ge alloys.[19] A salient feature of this mechanism is that once the first Shockley partial is emitted from a grain boundary, and cross slips onto another {111} plane, a deformation twin could nucleate and grow in both the primary and cross-slip planes without requiring the nucleation of additional Shockley partials from the grain boundary. However, high stress is required for this mechanism to operate. This mechanism is very appealing to nanocrystalline fcc metals because they generally lack the continuous partial dislocation sources that are required for some twinning mechanisms of coarse-grained metals.[20–25] In addition, nanocrystalline materials have much higher strength than their coarse-grained counterparts[26–28] and, therefore, deform under very high stress. This high stress helps overcome the extra energy requirement for the dislocation reactions involved in the recently proposed cross-slip mechanism.[19] Therefore, it is of scientific interest to investigate if this self-thickening, cross-slip twinning mechanism indeed operates in nanocrystalline fcc metals.

We choose electrodeposited nanocrystalline Ni film from Goodfellow Inc. as the material of study because it has been reported to deform by twinning extensively.[10] The as-purchased Ni film has an average grain size of 25 nm and a thickness of 100 μm. Growth twins were observed in the as-deposited nanocrystalline Ni but were rare. The nanocrystalline film was cryorolled to a thickness reduction of 45% at liquid nitrogen temperature to produce deformation twins. High-resolution electron microscopy (HREM) was used to study the deformation twins in the cryorolled nanocrystalline Ni.

Figure 1(a) is a transmission electron microscope image showing a nanocrystalline grain labeled A in electrodeposited Ni after cryorolling at liquid nitrogen temperature. The grain has twin plates in their interiors, with twin boundaries indicated by white arrows. Inside the white frame in Fig. 1(a) is a pair of cross-slip twins, and its clearer image is shown in Fig. 1(b), which is an enlarged HREM image of the framed area. Two intersecting twins are visible, labeled as T1 and T2, respectively. T1 is observed to extend all the way from the grain boundary on the right side. For the sake of discussion, we assume T1 is on the primary slip plane (111).

Interestingly, Fig. 1(b) shows that one end of T2 intersects with T1 in the area marked by *B*, while the other end terminates inside the grain (marked as *D*). The twin lamellar T2 was formed by Shockley partials on another slip plane $(11\bar{1})$. Figure 1(b) raises a question on how the twin lamellar T2 was formed. In other words, where did Shockley partials that formed T2 come from? They could not have come from the end that terminated inside the grain (end *D*), because it is extremely unlikely energetically or nearly impossible for partials to nucleate inside a grain area without crystalline defects. One theoretically possible, but statistically improbable, mechanism is that many full dislocations dissociated into stacking-fault ribbons on successive slip planes and they stacked together to form the twin lamellar T2. The probability that a thick twin like T2 forms under this mechanism is

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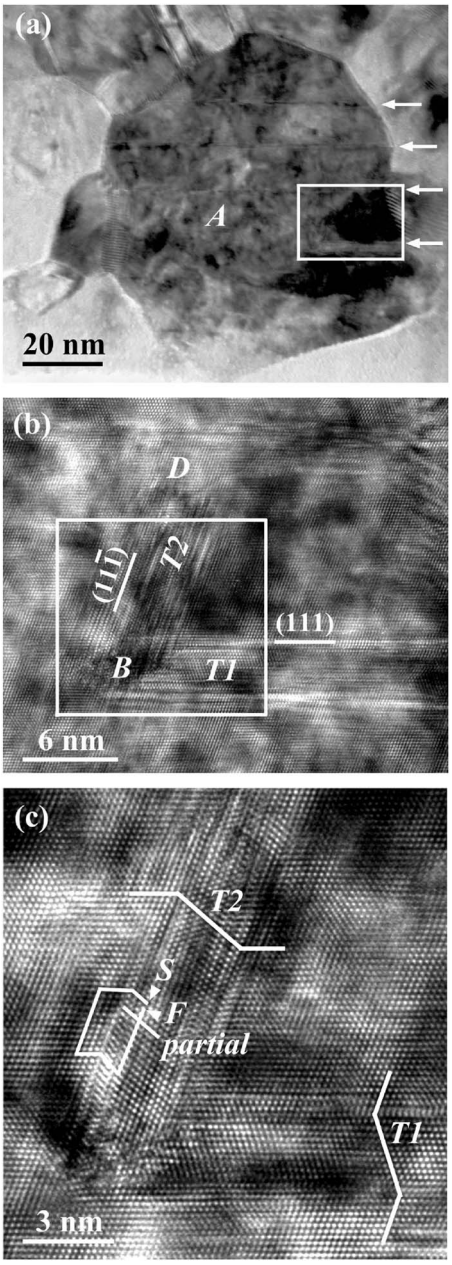

FIG. 1. A HREM image showing (a) a grain labeled A that contains a cross-slip twin in the white frame, (b) an enlarged image of the area in the white frame in (a), and (c) an enlarged image of the area in the white frame in (b).

extremely low. Only two-layer twins were previously reported in a nanocrystalline metal.[3]

The most likely location for the emission of Shockley partials on successive $(11\bar{1})$ slip planes to form the twin lamellar T2 is the area marked *B* where T1 and T2 meet. To have a closer look at this area, the image framed by the white frame in Fig. 1(b) is enlarged and shown in Fig. 1(c). A Burgers circuit starting at *S* and ending at *F* was drawn in Fig. 1(c). The magnitude and direction of the Burgers vector needed to connect the end to the beginning of the circuit give the Burgers vectors of the type $\mathbf{b}_p^{\text{Shockley}}=(1/6)[\bar{2}1\bar{1}]$. This indicates a Shockley partial dislocation. This Shockley partial in T2 is clearly emitted from the T1 and T2 intersection, providing direct evidence that the partials forming T2 are indeed from the intersection area of the two twins. Also shown in Fig. 1(c) is a severe lattice distortion at the area *B*

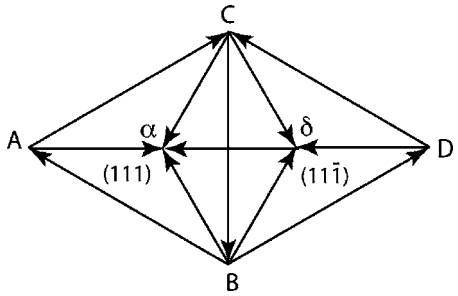

FIG. 2. Partial dislocation reactions and their reactions in the primary plane (111) and cross-slip plane $(11\bar{1})$.

and a lack of clear intersection boundary between T1 and T2, possibly caused by a high dislocation stress in the area. The above experimental observations are consistent with the self-thickening, cross-slip twinning mechanism that we proposed recently.[19]

The formation of T1 and T2 twins shown in Fig. 1 can be understood with the help of the self-thickening, cross-slip twinning mechanism. In the following discussion, Fig. 2 is used to illustrate the dislocation reactions in this mechanism. First, a Shockley partial was emitted from the grain boundary location and slipped to the left on the primary slip plane (111), producing a stacking fault on the (111) plane. We assume this partial as $\mathbf{C}\boldsymbol{\alpha}$ (see Fig. 2). This partial encounters an obstacle within the grain at the area marked *B* in Fig. 1(b). Upon encountering the obstacle, the partial $\mathbf{C}\boldsymbol{\alpha}$ splits into a perfect dislocation (**CB**), a twinning partial ($\mathbf{B}\boldsymbol{\delta}$) in the cross-slip plane $(11\bar{1})$, and a stair-rod dislocation ($\boldsymbol{\delta\alpha}$), i.e.,

$$\mathbf{C}\boldsymbol{\alpha} \rightarrow \mathbf{CB} + \mathbf{B}\boldsymbol{\delta} + \boldsymbol{\delta\alpha} \tag{1}$$

as shown in Fig. 2. The partial $\mathbf{B}\boldsymbol{\delta}$ slipped on the cross-slip plane $(11\bar{1})$ to produce a stacking fault on the $(11\bar{1})$ plane.

The perfect dislocation **CB** can cross slip on the $(11\bar{1})$ plane onto the adjacent (111) plane. After the cross slip, the perfect dislocation splits into two partials according to Eq. (2),

$$\mathbf{CB} \rightarrow \mathbf{C}\boldsymbol{\alpha} + \boldsymbol{\alpha}\mathbf{B}, \tag{2}$$

where the leading partial $\mathbf{C}\boldsymbol{\alpha}$ repeats the process in Eq. (1), emitting another $\mathbf{B}\boldsymbol{\delta}$ partial to slip on the $(11\bar{1})$, and the other partial $\boldsymbol{\alpha}\mathbf{B}$ slips back to the grain boundary on the (111) plane. This process nucleates the twinning nuclei on both the primary slip plane (111) and the cross-slip plane $(11\bar{1})$.

The dislocation reactions in Eqs. (1) and (2) provide a full reaction cycle that can be repeated to simultaneously thicken the twins on the successive primary slip planes and cross-slip planes without the need of the nucleation or emission of additional Shockley partial dislocations from the grain boundary. The repetition of the above process will produce both twin T1 and twin T2.

Note that a stair-rod dislocation $\boldsymbol{\delta\alpha}$ is left at each (111) and $(11\bar{1})$ slip plane intersection during each dislocation reaction cycle. This will produce severe lattice distortion, which is evident in Fig. 1(c). Also, the twin lamellar T2 terminated inside the grain because the $\mathbf{B}\boldsymbol{\delta}$ partials most likely did not receive a large enough resolved shear stress to drive them to slip further.

We would also like to point out that in our previous *in situ* study to observe the self-thickening, cross-slip twinning

mechanism in coarse-grained Cu–Ge alloy, HREM observation of lattice image was not possible. The HREM observations in the current study provide additional evidence to support the self-thickening, cross-slip twinning mechanism.

In summary, we have revealed that a recently proposed self-thickening, cross-slip twinning mechanism was indeed active in nanocrystalline Ni. The cross-slip twin, related lattice defects and distortions, and thickening of twins revealed by the HREM image are in full agreement with this mechanism.

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