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# Dislocation arrangements and crystallographic characterization of deformation band in fatigued copper single crystal

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ABSTRACT

This letter reveals the dislocation arrangements and crystallographic characterization of deformation bands (denoted DBII) in a copper single crystal fatigued at a high strain amplitude $\gamma_{pl} = 8 \times 10^{-3}$. The results show that the surface deformation morphology of the crystal displays the following features. (1) Primary slip bands (SBs) were formed after $2 \times 10^4$ cycles and these carried a relatively homogeneous and small plastic strain. (2) Secondary slip bands did not operate during cyclic deformation. (3) Deformation bands (DBs) with a width of 50 μm were homogeneously distributed over the whole surface of the crystal and were perpendicular to the SBs. (4) Dislocation patterns within the SBs often consisted of irregular structures, which did not show a persistent feature. The results indicate that these SBs are not typical persistent slip bands (PSBs). (5) Within the DBII, the microstructure can be classified into two types. One type consists of regular 100% ladder-like parallel PSBs. The other type is full of dislocation walls parallel to DB direction, which have not been reported previously. By crystallographic analysis of the DBII, it is shown that the habit plane of the DBII should correspond to the $(\bar{1}01)$ plane. Based on the observations above, it is suggested that the formation of DBII should be attributed to the local regularization of dislocation walls within primary slip bands.

## § 1. INTRODUCTION

In addition to the well known persistent slip bands (PSBs) in copper, the formation of deformation bands (DBs) seems to be another important feature induced by cyclic deformation. Much work has been done on the occurrence of DBs in fcc single crystals (Reid 1973, Crocker and Abell 1976, Flewitt and Crocker 1976). More recently, DBs have been often observed in fatigued copper single crystals. In general, two types of DB namely DBI and DBII, have been identified (Saletore and Taggart 1978, Gong *et al.* 1995, 1997, Li *et al.* 1998a, 1999a,b,c, 2000) in fatigued copper single crystals. Generally, the DBI is approximately parallel to the primary slip plane, while the DBII, namely kink bands, make a certain angle with the primary slip plane. It was observed that DBs are easily nucleated in fatigued copper single crystals with double- and multi-slip orientations. On the other hand, for copper single crystals with a single-slip orientation, the formation of DBs is confined to a

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very high strain amplitude (Gostelow 1971, Mughrabi 1978, Li *et al.* 1998b). In contrast with the PSBs, investigations on the DBs are relatively rare, so that the formation mechanism and microstructure of DBs are less understood. Owing to the difficulties in transmission electron microscopy (TEM) observations, it is not very clear whether the DBs have some habit planes with a low index, and what kinds of microstructure occur within DBs.

Recently, an electron channelling contrast (ECC) technique in conventional scanning electron microscopy (SEM) has received much interest for the study of dislocation arrangements in deformed materials (Zauter *et al.* 1992). This technique not only can provide a real and wide view of dislocation substructures conveniently, but also makes it possible to establish the relationship between dislocation arrangements and some special sites, such as PSBs, grain boundaries (Zhang and Wang 1998, 2000, Zhang *et al.* 1999) and the head of the crack (Wilkinson *et al.* 1997, Jia *et al.* 1999). By this special technique, Gong *et al* (1997) found that the microstructure within DBs in a fatigued copper [001] crystal consists of a typical labyrinth structure. Later, Li *et al.* (1998b) observed that the parallel dislocation walls are predominant microstructures within DBs in a fatigued copper $[\bar{1}35]$ crystal. It seems that the crystallographic orientation of crystals affects the formation mechanism of DBs and the associated slip systems as well as the resulting microstructures within DBs. To reveal further the formation mechanism and microstructures of DBs, a copper single crystal with a single-slip orientation was investigated in the present work in order to avoid the operation of secondary slip.

## § 2. Experimental procedure

Copper single crystals were grown from oxygen-free high-conductivity copper of 99.999% purity by the Bridgman method. A crystal with a typical single-slip orientation was selected to make a fatigue specimen with dimensions of $8\,\mathrm{mm} \times 3\,\mathrm{mm} \times 60\,\mathrm{mm}$ and a gauge section of $6\,\mathrm{mm} \times 3\,\mathrm{mm} \times 16\,\mathrm{mm}$ by electrospark cutting. The orientation of the crystal was determined by the electron back-scatter diffraction (EBSD) technique in a Cambridge S360 scanning electron microscope and is as follows:

$$\mathbf{G} = \begin{bmatrix} 0.5929 & -0.2148 & 0.7761 \\ 0.0478 & 0.5054 & -0.4316 \\ -0.02988 & 0.8362 & 0.4598 \end{bmatrix} = \begin{bmatrix} 3 & -5 & 5 \\ 5 & 12 & -3 \\ -2 & 20 & 3 \end{bmatrix}.$$

Here, $[-0.2148, 0.5045, 0.8362]$ or $[\bar{5}\,12\,20]$ correspond to the stress axis orientation of the specimen. The Schmid factors $\Omega$ of primary and secondary slip systems were calculated and are listed in table 1. It can be seen that the ratio values $Q = \Omega_S/\Omega_p$ of the Schmid factors of all the secondary slip systems with respect to that of the primary slip system $((111)[\bar{1}01])$ are smaller than 0.9. Clearly, the crystal is oriented for typical single slip in terms of the definition by Cheng and Laird (1981).

Table 1. Schmid factors of primary and secondary slip systems of the crystal.

| | B4 $(111)[\bar{1}01]$ | A3$(111)[101]$ | B5 $(111)[\bar{1}10]$ | C1 $(11\bar{1})[011]$ |
|---|---|---|---|---|
| $\Omega$ | 0.484 | 0.398 | 0.329 | 0.290 |
| $Q = \Omega_s/\Omega_p$ | 1 | 0.822 | 0.680 | 0.599 |

Therefore, it should be very difficult to activate secondary slip even at high strain amplitude.

Before the fatigue test, the specimen was electropolished to produce a strain-free surface for microscopy observations. Cyclic push–pull tests were performed on a Shimadzu servohydraulic testing machine under a constant plastic resolved shear strain amplitude $\gamma_{pl}$ of $8.0 \times 10^{-3}$ for $2 \times 10^4$ cycles at room temperature in air. A triangular wave with a frequency range 0.1–0.5 Hz was used. After cyclic deformation, the surface of the specimen was observed to examine the deformation morphology. Then, the crystal was polished again to remove the slip traces. The surface microstructures were observed by the SEM ECC technique in a Cambridge S360 scanning electron microscope to establish the relationship between dislocation arrangements and DBs. Here, an inverted imaging mode was adopted so that the bright areas in the ECC micrograph represent dislocation-poor regions, while the dark areas represent dislocation-dense regions. As a result, the observed dislocation patterns will be in accord with the transmission electron micrograph under bright imaging conditions.

## § 3. Results and discussion

### 3.1. *Surface deformation morphology*

Figure 1 shows the deformation morphology on the $[5\bar{3}3]$ surface of the crystal fatigued at a plastic strain amplitude $\gamma_{pl} = 8 \times 10^{-3}$ for $2 \times 10^4$ cycles. Two striking features on the surface can be clearly seen. One is that only primary slip bands (SBs) formed, as shown in the figure. Meanwhile, the SBs are separated by regular macroscopic DBs, also shown in the figure, and are distributed discontinuously. Since the ratio values $Q$ of the crystal are obviously smaller than 0.90, as listed in table 1, it should be very difficult for secondary slip systems to operate during cyclic deforma-

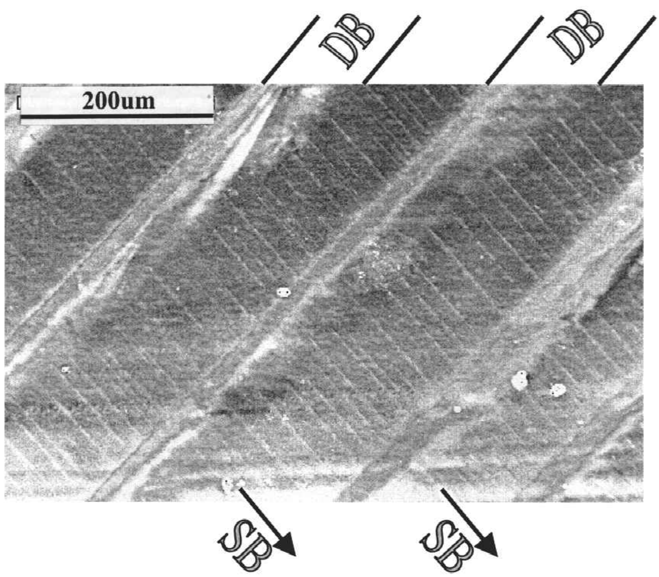


Figure 1. Deformation morphology on the $[5\bar{3}3]$ surface of a copper $[\bar{5}\,12\,20]$ single crystal fatigued at $\gamma_{pl} = 8 \times 10^{-3}$ for $2 \times 10^4$ cycles, showing SBs and DBs.

tion (Cheng and Laird 1981). In particular, the formation of DBs is a specially important deformation mode when the copper single crystal is cycled at a high strain amplitude (Gostelow 1971, Saletore and Taggart 1978, Mughrabi 1978, Gong *et al.* 1995). Those DBs spread over the whole surface and have a spacing of about 160–180 μm. The SBs between the DBs are perpendicular to but cannot penetrate into the DB. Those SBs are relatively thinner and are distributed homogeneously between the DBs. In contrast, DBs are very coarse and have an average width of 50–60 μm, which is about 15–20 times thicker than SBs. By comparing the present observation with the previous definition of DBs, it is suggested that the present DB belongs to the kink band (or DBII) in type. Meanwhile, we define the white region without SBs as DB, which is contrary to that used by Gong *et al.* (1997). The reason is that the white bands bulge or are extrusive from the crystal surface, indicating that they may carry more plastic strain than the SBs.

### 3.2. *Dislocation observations by the SEM ECC technique*

Figure 2 (*a*) shows the microstructure of the DB and the adjacent regions on the $[5\bar{3}3]$ surface of the crystal observed by the SEM ECC technique. It can be seen that the DB, which has a clear boundary with the other regions, interrupted the white SBs. The microstructure between the DBs consists of SBs and irregular dislocation loop patches, as shown in figures 2 (*b*) and (*c*). The boundary in figure 2 (*b*) is not distinct, indicating that the DB is propagating towards the adjacent region. However, as shown in figure 2 (*c*), several irregular dislocation bands terminate at the DB, whose boundary is more distinct. Those irregular dislocation bands should correspond to the SBs as indicated in figure 1 and figure 2 (*a*). Obviously, there are no regular ladder-like dislocation walls within the SBs. Therefore, the surface SBs between the DBs are not real ladder-like PSBs and do not show any persistent feature. Further observations show that the microstructure within the DB is composed of 100% ladder-like PSBs in parallel as shown in figure 2 (*d*). Hereafter, we define the DB containing the microstructures above as a developing DB. The developing DBs might be propagating towards each side, as shown in figure 3 (*a*). In the centre of the DB, the microstructure is composed of 100% ladder-like PSBs in parallel, similar to that shown in figure 2. Near the boundary of the DB, it seems that there exist transition regions, in which the ladder-like dislocation walls can extend over a certain distance from the DB and become irregular in comparison with those in the centre of the DB. It is suggested that the DB is not steady but evolves and propagates step by step. However, the microstructure within SBs is quite irregular, indicating that the SBs might be a hard phase in comparison with the ladder-like PSBs within the DB. Meanwhile, some SBs terminate at the boundary of the DB and show a one-to-one relationship with some PSBs within the DB. It is suggested that the ladder-like PSBs within the DB evolve from the SBs.

In addition to the microstructure described above, the dislocation arrangements within a DB on the same plane may display another feature, as in figure 3 (*b*). It can be seen that the microstructure within the DB consists of typical dislocation walls rather than ladder-like PSBs. Between the DB and upper region, there is a clear boundary, which prevents the SBs from passing through. However, between the DB and the lower area, it seems that there still exists a transition region, within which the SBs can penetrate into the DB over a certain distance. Meanwhile, some irregular ladder-like structure seems to be propagating towards the lower area, indicating that the DB can become wider by propagating towards the neighbouring region. In

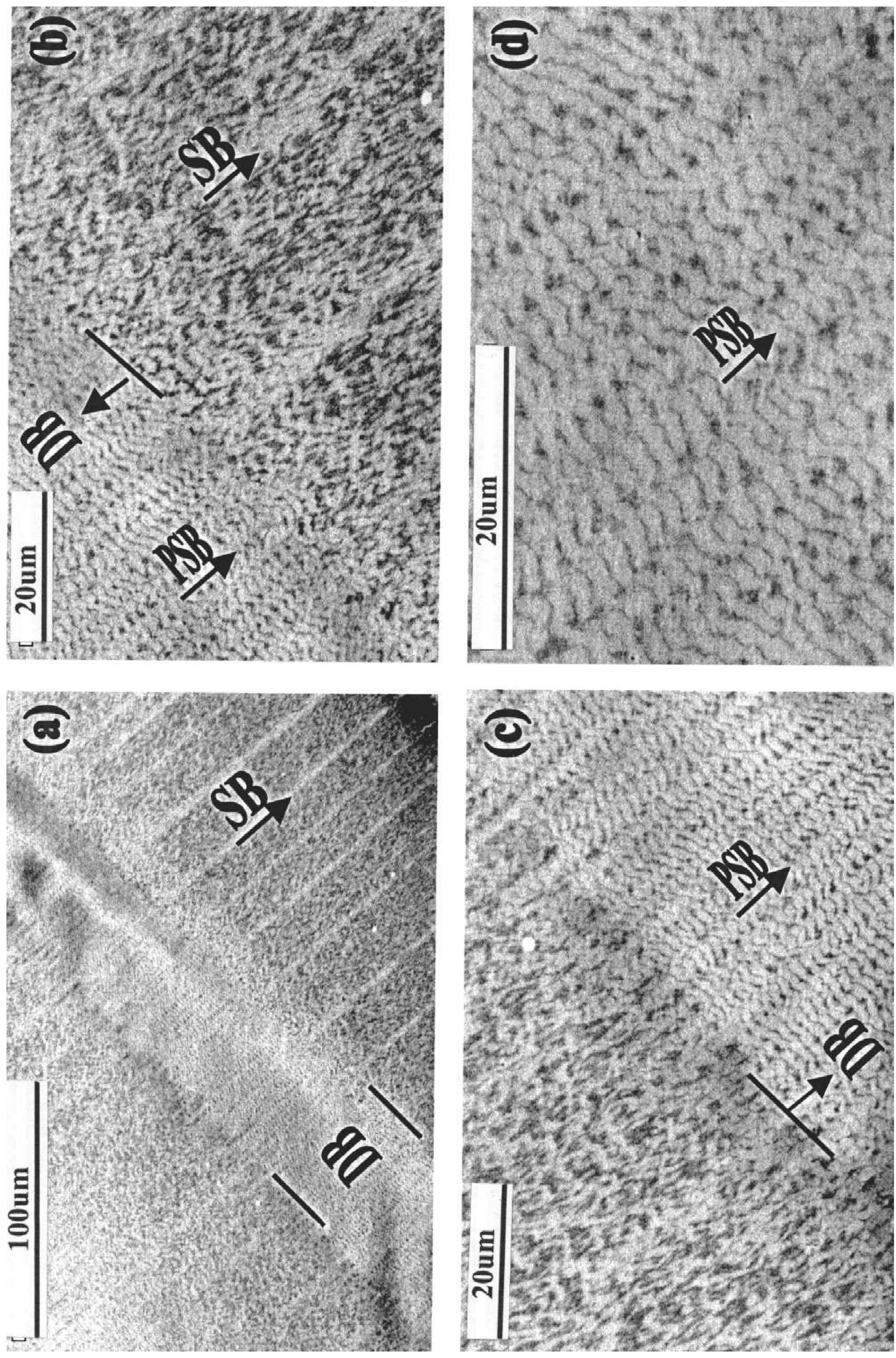


Figure 2. Dislocation arrangements near the developing DB observed on the $[5\bar{3}3]$ surface: (*a*) interaction between PSBs and the DB, showing discontinuous distribution of SBs; (*b*), (*c*) dislocation arrangements near the boundary of the DB; (*d*) dislocation arrangements within the DB.

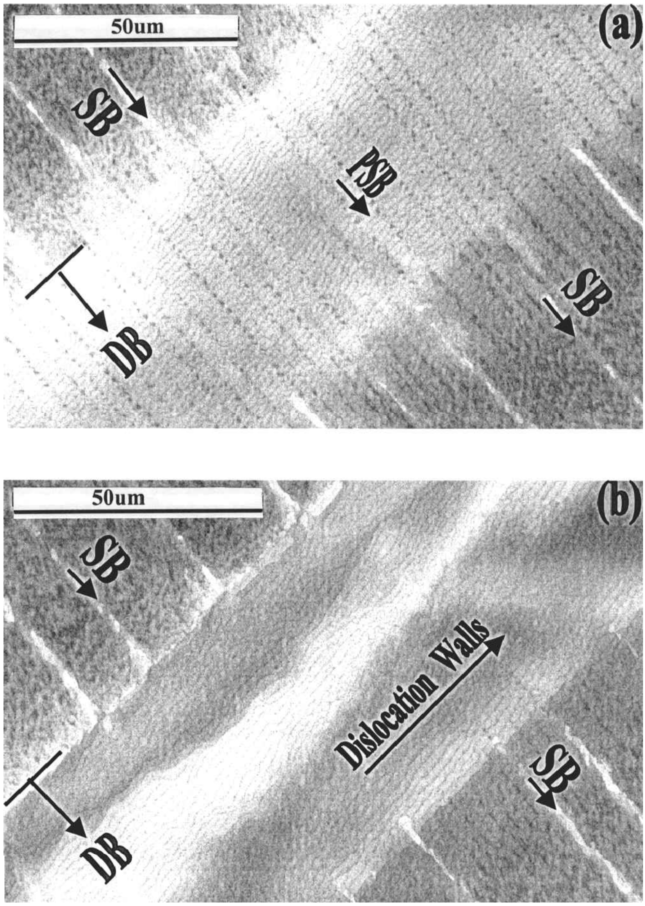


Figure 3. Comparison of microstructure within the developing and well developed DBs on the $[5\bar{3}3]$ surface of the crystal: (*a*) dislocation arrangements within the developing DB; (*b*) dislocation arrangements within the well developed DB.

addition, note that those dislocation walls are not perfectly parallel, which might be associated with the crystal rotation within the DB. We define the DB containing the microstructure above as a well developed DB, which should be steady in energy. From the observations above, it is suggested that the primary SBs did not show the persistent feature. This implies that the two-phase model of PSBs and veins is invalid when the copper single crystal is cycled at the strain amplitude beyond the plateau region of the cyclic stress–strain curve.

### 3.3. *Crystallographic characteristic of deformation bands*

It is well known that there exist two kinds of DB, namely DBI and DBII, in cyclically deformed fcc crystals. In general, DBI is almost parallel to the primary slip plane, and DBII makes an angle with the primary slip bands. However, the formation mechanism and crystallographic characteristics of DBs have not been clearly revealed. Vorren and Ryum (1988) showed that DBII is indexed as parallel to the $(31\bar{4})$ plane for most of orientations in fatigued aluminium single crystals, but away from the $(10\bar{1})$ plane. Saletore and Taggart (1978) concluded that, in a fatigued copper $[\bar{1}22]$ single crystal, the plane of DBII makes an angle of approximately 45° to the loading direction. This means that the DBII is formed on the plane of maximum shear stress and does not correspond to any low-index crystallographic plane. In addition, a dislocation avalanche mode was first proposed to explain the formation of DBs during cyclic deformation (Li *et al.* 1994). A lattice rotation of about 6° between macrobands and the matrix was detected in cyclically deformed aluminium single crystals using time-resolved acoustic microscopy (Zhai *et al.* 1995, 1996). It was suggested that this lattice rotation resulted from the accumulation of the irreversible slip in one direction in PSBs and was responsible for the formation of these macrobands. Recently, Gong *et al* (1995) and Li *et al* (1998a,b, 1999a,b,c, 2000) systematically investigated the formation of DBs in fatigued copper single crystals with different orientations. They found that DBI and DBII developed roughly along primary slip plane $\{111\}$ and the conventional kink plane $\{101\}$ respectively, and the habit planes of DBI and DBII are strictly perpendicular to each other. The formation of DBI and DBII was attributed to the local irreversible rotation of crystal during symmetrical push–pull loading. However, the investigations above did not provide a clear microstructural pattern of DBs, which is of special importance to understanding the formation and development of DBs.

From the above observations within DBs, it can be concluded that the present DB should belong to the DBII in type. Based on the dislocation arrangements within DBII above, we shall propose a new model to describe the crystallographic characteristics of the DBII. First, a three-dimensional crystallographic picture of a ladder-like PSB is illustrated in figure 4 (*a*). The ladder-like structure can be observed on the $(1\bar{2}1)$ plane, while parallel dislocation walls often appear on the primary slip plane (111). Those dislocation walls are perpendicular to the primary slip direction. By observing the dislocation arrangements within and near the DB again, it can be seen that the DB consists of dislocation rungs of ladder-like structure (figure 2 and figure 3 (*a*)). It is indicated that the habit plane of the DB should be consistent with the dislocation rungs of ladder-like PSBs. Consequently, it can be concluded that the habit plane of DBII is strictly along the $(\bar{1}01)$ plane, which is in accordance with the previous observations. The crystallographic relationships among primary slip planes, the developing DBII and dislocation walls are illustrated in figure 4 (*b*). For the well developed DB, it can be seen that there is a transition from the ladder-like structure on the $(1\bar{2}1)$ plane to the parallel dislocation walls along the DB direction (figure 3 (*b*)). The crystallographic relationships between the primary slip plane, the well developed DBII and dislocation walls are illustrated in figure 4 (*c*). Obviously, the formation of the well developed DBII must have evolved from the developing DBII through the disappearance of dislocation rungs on the $(1\bar{2}1)$ plane. Since the microstructure of DBII is characterized by ladder-like PSBs or parallel walls, as shown in figure 2 and figure 3, the DB should be a region with strong plastic strain localization. Furthermore, the dislocation arrangements within the DB should be favourable for carrying a large plastic strain. It is suggested that the formation of DBII may be attributed to the local regulation of dislocation walls on primary slip planes.

## § 4. Summary and conclusions

(1) DB is an important plastic deformation mode when a copper single crystal with a single-slip orientation is cycled at a high strain amplitude. The DB has

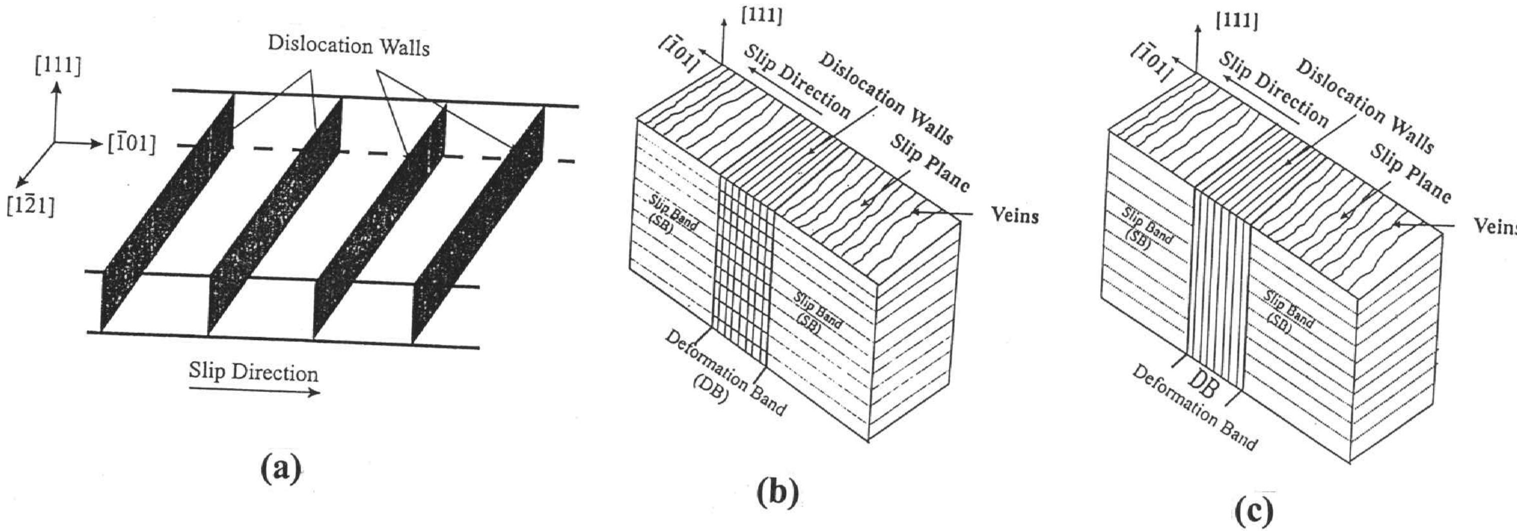


Figure 4. Sketches of the three-dimension structures of (*a*) the ladder-like PSB; (*b*) the developing DB and (*c*) the well developed DB.

a width of 50–60 μm and is homogeneously distributed over the whole crystal surface. However, the surface SBs did not show any persistent feature. It is indicated that the two-phase model of PSBs and veins might be invalid when the crystal is cycled at higher strain amplitude.

(2) The SEM ECC technique can reveal the dislocation arrangements within DBs, which are difficult to observe by the conventional TEM technique. The present study revealed the microstructure within DBII in a fatigued copper single crystal with a single-slip orientation, which has not been reported before. The results showed that there are two kinds of typical dislocation arrangement within the DBII. One consists of regular ladder-like PSBs in parallel within those developing DBs. Another is composed of regular and parallel dislocation walls within the well developed DBs. However, the microstructures between DBs are dislocation loop patches and irregular SBs. Therefore, those regions between DBs should be harder than the DBs. DBs must carry more plastic strain than SBs. Furthermore, the formation of DBII may be attributed to the local arrangement of dislocation walls on primary slip planes. By crystallographic analysis of DBII, it is clarified that the habit plane of the DBII is $(\bar{1}01)$.

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