Fundamental factors on formation mechanism of dislocation arrangements in cyclically deformed fcc single crystals

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This paper systematically summarizes the cyclic deformation behaviors of different kinds of face-centered cubic (fcc) single crystals, including copper, nickel, silver, as well as copper–aluminium, copper–zinc alloys in attempt to provide a historical perspective of the developments over the last several decades. Combined with plenty of previous research results, the influencing factors on cyclic deformation behaviors can be listed as follows: orientations, stacking fault energy (SFE), short-range order (SRO) and friction stress, or more generally, the ease of cross slip. Among them, the effect of orientations mainly reflects in the formation of the complex dislocation patterns, which depends on the activating secondary slip system. According to the effect of slip mode, the materials can be divided into two types: pure metals and alloys. For pure fcc metals, the effect of SFE is decisive. Due to the easy cross slip of screw dislocations, regular dislocation arrangements, e.g. veins, persistent slip bands (PSBs), labyrinth and cell patterns, are always to form. With increase in alloying element, antiphase boundary energy gradually replaces SFE to become a new decisive factor affecting the cyclic deformation behaviors of fcc alloy single crystals. The corresponding dislocation arrangements consist of dipole array and stacking faults (SFs) under the influence of planar slip. The relationship among several factors is well explained, which will help us better understand the nature of the fatigue damage of metallic materials and then improve the performance of the related materials.

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1. Introduction

Fatigue subjects have developed into one of the most important scientific research fields for materials scientists during the last 100 years. The study of the fatigue damage mechanisms linking with the micro-deformation can be retrospected to the early 20th century. Ewing and Rosenhain [1] and Ewing and Humfrey [2] demonstrated that fatigue-induced slip bands (SBs), which gradually developed into cracks, appeared on the surface of Swedish iron when subjected to repeated reversal of stress by bending in air at room temperature, which is the first insight into the mechanisms of metal fatigue and one of the earliest widely cited fundamental fatigue studies. As single crystal specimens were commonly used after the late 1950s, the most conclusive results on fatigue damage mechanisms began to be obtained from high-purity materials, especially from face-centered cubic (fcc) single crystals. Later than that, with the advent of sophisticated electronic control of electrohydraulic testing equipment, constant plastic strain amplitude control is widely practiced. Since the mid-1970s, the studies of low-cycle fatigue behaviors of single crystals have been steady and culminated.

On the other hand, since the dislocation theory was proposed by Taylor [3], Orowan [4] and Polanyi [5] in 1934 independently, the theory was applied to fatigue as the basis of metallurgical investigation. During the period of 1940–1960s, studies were devoted exclusively to the dislocation aspects of fatigue and a much more detailed understanding on the microstructural mechanisms was achieved. Accordingly, it becomes possible to establish the relationships between the microscopic dislocation mechanisms and the macroscopic deformation behavior in terms of the cyclic deformation behaviors, as discussed in this review. In the early times, some famous pioneers engaged in the study of the cyclic deformation behaviors of single crystals, for example, Forsyth [6,7], Thompson et al. [8], Mott [9], Cottrell and Hull [10], Wood [11] and May [12,13]. From the beginning of 1960s, numerous studies of cyclic deformation were carried out on various fcc single crystals including copper, nickel, silver, α copper–aluminium and α copper–zinc with the aim to clarify the nature of fatigue damage in crystalline materials by several research groups, for example, Laird at Philadelphia (USA), Mughrabi at Stuttgart (Germany), Neumann at Düsseldorf (Germany), Holste and Blochwitz at Dresden (Germany),
Lukáš at Brno (Czech Republic), Brown and Winter at Cambridge (UK), Basinski at McMaster (Canada), Wang ZR at Toronto (Canada) and Wang ZG at Shenyang (China). Most of the fundamental work in the last several decades can be found in several monographs [14–16] and in a number of excellent reviews [17–26] which emphasized different aspects of the discussed subject. By analyzing and thinking over the research results over the years, it will be necessary to treat a wider range of materials and consider fcc crystals as a whole.

Based on the brief introduction above, this paper will only deal with two topics. One is to summarize the fatigue phenomena in different fcc metals, where an attempt will be made to illustrate a historical picture of the developments over the last half of century. The other is to outline our present understanding on the microscopic mechanisms of cyclic deformation behaviors that lead to the evolution of dislocation arrangements in the bulk and the appearance of slip morphologies at the specimen surface of fcc single crystals.

2. Effect of orientation on cyclic deformation behaviors of fcc single crystals

Cyclic deformation behaviors of fcc single crystals are composed of three interrelated parts. They are the cyclic stress–strain (CSS) curve, the surface slip morphologies and the microscopic dislocation arrangements. If a fully annealed fcc single crystal is subjected to a cyclic plastic straining, it will harden rapidly during the first tens of cycles (see Fig. 1a). The cyclic hardening rate is related to the crystal orientation and the applied plastic strain amplitude [27,28]. The hardening increment per cycle decreases as the cyclic deformation continues and eventually it falls to zero in about 1000 cycles. When the hardening increment vanishes, the hardening curve becomes horizontal, and the crystal is said to be “in saturation”. “Saturation” describes a state of the specimen in which rapid hardening has been accomplished and essentially no further hardening then occurs. A plot of the resolved shear stress amplitude in saturation $\tau_s$ versus the resolved shear strain amplitude $\gamma_{pl}$ generates the cyclic stress–strain (CSS) curve, as shown in Fig. 1b. After cyclic deformation of an fcc single crystal, it is natural that fatigue damage including slip bands (SBs) and deformation bands (DBs) will appear at the surface. These bands result from the accumulation of cyclic slip irreversibilities and develop gradually with increasing cycles. Under these surface slip morphologies a variety of complicated dislocation structures can be observed, collectively known as dislocation

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Fig. 1. Cyclic stress–strain curve of fcc single crystal cycled at the constant plastic resolved shear strain amplitude: (a) hysteresis loop and (b) cyclic stress–strain curve.
arrangements. Dislocation arrangements are highly ordered spatial structures produced by dislocation reactions and rearrangements. Although they show all kinds of patterns, their basic feature is the spatial periodically ordered structure with the alternate appearance of dislocation-rich and dislocation-poor regions [29,30].

The crystal orientation is one of the most important factors on cyclic deformation behaviors of fcc crystalline materials. As shown in the stereographic triangle in Fig. 2a, different orientations have different slip systems. Firstly, for single-slip oriented fcc single crystals the slip system should be $(111)[101]$, in which slip plane is $(111)$ and Burgers vector along slip orientation $B = [101]a/2$. Secondly, conjugate double-slip systems include both $(111)[101]$ and $(111)[011]$. The two slip systems can transform into each other under cyclic loading, which is the reason why they are called conjugate double-slip systems. The critical double-slip systems include both $(111)[101]$ and $(111)[101]$ with their respective independence under cyclic loading. The coplanar double-slip systems have both $(111)[101]$ and $(111)[110]$ with the same slip plane. Finally $[011]$, $[111]$ and $[001]$ multiple-slip oriented fcc single crystals have four, six and eight slip systems, respectively. Due to difference of crystal orientations, the fatigue-induced classical dislocation arrangements show various types, such as veins, persistent slip bands (PSBs), labyrinth and cell patterns [18,31–33]. These typical patterns, especially PSB-ladder structure with equidistant rungs, were firstly found in cyclically saturated copper single crystals [34–36], which was one of the most interesting findings.

Now it has been known that for copper, nickel or silver single crystals, as shown in Fig. 2b, the orientation-dependent dislocation arrangements can be divided into three regions in the stereographic triangle: the $[011]$ region, the $[001]$ region and the $[111]$ region [37]. The common characteristics in the $[011]$ region are to form the PSB-wall structures in deformation band II (DBII), which is different from those in single-slip oriented ones. The $[001]$ region is made up of the orientations around the $[001]$ including itself. The common feature is that the labyrinth structure is easier to form. The $[111]$ region is composed of the orientations around the $[111]$ including itself. The vein structure at low strain amplitudes and the cell or wall structure at high strain amplitudes become the most familiar dislocation patterns [38].

The above division is derived from the extensive research in copper, nickel and silver single crystals with different orientations. The cyclic deformation behaviors of the three kinds of fcc metals will be introduced as follows, respectively.

![Fig. 2. Effect of orientation on dislocation arrangements: (a) easily activated slip systems in the stereographic triangle and (b) general principle of the effect of orientation on dislocation arrangements [37].](image-url)
2.1. Cyclic deformation behaviors of copper single crystals with different orientations

It is well known that the “persistent” phenomenon of fatigue-induced slip bands was defined for the first time in 1956 by Thompson et al. [8] in their study on the nature and properties of slip bands in fatigued copper mono- and poly-crystals. They found that these “persistent slip bands” (PSBs) reappeared at the old sites when the specimen was fatigued again after the previously formed slip bands had been polished away, which indicated that the microstructure in the bulk of PSBs is different from that of the surrounding matrix (see Fig. 3). Therefore, Winter [39] in 1974 firstly proposed a two-phase model to explain the cyclic saturation dislocation structures. As shown in Fig. 4, for a given shear strain amplitude \( \gamma_{pl} \) when the cyclic saturation occurs, the volume fractions \( f_{PSB} \) and \( f_{M} \) of PSBs and matrix, respectively, are constant. The plastic strain amplitudes carried by PSBs and matrix are \( \gamma_{PSB} \) and \( \gamma_{M} \), respectively; therefore the average shear plastic strain amplitude can be expressed as [39]

\[
\gamma_{pl} = \gamma_{PSB} f_{PSB} + \gamma_{M} f_{M}
\]  

Based on these early works, Mughrabi [40] established experimentally the famous CSS curve of copper single crystals oriented for single slip and found that the CSS curve clearly exhibited three regions A–C (see Fig. 4). For copper crystals cycled in region A (\( \gamma_{pl} < \gamma_{pl,AB} \)), the matrix veins form and the saturation stress increases with increasing the applied strain amplitude. The onset of region B is marked by the first appearance of PSBs [31,36]. As the strain amplitude \( \gamma_{pl} \) increases in region B (\( \gamma_{pl,AB} < \gamma_{pl} < \gamma_{pl,BC} \)), the volume fraction of the PSB’s increases accordingly [17,39,41,42] and the CSS curve shows a plateau behavior with the strain-independent stress amplitude of about 28 MPa. With further increasing strain amplitude up to region C (\( \gamma_{pl} > \gamma_{pl,BC} \)), the saturation resolved stress increases again and the dislocation structure is characterized by dislocation cells or labyrinth structures [43–46]. Soon after, Cheng and Laird [47,48] found that for most single-slip oriented copper single crystals their saturation resolved shear stress amplitudes at room temperature are in the range of 28–30 MPa.

Based on the single-slip oriented copper single crystals, different double-slip and even multiple-slip oriented crystals were investigated in succession. Jin and Winter [49–52] studied systematically

![Fig. 3. PSBs in fatigued copper. (a) SEM, surface relief and (b) TEM micrograph of PSBs with ladder structure between matrix of dipolar veins in a section perpendicular to the primary slip plane, containing primary Burgers vector b. Quoted from Mughrabi [43].](image-url)
the cyclic deformation behaviors of copper single crystals with three $[0\ 1\ 2]$, $[1\ 1\ 2]$ and $[1\ 2\ 2]$ double-slip orientations. They found that different regions in the stereographic triangle corresponded to different dislocation interactions, which resulted in different cyclic hardening rates, saturation stresses and dislocation arrangements. Firstly, the saturation stresses of $[0\ 1\ 2]$, $[1\ 1\ 2]$ and $[1\ 2\ 2]$ copper single crystals were found to be 28.3 MPa, 31.4 MPa and 35.8 MPa, respectively. Secondly, the dislocation arrangements of differently oriented copper single crystals depend on the specific responsible dislocation interactions. In the case of $[1\ 1\ 2]$ crystals, dislocations in the different systems occupy clearly separated regions, in each of which they form the structures very similar to those in single-slip oriented crystals. In $[0\ 1\ 2]$ crystals, dislocation interactions between the primary and the critical slip systems ($\{1\ 1\ 1\}\{1\ 0\ 1\}$ and $\{1\ 1\ 1\}\{1\ 0\ 1\}$) lead to the formation of sessile jogs. Dislocations in the different systems tend to occupy separated zones. The majority of one type of dislocation in each zone constitutes a structure strongly affected by the dislocations of the other system. According to the degree of influence, dislocation arrangements may be similar to those in single-slip oriented ones or be labyrinth-like. In single-slip oriented crystals, the two-phase structure consisted of matrix and PSBs is common. However, in $[1\ 2\ 2]$ crystals, multiple slip is active locally because the dislocations are coplanar and cell-like structures are easier to form, which lose their similarity with the two-phase structure.

As for the multiple-slip orientations, Lepisto et al. [53–55] and Gong et al. [56,57] investigated the cyclic deformation behaviors of $[1\ 1\ 1]$ and $[0\ 0\ 1]$ copper single crystals, respectively. The results showed that the CSS curve in both multiple-slip oriented copper single crystals did not show any plateau behavior. Furthermore, they found that cell and labyrinth structures appear in $[1\ 1\ 1]$ and $[0\ 0\ 1]$ copper single crystals after cyclic saturation, respectively. Later, Li et al. [58] summarized the cyclic hardening rates, CSS curves and the corresponding dislocation patterns of $[0\ 1\ 1]$ copper single crystals. They concluded that the cyclic deformation behavior of $[0\ 1\ 1]$ copper single crystals is quite different from those of $[1\ 1\ 1]$ and $[0\ 0\ 1]$ copper single crystals. The CSS curve of $[0\ 1\ 1]$ single crystals shows a clear plateau behavior, which is similar to that of single-slip oriented crystals. Based on the systematic investigations on the fatigue behaviors of copper single crystals with double- and multiple-slip orientations, Li et al. [59–62] have given a clear and comprehensive summary on the orientation effect of copper single crystals, which can be especially introduced as follows: on the $0\ 1\ 1/1\ 1\ 1$ side of the stereographic triangle, the CSS curves vary from a wide plateau of the $[0\ 1\ 1]$ crystals to no...
plateau for [1 1 1] copper single crystals. The quasi-plateau behavior of coplanar double-slip oriented copper single crystals can be regarded as a transition. On the 0 0 1/0 1 1 side, the CSS curves exhibit a narrower and narrower plateau region from [0 1 1] crystal to [0 0 1] crystal. On the 0 0 1/1 1 1 side of the stereographic triangle, the CSS curve of the [1 1 2] copper single crystal exhibits a clear plateau region, which is shorter than that of the single-slip crystals although the saturation resolved shear stress amplitudes are nearly the same. However, the CSS curve of [1 1 7] crystals does not show any plateau region, which is similar to that of [2 2 3] and [1 1 1] single crystals. Corresponding to the CSS curve, the dislocation patterns of differently oriented copper single crystals all have distinct characteristics. The PSB-ladder structure appears in copper single crystals with the single-, [1 1 2] double- and [0 1 1] multiple-slip orientations. As the orientations of copper single crystals change from [0 1 1] to [0 0 1] in the stereographic triangle, the corresponding dislocation patterns vary from PSB ladders to labyrinth structure. When the orientation changes from [0 1 1] to [1 1 1], PSB ladders will be gradually converted into the cell-like structure.

Until then, the investigation on the cyclic deformation behaviors of copper single crystal has been comparatively perfect. It should be said that the systematical research on copper single crystals not only established the theoretical basis for polycrystalline materials, but provided a reference for other fcc single crystals.

2.2. Cyclic deformation behaviors of nickel single crystals with different orientations

As early as the establishment of three-stage CSS curves of copper single crystals, Mughrabi and his colleagues carried out the studies on the cyclic deformation and fatigue damage of other typical fcc single crystals, including nickel single crystals. In ASTM symposium of 1979, Mughrabi et al. [18] proposed that the CSS curve of single-slip oriented nickel single crystal showed the same three stages with a higher plateau stress of 52 MPa as compared with copper single crystal. Although the authors did not directly present the figures about the dislocation arrangements of nickel single crystal, they clearly pointed out that the regular arrangements of dislocations, especially ladder-like PSBs similar to that in copper single crystals, also exist in nickel single crystals. In 1982, Blochwitz and Veit [63] further studied the CSS curves and the corresponding fatigued dislocation structures in single-slip oriented nickel single crystal. They found that the volume fraction of PSBs increased with the increment of the applied strain amplitude and confirmed again the three-stage character of the CSS curve in nickel single crystal with single-slip orientations. In the same year, Mecke and Blochwitz [64] and Mecke et al. [65] described and characterized qualitatively the saturation dislocation structures and the development of dislocation arrangements in fatigued nickel single crystals with different orientations. The dislocation arrangement in [0 0 1] nickel single crystal is characterized by the labyrinth structure with condensed and uncondensed (1 0 0) dislocation rungs. In region B of the CSS curve for [1 4 9] nickel single crystal, the saturation dislocation structure consists of PSBs and vein-like matrix. With increasing strain amplitude, besides the described vein-like structure, the so-called maze or labyrinth structure appears as another type of the saturation structure since Charsley [66] firstly discovered the “maze” structure in a fatigued copper–nickel alloy. The labyrinth structure consists of two sets of edge dislocation rungs, which actually is the labyrinth structure of [0 0 1] nickel single crystal. These labyrinth structures gradually evolve and finally form with the cyclic number, just like the PSBs. Later, Kahle [67] systematically investigated the cyclic deformation behavior of [1 1 1] nickel single crystal and found that its CSS curve did not show any plateau in the total strain amplitude range of $6.0 \times 10^{-5} \leq \epsilon_{pl} \leq 5.0 \times 10^{-4}$.

From the mid-1990s, the cyclic deformation behaviors of nickel single crystals were systematically discussed in terms of the effects of the temperature and orientation by Dresden group. On the effect of temperature, Bretschneider et al. [68] compared the stress–strain behaviors and the dislocation structures of [1 4 9] nickel single crystal cyclically deformed at room temperature and 77 K in order to explain the influence of low testing temperature on strain localization. They found that the saturation resolved shear stress amplitude of nickel single crystal at 77 K is about 50 MPa higher than that at room temperature. After low-temperature deformation, regularly arranged ladders were only found if the specimens were pre-deformed at room temperature and PSB layers have already formed in the crystal. Nearly the entire volume of the saturated dislocation structure of crystals deformed at 77 K
without pre-deformation consists of extended wall structures with irregularly arranged dislocation-dense regions. Subsequently, Schwab et al. [69] observed the saturation dislocation arrangement after cyclic deformation of nickel single crystals oriented for single slip at 300 K and 600 K, respectively. The imaging by the electron channeling contrast (ECC) technique showed that the well known two-phase dislocation structure consisting of PSBs and matrix vein appeared at both of temperatures. Only the mean channel width $d_c$ between the rungs is slightly different. It is well documented that the channel width increases with increasing the testing temperature, whereas the rung thickness remains constant [70], as shown in Fig. 5. On this basis, Bretschneider et al. [71] summarized the cyclic deformation behavior of single-slip oriented nickel single crystals at deformation temperatures between 293 K and 900 K. They found that PSBs with a typical ladder-like structure occurred at temperatures up to 800 K. At 600 K and 750 K the measured CSS curves exhibited extended ranges in which the saturation stress amplitude was nearly independent of the applied plastic strain amplitude. The plastic strain amplitudes at the upper and the lower borders of these plateaus in the CSS curves decreased with increasing temperature (see Fig. 6). Likewise, most of the measured structure lengths of the rung/channel arrangement in the PSB ladders and of the dipole arrangement in the PSB rungs have been found to become larger at elevated temperatures. The only exception observed is that the rung thickness remains constant independent of the temperature.

Besides the temperature, the orientation is another key factor that influences the cyclic deformation behavior of nickel single crystals. Based on studies including [1 1 1] nickel single crystal by Kahle

![Fig. 5. Saturation dislocation arrangement after cyclic deformation of nickel single crystals at (a) $T = 300$ K and (b) 600 K observed on the surface parallel to (1 2 1) plane with the strain amplitude of $\gamma_{pl} = 2 \times 10^{-3}$. Quoted from Schwab et al. [69].](image)

![Fig. 6. CSS curves of nickel single crystals oriented for single slip at 77, 293, 600 and 750 K: (......), upper and lower ends of the plateau. (The CSS curves are taken from the papers by Hollmann [169] and Bretschneider et al. [71].)](image)
[67] and [149] nickel single crystal by Bretschneider et al. [71], Buque [72] thoroughly summarized the effect of orientations on the CSS curves and dislocation arrangements of nickel single crystals. He found that the CSS curve of the [0 1 1] crystal showed a very pronounced plateau in the total strain amplitude range of $1.0 \times 10^{-4} < \gamma_{pl} < 5.0 \times 10^{-3}$ in which the saturation stress is similar to that of single-slip oriented nickel single crystal. In [0 0 1] nickel single crystal, the CSS curve only shows a quasi-plateau in which the corresponding saturation stress is slightly higher than those of single-slip oriented and [0 1 1] crystal. The CSS curve of [1 1 1] crystal shows apparently no plateau. Correspondingly, Buque [72] found that the occurrence of persistent DBs is the most typical feature of the surfaces of fatigued [0 1 1] crystal. A DB is characterized by a stack of densely distributed and strongly localized slip markings which belong to a certain slip system. The dislocation structures of [0 1 1] crystal are typically grouped in structure bands which consist of short ladder-like substructures (short PSBs), wall-like dislocation configurations and patch dislocation patterns in the plateau region. At the total strain amplitude $\gamma_{pl} \geq 5.0 \times 10^{-3}$ elliptical cells exist in the entire crystal. The surfaces of cycled [1 1 1] crystal are typically covered by two sets of slip traces crossing one another. DBs with the characteristics of conventional kink bands were also found but as a minor case. The dislocation structures of [1 1 1] crystal consist of long condensed wall-like dislocation configurations with the longest extension lying perpendicular to the tensile axis. The observed structure types fall in line with those subsequently found in corresponding oriented grains of cycled nickel polycrystals including the research works from Buque et al. [73–76], which in essence are still attributed to the orientation effect. In these studies, Buque et al. [72–75] made a good summary of the typical dislocation arrangements in differently oriented nickel single crystals and displayed them in the stereographic triangle. In 2004, Holste [77] summarized comprehensively the results obtained by himself and co-workers during the last two decades in studies on the effects of temperature and orientation on the CSS curve, the saturation stress and the dislocation structure of nickel single crystals. Thus, it should be said that, similar to the case of copper single crystals, nickel single crystals are also among those fcc single crystals in which the cyclic deformation and fatigue damage have been investigated most systematically.

2.3. Cyclic deformation behaviors of silver single crystals with different orientations

In addition to copper and nickel, silver is another typical fcc metal. Until now the studies on the cyclic deformation behaviors of silver single crystal are still rare. The previous main researchers are Sastry et al. [78] and Mughrabi et al. [18]. In 1976, Sastry et al. [78] studied the defect microstructure at saturation of [128] silver single crystals subjected to constant strain amplitude in the range of $2.5 \times 10^{-3} < \gamma_{pl} < 2.5 \times 10^{-2}$. They found that the defect microstructure at saturation can be characterized by the density and distribution of point-defect clusters and the dislocation rung spacing along the primary Burgers vector direction. The saturation stress is inversely proportional to the dislocation rung spacing along the primary slip direction and to the spacing of the point-defect clusters on the primary slip plane. Later in 1977, Sastry and Ramaswami [79] further studied the annealing behavior of dislocation structures and point-defect clusters in fatigued silver single crystals. They found when the annealing temperature exceeded 500°C, the dislocation cell structure would almost completely annihilate with no indication of polygonization and recrystallization. Besides copper and nickel single crystals, Mughrabi et al. [18] also investigated the cyclic deformation behavior of silver single crystal at the same stage. Compared with copper and nickel single crystals, they found that the characteristic thresholds for the formation of a PSB-ladder structure at room temperature, expressed as $\tau_s/G$, had a very similar value of $(6.55 \pm 0.5) \times 10^{-4}$ (see Fig. 7) for the three kinds of fcc single crystals, which is worth further considering.

Recently, our group [80–83] have systematically investigated the cyclic deformation behaviors of silver single crystals with the orientations including [239], [1818], [459], [1414], [233] and [0 1 1]. The fatigue testing conditions and data on cyclic saturation in various silver single crystals are summarized in Table 1, where $\gamma_{pl}$ and $\tau_s$ are the plastic strain amplitude and saturation resolved shear stress, respectively. The detailed results can be enumerated one by one as follows.

As shown in Fig. 8, for differently oriented silver single crystals, it seems that all the data can be plotted as two curves with clear plateaus. One of the two plateaus is composed of the data from [1414] and [233] silver single crystals. The other plateau consists of the data from the orientations
Table 1
Fatigue testing conditions and data for silver single crystals with different orientations.

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<tr>
<th>Specimen no.</th>
<th>( \gamma_{pl} ) (MPa)</th>
<th>Cycles</th>
<th>( \tau_s ) (MPa)</th>
<th>Specimen no.</th>
<th>( \gamma_{pl} ) (MPa)</th>
<th>Cycles</th>
<th>( \tau_s ) (MPa)</th>
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Fig. 7. Cyclic stress–strain data for nickel, copper and silver single crystals in the region of very low \( \gamma_{pl} \). The data pertaining to the formation of the first PSB’s fall into the indicated rectangle. Quoted from Mughrabi et al. [18].

Fig. 8. CSS curves of differently oriented silver single crystals. The CSS curve data of silver single crystals with a Schmid factor \( \Omega = 0.5 \) are from unpublished work of Mughrabi. Quoted from Mughrabi [84].
of [239], [459], [1818] and [0 1 1], in combination with the data from Mughrabi et al. [18,84]. Fig. 8 shows that in the latter plateau, up to a shear strain amplitude of about $1.0 \times 10^{-3}$, the saturation stresses are somewhat lower than those at the higher amplitudes. Therefore it should be emphasized that the saturation stresses corresponding to the two plateaus are about 18–21 MPa and 25–26 MPa, respectively. Based on the formation of the plateau, the CSS curves of silver single crystals can be divided into three regimes similar to those of copper and nickel single crystals [40,71].

Fig. 9 shows the surface slip morphologies and the corresponding typical dislocation patterns of [239] silver single crystals at different magnifications. In Fig. 9a, the slip bands (SBs) distribute homogeneously on the specimen surface. Meanwhile, typical PSB ladders appear in [239] silver single crystals (see Fig. 9b), and correspond well with the surface SBs (see Fig. 9c). At higher magnification, the existence of ladders built up of rungs [85], can be confirmed. In fact, it is clear that each PSB ladder is composed of a number of roughly equidistant rungs (see Fig. 9d).

Fig. 10 presents the changes of dislocation arrangements in [1818] silver single crystals with increasing strain amplitude. When $\gamma_{pl} = 1.0 \times 10^{-3}$, a PSB-ladder structure starts to appear (see Fig. 10a). However, most PSB ladders only become clear at a strain amplitude of $\gamma_{pl} = 2.0 \times 10^{-3}$, (see Fig. 10b). As the strain amplitude increases to $4.0 \times 10^{-3}$ (see Fig. 10c), PSB ladders corresponding to the secondary SBs begin to appear on the specimen surface, with two sets of PSB ladders interlaced with each other. With further increasing the strain amplitude to $8.0 \times 10^{-3}$ (see Fig. 10d), the dislocation patterns become dominated by a labyrinth structure.

Fig. 11 shows PSBs in a [459] silver single crystal. These PSB walls are approximately perpendicular to the primary Burgers vector and to the primary slip plane and they lie approximately on the $(101)$ plane. Although the orientation of [459] silver single crystals is close to the conjugate double-slip orientation, they behave like single-slip oriented crystals. Besides, the influence of cyclic number on the dislocation arrangements was primarily studied for [1414] silver single crystals close to critical double-slip oriented ones. Fig. 12 shows that the interactions between primary and secondary PSB ladders still play an important role in the cyclic deformation although the primary and secondary PSB ladders have their respective regions (see Fig. 12d). Meanwhile, it was found that with increasing cyclic

![Image](https://example.com/image.png)

**Fig. 9.** Surface slip morphologies and dislocation patterns of [239] silver single crystal at the plastic strain amplitude of $\gamma_{pl} = 2.1 \times 10^{-3}$. (a and c) surface slip morphologies; (b and d) PSB-ladder structures. Viewed from $(121)$. 
number the proportion between primary and secondary PSB regions gradually changes from a structure dominated by primary PSB regions to a structure with an equal proportion of primary and secondary ones.

In silver single crystal of [2 3 3] orientation, coplanar double-slip is favored. In the previous study on [2 3 3] copper single crystal, Li et al. [61, 86] found that its CSS curve showed a quasi-plateau and the corresponding saturation dislocation arrangements are filled with cell structure. Likewise, a cell structure also appears in fatigued [2 3 3] silver single crystal. As shown in Fig. 13a, for [2 3 3] silver single crystal at low strain amplitude a dislocation vein microstructure can be clearly seen. With increasing strain amplitude \( \gamma_{pl} \), the specimen surface consists entirely of cell structure (see Fig. 13b), which further demonstrates that orientation must play an important role in the cyclic deformation behavior of silver single crystals.

![Fig. 10. Dislocation structures of [1818] silver single crystal at different plastic strain amplitudes. (a) \( \gamma_{pl} = 1.0 \times 10^{-3} \); (b) \( \gamma_{pl} = 2.0 \times 10^{-3} \); (c) \( \gamma_{pl} = 4.0 \times 10^{-3} \) and (d) \( \gamma_{pl} = 8.0 \times 10^{-3} \). Viewed from (2 2 1).](image)

![Fig. 11. Dislocation patterns of [459] silver single crystal under different magnifications at the plastic strain amplitude of \( \gamma_{pl} = 4.5 \times 10^{-3} \). (a) PSBs and (b) PSB-wall structure. Viewed from (1 1 1).](image)
Interesting orientation also includes [0 1 1] multiple-slip orientation. Li et al. [58] and Buque [72] systematically summarized the cyclic deformation behaviors of [0 1 1] copper and nickel single crystals, respectively. Firstly, the CSS curves of [0 1 1] single crystals, of both copper and nickel, are approximately in accordance with those of single-slip oriented ones. Secondly, the classical DBs appear in both copper and nickel [0 1 1] single crystals. Thus, it is natural to speculate whether the similar behavior also appears in [0 1 1] silver single crystal or not. Fig. 14 presents typical dislocation patterns of [0 1 1] silver single crystal at a plastic strain amplitude of $\gamma_{pl} = 6.1 \times 10^{-4}$. As shown in Fig. 14a, regular dislocation walls form within DBs similar to the arrangements in copper and nickel single crystals [72,87]. Compared with Fig. 11, it can be seen that the wall structure in [0 1 1] silver single crystal is mainly located in DBs, with a two-phase structure composed of PSB and vein on both sides of the DBs. However, the PSB-wall structure in [459] silver single crystal is not located in DBs.

Fig. 12. Dislocation structures of [1414] silver single crystal under different number of cycles at the same plastic strain amplitude of $\gamma_{pl} = 5.0 \times 10^{-4}$. (a) 3000 Cycles; (b) 5000 cycles; (c) 10,000 cycles and (d) 20,000 cycles. Viewed from (6 2 1).

Fig. 13. Dislocation configurations of [233] silver single crystal. (a) Vein at the plastic strain amplitude of $\gamma_{pl} = 1.35 \times 10^{-4}$ and (b) cell at the plastic strain amplitude of $\gamma_{pl} = 8.1 \times 10^{-3}$. Viewed from (3 3 1).
The walls themselves correspond to SBs, thus there is only vein structure on both sides of the walls. In addition, PSB ladders are the other major dislocation structure in [0 1 1] silver single crystal. Fig. 14d shows a classical PSB-ladder structure, but in which the walls are not as regular as those in PSB ladders in [½ 1/2 239] silver single crystal, compared with Fig. 9d. Therefore, “wall” structure includes two types, one corresponds to the formation of DBs; the other appears on the (1 1 1) plane corresponding to the formation of PSBs.

All in all, on the one hand, differently oriented silver single crystals show their respective dislocation arrangement characters. PSB ladders appear in [2 3 9], [0 1 1] and [4 5 9] silver single crystals. The labyrinth structure is formed in [½ 1/2 233] silver single crystal at high train amplitude. And the vein and cell are the main structures in [½ 1/2 233] silver single crystal at low and high strain amplitudes, respectively. The interaction between the primary and secondary PSBs appears in [1 4 1 4] silver single crystal. On the other hand, combined with the results of copper and nickel single crystals, the effect of the orientations on the cyclic deformation behaviors of the three typical single crystals is found to follow a general principle, which is much interesting and will be further discussed as follows.

2.4. Similarities of orientation effect among copper, nickel and silver single crystals

Based on plenty of experimental data from Mughrabi [40], Cheng and Laird [48], Jin and Winter [49,50], Lepisto et al. [53,55], Gong et al. [56,88,89] and Li et al. [58,86,90,91], Li et al. [59,61] summarized the plateau behavior in the CSS curves of copper single crystals with different orientations. As shown in Fig. 15a, with the variation of the orientation, the CSS curves of copper single crystals show significant difference, which can be specially introduced as follows: (1) the similar saturation plateau behaviors appear in the single-, conjugate double- and [0 1 1] multiple-slip oriented copper single crystals; (2) the coplanar double-, critical double- and [0 0 1] multiple-slip oriented copper single crystals have obviously higher saturation stress in their CSS curves without plateau region. Corresponding to the CSS curves, the dislocation configurations of copper single crystals with different
orientations also show certain regularity. As shown in Fig. 15b, the classical PSB-ladder structure appears in copper single crystals with the single-, conjugate double- and [0 1 1] multiple-slip orientations. As the orientations of copper single crystals change from [0 1 1] to [0 0 1] in the stereographic triangle, the corresponding dislocation patterns vary from PSB ladders to labyrinth structure. When the orientation changes from [0 1 1] to $\frac{1}{2}[1 1 1]$, PSB ladders will be gradually converted into the cell-like structure.

The effect of the orientation on the cyclic deformation behaviors of nickel single crystals has a similar principle. In comparison with the results from Bretschneider et al. [71] and Kahle [67], Buque et al. [72,74] systematically summarized the CSS curves and the classical dislocation configurations of nickel single crystals with different orientations. As shown in Fig. 16, there are also apparent plateau regions and classical PSB ladders in $\frac{1}{2}[1 4 9]$ and [0 1 1] nickel single crystals. Differently, labyrinth structure appears in the [0 0 1] nickel single crystal and PSB-wall structure forms in the $\frac{1}{2}[1 1 1]$ nickel single crystal. As the orientation changes from $\frac{1}{2}[1 4 9]$ or [0 1 1] to [0 0 1], more to [1 1 1], the CSS curve of nickel single crystals shows a continuous elevation or inclination as a rule.

Similar to copper and nickel single crystals, the typical PSB-ladder structure also appears in the single-slip oriented silver single crystals, which corresponds to the occurrence of the plateau behavior, as shown in Fig. 17. But the difference is only that the plateau stress of silver single crystals is about 20 MPa. Likewise, in [45 9] and [0 1 1] silver single crystals, the plateau behavior and PSB walls are the main feature. The labyrinth structure is much obvious in silver single crystals with the orientation

Fig. 15. CSS curves and dislocation configurations of copper single crystals with different orientations [59,61]. (Note: Fig. 15b is drawn according to the extensive researches [49,50,53,55,56,58,62,86,88–91].)

Fig. 16. CSS curves and dislocation configurations of nickel single crystals with different orientations. Quoted from Buque [72,74].
close to [0 0 1]; the cell structure is much evident in silver single crystals with the orientation close to [1 1 1]. For [0 0 1] and [1 1 1] silver single crystals, it can be judged that the plateau behavior of the CSS curves will also become no longer apparent.

As shown in Fig. 18, Li et al. [59] summarized the principle on the effect of the orientations on the plateau behaviors. They found that the rules can be obeyed as follows: on the 001/111 side [91], considering [1 1 2] as the center, with the transition of the orientation to [0 0 1] and [1 1 1], respectively, the plateau behaviors of the CSS curves gradually disappear. On the 0 0 1/0 1 1 side [90], from [0 1 1] to [0 0 1], the CSS curves change as follows: clear plateau → shorter plateau → quasi-plateau → no plateau. On the 011/111 side [86], from the wide plateau of the [0 1 1] copper single crystal to the no plateau of the [1 1 1] copper single crystal, the quasi-plateau behaviors of coplanar double-slip-oriented copper single crystals can be regarded as a transition.
Both dislocation structures and plateau behavior of copper, nickel and silver single crystals have been divided into the three regions in the stereographic triangle, so the investigation on the orientation effect can be summarized as the investigation on the \([0 0 1]\), \([0 1 1]\) and \(\frac{1}{2} [1 1 1]/C_{22}\) fcc single crystals, respectively. Then what role does the orientation play in the cyclic deformation behaviors of fcc single crystal? To reveal the physical nature of the orientation effect, \([0 0 1]\) and \(\frac{1}{2} [1 1 1]/C_{22}\) copper single crystals have been selected as model materials and further discussions will be made by systematically investigating their formation mechanisms of dislocation arrangements during cyclic deformation.

2.5. Intrinsic relationship between dislocation arrangements and crystal orientations

2.5.1. CSS curves of copper single crystals with multiple-slip orientations

Fig. 19 summarizes the CSS curves of the three multiple-slip oriented copper single crystals, among which \(\frac{1}{2} [1 1 1]/C_{22}\) copper single crystal has the highest saturation stress amplitude. From Fig. 19, the similarity and difference of the CSS curves between three multiple-slip orientations and single-slip ones can be clearly identified. In agreement with the same oriented nickel and silver single crystals \([72,83]\), the CSS curve of \([0 1 1]\) copper single crystal also shows an obvious plateau behavior except that the plateau stress is slightly higher than that of single-slip oriented copper single crystal. However in \([0 0 1]\) and \(\frac{1}{2} [1 1 1]/C_{22}\) copper single crystals, there is no plateau region in their CSS curves (see Fig. 19). Their respective saturation stresses increase monotonically with increasing the strain amplitudes. It can also be seen that the data from Gong et al. \([56]\) and Li et al. \([58]\) are consistent with the data from the present \([0 0 1]\) and \([1 1 1]\) copper single crystals very well, which further strengthens our understanding of the cyclic deformation characteristics in copper single crystals with different orientations.

The macroscopic deformation behaviors are dominated by their microscopic dislocation patterns. For example, the appearance of the plateau region corresponds to the formation of PSBs. Therefore, for the three typical multiple-slip oriented copper single crystals, various dislocation patterns are just the reason for the similarity and difference of the CSS curves. In the following, the formation mechanisms of different dislocation patterns will be discussed by taking into account the orientation effects, respectively.

2.5.2. Formation mechanism of wall structure from \([0 1 1]\) orientation

Fig. 20 presents the dislocation patterns of \([0 1 1]\) copper, nickel and silver single crystals. In addition to the PSB-vein structure, the wall structure is also frequently seen. The common feature in the three kinds of single crystals is the appearance of DBII, whose habit plane is \((1 0 1)\) or close to \((1 0 1)\) \([92]\). In other words, the habit plane of dislocation walls in DBII is also close to \((1 0 1)\). How do these...
wall structures in DBII form? And why is the wall structure made up of DBII easier to be seen in [0 1 1] fcc single crystals?

The primary slip system in fcc crystal is \( (111) \{110\} \) and the habit system of DBII is \( (101) \{111\} \) (see Fig. 21a), which manifests the conjugate relationship in crystallography between slip system and habit system. Based on the orientation relationship above (Fig. 21b), the resolved shear stress along the slip plane and slip direction \( \tau_s \) and the one along the habit plane and habit direction \( \tau_h \) can be expressed as follows, respectively:

**Fig. 20.** Dislocation configurations of [0 1 1] nickel, copper and silver single crystals: (a) nickel single crystal cycled at \( \gamma_{pl} = 6.1 \times 10^{-5} \), viewed from \( (111) \) plane; (b) copper single crystal cycled at \( \gamma_{pl} = 1.22 \times 10^{-5} \), viewed from \( (122) \) plane and (c) silver single crystal cycled at \( \gamma_{pl} = 6.1 \times 10^{-4} \), viewed from \( (2 1 1) \) plane.

**Fig. 21.** Schematic map of the resolution axial stress: (a) the habit planes of SBs and DBII, respectively and (b) decomposition of the axial stress \( \sigma \) based on different habit planes.
\[ \tau_a = \sigma \cos \varphi \cos \phi = \Omega \sigma \]  
\[ \tau_b = \sigma \cos \varphi \cos \psi = \Omega \sigma \]

where \( \varphi \) is the interacting angle between the loading direction and the normal direction of the glide plane; \( \phi \) is the interacting angle between the loading direction and the slip direction; \( \psi \) is the interacting angle between the loading direction and the normal direction of the best observation plane \((1\overline{2}1)\); \( \Omega \) is the Schmid factor. Comparing Eqs. (2) and (3), a basic judgment can be made that \( \tau_{DBII} = \tau_b = \tau_a = \tau_{SB} \) because \((1\overline{1}1)\)(\(101\)) and \((101)\)(\(1\overline{1}1\)) systems in fcc crystals have the identical resolved shear stress under the same axial stress. There are four equivalent slip systems with the same activating possibility in \([0\ 1\ 1]\) crystals, but one slip system will be always firstly activated. Assuming that the activating slip system is \((1\overline{1}1)\)(\(10\overline{1}\)), the formation of wall structures in DBII can be imagined as follows: under the action of the shear stress \( \tau_a \), the \([0\ 1\ 1]\) crystal begins to glide along the primary slip plane. With constant cyclic loadings, regular dislocation patterns – PSB-ladder structure finally forms. If the strain amplitude is further increased, a single PSB will not be able to carry the higher plastic strain and thus more PSBs will form. However, when PSBs cannot carry more plastic strain, the crystal makes adjustment by activating the secondary slip system, thereby the labyrinth or cell structures appear in the crystal. But sometimes the secondary slip system cannot be activated timely, especially in some local regions. In this case, the crystal will adopt another way to carry a greater plastic strain, leading to the formation of DBII [87]. As shown in Fig. 22a, during the development of DBII the SBs closely align, so the ladder structures in these S Bs move forth and back along the slip plane (see Fig. 22b). Under the action of the shear stress \( \tau_b \), there is always a moment for the PSB ladders to connect with each other to constitute shear plane, as shown in Fig. 22d. Along the habit plane of DBII, the shear strain \( \gamma_b \) appears and subsequently the wall structures form in DBII (see Fig. 22c).

Zhang et al. [87] found that DBII did bear much plastic strain than S Bs. In terms of the volume fraction of DBII, this deformation was roughly estimated to be \(7.5 \times 10^{-3} - 2 \times 10^{-2}\). Meanwhile the intrusion and extrusion caused by DBII are stronger than those caused by S Bs, which is in well agreement with the formation of wall structures in DBII. Until now, the formation of wall structure in DBII has been well explained. But why is DBII composed of the walls more prone to appear in \([0\ 1\ 1]\) crystals?

For \([0\ 1\ 1]\) or near \([0\ 1\ 1]\) fcc single crystals, Li [93] has pointed out that the habit plane of DBII is close to \((1\ 0\ 1)\) or \((0\ 1\ 1)\) or \((1\ 1\ 0)\). Furthermore, Li et al. [92] systematically summarized the angles between the habit planes of a variety of DBs and the loading axes in different oriented copper single crystals (see Table 2). The habit plane of DBI is the same as that of slip bands (S Bs). Accordingly, Table 2 provides such data as that in \([0\ 1\ 1]\) orientation, the angle between the habit plane of S Bs or DBII and

\[ \begin{align*}  
\tau_a &= \sigma \cos \varphi \cos \phi = \Omega \sigma \\
\tau_b &= \sigma \cos \varphi \cos \psi = \Omega \sigma 
\end{align*} \]

Fig. 22. Illustration maps of the formation of DBII: (a) developing DB; (b) well-developed DB; (c) PSB ladders in developing DB and (d) PSB walls in well-developed DB. (a and b) were quoted from Zhang et al. [87].
the loading axis has the maximum value of 60° and the habit plane of DBII and the loading axis has the minimum value of 24°. Li et al. [92] adopted the concept of dislocation avalanche factor to well explain the formation of DBII. They believed that the lattice distortion due to the tension–compression asymmetry is the main driving force to form DBII. And they also believed that DBII is more prone to appear in fcc single crystals with double- or multiple-slip orientations than single-slip orientation.

Wang et al. [94] and Wang and Laird [95] proved that when the primary slip system operated, the slip resistance of the secondary slip system would be greatly increased because of the latent hardening. Thus, generally speaking, only at higher strain amplitudes, significant secondary SBs can occur, our previous works [80] also proved this judgment.

For [0 1 1] single crystal, the saturation dislocation patterns are characterized by wall structure in DBII (see Fig. 20). As shown in Fig. 23a, in the stereographic triangle, no matter what kind of orientation, slip represents the movement mode of individual dislocation, however, shear aims at the movement of dislocation patterns, which is the main difference between the two deformation modes in [0 1 1] single crystal. Therefore, DBII consisting of dislocation patterns form, no more space on the primary slip system takes on the higher plastic strain. When the normal stress along [1 1 1] always imposes the action on the crystals. It can be imagined that firstly a large number of dislocations are produced on the primary slip system, but after the regular time the normal stress along [1 1 1] always imposes the action on the crystals. Therefore, DBII consisting of dislocation walls carries a higher plastic strain by the shearing motion in the absence of critical and coplanar secondary slip systems.

2.5.3. Formation mechanism of labyrinth structure from [0 0 1] orientation

Fig. 24 presents the dislocation patterns of [0 0 1] copper single crystal cyclically deformed at different plastic strain amplitudes. When \( \gamma_{\text{pl}} \) is \( 1.22 \times 10^{-4} \), the dislocation patterns consist of veil

<table>
<thead>
<tr>
<th>Loading axis</th>
<th>Habit plane of DBs</th>
<th>Angles between DBI and loading axis (( \varphi ))</th>
<th>DBII and loading axis (( \Phi ))</th>
<th>DBI and DBII</th>
</tr>
</thead>
<tbody>
<tr>
<td>[0 1 1]</td>
<td>DBI (0.49, 0.67, 0.56) DBII (−0.63, −0.18, 0.76)</td>
<td>60°</td>
<td>24°</td>
<td>90°</td>
</tr>
<tr>
<td>[0 3 4]</td>
<td>DBI (0.54, 0.64, 0.56) DBII (−0.59, −0.21, 0.78)</td>
<td>56°</td>
<td>30°</td>
<td>89°</td>
</tr>
<tr>
<td>[2 5 5]</td>
<td>DBI (0.70, 0.57, 0.50) DBII (−0.65, −0.76, 0.00)</td>
<td>30°</td>
<td>44°</td>
<td>89°</td>
</tr>
<tr>
<td>[2 3 3]</td>
<td>DBI (0.65, 0.57, 0.50) DBII (−0.44, −0.25, 0.86)</td>
<td>12°</td>
<td>35°</td>
<td>90°</td>
</tr>
<tr>
<td>[1 1 2]</td>
<td>DBI (0.65, 0.51, 0.57) DBII (−0.64, −0.07, 0.77)</td>
<td>24°</td>
<td>59°</td>
<td>89°</td>
</tr>
<tr>
<td>[2 2 3]</td>
<td>DBI (0.59, 0.60, 0.54) DBII (−0.64, −0.02, 0.77)</td>
<td>23°</td>
<td>59°</td>
<td>88°</td>
</tr>
<tr>
<td>[1 1 7]</td>
<td>DBI (0.65, 0.60, 0.46) DBII (−0.01, −0.63, 0.77)</td>
<td>26°</td>
<td>42°</td>
<td>88°</td>
</tr>
</tbody>
</table>
structures (see Fig. 24a). But the vein in [0 0 1] copper single crystals shows some directivity, which is different from that of single-slip oriented ones. On the other hand, similar to the single-slip oriented crystals, ladder-like structures also appear in some local regions of [0 0 1] copper single crystal, as shown in Fig. 24b. However, these PSB ladders disappear at higher strain amplitudes very quickly. For example, at the plastic strain amplitude $\gamma_{pl} = 2.5 \times 10^{-4}$, [0 0 1] copper single crystal is filled with the irregular labyrinth structures (see Fig. 24c). The emergence of these labyrinth-like structures hinders the formation of two-phase structure. With increasing the strain amplitudes, these dislocation structures will finally evolve into the regular labyrinth ones, as shown in Fig. 24d. In the (0 1 0) longitudinal section regularly spaced thin rungs parallel to (0 0 1) and (1 0 0) were found [51]. Both of sets have the same spacing of $\sim 0.50 \mu m$ and their intersection with (0 1 0) plane results in the labyrinth structure. Meanwhile, Jin and Winter [96] found that the walls parallel to (0 0 1) were much more common than that parallel to (1 0 0). The same phenomenon can also be seen in Fig. 24d. This indicates that the labyrinth structure has its own distinctive features – two sets of mutual perpendicular rungs. Then how do these rungs form?

Fig. 2a indicates the position of [0 0 1] orientation in the stereographic projection. It is known that [0 0 1] orientation has eight sets of equivalent slip systems with the same Schmid factor of 0.404, which can be activated simultaneously. Usually the more easily activated secondary slip systems are only conjugate and critical slip systems. However, Jin and Winter [49,50] pointed out that
conjugate dislocation reactions produced stable and non-slip Lomer–Cottrell dislocation locks. The reaction is so strong that the subsequent dislocation movement is obstructed. The obstruction made the dislocations in the different slip systems occupy clearly separated regions, in each of which a set of two-phase structure was formed. Therefore, the effect of the conjugate dislocation reaction on the formation of labyrinth structure can be negligible. On the other hand, our recent research [82] demonstrated that critical dislocation reactions produced sessile jogs, which also hindered the subsequent dislocation movement. But different from dislocation locks, the reaction intensity of sessile jogs is weak. Thus the dislocations belonging to the respective slip systems will overcome the resistance to keep the interaction with each other. They will not be strictly limited to their respective regions and affect each other in the process of the development. The formation of the labyrinth structure results from this effect.

Now it is clear that the formation of labyrinth structure is related to the activating of critical slip system and how the labyrinth forms by dislocation reactions. Fig. 25 illustrates the formation of labyrinth structure [96]. It is well known that (0 1 0) plane is the best plane to observe the labyrinth structure. On the (0 1 0) plane, two sets of PSB ladders belonging to the primary and critical second slip systems have formed, which accords with the movement and development of dislocations in their respective regions (see Fig. 25a). With cyclic deformation, the reaction between dislocations along \([\frac{1}{2}101]\) and \([101]\) will take place, as shown in Fig. 25b. The result of the reactions can be expressed below:

\[
\frac{1}{2}[101] + \frac{1}{2}[101] = [001]
\]

\[
\frac{1}{2}[101] + \frac{1}{2}[101] = [100]
\]

From Fig. 2a, it is well known that the slip directions of the primary and critical secondary slip systems are \([101]\) and \([1 0 1]\), respectively. The two slip directions correspond to the Burgers vectors of their
respective dislocations. By summing the above vectors, new Burgers vectors perpendicular to (0 0 1) and (1 0 0) planes will form (see Fig. 25b). Nowadays it should be clear that the activating of the critical secondary slip system and its interaction with the primary slip system are the basic processes for the formation of the labyrinth structure in [0 0 1] crystals. However, why (0 1 0) is the most suitable plane to observe the labyrinth structure? It can be seen from Fig. 2a that among all the orientations the normal direction of only (0 1 0) plane is perpendicular to the two slip directions of \( \frac{1}{2} C_{22}^{101} \) and \([1 0 1]\) simultaneously. Jin and Winter [96] further determined that rungs parallel to (0 0 1) were much more common than (1 0 0) rungs (see Fig. 25c), even some specimens showed (0 0 1) rungs exclusively (see Fig. 25d). The reason for the prevalence of (0 0 1) rungs is still unclear and further understanding will be necessary.

2.5.4. Formation mechanism of cell structure from \([\overline{1} 1 1]\) orientation

The dislocation patterns of \([\overline{1} 1 1]\) single crystal are very different from those of [0 0 1] or [0 1 1] ones. Fig. 26 shows the classical dislocation patterns in \([\overline{1} 1 1]\) copper single crystal. At low strain amplitude, the dislocation patterns mainly consist of vein structures (see Fig. 26a). With increasing the strain amplitude, the cell structure becomes the dominant dislocation patterns directly instead of the vein structure without the intermediate formation of PSB ladders (see Fig. 26b), which is the most different from that of single-slip oriented one. These vein and cell structures in fatigued \([\overline{1} 1 1]\) copper single crystal have a common feature that they show more wall structure on the observation plane \([1 3 2]\) (see Fig. 2a), which thus can be called “vein wall” and “cell-wall” structure, respectively, as shown in Fig. 26c and d. But these cell walls are not the same as those wall structures on \([1 1 1]\) plane. From Fig. 26d, it can be also seen that the cell structures with a certain direction distribute basically along the SBs. Then how do these cells form? The detailed discussion follows again from the stereographic triangle.

As shown in Fig. 2a, in the stereographic triangle, the primary slip system of \([\overline{1} 1 1]\) single crystal is also firstly activated, followed by the operating of conjugated and coplanar secondary slip systems.
Likewise, after excluding the effect of conjugated slip system, coplanar slip system will affect the formation of cell structure. Naturally, the second question is how the primary and coplanar slip systems interact with each other to eventually form the cell structure on (1 1 1) plane.

As shown in Fig. 27 a, the cell structures on (1 1 1) plane display more anisotropy. It seems that there is no rule to follow. In the discussion earlier, it has been mentioned that the orientation of [0 0 1] single crystal can be regarded as single-slip orientation fallen in one stereographic triangle because in the actual deformation process one primary slip system is always first to be activated and then followed by some secondary slip systems. It depends on the specific orientation of the crystal what kind of secondary slip system will be activated. For [1 1 1] single crystal, the easily activated secondary slip systems are the conjugated and coplanar slip systems. Also in the discussion earlier, it has been made clear that the activating of the conjugated slip system does not change the appearance of the plateau region and the formation of PSB ladders. Therefore, the coplanar slip system should play the main role in the formation of the cells. Fig. 2a shows that the primary and coplanar slip systems are (1 1 1)[0 1 1] and (1 1 1)[1 1 0], respectively. Jin and Winter [49,50] found that the dislocations belonging to these two slip systems will form the other coplanar dislocation on (1 1 1) plane as follows:

\[
\frac{1}{2}[101] + \frac{1}{2}[110] = \frac{1}{2}[011]
\]

Three sets of dislocations move and proliferate on (1 1 1) plane together and then their corresponding PSB walls along respective slip directions are produced firstly (see Fig. 27b). Meanwhile, it can be seen from Fig. 27b that the flip–flop motion of edge dislocations respectively with three Burgers vectors \( b_1, b_2 \) and \( b_3 \) is equivalent to the lattice rotation with [1 1 1] axis. It is well known that lattice rotation is coordinated by the generation of screw dislocations, which will form the twist boundary. Similarly, the screw dislocations appear along the edge of wall structures due to lattice rotation. These screw dislocations interact to form dislocation tangles as \( S_1 \) and \( S_2 \) shown in Fig. 27b, which will...
connect the edge dislocation walls $E_1$ and $E_2$ into a whole. By analogy, these wall structures will further form the network structure under the action of different types of screw dislocations, leading to the final formation of the cell structure with [1 1 1] axial rotation. Furthermore, just as (0 1 0) plane is the most suitable plane to observe the labyrinth structure, {1 1 1} plane can be regarded as the most suitable plane for the observation of the cell structure. Thus looking along $\frac{1}{2}[1 2 1]$ direction, only projection of the cell appears on $(1 1 1)$ plane as a result of the change of the observation plane, which is the reason why the cell-wall structures are observable or not (see Fig. 27c and d).

In summary, it has been confirmed that the cyclic deformation-induced classical dislocation patterns vary according to the different orientations of fcc single crystals. Compared with the structure of the ladder-like PSBs in single-slip oriented copper crystals, the labyrinth structure consists of two dislocation rungs (0 0 1) and (1 0 0) perpendicular to each other and (0 0 1) rung structures are usually found in the [0 0 1] crystals. In [1 1 1] crystals, the cell-wall structures are the most basic dislocation pattern and in [0 1 1] crystals, the wall structures in DBII are more prone to be observed. The habit plane of these walls is the same (1 0 1) plane as DBII. It should be also pointed out that the formation of the complex dislocation patterns depends on the activating slip system. If the critical secondary slip system is easier to operate, the labyrinth is more likely to be observed. If the coplanar secondary slip system is activated, the cell structure is more commonly seen. If there is no secondary slip system to be activated, the wall structure is more prone to appear. In short, for the dislocation patterns, in addition to the traditional slip systems, the shear systems with non-close-packed planes should be equally considered, which will help us better understand the effect of orientations on cyclic deformation behaviors, especially the formation of dislocation patterns, of fcc crystals.

3. Effect of slip modes on cyclic deformation behaviors of fcc single crystals

The cyclic deformation behaviors of different fcc pure metals have been introduced in Part 2. It can be found that the cyclic deformation behaviors of copper, nickel and silver single crystals with the

![Fig. 27. Cell structure of [1 1 1] fcc single crystal: (a and b) cell structure and its formation mechanism, (1 1 1) foil; (c) cell-wall structure, (1 3 2) foil; (d) illustration of formation of cell-wall structure, (1 2 1) foil. Quoted from Li et al. [38] and Lepisto et al. [53].](image)
same orientation indeed show more similarities. However, fcc alloy single crystals have their characters of the cyclic deformation behaviors. In the classical textbook “Fatigue of Materials”, Suresh [16] systematically summarized the research results of different kinds of fcc crystals in the past three decades [18,31,40,71,97–103], as listed in Table 3. At room temperature the saturation plateau stresses of copper, nickel single crystals are 27.5 MPa and 50 MPa, respectively, while the saturation plateau stress of silver single crystal is only 20 MPa [83]. These data show the obvious differences in their cyclic saturation stresses, which is just the external manifestation of these metals.

On the other hand, PSB-ladder structure corresponding to the plateau behavior originates from the metastable multipolar vein structure. The early experimental evidence for the metastability can be traced back to the research from Neumann and Neumann [104,105] and Basinski et al. [27]. The formation of the edge multipole veins can be attributed to the effective elimination of screw dislocations, depending on different metallurgical parameters, to represent a more suitable material parameter. In 2001, Hong [115] summarized the relationships between the CSS response and slip mode of copper–aluminium and copper–zinc alloys, with different aluminium and zinc contents, respectively. He suggested that the planarity of slip was promoted not only by low SFE but also by the increase in shear modulus, atomic size misfit and solution content. Accordingly, he developed a model for discriminating wavy and planar slip behaviors quantitatively. In the face of plenty of research results over nearly half a century [20,31,40,101–127], Wang [128] further concluded that although on many occasions, the value of SFE can be employed as the criterion for distinguishing wavy slip from planar slip, one cannot always obtain a satisfactory result if such a criterion for the transition is employed alone.

### Table 3

Cyclic stress–strain characteristic data of fcc metals. Quoted from Suresh [16].

<table>
<thead>
<tr>
<th>Materials</th>
<th>(\gamma_{pl,AB})</th>
<th>(\gamma_{pl,BC})</th>
<th>(\tau_0^{\text{pl}}/\text{MPa})</th>
<th>Reference</th>
</tr>
</thead>
<tbody>
<tr>
<td>Cu (523 K)</td>
<td>(1.0 \times 10^{-4})</td>
<td>(1.0 \times 10^{-3})</td>
<td>14.0</td>
<td>Li and Laird [99]</td>
</tr>
<tr>
<td>Cu (295 K)</td>
<td>(6.0 \times 10^{-5})</td>
<td>(7.5 \times 10^{-3})</td>
<td>27.5</td>
<td>Mughrabi et al. [18]</td>
</tr>
<tr>
<td>Cu (77.4 K)</td>
<td>(-)</td>
<td>(8.0 \times 10^{-3})</td>
<td>48.0</td>
<td>Basinski et al. [98]</td>
</tr>
<tr>
<td>Cu (4.2 K)</td>
<td>(-)</td>
<td>(-)</td>
<td>73.0</td>
<td>Basinski et al. [98]</td>
</tr>
<tr>
<td>Cu–2.0Al (at.%)(295 K)</td>
<td>(1.0 \times 10^{-4})</td>
<td>(3.0 \times 10^{-3})</td>
<td>33.0</td>
<td>Wilhelm and Everwin [100]</td>
</tr>
<tr>
<td>Cu–5.0Al (at.%)(295 K)</td>
<td>(-)</td>
<td>(-)</td>
<td>32.0</td>
<td>Woods [31]</td>
</tr>
<tr>
<td>Cu–16.0Al (at.%)(295 K)</td>
<td>(-)</td>
<td>(-)</td>
<td>20.0–25.0</td>
<td>Li and Laird [99]</td>
</tr>
<tr>
<td>Cu–2.0Co (at.%)(295 K)</td>
<td>(3.0 \times 10^{-4})</td>
<td>(5.0 \times 10^{-3})</td>
<td>27.5</td>
<td>Wilhelm and Everwin [100]</td>
</tr>
<tr>
<td>Ni (295 K)</td>
<td>(1.0 \times 10^{-4})</td>
<td>(7.5 \times 10^{-3})</td>
<td>52.0</td>
<td>Mughrabi et al. [18]</td>
</tr>
<tr>
<td>Ni (293 K)</td>
<td>(1.0 \times 10^{-4})</td>
<td>(8.0 \times 10^{-3})</td>
<td>50.0</td>
<td>Bretschneider et al. [71]</td>
</tr>
<tr>
<td>Ni (600 K)</td>
<td>(7.5 \times 10^{-5})</td>
<td>(5.0 \times 10^{-3})</td>
<td>20.5</td>
<td>Bretschneider et al. [71]</td>
</tr>
<tr>
<td>Ni (750 K)</td>
<td>(-)</td>
<td>(-)</td>
<td>12.0–16.0</td>
<td>Bretschneider et al. [71]</td>
</tr>
<tr>
<td>Ag (295 K)</td>
<td>(6.0 \times 10^{-5})</td>
<td>(7.5 \times 10^{-3})</td>
<td>17.5</td>
<td>Mughrabi et al. [18]</td>
</tr>
<tr>
<td>Al–1.6Cu (at.%)(295 K)</td>
<td>(1.5 \times 10^{-5})</td>
<td>(1.5 \times 10^{-3})</td>
<td>95.0</td>
<td>Lee and Laird [101]</td>
</tr>
<tr>
<td>Fe–11Ni–16Cr–2Mo (wt.%)(295 K)</td>
<td>(-)</td>
<td>(-)</td>
<td>59.0</td>
<td>Yan et al. [102]</td>
</tr>
<tr>
<td>Fe–19Ni–11Cr–2Mo (wt.%)(295 K)</td>
<td>(-)</td>
<td>(-)</td>
<td>59.0</td>
<td>Kaneko et al. [103]</td>
</tr>
</tbody>
</table>
On the basis of plenty of research results, the transition of slip mode can be identified as a key to determine the cyclic deformation behaviors of fcc single crystals. Slip mode includes planar slip and wavy slip. Compared with copper, nickel and silver characterized by wavy slip, planar slip is much common in concentrated solid solutions of copper such as copper–aluminium and copper–zinc. Then we will summarize the cyclic deformation behaviors of the two copper alloys, respectively.

3.1. Cyclic deformation behaviors of \( \alpha \) copper–aluminium single crystals

Apart from pure metals, fcc crystals also include alloys, such as copper-based alloy, nickel-based alloy, silver-based alloy and so on. The studies on the cyclic deformation behaviors of fcc alloy single crystals have primarily focused on copper-based alloys, including \( \alpha \) copper–aluminium, \( \alpha \) copper–zinc and others. Copper–aluminium single crystals have been studied most systematically and will be discussed first. Fig. 28 shows the binary phase diagram of copper–aluminium alloy [129]. When the aluminium content is less than 16 at\%, copper–aluminium alloys are single-phase solid solution. Depending on the aluminium content, the research works on copper–aluminium single crystals fall into three categories: copper–aluminium single crystal with low (\( \leq 5 \) at\%), medium (5 at\%–10 at\%) and high (\( \geq 10 \) at\%) aluminium content, respectively.

With respect to the CSS curves, copper–aluminium single crystals with low aluminium content show the cyclic deformation behaviors similar to copper single crystals. As mentioned earlier, Mughrabi [40] established the CSS curve of copper single crystals at the strain amplitude of \( 6.0 \times 10^{-5} < \gamma_{pl} < 7.5 \times 10^{-3} \) and found that the CSS curve showed a clear plateau region. Later, Wilhelm and Everwin [100] found that the strain amplitudes corresponding to both ends of the plateau in fatigued copper–2 at\% aluminium single crystals were respectively \( \gamma_{pl,AB} = 1.0 \times 10^{-4} \) and \( \gamma_{pl,BC} = 3.0 \times 10^{-3} \). Furthermore Woods [31] found that the CSS curve of fatigued [123] copper–5 at\% aluminium single crystals also showed the plateau behavior but the plateau region was shorter (see Fig. 29a). In cyclically saturated copper–5 at\% aluminium single crystals the ladder-like PSBs were still seen, as shown in Fig. 29b. For the saturation stress amplitude, there are different opinions. Woods [31] believed that the saturation stress of copper–5 at\% aluminium single crystals was the same as that in copper single crystals. However, Abel et al. [118] suggested that the saturation stress of copper–aluminium single crystals with low aluminium content was a little higher than that of copper single crystals, which resulted from the studies on cyclic response of copper–aluminium with aluminium contents of 2 at\%, 5 at\%, 7 at\%, 11 at\% and 16 at\%. They found that the initial cyclic...
hardening rates decreased and the saturation stresses increased with decreasing SFE. At lower SFE, fatigue fracture may develop prior to the occurrence of cyclic saturation. The copper–7 at% aluminium alloy exhibited higher stress amplitudes and hardening rates with increasing applied strain amplitude. At low strain amplitudes, fatigue fracture commenced before saturation stress reached. At the intermediate strain amplitudes, cycling tended to develop a horizontal segment in the hysteresis loop. At high strain amplitudes the cyclic softening occurs immediately after the cyclic hardening. This behavior is similar to that of copper single crystals, although at a higher stress level. In the case of the copper–11 at% aluminium and copper–16 at% aluminium alloys, cycling did not lead to saturation. The high cyclic stresses and the large energy absorption values to failure indicate that copper–aluminium alloys with high aluminium contents have a better fatigue performance than any of those with higher SFE [118].

Apart from Abel et al. [118], Wu et al. [130,131] systematically investigated the cyclic deformation behavior of copper–aluminium alloys with medium aluminium contents. In the range of $\gamma_{pl} = 1.1 \times 10^{-4} - 7.2 \times 10^{-3}$, copper–7 at% aluminium single crystals were cyclically deformed at room temperature. The results showed, similar to the case of copper single crystals, that the CSS curve of copper–7 at% aluminium single crystals exhibited a plateau region in the strain range from $1.1 \times 10^{-3}$ to $4.5 \times 10^{-3}$ (see Fig. 30a). The saturation stress of the plateau is about 27 MPa. Regular strain bursts occurred during

![Fig. 29. Cyclic hardening curves and saturation dislocation configurations of copper–5 at% aluminium single crystals: (a) cyclic hardening curves at different plastic strain amplitudes; (b) ladder-like dislocation structure, the foil $\|\{121\}$. Quoted from Woods [31].](image)

![Fig. 30. Cyclic hardening curves and saturation dislocation configurations of copper–7 at% aluminium single crystals: (a) cyclic hardening curves at different plastic strain amplitudes and (b) dislocation tangles, the foil $\|\{121\}$. Quoted from Wu et al. [130].](image)
cyclic deformation of the copper–7 at% aluminium single crystals. The occurrence of strain bursts strongly depended on the strain amplitude applied. At low strain amplitude the surface slip morphologies of copper–7 at% aluminium single crystals are similar to those of copper–16 at% aluminium single crystals. The slip traces are all smooth and straight. The slip appearance is characterized by a planar slip mode. The dislocation structure consists mainly of dipoles and multipoles (see Fig. 30b), which would induce the continuous cyclic hardening without saturation. At high strain amplitude, the slip morphologies are close to those of copper single crystals. The slip trace is tortuous and the three-dimensional dislocation structures induced by the wavy slip appear in the specimens. At intermediate strain amplitudes the slip appearance is characterized by both wavy and planar slip modes. Their interaction with each other is responsible for the cyclic saturation observed in the copper–7 at% aluminium single crystals.

In copper–aluminium alloy systems, the most systematic work was carried out on copper–16 at% aluminium single crystals. The early investigations on the fatigue behavior of copper–aluminium alloys can be retrospected to the 1960s. Feltner and Laird [106,107] found that the CSS curve of polycrystalline copper–7.5 wt% aluminium (~16 at% aluminium) alloy is sensitively dependent on the deformation and thermal history, in contrast to the history-independent CSS curve of copper polycrystals. The authors concluded that, whereas the cell structures in copper were almost independent of the initial condition of the material, the structure in copper–7.5 wt% aluminium depended appreciably on the initial material condition. In copper–7.5 wt% aluminium, the formation of cell structures is completely inhibited, and the dislocation structures are comprised of both planar arrays and dislocation dipoles and multipoles. Here, it is suggested, in consonance with [108,109], due to the lower SFE, the nature of planar slip, related to the difficulty of cross slip, in copper–7.5 wt% aluminium prevents the evolution of stable history-independent dislocation arrangements from the initial dislocation structures.

Since the 1980s, Yan et al. [102] and Buchinger et al. [132] began to regard copper–16 at% aluminium single crystals as the model material and carefully studied their CSS curves, surface slip morphologies and dislocation structures. Yan et al. [102] pointed out that copper–16 at% aluminium single crystal had a regular CSS curve with a plateau and the plateau stress was the same as that of copper at the strain amplitudes of \( \gamma_p = 1.5 \times 10^{-3} \)–\( 5.0 \times 10^{-3} \), which is somewhat similar to the later results obtained on [123] copper–16 at% aluminium single crystal by Wu et al. [130] (see Fig. 31a). It is interesting that the plateau stress at 77.4 K is roughly the same as that of copper at the same temperature. The strain localization is also found to occur in the alloy and shows itself as persistent Lüders bands (PLBs), which differ from the PSBs of wavy slip metals because of their different dislocation structure (see Fig. 31b). In the previous studies, Abel et al. [118] indicated that when the aluminium content exceeds 7 at%, the CSS curve does not show a clear plateau. And when the aluminium content is more

![Fig. 31. Cyclic hardening curves and saturation dislocation configurations of copper–16 at% aluminium single crystals: (a) cyclic hardening curves at different plastic strain amplitudes. Quoted from Wu et al. [130] and (b) persistent Lüder's bands (PLBs), the foil ||(1 2 1). Quoted from Laird and Buchinger [135].](image-url)
than 10 at%, the cyclic saturation and thus the plateau behavior would also disappear. Just as mentioned by Feltner and Laird et al. [106,107], the cyclic deformation behavior of copper–aluminium alloy with high aluminium contents depended strongly on the prior history. This is the key reason why the research results obtained can differ, depending on the alloy composition. In addition, based on the research of Yan et al. [102], Buchinger et al. [132] further investigated the dislocation structure of copper–16 at% aluminium at intermediate strain amplitudes. Except for PLBs mentioned by Yan et al. [102], they also found a finer dislocation structure – dipole array. These dipole arrays appeared on the (121) plane and arranged along the slip plane in patterns, also called glide plane structure.

At the beginning of 1990s, Hong and Laird [120–122,133] extensively studied the cyclic deformation behaviors of copper–16 at% aluminium single crystals including strain burst behavior, friction stress, back stress and surface slip morphologies. The results showed: (1) strain burst behavior existed widely in copper–16 at% aluminium alloy at various strain amplitudes and the cyclic hardening resulted from the strain burst; (2) the active life of each slip band in copper–16 at% aluminium single crystals was much shorter than that of copper single crystals. Since the localized strain moved around the gauge section because of the short active life of the slip bands, the overall deformation finally became rather homogeneous; (3) irrespective of strain amplitude, the gauge section eventually became completely filled with slip bands; (4) from the analyses of fatigue hysteresis loops, it is found that in the initial stage of cyclic deformation the friction stress (12 MPa) is larger than the back stress (7 MPa). With accumulation of cycles the back stress increases and eventually exceeds the friction stress. The cyclic hardening in this alloy is mostly caused by the increase of the back stress, which seems to be caused by the accumulation of multipoles and dislocation pile-ups frequently observed in this alloy. Based on these results, Inui et al. [134] further observed the dislocation structure of copper–16 at% aluminium single crystals. They found that at all stages of deformation, primary edge dislocations dominated the structure, forming dipolar and multipolar arrays which sometimes showed evidence of intersection along the traces of the cross-slip planes. With increasing stress level, the density of both primary and secondary dislocations increases and the spacing of the activated primary slip planes decreases. Primary screw dislocations are rarely observed to form multipoles. Numerous small prismatic loops are formed at the higher stress levels. Except for these loops, the dislocation structure of cyclically deformed copper–16 at% aluminium single crystal is similar to that of copper–aluminium alloy in monotonic deformation. The strain localization also occurs in copper–16 at% aluminium single crystals but in a way different from PSBs of copper single crystals, which might be the main reason for the CSS responses without saturation and plateau behaviors.

In 2001, Wu et al. [130,131] published detailed results on the cyclic deformation behaviors of copper–16 at% aluminium single crystals with different orientations including [123], [117] and [023]. Only when the strain amplitude is low enough (γpl ≥ 6.3 × 10⁻³), strain bursts are detected for all the three oriented single crystals. The frequency of strain bursts increased first and then decreased with increasing strain amplitude for all the three orientations. During the in situ observation, new SBs continued to form with strain bursts. The cyclic hardening rate is very low for all the crystals and the orientation and plastic strain amplitude have no effect on it. In the strain amplitude range of γpl = 1.1 × 10⁻⁴–7.0 × 10⁻³ the cyclic hardening curves of three different oriented copper–16 at% aluminium single crystals showed the saturation behavior. The cyclic hardening curves had a longer non-hardening stage than that of copper–7 at% aluminium single crystals. With increasing the strain amplitudes, the saturation stage disappeared and the crystals continued hardening. The deformation character above is the same for differently oriented crystals, so the effect of orientations can be ignored. The planar dislocation structure was observed in the three crystals. At low strain amplitudes, the space between SBs is so large that the interaction of dislocations on different slip planes can be neglected. At higher strain amplitudes, the space between SBs is smaller, the interaction of dislocation on different slip planes is stronger and saturation is no longer attained. On the whole, the cyclic deformation behaviors of copper–aluminium alloys exhibit more deviations from those of copper single crystals, as the aluminium content increases. A similar picture is obtained in the case of copper–zinc alloy in dependence on the zinc content.

Recently, the present authors studied the fatigue-induced dislocation structure of polycrystalline copper–16 at% aluminium alloy. As shown in Fig. 32, PLBs along a certain direction are most commonly seen in copper–16 at% aluminium alloy. Then the question is: how do these PLBs form? Since
the discovery of the PLB structure [102], there has been continued research on this particular structure. Laird et al. [132,134,135] considered that the dislocation structure in fatigued copper–aluminium alloy was the glide plane structure consisting of dipolar arrays of primary dislocations. At higher strain amplitudes, these structures coexist with more complex structures which have nucleated from the dipolar arrays, giving rise to persistent slip in PLBs, differing from regular PSBs composed of dipolar rungs. It can be seen from the comparison between Fig. 32a and b that the PLB structure is mainly composed of a finer substructure with stacking faults (SFs). These SFs interacted with much secondary dislocation activity and internal stress field, which eventually formed the PLB structure. Feltner and Laird [108] has mentioned that these PLBs corresponded well with the surface SBs. Therefore, it can be concluded that in copper single crystals with wavy slip characteristic PSBs are the classical dislocation structure corresponding to the surface SBs. With decreasing SFE, in copper–aluminium single crystals with planar slip features, PLBs become the most typical dislocation structure also corresponding to SBs.

In short, PLBs with persistent characteristics are the most commonly seen phenomenon in fatigued copper–16 at% aluminium alloy. Further analysis shows that these PLB structures consist of the dipolar arrays described by Buchinger et al. [132] or SFs proposed by the present authors, which represent the
finer substructure of PLBs. As shown in Fig. 33, the two types of substructures are the basis constituting PLB structure. So the question arises: what is the relationship between the SFs and dipolar arrays? This issue will be answered later.

3.2. Cyclic deformation behaviors of $\alpha$ copper–zinc alloys

Both single-phase copper–zinc and copper–aluminium alloys are copper-based solid solutions. The SFE values of both alloys are characterized by a significant decrease with the increase in alloying

Fig. 33. Two apparently different types of dipole arrays observed in polycrystalline copper–16 at% aluminium cycled at the plastic strain amplitude of $\varepsilon_{pl} = 3.0 \times 10^{-2}$: (a) regularly spaced array of dipoles and (b) stacking fault bands. The foil $[\{110\}$.

Fig. 34. The binary phase diagram of copper–zinc alloy [129].
element, which will lead to the change in slip mode. Thus copper–aluminium and copper–zinc alloys exhibit considerable similarities in their cyclic deformation behaviors.

Fig. 34 shows the binary phase diagram of copper–zinc alloy [129]. As the atomic weights of copper and zinc are roughly the same, mass percent can be approximately regarded as atomic percent. For the sake of consistency with previous scientific presentations, the alloy content of the copper–zinc alloys discussed in this paper will be stated in mass percent. As shown in Fig. 34, when zinc content is less than 30 wt%, copper–zinc alloy becomes replaceable copper–zinc solid solution. The relevant research on the cyclic deformation behavior of copper–zinc single crystals is still not much. The representative researchers are Lukáš et al. [109,123,136–138], Wang et al. [124,139–141], Gong et al. [142], Hu et al. [143–145] and Carstensen [146]. Their works will be introduced according to the time sequence.

Early research began with copper–30 wt% zinc single crystals. As in copper–aluminium alloys, Youssef [147], as well as Lukáš and Klesnil [109,136] did not find the classical PSB structure in copper–zinc alloy and suggested that the CSS curve of these alloys did not show a plateau region. For copper–22 wt% zinc single crystals, Lukáš et al. [123,137,138] indicated that there were four stages in their CSS curve and compared the curve with that of copper single crystals as shown in Fig. 35. The dislocation structures at different stages are all different. At stage I, only primary slip system is operated and on the primary glide plane there are only edge dislocations to interact. The CSS curve with a

Fig. 35. CSS curve of copper–22% zinc single crystals presented together with the CSS curve of copper single crystals; ● constant-amplitude tests, Amsler testing; ○ ascending step test, Amsler testing; ▲ constant-amplitude tests, MTS testing. Quoted from Lukáš et al. [137].

Fig. 36. Cyclic hardening curves and dislocation structures of copper–30% zinc single crystals deformed at various plastic shear strain amplitudes. Quoted from Gong et al. [141,142].
small slope at this stage is not sensitive to the strain amplitude. In stage II, besides the primary slip system, non-coplanar secondary slip system is also activated. At this stage, the alloy has a high dislocation density and the CSS curve has a larger slope than that in the first stage. In stage III, the activation of slip systems is similar as in the second stage, but SBs are produced by the primary slip system. These SBs gradually increase with increasing strain amplitudes, which corresponds to the plateau region at this stage. In stage IV, besides the SBs induced by the primary slip system, the secondary slip system also forms the SBs. The interaction between the two sets of SBs increases the slope of the CSS curve greater. In all stages, the dislocation patterns are of typical planar structure.

Hu et al. [143–145] systematically summarized the strain burst behaviors of copper–zinc single crystals with different zinc contents. They found that strain bursts were mainly associated with new slip which formed immediately adjacent to the previously operating slip. The hysteresis loops

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**Fig. 37.** Dislocation microstructure at the end of the primary hardening stage ($e_{cum} = 1$) of copper–30% zinc cycled at strain amplitude of $e_{pa} = 3.4 \times 10^{-3}$: (a) two apparently different types of dipole arrays; (b) the array marked with A in Fig. 30a is a regularly spaced array of dipoles and (c) the array marked with B in Fig. 30a appears to be a stacking fault band. Quoted from Carstensen [146].
can be divided into three different types of shapes I, II and III, at low, intermediate and large plastic resolved shear strain amplitudes, respectively. The authors also observed the PLBs and proposed that the propagation of coarse slip bands was associated with Lüders bands extension from several places along the specimen. Wang et al. [124,139,140] investigated mostly the multiple slip and cross slip phenomena and the corresponding slip morphologies. On this basis, Gong et al. [141,142] focused on the cyclic hardening curve and typical dislocation structure in fatigued copper–30 wt% zinc single crystals (see Fig. 36). At low strain amplitudes, the dislocation structure is characterized by dislocation segments and multipoles, very similar to that formed in stage I of tensile deformation in the same material. At high strain amplitudes, however, the dislocation structure is dominated by planar

Fig. 38. A closely spaced dipole array (a) with Moiré fringes. At higher magnification (b) the dipole structure perpendicular to the Moiré fringes can be seen revealing the characteristic “finger print”-structure. In (c) the structure is viewed at a weak beam condition to narrow the width of the dislocation images (the contrast has been inverted to compare directly with the bright field contrast in (b)). Faint traces of the dislocations in the lower slip plane appear halfway between the dislocations in the upper slip plane. Quoted from Carstensen [146].
dislocation loops and tangles. In this case a small amount of dislocation multipoles and a special zigzagged structure were also detected. For low strain amplitudes ($\gamma_{pl} < 3 \times 10^{-4}$), cyclic saturation occurred after an initial cyclic hardening stage, but at high strain amplitudes ($\gamma_{pl} \geq 6.0 \times 10^{-4}$) saturation could not be reached until fatigue failure. The initial cyclic hardening rate ($\theta_{0.2} = \Delta\tau/\Delta\gamma_{pl,cum}$, $\Delta\gamma_{pl,cum} = 0.2$) was found to decrease with increasing the applied strain amplitude.

In both copper–16 at% aluminium and copper–30 wt% zinc single crystals, the cyclically deformed dislocation structure is mostly a PLB structure composed of dipolar arrays or SFs. Concerning the relationship between the SFs and the dipolar arrays, Carstensen [146] gave some meaningful clues. He observed two apparently different types of dipole arrays in cyclically deformed copper–30% zinc alloy, which are marked with A and B respectively in Fig. 37 a. As shown in Fig. 37b and c at higher magnification, the dislocation array shown in A is a regularly spaced array of paired dislocations. Each pair is a dipole which can be established by looking at the fringes on the side of the individual dislocations. These fringes appear on one side of each dislocation due to the dissociation of the partials. The fringes are on opposite sides of the two dislocations constituting a pair, indicating that the dislocations are of opposite sign. Hence it appears that the paired dislocations are dipoles. The array shown in B displays a characteristic fringe pattern and thus is probably an SF band. However, a closer look at the structure leads to a different conclusion, as discussed below.

Thus, Carstensen [146] found that the fringe pattern like SFs represented an interesting structure perpendicular to itself at a higher magnification, as shown in Fig. 38. Fig. 38a suggests that the SFs are also an array of dipoles, but the dipoles are very closely spaced. Fig. 38b shows a bright field image of the fringe pattern. The resolution is not sufficient to enable a distinction between the individual dislocations in the dipoles, i.e. each dipole appears as if it was a single dislocation. By using the TEM weak beam technique the width of the dislocation images is reduced and the individual dislocations may be resolved as seen in the micrograph on Fig. 38c. The dislocations in the lower slip plane appear as faint lines halfway between the dislocations in the upper slip plane. The average dipole spacing, measured directly on the micrograph, is approximately 15 nm and thus the dipole height is assumed to be 7.5 nm. This assumption is based on a 45° dipole configuration, which corresponds to the position of minimum interaction energy of the dipole dislocations.

When comparing the dipole arrays in Figs. 37 and 38 differences in the dipole spacing are evident. The smaller the spacing between the slip planes (i.e. the dipole height) is, the closer the dislocations pack together. Perhaps it is just the difference in the dipole spacing that leads to different evolution track of dislocations and the eventual formation of PSB or PLB structure in various fcc crystals.

3.3. Effect of slip mode on dislocation arrangements in fcc single crystals

As shown in Fig. 39 [148], the planar slip and wavy slip are two extreme cases and intermediate cases could be possible. For planar slip, glide plane softening effect is the main reason why the

![Fig. 39. Stacking fault energy effects – wavy slip and planar slip. Quoted from Lawrence [148].](image)
dislocations follow other on the same plane. For wavy slip, glide spreads over several planes by independent cross-slip events. The features of cyclic deformation behaviors of typical planar slip materials, including copper–aluminium and copper–zinc alloys, have been investigated systematically as mentioned above [133,134,141,142]. It is assumed that the reduction of SFE value is the origin of planar slip behavior. This explanation has been widely accepted because a reduced SFE will lead to the difficulty of screw dislocations to cross slip onto other slip planes.

However, the systematic studies on the slip mode have shown that in fcc solid solutions with high solution content such as copper–16 at% aluminium and copper–30 wt% zinc and precipitation-hardened materials with a finely dispersed second phase such as nickel–aluminium and aluminium–lithium, in addition to the SFE, the SRO and the friction stress even become the primary factors affecting the slip mode. Hence, at different stages in the evolution of dislocation arrangements, a variety of factors play the decisive role, respectively. Taking pure metals and copper alloys for example, the fundamental formation mechanisms between their different dislocation arrangements will be discussed and compared according to the evolution steps.

3.3.1. Cross slip of screw dislocations and SRO

Before the more detailed discussion, first of all, dislocation distribution will be given for planar and wavy slip behaviors. As shown in Fig. 40a [149], the dislocation configurations of pure nickel are mainly edge dipoles which are arranged in bundles. In contrast, in Fig. 40b [149], dislocation structures of copper–10 at% aluminium are arranged in extended planar pile-up and multipole configurations. The difference of dislocation distribution depends on cross slip of screw dislocations. Back to Fig. 39, in the case of planar slip the glide plane softening effect occurring on both the primary and the cross-slip plane is the main reason why all dislocations emitted from the source S will follow each other even after a double cross-slip process. For comparison, in the case of wavy slip, cross slip of screw dislocations following each other now can occur on any plane.

Table 4 lists the ratio $c/l$ (SFE $c$ and shear modulus $l$), the occurrence of SRO and slip mode of three kinds of metals (silver, copper, nickel) and some of their solutions. It gives clear evidence that the occurrence of planar slip correlates with well-developed SRO and not with small values of $c/l$. Cohen

![Fig. 40. TEM micrograph of dislocation structures in wavy and planar slip materials: (a) nickel, (1 1 1) foil [170] and (b) copper–10 at% aluminium, (1 1 1) foil [149].](image)
and Fine [150] explained that the correlation between SRO and planar slip can be explained in the following way: the first dislocation moving through the lattice destroys this SRO. Since the order is only of a short range the following dislocations are not able to restore this order. Therefore, the first dislocation faces a higher resistance against slip than all the others.

Multipolar debris can be created by the cross slip of screw dislocations, as shown in Fig. 41. The dipole arrays are considered to be the first form on the “softest” slip planes, i.e. the slip planes with the lowest friction stress, and extend until these slip planes become “hard”. Then the deformation shifts to other primary slip planes to form more dipole arrays. The formation of the dipole arrays is suggested to be promoted by the presence of SRO domains. The ordered domains are in a lower energy level than the surrounding matrix so that if a dislocation passes through such a domain the order will be destroyed and the crystal structure shifts to a higher energy level.

As a quantitative parameter, the diffuse antiphase boundary energy $\gamma_{\text{APB}}$ can be used. Its calculation requires the knowledge of SRO parameters and pair interaction potentials. Schwander et al. [151] gave the generalized calculation scheme. Firstly, a short-range ordered crystalline binary alloy A–B can be described by giving the average concentration of B atoms, $c_B$, and the Warren–Cowley short-range order parameters $\alpha_i$ [152].

**Table 4**
The ratio $\gamma/\mu$, SRO and slip behavior of three metals (silver, copper, nickel) and their solid solutions [149].

<table>
<thead>
<tr>
<th>Alloy</th>
<th>Solution concentration (at.%)</th>
<th>$\gamma/\mu$ (10$^{-13}$m)</th>
<th>SRO</th>
<th>Slip behavior</th>
<th>Reference</th>
</tr>
</thead>
<tbody>
<tr>
<td>Ni–Fe</td>
<td>25</td>
<td>13</td>
<td>Yes</td>
<td>Planar</td>
<td>Karnthaler and Schügerl [111]</td>
</tr>
<tr>
<td>Ni–Cr</td>
<td>30</td>
<td>9.5</td>
<td>Yes</td>
<td>Planar</td>
<td>Clément [112]</td>
</tr>
<tr>
<td>Cu–Mn</td>
<td>12</td>
<td>10.3</td>
<td>Yes</td>
<td>Planar</td>
<td>Steffens et al. [171]</td>
</tr>
<tr>
<td>Cu–Al</td>
<td>15</td>
<td>1.1</td>
<td>Yes</td>
<td>Planar</td>
<td>Karnthaler et al. [172]</td>
</tr>
<tr>
<td>Ag–Li</td>
<td>25.2</td>
<td>5.25</td>
<td>Yes</td>
<td>Planar</td>
<td>Clément [112]</td>
</tr>
<tr>
<td>Ni</td>
<td>0</td>
<td>16.4</td>
<td>No</td>
<td>Wavy</td>
<td>Karnthaler and Fischer [170]</td>
</tr>
<tr>
<td>Ni–Co</td>
<td>0–68</td>
<td>16–1.6</td>
<td>No</td>
<td>Wavy</td>
<td>Mader et al. [173]</td>
</tr>
<tr>
<td>Ni–Cr</td>
<td>10</td>
<td>11.2</td>
<td>No</td>
<td>Wavy</td>
<td>Clément [112]</td>
</tr>
<tr>
<td>Cu–Mn</td>
<td>3</td>
<td>10.3</td>
<td>No</td>
<td>Wavy</td>
<td>Clément [112]</td>
</tr>
<tr>
<td>Ag</td>
<td>0</td>
<td>7</td>
<td>No</td>
<td>Wavy</td>
<td>Moon and Robinson [174]</td>
</tr>
<tr>
<td>Ag–Li</td>
<td>2.5</td>
<td>5</td>
<td>No</td>
<td>Wavy</td>
<td>Clément [112]</td>
</tr>
<tr>
<td>Cu</td>
<td>0</td>
<td>10.3</td>
<td>No</td>
<td>Wavy</td>
<td>Mughrabi [175]</td>
</tr>
</tbody>
</table>

**Fig. 41.** Perspective sketch showing how multipolar debris can be created by the cross slip of a procession of dislocations [176].
where \( P_{BB}^i \) denotes the conditional probability of finding a B atom in the \( i \)th shell of neighbors provided that there is a B atom in the center. Short-range order parameters may be determined from diffuse X-ray or neutron scattering data.

Furthermore, Varschavsky and Donoso [153] indicated that the theoretical value of APB energy on a \( \{1 1 1\} \) plane can be estimated as follows:

\[
\gamma_{APB} = 0.7 kT_c \xi^2 \frac{\gamma_{i}^2}{b^2}
\]

where \( k \) is the Boltzmann constant. Taking \( b = 0.26 \text{ nm} \) for copper–15 at% aluminium, we obtain \( \gamma_{APB} = 34 \text{ mJ/m}^2 \).

Table 5 shows that for copper–aluminium and copper–zinc alloys, the SFE value decreases with increasing the content of alloying element, but the antiphase boundary energy caused by SRO increases. The reduction of the former increases the difficulty of the mergence of partial screw dislocation into the perfect dislocation; the increase of the latter increases the resistance to the dislocation motion. Both cooperative effects make it difficult for screw dislocations to cross slip and dislocation motion is based on planar slip. Hence, the degree of difficulty of cross slip will determine the subsequent evolution of dislocation configurations and the final dislocation distribution.

3.3.2. Annihilation of screw dislocation and formation of edge dislocation dipole

Based on plenty of discussions on the role of cross slip [17–24,27,33,43,128], Mughrabi [26] summarized that cross slip is important in two ways: elimination of the screw dislocations and formation of the elongated dipole loops into shorter segments. First of all, the annihilation mechanism of screw dislocation will be introduced. Dislocations with opposite directions attract each other with elastic forces and if they are close enough to each other they can form a stable couple, i.e. a dipole. In Fig. 42, a schematic representation of a screw dislocation dipole is shown. The spacing between the
two dislocations is called the dipole height, and dipoles below a certain height collapse due to the mutual attraction of the dislocations. This height is called the annihilation distance, $y$. Screw dislocations collapse very easily as they are able to cross slip. The screw annihilation distance, $y_s$, is thus larger than the edge annihilation distance, $y_e$.

Taylor and Christian [154] proposed a model that cross slip was controlled by the interatomic plane distance $y_s$ for a double cross-slip process, which would lead to annihilation of screw dislocations. In this model, the larger the $y_s$ value is, the easier is the double cross slip leading to a wavy-slip mode. They suggested that the value $y_s$ was dependent on both the shear modulus and the critical resolved shear stress (CRSS). Mughrabi and Wang [114] further elaborated this idea and expressed the annihilation distance $y_s$ as

$$y_s = \frac{Gb \sin \alpha (1 - \cos \alpha)}{2\pi \sigma_y \left(\phi_p - \phi_c\right)}$$

Here, $G$ is the shear modulus, $b$ is the Burgers vector, $\alpha$ is the angle between the primary slip plane and the cross-slip plane, and $\phi_p$ and $\phi_c$ are the Schmid factors of the primary slip system and the cross-slip system respectively. The yield stress $\sigma_y$ takes into account the effect of alloying via the dislocation-solute atom interaction in single-phase alloys.

Cross slip is frequently induced by internal stresses and related in particular to the mutual attraction and annihilation of unlike screw dislocations on glide planes spaced less than a critical distance apart [155,156]. This distance, the screw annihilation distance, decreases not only with decreasing SFE, increasing strain rate, and decreasing temperature, but also with increasing yield stress and the presence of SRO. The parameter $y_s$ is thus considered to be more suitable for characterizing the slip mode (wavy or planar) than just the SFE [157]. In copper $y_s$ has been estimated to $\sim 50 \text{ nm}$ [158], whereas in Cu–30% Zn it is considered to be $\sim 5 \text{ nm}$ [114].

On the other hand, the cross slip of screw dislocations will bring the occurrence of edge dislocation dipole. During cyclic deformation the loading direction is continuously reversed and thus the dislocations are moving back and forth. A plausible multiplication process is shown in Fig. 43 [146]. A constriction enables an extended screw dislocation, moving in a $(111)$ primary slip plane, to cross slip onto a $(111)$ cross-slip plane. The dislocation spreads out in the cross-slip plane until a second constriction enables cross slip back to another $(111)$ primary slip plane. This double cross-slip process is shown in Fig. 43a. Extended screw dislocations are now spreading out in two $(111)$ planes under the action of a shear stress $\tau_0$ applied in the $(110)$ direction. The edge segments of the extended dislocations have line vectors of opposite signs in the two $(111)$. This results in two edge dislocation dipoles with a dipole height, $t$, corresponding to the spacing between the two $(111)$ planes. The dipole formation is shown in Fig. 43b.

Fig. 43. Multiplication mechanism for dislocations during cyclic deformation [146]: (a) an extended screw dislocation performs double cross slip from one primary slip plane to another and (b) extended dislocations now spread out in two primary slip planes under the action of an applied shear stress resulting in two edge dislocation dipoles.
3.3.3. Aggregation of edge dislocation dipoles and formation of PSBs

As mentioned above, in copper $y_e$ has been determined to be $\sim50$ nm at room temperature. The corresponding annihilation distance $y_e$ for pairs of unlike edge dislocations is about 1.6 nm [158]. Naturally cross slip will eliminate the screw parts so that the edge parts are mainly left [159]. Therefore, the dipole segment entirely composed of the edge dislocations will firstly form, as illustrated in Fig. 44a. With further cyclic deformation, some dipole segments aggregate continuously and spontaneously to form the more stable vein structure (see Fig. 44b). These veins were hardened to the extent until they can no longer accommodate the imposed plastic strain amplitude. Then parts of the edge dislocations in the veins decompose and glide to-and-fro and ultimately rearrange into the PSB-ladder structure, as shown in Fig. 44c.

The evolution processes of the dislocation patterns and the final formation of PSB-ladder structure were described above on a macroscopic scale. How is the evolution process of the corresponding dislocation patterns on a microscopic scale? Here, the formation of edge dipole segments should be firstly mentioned. Due to the formation of the dipole segment, the extended dislocation is constrained in a narrow region, which will definitely lead to the reduction of the width of extended dislocation to some extent. Next, as the macroscopic dislocation patterns transform from the dipole segment to the vein structure, the dislocation density is increasing [160]. Supposing that more dislocations will join the dipole segment, the width of the extended dislocation will be further compressed. Interestingly, the trap distance between the positive and negative edge dislocations will become smaller. As mentioned by Mughrabi [29], the local shear stress in the channel $\tau_{loc}$ is about 20 MPa and significantly smaller than the applied stress $\tau_s$. The local stresses in the vein and rung are about 40 and 60 MPa, respectively. Therefore, according to the estimation by Antonopoulos et al. [161], the primary edge dipoles are about 15 nm in height. And the average height of faulted dipoles was found to be about 5 nm. When the dipole segment transforms into the vein structure, the dipole includes the primary edge dipole and faulted dipole. When the vein evolves into PSB ladder, the PSB ladder is entirely composed of the faulted dipole.

3.3.4. Composition of classical dislocation patterns in fcc crystals

Based on the previous introduction, it can be found that the cyclic deformation behaviors of copper, nickel and silver single crystals have a great similarity. As shown in Fig. 45, the CSS curves of the three kinds of fcc single crystals with medium SFE usually have three regions A–C. The classical dislocation patterns corresponding to the plateau region are PSB-ladder structures [40,71,83]. In region A with
low strain amplitudes the vein or loop patch structure is commonly seen. And in region C with high strain amplitudes the corresponding dislocation structure is the labyrinth or cell [83]. The three plateau regions become gradually shorter and the corresponding plateau stress amplitudes become smaller in turn from 50–52 MPa for nickel [18,71], 28–30 MPa for copper [40,48], to 18–20 MPa for silver [83,84], respectively.

Since the macroscopic [26] deformation behavior of fcc single crystals is determined by the microscopic dislocation structure, it is necessary to understand the structural features of these dislocation arrangements. The first step is to grasp the characteristics of the PSB ladder, which is composed of the rungs with high dislocation densities and the channels with low dislocation densities. In the channels, screw dislocations straddle from rung to rung glide to-and-fro [17–24,31,33,40,41]. It is well known that dislocation movement in fcc metals is dominated by the dissociated dislocations. In the dipole rungs on the microscopic scale, each extended dislocation is composed of two dissociated partial dislocations. The spacing $w_s$ between the dissociated dislocations under such array can be estimated by the dislocation density $(10^{15–16} \text{m}^{-2})$ of the dipole rung. The results showed that the spacing and the width of the extended dislocations are $\sim 5$ nm and $3–5$ nm, respectively. On the macroscopic scale, the geometric structure parameters of PSBs will become important. In the PSBs of copper and silver single crystals: i.e. channel width $d_c$ of $1 \pm 0.2 \text{µm}$; rung thickness $d_w$ of $0.12 \text{µm}$ and PSB height $h_{PSB}$ of about $2–3 \text{µm}$. The similar structure parameters also appear in fatigued nickel single crystal [69].

Eventually, for the stable dislocation arrangements, as pointed out by Mughrabi [19,26], the bowing-out edge segments lead to an increase in the dislocation density by producing new screw segments across the channels, which then glide along the channels and deposit new edge dislocations in the rungs, as illustrated schematically in Fig. 46. At this time, the bowing out and annihilation of pairs of unlike screw dislocations and the occurrence of new edge dislocations and the annihilation of narrow edge dislocation dipoles in the rung constitute two kinds of dynamic equilibriums [43].

3.3.5. Criterion for formation of dislocation arrangements in fcc metals and alloys

Based on plenty of research results over nearly half a century [20,31,40,100,116–127], Wang [128] summarized the SFE value of various fcc metals in Table 6. The data quoted in this paper all derived
from Table 6. Firstly, the SFE and the extended dislocation of pure fcc metals will reach the stability and its width will also be determined according to the following relationship:

$$d_{\text{extend}} = \frac{Gb^2}{8\pi\gamma_{\text{SF}}} \left( \frac{2 - \nu}{1 - \nu} \left( 1 - \frac{2\nu}{2 - \nu} \cos 2\theta \right) \right)$$

(10)

Here, $\nu$ is the Poisson ratio, $b$ denotes the length of the Burgers vectors of the partials, $\theta$ is the angle between the Burgers vector of the perfect dislocation and the dislocation line. Hence, the SFE value of screw dislocations can be expressed as:

$$\gamma_{\text{SF}} = \frac{Gb^2(2-3\nu)}{8\pi d_{\text{ex}}(1-\nu)}$$

(11)

Secondly, considering the effect of the frictional force on slip mode, taking copper–15 at% aluminium as an example, the width of the extended dislocation in the copper–aluminium alloys can be expressed as:

![Dislocation distribution in the ladder structure in fcc metals](image-url)
where $\nu = 1/3$, the corresponding extended widths of screw dislocations can be obtained as follows:

$$d_{\text{extend}} \approx \frac{Gb^2}{8\pi(\gamma_{sf} - F_f)} \frac{2 - \nu}{1 - \nu} \left(1 - \frac{2\nu}{2 - \nu} \cos 2\theta\right)$$  \hspace{1cm} (12)

It is undeniable that the difference of the extended dislocation width causes directly the fundamental changes in the motion mode of dislocations. Meanwhile, the SRO effect should be also considered. Perevalova et al. [162] and Unal and Mitchell [163] have pointed out that the antiphase boundary $\gamma_{\text{APB}}$ energy can be calculated as:

$$\gamma_{\text{APB}} = \frac{Gb^2}{2\pi d_{\text{APB}} K}$$  \hspace{1cm} (14)

where $K$ is equal to 1 in the case of screw dislocation, $K = 1 - \nu$ in the case of edge dislocation, $d_{\text{APB}}$ is the distance between partial dislocation of a super dislocation. Compared with Eq. (8) based on the thermodynamic calculation, Eq. (14) derives from the interaction between the super dislocation pair. Therefore, the distance of screw partial dislocations can be shown as below:

$$d_{\text{APB}} = \frac{Gb^2}{2\pi \gamma_{\text{APB}}}$$  \hspace{1cm} (15)

Next, considering a basic fact that for pure copper, the extended width of screw dislocation is relatively narrow and screw dislocation is easy to merge and cross slip without the effect of antiphase boundary. For copper–15 at% aluminium, the extended width of screw dislocation is so wide that it is very difficult for screw dislocation to merge into the perfect dislocation. When these screw dislocations slip and face with the antiphase boundary, the dislocation will gradually pile up, which gives the dislocations opportunity to adjust their extended widths. As shown in Fig. 47, we can use a new parameter to describe the piling-up of dislocations as blow

$$F_{2-1}(d') = \gamma_{\text{SF}} + F_{2-\text{pu}}$$  \hspace{1cm} (16)

Here $F_{2-1}$ is the repulsive interaction force between partials 1 and 2 at a distance $d'$, and $F_{2-\text{pu}}$ is the resultant on the partial 2 of the forces due to the presence of the piling up. The dissociation distance is now $d'$ ($d' \neq d$). The effect of dislocation piling-up is equivalent to the increase of SFE value, so that the extended dislocation width will decrease and screw dislocations will have the opportunity to merge

---

**Fig. 47.** Effect of the piling-up on the dissociation width. (a) Equilibrium of one isolated dissociated dislocation and (b) equilibrium of the same dislocation at the head of a piling-up assuming that partial 1 is locked. Quoted from Pettinari et al. [178].
and cross slip on another slip plane. But on the other hand, different from the free cross slip on any slip plane of screw dislocations in pure metals, due to the piling-up of dislocations, the screw dislocation in copper–aluminium alloy will cross slip with the same height, which is just a reflection about the nature of planar plane.

Based on the above reasons, the two parameters, the extended dislocation width and the width of the antiphase boundary, can be compared. When the two parameters are equal, referring to Eqs. (13) and (15), the relationship among SFE, APB and frictional stress can be explored as:

$$\frac{\gamma_{SF} - F_f}{\gamma_{APB}} = \frac{3}{8}$$

Assuming that the resistance of dislocation movement is equivalent to the antiphase boundary energy, Eq. (17) can be simplified as follows:

$$\frac{\gamma_{SF}}{\gamma_{APB}} = 11/8$$

Compared with Table 5, it is found that the ratio exactly corresponds to $\gamma_{SF}/\gamma_{APB}$ from the copper–6.5 at% aluminium alloy, which is just such copper alloy with the slip feature from wavy to planar slip. Hence, as shown in Fig. 48, the following judgments on the ratio $\gamma_{SF}/\gamma_{APB}$ can be made: the competition between the two parameters directly results in the change of slip mode. Now, using another way, subtracting the two parameters, we will get:

$$\Delta \gamma = \gamma_{SF} - \gamma_{APB}$$

As illustrated in Fig. 48, first of all, $\gamma_{APB}$ of pure metals is zero, the slip mode of fcc metals is dominant by the SFE value. At the right side of Fig. 48, the SFE value can be regarded as the sole criterion on the formation of regular dislocation arrangements. Screw dislocations with high SFE cross slip easily

Fig. 48. New criterion on the formation of regular dislocation arrangements or not in different kinds of fcc metals or alloys by taking into account the effect of the slip mode and the energy difference $\Delta \gamma$. 
and PSBs, labyrinth and cell structures usually appear in fcc single crystals such as copper, nickel, silver and so on. At the left side of Fig. 48, with the increase of aluminium content, the antiphase boundary energy increases and the SFE value decreases. At this time, the slip mode of fcc alloys is affected by two factors of SFE and SRO. So dislocation movement has the features of both planar slip and wavy slip. The dislocation structures transforms from regular dislocation configurations to planar dislocation structures. With further increase of aluminium content, the SFE value continues to decrease and the dislocation movement depends on the planar slip under the influence of antiphase boundary energy. The dislocation arrangements are entirely composed of dipolar arrays or the SFS.

4. Conclusion and perspective

It is well known that the formation of PSB-ladder structure and the appearance of the corresponding plateau region in the CSS curves are two most typical features in cyclic saturated copper single crystals. Fcc metals with similar characteristics also include silver, nickel and copper–aluminium alloy with low aluminium content. However, considering more fcc metals or alloys (such as aluminium or copper–aluminium alloy with high aluminium content), it is found that both PSB ladders and the plateau disappear, which is distinctly different from the foregoing fcc metals. As a result, the fundamental factors on formation mechanism of dislocation evolution in fcc single crystals can be summarized as follows.

(1) **Effect of orientation**: Combined with the results of copper, nickel and silver single crystals, it can be concluded that the effects of the orientations on the cyclic deformation behaviors follow a general principle: the orientation-dependent dislocation configurations can be divided into three regions in the stereographic triangle, including the central region, [0 0 1] and $\frac{\sqrt{3}}{2}$111$ regions. Based on further research about the three multiple-slip orientations, it can be found that the formation of more complicated labyrinth, cell or wall structures depends on which slip system is priority to activate. The easy operation of the critical secondary slip system will contribute to the formation of labyrinth structure. The activation of coplanar secondary slip system will be beneficial to the formation of cell structure. If no secondary slip system is activated, the wall structure is more prone to appear.

(2) **Effect of slip mode (SFE and SRO)**: The CSS curves of copper–aluminium alloys have a shorter plateau with aluminium content increase. Meanwhile, the dislocation configurations change from PSB ladders in copper–aluminium crystal with low aluminium content ($\leq$ 5 at% aluminium) to PLBs in copper–aluminium crystal with medium, even high aluminium content ($\geq$ 8 at% aluminium). Studies show that PLBs mainly consist of the dipole array or SFS by dislocation reactions and interactions with other secondary dislocations. The cyclic deformation behaviors of copper–zinc alloy single crystals are similar to those of copper–aluminium alloys. Considering the pure metals and alloy together, it could be found that the factors on the formation of dislocation arrangements in fcc single crystals are different. For pure metals, the SFE value is the decisive factor. For fcc alloy crystals, with the increase of alloying element, the impact of the antiphase boundary energy gradually raises and becomes dominant. And ultimately the dislocation movement in fcc crystals changes from wavy slip to planar slip.

(3) **Perspective**: On the basis of the judgment that the SFE and SRO can be regarded as the core criterion on the formation of dislocation arrangements, the effect of temperature and frequency can be further considered as the external loading conditions. Particularly, the change of temperature affects both the SFE value and the value of antiphase boundary energy. For fcc crystals, the effect of temperature is not able to be ignored. Undoubtedly the most typical representative is aluminium. Pure aluminium single crystals with high SFE value is much easy to cross slip. The dislocation arrangements are composed of cell structure. However, recently the similar PSBs are observed in aluminium single crystals at very low temperatures [164], which should be caused by the effect of temperature because the SFE value changes with the alteration of temperature [165–167]. Thereby the aggregation mode of dislocation will be different. The related research
will be discussed in the next paper in order to obtain a unified model on explaining the formation mechanisms of dislocation arrangements in a variety of fcc metals and alloys in the near future [168].

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