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Low-cycle fatigue-cracking mechanisms in fcc crystalline materials

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The low-cycle fatigue (LCF) cracking behavior in various face-centered-cubic (fcc) crystalline materials, including Cu single crystals, bicrystals and polycrystals, Cu–Al and Cu–Zn alloys, ultrafine-grained (UFG) Al–Cu and Cu–Zn alloys, was systematically investigated and reviewed. In Cu single crystals, fatigue cracking always nucleates along slip bands and deformation bands. The large-angle grain boundary (GB) becomes the preferential site in bicrystals and polycrystals. In addition, fatigue cracking can also nucleate along slip bands and twin boundaries (TBs) in polycrystalline materials. However, shear bands and coarse deformation bands are observed to the preferential sites for fatigue cracking in UFG materials with a large number of GBs. Based on numerous observations on fatigue-cracking behavior, the fatigue-cracking mechanisms along slip bands, GBs, TBs, shear bands and deformation bands were systematically compared and classified into two types, i.e. shear crack and impingement crack. Finally, these fatigue-cracking behaviors are discussed in depth for a better understanding of their physical nature and the transition from intergranular to transgranular cracking in various fcc crystalline materials. These comprehensive results for fatigue damage mechanisms should significantly aid in obtaining the optimum design to further strengthen and toughen metallic materials in practice.

Keywords: fcc crystals; fatigue crack; dislocation; grain boundary; twin boundary; slip band; shear band; deformation band

1. Introduction

Face-centered-cubic (fcc) crystalline materials are widely used in commercial applications and are often grown to single-, bi- or poly-crystals to study their low-cycle fatigue (LCF) cracking mechanisms [1–4]. In these studies, fatigue-cracking behavior plays an important role in the fatigue failure processes, and these cracks were found to nucleate along slip bands, deformation bands, large-angle grain boundaries (GBs) and twin boundaries (TBs), respectively [2,4,5]. On the other hand, over the past two decades, nano-crystalline or ultrafine-grained (UFG) materials, produced by severe plastic deformation (SPD) method, have attracted increasing interest and have the potential of revolutionizing traditional materials due to their excellent mechanical properties [6,7]. These materials have strengths exceeding those
of their coarse-grained counterparts [7], but exhibit a lower fatigue life at room temperature, which limits their practical applications [8–10]. It is believed that fatigue damage and fatigue life strongly depend on the formation of shear bands in the UFG materials, i.e. the deterioration in LCF life can be attributed to the change of fatigue-cracking site from GBs or slip bands to shear bands [11–15]. To improve the fatigue properties of the UFG materials, the appearance of these interfaces and deformation areas, where it is easy to initiate fatigue cracks, such as shear bands, should be prevented. Therefore, two questions arise: what types of interfaces are preferential for the nucleation of fatigue cracks and why are the interfaces for fatigue cracking changed? However, to the best of our knowledge, these questions have not been answered satisfactorily by previous studies. Fatigue failure of industrial engineering materials or components often results in catastrophic events and, over the last 100 years or so, a large amount of research has been devoted to identifying fatigue damage behavior [11,16–20]. These studies have involved many materials from single crystals to polycrystals and even UFG materials.

In 1903, Ewing and Humfrey [21] were the first to observe the slip bands on the surfaces of rotating–bending Swedish iron specimens. These interesting observations led to other studies on the fatigue damage behavior of various crystalline materials. One important finding showed that, in pure Cu, the preferred location for fatigue cracking is the interface between the matrix and persistent slip bands (PSBs) [17,18,22–25]. Another study showed that fatigue cracking can nucleate along deformation bands (DBs) at high strain amplitude [26]; whereas, in previous studies, this cracking behavior was attributed to the plastic strain localization and high stress concentration induced by PSBs or DBs [16,23].

Fatigue cracking in polycrystalline materials and the cracking mechanisms have also been the subject of numerous studies over the past 30 years [1,27,28]. In addition to observations on fatigue cracking along slip bands, it was reported that GBs are often the preferential sites for fatigue-crack formation [1,28]. To better understand the fatigue-cracking mechanism along GBs, Kim and Laird [27,28] employed an optical microscopy and two-beam interferometry technique to study fatigue-crack nucleation in high strain fatigue of oxygen-free high conductivity (OFHC) Cu polycrystals, proposing a step mechanism for fatigue cracking at GBs. Mughrabi et al. [16,19] suggested another mechanism based on the interactions between PSBs and GBs in polycrystalline Cu. They hypothesized that the piling-up of dislocations at GBs is the reason for fatigue-crack initiation, due to irreversible slip during cyclic deformation. Recently, using various Cu bicrystals to clarify the GB effect on fatigue damage, a PSB–GB mechanism has been suggested and documented [20,29–32].

In addition to large-angle GBs, TBs, as important interfaces in polycrystalline materials, also play an important role in fatigue-cracking processes. Several studies have found that fatigue cracks are difficult to nucleate along TBs in polycrystalline copper [5,33,34]. Boettner et al. [5] considered that the formation of TB cracks during fatigue requires the operation of slip planes parallel to the TB, and that these types of fatigue cracks are rarely observed. Recent findings have shown that fatigue cracking can nucleate along TBs easily in materials with low stacking fault energy (SFE), such as Cu–Zn [35] or Cu–Al alloys [36]. However, TB cracking does not always occur in Cu–Al and Cu–Zn alloys and is strongly dependent on the amount of Al or Zn [35,36]. This poses another interesting question: are TBs strong or weak in
resisting fatigue cracking in Cu alloys? However, to date, there are no detailed studies on the fatigue-cracking mechanism along TBs.

Shear bands have also attracted much attention in nano-crystalline and UFG materials. Shear bands in cyclically deformed UFG Cu were first reported by Agnew and Weertman [11]. Other studies have shown that shear bands often orient at $\sim 45^\circ$ with respect to the cyclic loading axis, accordingly, fatigue cracks often initiating and propagating along the shear bands [15]. Wu et al. [14] recorded two types of shear bands in fatigued UFG Cu. They found that most shear bands appeared in the form of extrusions/intrusions, while other bands were step-like in appearance. The latest study has shown that a shear band is simply a thin sheet of tangled dislocations having different Burgers vectors [12]. However, the formations of shear bands and the related fatigue-damage mechanism are still unclear. The popular view attributes shear band formation to the oriented distribution of defects along the shear plane during the last pass of equal channel angular pressing (ECAP) [37]. However, it is unclear whether the streamline plane influences the formation of shear bands. Fatigue cracking can also nucleate along DBs in UFG materials at high strain amplitude [15]. To distinguish the DBs formed in polycrystalline materials, DBs in polycrystalline and UFG materials are designated as either slip- or shear-DBs in the present work.

As mentioned above, the position of fatigue cracks can vary among single crystals, bicrystals, polycrystals and nano-crystalline or UFG materials. To date, there has been no systematic investigation on the different fatigue-cracking mechanisms in fcc crystalline materials with different microstructures nor discussion on how fatigue-cracking interfaces vary between microstructures. This study systematically evaluates the various fatigue-cracking mechanisms associated with plastic strain localization and the different interfaces in fcc crystalline materials.

2. Experimental procedure

The materials used in this investigation are OFHC Cu, commercially purity Cu, Cu–10 wt% Zn, Cu–32 wt% Zn, Cu–5 at% Al, Cu–8 at% Al, Cu–16 at% Al and Al–0.7 wt% Cu alloys. Cu single- and bi-crystals were grown from the OFHC Cu of 99.999% purity by the Bridgman method in a horizontal furnace [4]. Polycrystalline Cu, Cu–10 wt% Zn, Cu–32 wt% Zn, Cu–5 at% Al, Cu–8 at% Al, Cu–16 at% Al alloys were obtained as cold-rolled/forged plates annealed at 800°C for 2 h in an Ar atmosphere. To obtain high dislocation density and UFG materials, pure Cu and Cu–32 wt% Zn and Al–0.7 wt% Cu rods were subjected to ECAP for one or four passes through a solid die with an angle of 90° between the two channels [15].

The samples of Cu single- and bi-crystals, polycrystalline Cu, Cu–Zn alloys and Cu–Al alloys were cut into fatigue specimens with a gauge size of $16 \times 5 \times 4$ mm, and the ECAPed Al–Cu alloys were cut into fatigue specimens with a gauge size of $16 \times 5 \times 5$ mm. These specimens were tested on a Shimadzu fatigue-testing machine under constant plastic strain amplitude control (controlled by an extensometer) with a strain ratio of $R = -1$ at room temperature in air. The samples of ECAPed Cu and Cu–32 wt% Zn alloy were cut into fatigue specimens with a gauge size of $6 \times 3 \times 2$ mm, and these specimens were tested on an Instron 8871 fatigue-testing
machine under constant stress amplitude control with a stress ratio of $R = -1$ at room temperature in air.

Before fatigue tests, all the specimens were polished to produce a mirror-like surface for fatigue-damage observations. After fatigue tests, all the specimen surfaces were observed with a LEO Supra 35 field-emission scanning electron microscope (SEM). Using an electron-channeling contrast (ECC) technique in SEM, dislocation arrangements of Cu single- and bi-crystals were observed and revealed. In addition, Cu–5 at% Al and Cu–16 at% Al samples were observed on a JEM-2000FX II transmission electron microscope (TEM) to reveal the interactions between dislocations and GBs or TBs in low- and high-SFE materials.

3. Experimental results

3.1. Fatigue cracking behaviors of Cu single crystal

In Cu single crystals, plastic strain is mainly concentrated in PSBs during fatigue [2], as shown in Figure 1a. With further cyclic deformation, those PSBs often become the preferential sites for fatigue-crack nucleation [38], as shown in Figure 1c. The SEM–ECC technique is a convenient way to display dislocation arrangements on a large area or specific sites [31,39], and the ladder-like dislocation pattern within PSBs is clearly visible in Figure 1b.

Fatigue cracking can also nucleate along slip-DBs at high strain amplitude [26]. When a Cu single crystal, oriented for single slip, was cycled at a near-maximum plastic strain amplitude in the cyclic stress–strain (CSS) curve, some intense slip-DBs of 50–60 μm in width were homogeneously distributed over the whole crystal surface, as shown in Figure 2a. To reveal the microstructure of the slip-DB in detail, a high magnification photomicrograph of a region in Figure 2a is shown in Figure 2b. As can be seen, the density of the slip bands within the slip-DB differs from that in Figure 1.
the matrix. The SEM–ECC technique shows that the dislocation arrangements within slip-DBs also differ from that in the matrix, as shown in Figure 2c. Normally, the slip-DB is a soft phase and carries more applied plastic strain than slip bands [20]. With continuous cyclic deformation, the fatigue cracks frequently nucleated along those slip-DBs, as shown in Figure 2d. Based on observations on fatigue cracking within slip-DBs, a new deformation mechanism along slip-DB was proposed to describe the strain localization in a Cu single crystal cycled at high strain amplitude [26]. It suggests that fatigue-crack initiation during cyclic deformation of Cu single crystals is always accompanied by intense plastic strain localization, either in PSBs or in DBs.

3.2. Fatigue-cracking behavior of Cu bicrystals

To investigate fatigue-cracking behavior along large- and low-angle GBs, we employed different Cu bicrystals with a GB perpendicular, parallel or inclined to the loading direction [4,40–42]. Surface observations revealed that PSBs could pass through low-angle GBs and the dislocation arrangement near low-angle GBs is consistent with the surface slip morphologies (Figures 3a and b). On the other hand, many PSBs were activated and terminated at the large-angle GBs, and the dislocations within PSBs are always piled-up near the large-angle GBs (Figure 3c). With further cyclic deformation, micro-cracks nucleated at the intersection of PSBs and GBs, as shown in Figure 3d. Figures 3e and f show typical intergranular fatigue

![Figure 2. Characteristics of cyclic deformation in a fatigued Cu single crystal II: (a) coarse slip-DBs and fine PSBs in a single crystal fatigued at a high strain amplitude of $\gamma = 8 \times 10^{-3}$, (b) magnified image of a, (c) dislocations pattern of slip-DBs observed by SEM-ECC and (d) fatigue cracks along the slip-DBs observed by SEM-ECC.](image-url)
cracking of the bicrystal with a perpendicular GB. Note that there are only primary slip bands near the GB with an obvious crack. With further cyclic deformation, fatigue cracks continuously propagated along the GB until fatigue failure occurred [4]. However, PSBs are never the sites of fatigue cracking. Thus, it can be concluded that intergranular fatigue cracking is the dominant damage mode; in other words, the existence of the large-angle GB will significantly decrease the fatigue lives of bicrystals due to the early occurrence of intergranular fatigue cracking.

3.3. Fatigue cracking behavior in polycrystalline Cu, Cu–Al and Cu–Zn alloys

Fatigue cracks on the surface of polycrystalline Cu, Cu–Al and Cu–Zn alloys, induced by cyclic deformation at different axial plastic strain amplitudes, are shown
in Figures 4, 5 and 6, respectively. The fatigue cracks nucleated frequently along the large-angle GBs, as shown in Figures 4a, 5a and 6a. Sometimes, fatigue cracks nucleated and propagated along slip bands, as shown in Figures 4b, 5b and c, and 6b–d. This phenomenon is in agreement with results in Cu bicrystals [4] and Cu polycrystals [1,43,44]. During fatigue-crack propagation, a GB crack occasionally propagated to a site where preferential cracking occurred along slip bands, changing its path and becoming a slip-band crack. Conversely, a slip-band crack propagated to a GB, becoming a GB crack. Via this mechanism, fatigue cracks can propagate along either intergranular or transgranular paths with the possibility of preferential cracking.

TBs are another special boundary across which atoms often have very good coherent relationships. A recent series of mechanical tests and microstructural observations have revealed that TBs can significantly improve the strength and ductility of nano-structured Cu with a high density of nano-scale twins [45–48]. Fatigued polycrystalline Cu, Cu–Al and Cu–Zn alloys were systematically observed to elucidate whether the TBs in these specimens with different Al and Zn contents were strong enough to resist fatigue cracking. Fatigue cracking along the TBs was not observed in pure Cu polycrystals; however, some fatigue cracks always nucleated along slip bands near TBs, due to strain concentration. This is consistent with previous results [5,33,34]. Fatigue-cracking behavior in Cu–Zn alloys is strongly dependent on Zn content. For example, in Cu–10 wt% Zn alloy, fatigue cracks

Figure 4. Fatigue cracking along (a) large-angle GBs ($\varepsilon_p = 1 \times 10^{-3}, N = 15,000$ cycles) and (b) slip bands in fatigued polycrystalline Cu ($\varepsilon_p = 4 \times 10^{-3}, N = 1000$ cycles).
initiating along TBs were rarely observed, but mainly originated from the slip bands near TBs, as shown in Figure 5c. However, for Cu–32 wt% Zn alloy, fatigue cracks were frequently observed to nucleate along the TBs, as shown in Figures 5e and f. In addition, there are no slip bands parallel to TBs near the TB cracks, which differs from the result of Boettner et al. [5]. On the other hand, micro-cracks were observed near the TBs, as shown in Figure 5d. Our previous studies have suggested that fatigue cracking is attributed to the distribution of these micro-cracks along the TB, whether they nucleated along TBs or not [35].

Fatigue-cracking behavior and the related dislocation patterns in Cu–Al alloys with high and low SFE were also observed, as shown in Figures 6c–f and 7. For example, in Cu–5 at% Al alloy, most dislocations could easily transfer through TBs by cross slip, as shown in Figures 7a and b. Fatigue cracks mainly nucleated along
these severely plastic-deformed slip bands, as shown in Figure 6c, which is consistent with the results for single- and poly-crystalline Cu. With increasing Al contents, fatigue cracks not only nucleated along slip bands near the TBs, but also appeared along the TBs in Cu–8 at% Al alloy, as shown in Figures 6d and e. However, for Cu–16 at% Al alloy, most dislocations were difficult to cross slip and were piled-up near the TBs, due to the decrease in their SFE, as shown in Figures 7c and d. The behavior of dislocations could be explained by the relative ease of cross slip in these materials. As shown in Figure 8, the slip planes in the matrix and twin are symmetrically distributed beside TBs. If the dislocations slip from matrix to twin, the slip direction will be changed by the cross slip. Therefore, the behavior of dislocations near TBs was affected by cross slip and fatigue cracks mainly nucleated along the TBs, as shown in Figure 6f. These results are similar to those for Cu-Zn.

Figure 6. Fatigue cracking behaviors in Cu-Al alloys: (a) LAGBs cracking in Cu–8 at% Al ($\varepsilon_{pl} = 5 \times 10^{-4}$, $N = 10,000$ cycles), (b) slip bands cracking in Cu–8 at% Al ($\varepsilon_{pl} = 4 \times 10^{-3}$, $N = 1000$ cycles), (c) slip bands cracking near TB in Cu–5 at% Al ($\varepsilon_{pl} = 1 \times 10^{-3}$, $N = 5000$ cycles), (d) slip bands cracking near TB in Cu–8 at% Al ($\varepsilon_{pl} = 4 \times 10^{-3}$, $N = 1000$ cycles), (e) TB cracking in Cu–8 at% Al ($\varepsilon_{pl} = 4 \times 10^{-3}$, $N = 2700$ cycles) and (f) TB cracking in Cu–16 at% Al ($\varepsilon_{pl} = 4 \times 10^{-3}$, $N = 2700$ cycles).
alloys and are related to different interactions between the dislocations and TBs in these alloys.

From these observations, it is concluded that fatigue cracks can nucleate in the following locations: (i) along large-angle GBs; (ii) along slip bands in the grain interior; (iii) along some TBs in Cu–Al or Cu–Zn alloy with high Al or Zn content. There is only one type of cracking interface in the grain interior, i.e. slip bands, and fatigue cracks always nucleate along such an interface. Locations (i) and (ii) have two different types of cracking interfaces, slip bands and GBs/TBs. Nucleation of fatigue cracks along GBs is a result of competition between slip bands and GBs, which can be explained by the PSB–GB mechanism [16,19]. There is also competition between slip bands and TBs for TB cracking, a result which is affected by the SFE and will be discussed further in Section 4.1.

Figure 7. TEM images of dislocation patterns: (a, b) Cu–5 at% Al ($\varepsilon_{pl} = 4 \times 10^{-3}$, $N = 1500$ cycles) and (c, d) Cu–16 at% Al ($\varepsilon_{pl} = 4 \times 10^{-2}$, $N = 2500$ cycles).

Figure 8. Sequence of the {111} plane near TB and the relationship of slip systems between twin and matrix.
3.4. Fatigue-cracking behavior in UFG Cu, Cu–Zn and Al–Cu alloys

A crystalline material pressed by ECAP has high density of dislocations and very fine grains, as well as numerous GBs [7], leading to complex fatigue-cracking mechanisms as slip bands are often restrained during cyclic deformation. Therefore, our understanding of fatigue damage and cracking mechanisms in such materials is limited. Here, the fatigue-cracking behavior in pure Cu, Cu–32 wt% Zn and Al–0.7 wt% Cu alloys subjected to ECAP was studied to elucidate their fatigue-cracking mechanisms.

Figure 9 shows the fatigue-cracking behavior of pure Cu and Cu–32 wt% Zn alloy after one-pass ECAP. There are two types of shear bands, i.e. PSB-like shear bands with extrusion/intrusion (Figures 9a and d) and normal shear bands without extrusion/intrusion (Figures 9b and e), similar to the observations by Wu et al. [8]. Many fatigue cracks also nucleated along the shear bands after cyclic deformation, as shown in Figures 9c and f. There are always obvious streamlines and a high density of dislocations, owing to the severe shear deformation, in the materials subjected to ECAP and the streamline often makes an angle of $\sim 27^\circ$ with respect to the extrusion direction [49]. However, the fatigue cracks shown in Figure 9 make an angle of $\sim 45^\circ$ with respect to the extrusion direction. This indicates that fatigue cracking in UFG materials did not originate from the streamlines caused by ECAP and is mainly associated with the maximum shear stress.

After cyclic deformation of the Al–0.7 wt% Cu alloy with four passes, fatigue cracks also nucleated along the shear bands at a plastic strain amplitude below $10^{-3}$, the direction of these bands makes an angle of $\sim 45^\circ$ with respect to the tensile direction, as shown in Figures 10a and b. After four-pass ECAP, the grain size of the Al–0.7 wt% Cu alloy was in the submicrometer range [15]. Nevertheless, these shear bands traversed the whole specimen and were at least several hundreds of
micrometers in length. Therefore, it can be concluded that the nucleation of shear bands is independent of the crystallographic orientation of the ultra-fined grains and that these shear bands cannot terminate at GBs, since their length is much longer than the grain size. On the other hand, the cyclic deformation morphology changed at a plastic strain amplitude greater than $10^{-3}$. A group of coarse shear-DBs can be seen on the surface of the specimen in Figure 10c. Fatigue cracks can also nucleate within coarse shear-DBs, as shown in Figure 10d. The coarse shear-DBs are perpendicular to the shear bands and are composed of a large number of parallel shear bands. These features are similar to those of slip-DBs in fatigued Cu single crystals [20], and the formation of the latter can be explained by locally irreversible rotation of the crystal [50]. However, the coarse shear-DBs in the ECAPed Al–0.7 wt% Cu alloy have no crystallographic feature and their formation only depends on the applied strain amplitude rather than local crystallographic rotation.

From the above observations, fatigue cracking can take place either along the shear bands at low or medium strain amplitudes, or along coarse shear-DBs at high strain amplitudes. Both of mechanisms can be regarded as a manifestation of strain localization and the GBs can not block propagation of shear bands. Therefore, GBs are slightly damaged under cyclic deformation and the fatigue cracks finally nucleated along shear bands or shear-DBs. In addition, these bands are independent of crystallographic orientation and parallel to the maximum shear stress plane of the specimens, which fulfils the Tresca criterion [51].

Figure 10. Fatigue cracking along: (a, b) shear bands at a plastic strain amplitude below $10^{-3}$ and (c, d) shear-DBs at a plastic strain amplitude over $10^{-3}$ in ECAPed Al–0.7 wt% Cu.
4. Discussion

4.1. Classification of fatigue cracking in fcc crystalline materials

Based on the above results, there are several fatigue-cracking interfaces, i.e. slip bands, slip-DBs, GBs, TBs, shear bands and shear-DBs. The cracking mechanisms along these interfaces are discussed in the previous publications [4,7,16,19,23,35]. For a better understanding of the fatigue-cracking behavior, it is necessary to classify the fundamental cracking mechanisms. In terms of cracking modes, the fatigue cracks can be classified into two types: i.e. shear crack and impingement crack. Shear cracks can occur along slip bands, shear bands and two types of DBs during cyclic deformation. Normally, the nucleation of shear cracks is associated with shear strain localization during the fatigue process. Impingement cracks often originate from interfaces, such as GBs or TBs, during cyclic deformation, induced by piling-up of dislocations within slip bands [4,19,45,52,53]. Thus, these interfaces would be impinged by slip bands under cyclic deformation and fatigue damage could form along them. The fatigue cracks nucleated along these interfaces were induced by the accumulation of such damage. Following the above classification, two types of fatigue-cracking mechanisms will be discussed in detail below.

4.1.1. Shear cracking

There are four different cracking interfaces in this type of fatigue-cracking behavior; the cracking mechanism along slip bands have been fully discussed [16–18,22–24]. In general, the fatigue-cracking mechanism along slip bands was explained by the stress concentration caused by extrusions/intrusions [16,23]. However, it is unclear why fatigue cracks nucleate along bands without obvious extrusions/intrusions, such as slip-DBs, shear bands or shear-DBs. Therefore, the similarities and differences between fatigue-cracking behavior needs to be discussed further.

Firstly, shear bands in ECAPed materials do not always nucleate with extrusions or intrusions, as in fatigued Cu single crystals [16,23]. Sometimes, shear bands are formed without obvious extrusions/intrusions and fatigue cracks could nucleate along them [14]. Even in shear bands with extrusions/intrusions, their extent is much smaller than in slip bands, as reported previously [16,23]. Therefore, this shear cracking cannot be fully explained by stress concentration near the extrusion/intrusion. Secondly, slip bands are always associated with the formation of extrusions/intrusions and different dislocation patterns are found in the slip bands and matrix [2]. Differences in microstructures could also contribute to fatigue cracking. As illustrated in Figures 11a and c, the microstructures and properties of slip and shear bands differ from the matrix, such as in the two-phase model of slip bands and matrix [2]. Thus, plastic strain localization within slip bands/shear bands is inconsistent with that of the matrix and this difference should drive fatigue cracking along slip bands/shear bands.

With increasing strain amplitude, the distribution of slip bands/shear bands is changed and some slip-DBs/shear-DBs (bands with a high density of slip bands/shear bands) are formed [15,26,50,54]. Because the density of slip bands/shear bands in slip-DBs/shear-DBs differs from that of the matrix, the microstructures and properties of these DBs also change, as shown in Figures 11b and d. The effect of
these DBs during the fatigue process is similar to that of slip bands/shear bands, which may induce plastic strain incompatibility and, finally, the nucleation of fatigue cracks.

As mentioned above, slip bands, shear bands and two types of DBs are formed during cyclic deformation, and plastic deformation in the fatigue process is exhibited by the formation of these interfaces. In addition, these shear cracks always induce different deformation behavior between the bands and matrix, which could cause plastic strain incompatibility near the interfaces and cause relative glide between them. Shear stress induced by inconsistent deformation makes a sizeable contribution to fatigue cracking and, thus, such cracking behavior is classified as shear crack.

4.1.2. Impingement cracking
Plastic strain is mainly concentrated in the slip bands/shear bands during fatigue, but GBs and TBs are hard to deform. However, as illustrated in Figure 12, GBs or TBs always act as barriers to block the movement of dislocations carried by slip bands, leading to differences in the piling-up of dislocations at GBs and TBs [4,20,40,45,52,53]. In this section, the impingement cracking mechanism will be discussed in terms of the different blocking effects of GBs or TBs during cyclic deformation.

Piling-up of dislocations at large-angle GBs has been widely observed in fatigued polycrystals; therefore, it is natural that the crystallographic relationship between GBs and slip systems of two neighboring grains can be illustrated as in Figure 12a. The activated slip bands can reach the GB, but cannot pass through it because the intersections between the GB and slip planes are not colinear [35,39]. Slip bands may become a carrier and channel to transport residual dislocations and vacancies from the interior of grains to the GBs. When the residual dislocations and vacancies accumulated at a large-angle GB have a sufficiently high density, intergranular
cracking along the large-angle GB will occur under external cyclic stress [20]. Sometimes micro-cracks nucleate along the GB, as shown in Figure 3b, and fatigue cracking along the GB will occur by coalescence of micro-cracks, as illustrated in Figure 12b. Consequently, intergranular fatigue cracking is attributed to the irreversible interactions of slip bands with GB, forming a piling-up of dislocations or accumulation of vacancies. In other words, the PSB-GB mechanism dominates intergranular fatigue cracking [4,20].

Figures 12c and d show that the TBs could also block the motion of dislocations, as in the case for conventional GBs [45,52,53]. Activated slip bands reach TBs, but have difficulty in crossing them because the slip bands need to change their slip planes, as illustrated in Figure 12c. Since slip bands play the same role near TBs, they could carry the dislocations and vacancies from the grain interior to the TBs. However, fatigue-cracking behavior near TBs changes with decreasing SFE and are mainly affected by the ability of dislocations to be piled-up.

To study the physical nature of the impingement cracking mechanism, interactions between slip bands and GBs/TBs need to be considered based on observations of dislocation patterns. There are four types of boundaries, low-angle GBs, TBs in high- or low-SFE materials and high-angle GBs, and their fatigue-cracking behavior is illustrated in Figure 13. The orientation relationship beside the low-angle GB is seen clearly in Figure 13; the slip plane only changes slightly when slip bands transfer through the boundary. The screw dislocations could easily cross it, but a new partial dislocation will be formed along the GB when the edge dislocations cross it, which is made more difficult by the increasing misorientation of the GB. When the

Figure 12. Illustration of impingement crack: (a) interactions of adjacent slip planes with a large-angle GB, (b) dislocations piled-up at GB or micro-cracks nucleating along GB, (c) interactions of the adjacent slip planes with a TB and (d) dislocations piled-up at TB or micro-cracks nucleating along TB.
misorientation is relatively low, the dislocations could cross low-angle GB freely, forming a continuous dislocation pattern beside it [40], i.e. the percentage of piled-up dislocations is approximately zero. Correspondingly, the orientation relationship and dislocation pattern beside a large-angle GB is qualitatively shown in Figure 13. The slip bands beside this type of boundary are obviously changed and the dislocations carried by slip bands could not cross this boundary, i.e. the percentage of piled-up dislocations is approximately 100% [40]. Between the two types of boundaries, the orientation relationship near TBs is also shown in Figure 13. The slip planes beside the TB are changed 60°/C14 and intersect at a straight line in TB. Therefore, the percentage of piled-up dislocations is strongly affected by their ability to change their slip plane. The perfect dislocation disassociate into two partial dislocations and the distance between them decreases with increasing SFE [55]. On the other hand, partial dislocations must be combined to perfect dislocations when they were piled-up at TBs via the following dislocation reaction:

$$\frac{1}{6}[\bar{1}\bar{1}2] + \frac{1}{6}[\bar{2}11] + SF \rightarrow \frac{1}{2}[\bar{1}01].$$

Figure 13. Illustration of fatigue-cracking behavior, orientation relationships, dislocation patterns and percentage of piled-up dislocations beside different GBs and TBs.
The difficulty of this reaction increases with decreasing SFE. To cross the TB, a perfect dislocation must dissociate into one dislocation in the twin and a partial dislocation at the TB [56–57]. For each dislocation, two reactions are possible:

\[
\frac{1}{2}[110] \rightarrow \frac{1}{2}[110]_T + 2 \times \frac{1}{6}[11\bar{2}], \tag{2}
\]

\[
\frac{1}{2}[110] \rightarrow \frac{1}{2}[101]_T + \frac{1}{6}[21\bar{1}]. \tag{3}
\]

These reactions require an increase in self-energy: 133% for Reaction (2) and 33% for Reaction (3) [56–57]. The energy required in Reactions (2) and (3) makes it more difficult for dislocations to pass through TBs than low-angle GBs. There is no difference between Cu and Cu alloys, however; the ability of dislocations to be piled-up along TBs in these materials is mainly affected by SFE. Therefore, partial dislocations in high-SFE materials could pass through TBs easily, i.e. the percentage of piled-up dislocations near a TB is low. However, partial dislocations find it hard to pass through TBs in low-SFE materials; accordingly, the percentage of piled-up dislocations is high, which easily induces TB cracking.

From PSB-GB mechanisms, fatigue cracking along GBs is mainly affected by piled-up dislocations. The percentage of piled-up dislocations increases from low-angle GBs, to TBs in high- and low-SFE materials and large-angle GBs, as shown above. Therefore, impingement cracking is mainly affected by piled-up dislocations, which are influenced by the orientation relationship beside the boundary and the ability to cross slip.

4.2. Relationship between fatigue-cracking behavior in various fcc crystals

Following the classification of fatigue-cracking types, the relationship and transition between their behavior will be discussed in this section. Plastic deformation involves slip and shear in these fcc crystalline materials during cyclic deformation. Fatigue cracking behavior is mainly caused, directly or indirectly, by plastic localization. Since the main differences in these fcc materials are grain sizes and density of dislocations, the relationship between slip/shear stress and grain size or density of dislocations should be discussed.

Firstly, the relationship between the applied stress and grain size is given by the Hall–Petch relationship [58,59]:

\[
\sigma_s = \sigma_0 + k \cdot d^{-\frac{1}{2}}. \tag{4}
\]

If the applied normal stress in changed to resolved shear stress, one can get

\[
(\tau_{\text{slip}} - \tau_0) = Kd^{-\frac{1}{2}}, \tag{5}
\]

where \(\sigma_s\) is the yield stress, \(\sigma_0\) and \(\tau_0\) are constants, \(k\) and \(K\) are two coefficients, \(d\) is the grain size and \(\tau_{\text{slip}}\) is the slip stress. In addition, the relationship between slip stress and the density of dislocations could be derived from the following
classic relation [60,61]:

$$\sigma = \sigma_0 + M\alpha Gb\sqrt{\rho}. \quad (6)$$

Similarly, one can also transfer the normal stress into the resolves shear stress as

$$\tau_{\text{slip}} - \tau_0 = C\sqrt{\rho}, \quad (7)$$

where $\sigma$ is the yield stress, $\sigma_0$ and $\tau_0$ are constants, $M$ is the average Taylor factor, which evolves in the process of straining (hereafter, $M$ is considered constant for simplicity), $b$ is the magnitude of the Burgers vector, $G$ is shear modulus, $\rho$ is the density of dislocations, and $C$ is a constant. Secondly, the stress driving a shear band cannot be expressed exactly by the grain size or density of dislocations, but their relationship can be deduced from the plastic deformation modes in different materials, as follows:

$$\tau_{\text{shear band}} \begin{cases} > \tau_{\text{slip}}, \quad \text{Single-, bi- and poly-crystals,} \\ \leq \tau_{\text{slip}}, \quad \text{ECAPed, UFG and nano-materials,} \end{cases} \quad (8)$$

where $\tau_{\text{shear band}}$ is the critical stress in forming one shear band. According to Equations (4)–(8), slip stress increases as grain size decreases or the density of dislocations increases. The shear stress driving the shear band should be higher than the slip stress in single-, bi- and poly-crystals, but smaller than the slip stress in the UFG- and nano-materials. Therefore, the relationship between fatigue-cracking behavior in various fcc crystals can be illustrated as in Figure 14. In this schematic, the competition between slip and shear bands influences the plastic deformation

Figure 14. Transformation of fatigue-cracking behavior in various fcc crystals: cracking transformed from shear crack to impingement crack, and then to shear crack.
modes and, finally, dominates the fatigue-cracking mechanisms. It can be seen from Figure 14 that shear cracking will occur in two zones: i.e. the slip band zone (plastic strain concentration within slip bands) and the shear band zone (plastic strain concentration within shear bands). In the slip band zone, GBs/TBs contribute significantly to fatigue cracking and the dislocations carried by slip bands could be piled-up by them. Therefore, the cracking behavior displays different modes in single-, bi- and poly-crystals [4,20,45,52,53]. In single crystals, fatigue cracking nucleates along the slip bands or slip-DBs in a shear cracking mode without the resistance of GBs/TBs. Due to the existence of GBs/TBs, however, fatigue cracking mainly nucleates along the large-angle GBs for impingement cracking mode in bi- and poly-crystals. In addition, fatigue cracking also nucleates along slip bands in the interior of grains and TBs in low-SFE materials [35,36]. In shear band zone, GBs/TBs no longer influence the nucleation and propagation of shear bands and these shear bands could cross GBs/TBs. Therefore, fatigue cracking often nucleates along shear bands or shear-DBs in shear cracking mode in SPDed materials.

As mentioned above, the LCF properties of different crystalline materials are strongly affected by fatigue-cracking sites. To improve the fatigue properties of various fcc crystalline materials, two aspects should be considered. Firstly, the grain size and density of dislocations should be controlled to avoid shear bands during the fatigue process. Secondly, increasing specific boundaries, such as TBs, will improve the resistance to fatigue cracking. In terms of recent reports that TBs simultaneously strengthen and improve the ductility of materials [44-47]; therefore, increasing the TBs in materials may have the potential to effectively resist fatigue cracking [35,36].

5. Conclusions

Based on observations of fatigue-cracking behavior in Cu single-, bi- and poly-crystals and some Al–Cu, Cu–Al and Cu–Zn alloys, the following conclusions can be drawn:

(1) In various fcc crystals, fatigue cracks nucleate along slip bands, slip-DBs, GBs, TBs, shear bands and shear-DBs, and the behavior can be classified into two types: shear cracking and impingement cracking.

(2) Impingement cracking behavior is mainly affected by piled-up dislocations, which is explained by the PSB-GB mechanism. The percentage of piled-up dislocations increases in the following sequence: (a) low-angle GBs, (b) TBs in high-SFE materials, (c) TBs in low-SFE materials and (d) large-angle GBs. Correspondingly, the possibility of fatigue cracking follows an increasing trend from low-angle GBs, to the TBs and the large-angle GBs in various fcc crystalline materials.

(3) Fatigue cracking behavior is influenced by competition between slip stress and shear stress. With decreasing grain size or increasing density of dislocations, the main fatigue-cracking modes in various fcc crystals change from shear crack to impingement crack, and then to shear crack.

(4) When critical shear stress is lower than slip stress, fatigue cracking will nucleate along shear bands. In contrast, lower slip stress facilitates the
nucleation of slip bands, and fatigue cracks finally nucleate along slip bands, slip-DBs in single crystals and GBs/TBs in bi- and poly-crystals.

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References