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Physics and model of strengthening by parallel stacking faults

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We have recently reported that parallel stacking faults (SFs) can tremendously increase the strength of a magnesium alloy. The strengthening is found to increase linearly with the reciprocal of the mean SF spacing, d. In this study we analyze dislocation interactions with SFs, and then propose a physics-based model to explain the observed relationship between yield strength and SFs spacing. Similar to the empirical Hall-Petch relationship for grain size, it is expected that this strengthening mechanism will hold true for a variety of materials engineered with parallel spaced stacking faults over a wide range of fault spacing. © *2013 AIP Publishing LLC*. [http://dx.doi.org/10.1063/1.4822323]

Traditionally, one or more of the following mechanisms have been used to strengthen metals and alloys: solid solution, precipitates, grain boundary, and cold-working. The precipitates, boundaries, or defects introduced by these strategies serve as barriers to dislocation motion, and consequently, increase the macroscopic strength of the material. Grain boundary strengthening has historically followed the empirical Hall-Petch relationship, where within a certain range of grain sizes (e.g., 20 nm to hundreds of micrometers[1–9]), the yield strength of metals and alloys increases linearly with the reciprocal of square root of grain size,[9–13] as represented by the equation $\sigma_{0.2} = \sigma_0 + \frac{k}{\sqrt{d}}$.[14,15]

More recent studies have shown that a twin boundary, which is a type of coherent internal boundary, can also be effective in blocking dislocation motion and when present in sufficient volume densities, enable increased strain hardening.[14] Lu *et al.* showed that the twin thickness (the spacing between two adjoining twin boundaries) can also be taken as a characteristic structural dimension that affects the yield strength in the same manner as grain boundaries for nanocrystalline Cu, i.e., follows the Hall-Petch relationship.[16–18]

In heavily deformed metals with sufficiently low stacking fault energies (SFE), another type of coherent planar defect, stacking faults (SFs), is commonly observed.[19–23] This is especially true for face-centered cubic metals when the grain sizes are smaller than a critical nanosize.[24,25] The formation of SFs involves dissociation of a full dislocation into partial dislocations that bound a planer stacking fault ribbon,[26] or emission of partial dislocations from grain boundaries.[25] Analogous to coherent twin boundaries, stacking faults should impede dislocation movement and thus strengthen the material. However until recently, there were no systematic studies or reports indicating this as a tractable mechanism.

Recently, a high density of SFs with nano-scale spacing was introduced in a Mg-8.5Gd-2.3Y-1.8Ag-0.4Zr (wt. %) alloy through conventional hot rolling.[27–30] Addition of certain alloying elements, such as Gd and Y, can significantly decrease the SFE of Mg alloys and, thus, promote the formation of SFs during deformation.[31,32] TEM investigation showed that the mean spacing of the SFs, *d*, decreased as the rolling strain increased. The nano-spaced stacking faults were demonstrated to make the main strengthening contribution to the observed ultrahigh strength (600 MPa ultimate tensile strength and 575 MPa tensile yield strength). More interestingly, as shown in Fig. 1, the room temperature yield strength of the Mg alloy, $\sigma_{0.2}$, increased linearly with the reciprocal of d (d^{-1} dependence),[27] expressed as $\sigma_{0.2} = \sigma_a + k \cdot d^{-1}$, instead of the commonly observed Hall-Petch relationship ($d^{-1/2}$ dependence) and where σ_a is the flow stress stemming from all other strengthening mechanisms other than parallel stacking faults, as marked by the blue arrow in Fig. 1.

In this paper, we elucidate the fundamental deformation physics that underlies this difference by first identifying the dislocations activated during the room temperature tensile testing of the Mg-alloy with nano-spaced SFs on the basal plane, and, based on the experimental observation, we propose a physical model to explain the d^{-1} dependent relationship between yield strength and SF spacing.

Hot rolling typically results in a strong texture in Mg alloys with the close packed basal plane parallel to the rolling direction.[33,34] Since the tensile testing direction is parallel to the basal plane in this study, deformation by contributions from basal and prismatic slip as well as extension twinning are minimized,[35] and ⟨c+a⟩ dislocations (Fig. 2) are necessary for slip on the pyramidal slip planes to accommodate overall plastic deformation.[36] Figs. 2(a)–2(c) show detailed TEM characterization using the two-beam technique on the 88% rolled sample after tensile testing. According to the visibility criterion, $\vec{g} \cdot \vec{b} \neq 0$, the ⟨a⟩ dislocation should be visible in a TEM image when the incident beam tilts to the two-beam condition with $\vec{g} = [11\bar{2}0]$ and ⟨c⟩ dislocation should be visible in the TEM image with $\vec{g} = [0002]$. Furthermore, the ⟨c+a⟩ dislocations should be visible in both the $\vec{g} = [11\bar{2}0]$ and $\vec{g} = [0002]$ conditions.[37,38] With this analysis, the TEM images show that ⟨c+a⟩

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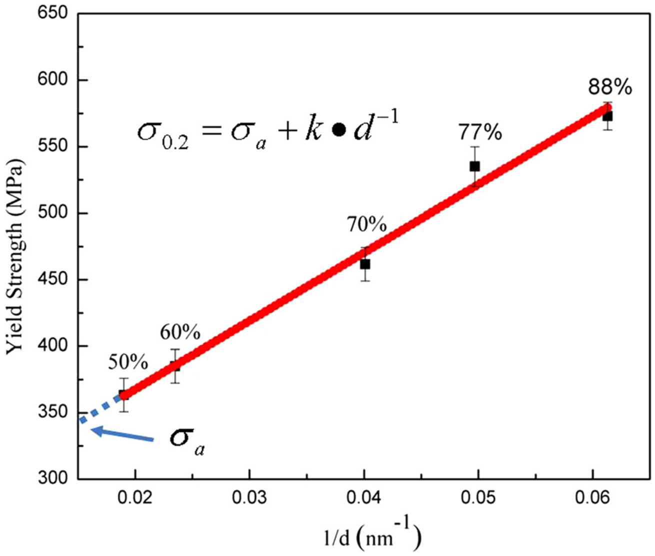

FIG. 1. Yield strength, $\sigma_{0.2}$, vs. the reciprocal of mean spacing between SFs, d, of rolled samples with different thickness reduction. The number by each data point indicates the thickness reduction by hot rolling. The yield strength of rolled Mg alloy can be expressed as $\sigma_{0.2} = \sigma_a + k \cdot d^{-1}$, where σ_a is the flow stress stemming from all other strengthening mechanism other than parallel stacking faults.

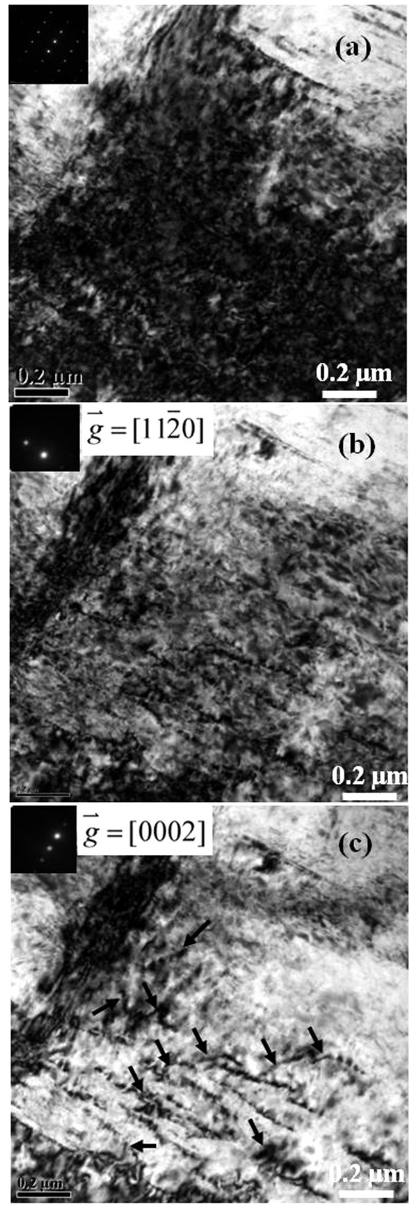

FIG. 2. TEM image of 88% hot rolling Mg-8.5Gd-2.3Y-1.8Ag-0.4Zr (wt.%) alloy samples: (a) after tensile test (b) and (c) two-beam bright field images with different g vectors indicated by the diffraction patterns. The non basal ⟨c+a⟩ are marked by short black arrows. Note that some and dislocations were also activated and this may due to slight orientation alteration of Mg grain during tensile testing.

dislocations (Fig. 2) were activated and cut the SFs on the basal plane during the tensile testing.[27] Recently, the observation of {c+a} dislocations being the main dislocations activated during deformation of a similar alloy has been reported, which validates our observations of the activation of ⟨c+a⟩ dislocations.[39]

In Mg alloys, ⟨c+a⟩ dislocations, such as $\frac{1}{3}\langle 1\bar{2}13\rangle$ type, usually slip on the pyramidal plane, $\{\bar{1}2\bar{1}2\}$,[36] as shown in Fig. 3(a). Here we assume that ⟨c+a⟩ dislocation is an edge dislocation and its dislocation line is straight. When a ⟨c+a⟩ dislocation encounters a SF and cuts through it (Fig. 3(c)), extra energy per unit length of dislocation line, E_S, is consumed.

For simplicity, assuming a dislocation with a unit length and a Burgers vector b glides a distance of x on a pyramidal slip plane (e.g., $\{\bar{1}2\bar{1}2\}$) as shown in Fig. 3. During this process, the dislocation cuts n SFs with an average spacing of d. According to Fig. 3, n can be calculated as

$$n = \frac{x}{d/\sin\theta}. \tag{1}$$

The extra energy consumed by cutting the n SFs can be calculated as

$$\Delta E = nE_S = \frac{xE_S}{d/\sin\theta}. \tag{2}$$

This extra energy is supplied by the extra work done by the extra applied resolved shear stress, $\Delta\tau$, to move the unit dislocation for the distance x, which can be described by

$$\Delta W = \Delta\tau x. \tag{3}$$

The extra work should be equal to the extra energy consumed by cutting the n SFs, i.e., $\Delta W = \Delta E$, which leads to

$$\Delta\tau x = \frac{xE_S}{d/\sin\theta}. \tag{4}$$

Simplifying Eq. (4), the extra applied shear stress that is needed to cut the stacking faults can be described as

$$\Delta\tau = k/d, \tag{5}$$

where $k = E_S \sin\theta$ is a constant.

Thus, the extra applied shear stress that is needed for a dislocation to move on a non-basal slip plane and cut through SFs is linearly proportional to the reciprocal of the mean spacing between adjacent SFs, d. Assuming that the shear stress needed for the dislocation to overcome all other sources of resistance, including lattice friction (Peierls stress),

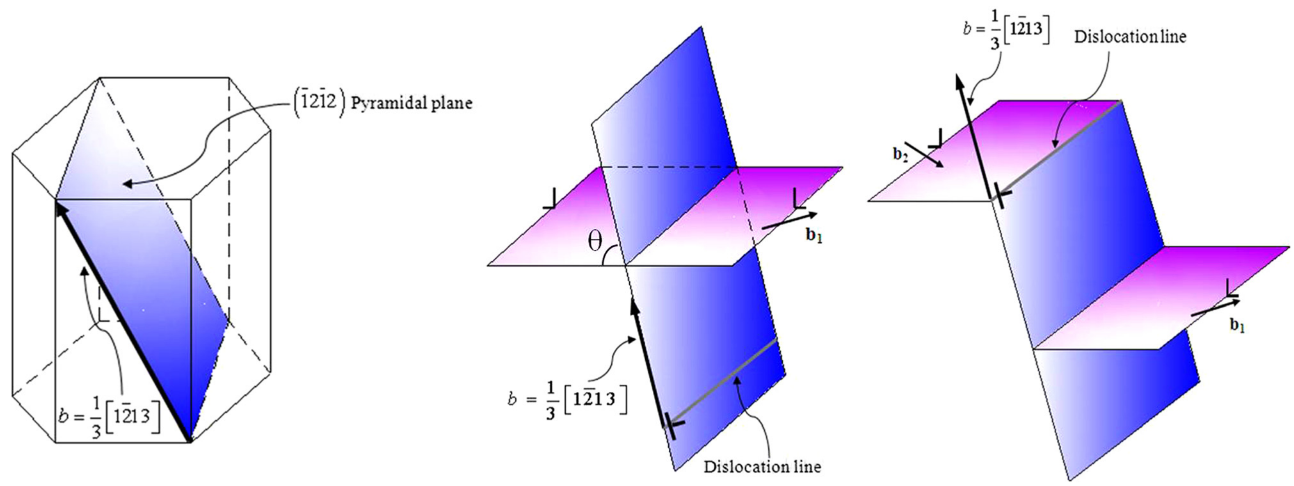

FIG. 3. Schematic illustration of (a) a ⟨c+a⟩ dislocation, (b) ⟨c+a⟩ dislocation motion and interaction with a basal stacking fault, and (c) ⟨c+a⟩ dislocation cutting through the stacking fault.

grain boundary, etc., is τ_a, the strength of an alloy with high density of SFs can be described as

$$\tau = \tau_\alpha + k/d. \tag{6}$$

Equation (6) can be also expressed in the form of normal yield strength

$$\sigma = \sigma_\alpha + k/d. \tag{7}$$

Equation (7) is similar to the Hall-Petch relationship except that the yield strength is dependent on d^{-1} instead of $d^{-1/2}$ as in the latter.

Note that the condition for Eq. (7) to be valid is that dislocations interact and cut through SFs. This equation is applicable to hcp metals, where the ⟨c+a⟩ dislocation slip can accommodate plastic deformation, and more so in our experiments where plasticity by basal slip and extension twinning is suppressed by the strong basal texture and the geometry of the loading conditions.[40,41] For this physical model, we assume that bulk plasticity occurs via the emission of ⟨c+a⟩ dislocations that subsequently cut the SFs on the basal plane. The conditions of validity for Eq. (7) may also only be when the grain size is relatively large ($> 1\,\mu$m). It is generally reported that as the grain size of an alloy is refined to the nanometer range (<100 nm), conventional lattice dislocation slip is suppressed and that plasticity is significantly mediated by deformation mechanisms that occur through the motion of grain boundary (GB) defects such as grain boundary sliding.[42–44] Theoretically, Eq. (7) should also be applicable to other crystal systems such as face-centered cubic (fcc) and body-centered cubic alloys as long as the stacking fault energy is sufficiently low, and the cutting of the generated SFs by dislocations is a significant deformation mechanism. Further investigations are ongoing to clarify these issues.

It should also be noted that some dislocations accumulated between segmented SFs during the room temperature tensile testing, as opposed to all cutting through and also, some dislocations were observed when the two-beam condition was set to $\vec{g} = [11\bar{2}0]$ (Fig. 2).[27] These observations suggest that dislocation pileup may also play some roles in the deformation and strengthening of the alloy. If the dislocation piling up near the SFs becomes dominant, the SFs become similar to twin boundaries in terms of impeding dislocation motion and enhancing dislocation accumulation, in which case the Hall-Petch relationship should apply.[16–18] It appears that this is not the case in the Mg alloy studied here.

In summary, we have posited a physical model for strengthening in materials with parallel stacking faults. This model is based on the physics of dislocations cutting stacking faults. It describes a linear relationship between the applied stress and the reciprocal of the mean spacing between stacking faults, and thus explains the d^{-1} dependence (d is the average spacing between stacking faults) observed in an Mg alloy with high density of stacking faults. It is expected that this model will be valid for any metallic alloy where cutting of stacking faults by dislocations is a significant deformation mechanism. This also points out a new strengthening and toughening mechanism for hcp metals and alloys: lowering stacking fault energy and choose appropriate processing parameters to form a high-density of parallel spaced stacking faults.

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