Source-controlled yield and hardening of Cu(1 0 0) studied by in situ transmission electron microscopy

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Abstract

In the present work we investigate the mechanical properties of multiple slip oriented single crystal Cu(1 0 0) compression samples to shed light on size-dependent yield and hardening behavior at small-scales. Samples with diameters ranging from 90 nm to 1700 nm were fabricated using focused ion beam milling and tested in situ in a transmission electron microscope. The results demonstrate a dislocation source-limited size-dependent yield strength, as evidenced by size-dependent changes in the deformation morphology. Moreover, we report size dependency and strain dependency in the hardening behavior at these dimensions, where higher hardening is observed for smaller samples and at lower strains. This is explained by the source-limited nature of plasticity in small dimensions, which we demonstrate affects not just yield but also the hardening behavior in the nanopillars.

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1. Introduction

In recent years, a great number of investigations have studied the effect of sample size on the yield strength and flow behavior of single crystal samples in the micro- and sub-micrometer range [1–3]. Ever since the first report of systematic studies using focused ion beam (FIB)-based sample fabrication [4], three striking phenomena have been commonly observed: (i) size-dependent (yield) strength, (ii) intermittent flow, and (iii) high hardening. Several computational studies and in situ experiments analyzed possible mechanisms and gave further insight. In the following, we will give a short summary of the current understanding, focusing on face-centered-cubic (fcc) metals as investigated in the present study. (i) The size-dependent yield strength is commonly explained by the source truncation effect or the source starvation/exhaustion effect. The former essentially argues that on average smaller samples contain smaller sources, while the latter argues that smaller samples contain fewer sources. The source truncation effect was proposed using a single-slip model by Parthasarathy et al. [5] and further developed for multiple slip by Ng and Ngan [6]. Experimental evidence for this was gleaned from in situ scanning electron microscopy (SEM) by Kiener et al. [7] and more powerfully using transmission electron microscopy (TEM) [8]. Oh et al. [8] directly showed the operation of single-ended spiral sources as predicted from the simulations. The source starvation/exhaustion effect has been deduced from both experimental [9,10] and computational studies [11]. (ii) The intermittent flow is manifested by load drops or strain bursts during the test and was identified as dislocation avalanches using statistical methods on experimental data [12,13] as well as simulation results [14]. Again, in situ TEM investigations [8,10] confirmed this explanation by direct observation. (iii) Finally, there is the notion of strong hardening for small sample sizes. This was reported for Ni by Dimiduk et al. [15] and Frick et al. [16], for Au by Volkert and Lilleodden [17], and for Cu by Kiener...
et al. [18]. Computational studies are limited, as three-dimensional discrete dislocation dynamics (3D-DDD) simulations are restricted in sample size and attainable strain. Nevertheless, Rao et al. [19] were able to identify the shutdown of dislocation sources as a main reason for what they termed exhaustion hardening. The limitations of 3D-DDD can be partly overcome using 2.5D-DDD, as such computations can capture the hardening at higher strains. Benzerga [20] reported hardening in the bulk stage II regime for micro-pillars and a transition from Taylor hardening to exhaustion hardening due to a reduction of the dislocation source density. Unfortunately, only a limited number of in situ experiments focused on the hardening in small dimensions. Exhaustion hardening has been reported in Ni [10], and a recent contribution by Maaß et al. [21] used micro-Laue to investigate Au and Ni pillars and stressed the importance of a proper yield-point determination to clearly differentiate between size effects on yield and hardening.

It is the primary aim of this study to investigate the hardening mechanisms and possible size dependencies using quantitative in situ pillar compression in a TEM.

2. Experimental

In this section we will describe the sample fabrication and present typical in situ compression experiments indicative of the observed deformation morphologies. In addition, the data processing to determine the yield point and true stresses during the test, taking into account geometric sample deficiencies resulting from the FIB machining, will be described in detail.

2.1. Sample fabrication

A thin piece of Cu was cut from a bulk (1 0 0) single crystal, grinded to a wedge and electro-chemically etched to remove the deformation layer and produce a sharp wedge that could be fixed to a sample support. A dual beam FIB workstation (FEI Strata 235; FEI, Hillsboro, OR, USA) operated at 300 keV. The compression tests were performed in displacement-controlled mode with nominal displacement rates in the range of 0.5–2 nm s$^{-1}$ resulting in strain rates of $\sim 5 \times 10^{-3}$ s$^{-1}$. The load–displacement data were read out with 1 kHz, and a video of each test was recorded using a charge-coupled device (CCD) camera (Gatan Orius SC200D; Gatan, Pleasanton, CA, USA) with a resolution of 480 pixel $\times$ 480 pixel running at 30 frames per second. Some pillars were unloaded and reloaded multiple times to record higher quality still images of intermediate deformation stages.

2.2.1. In situ compression test of a 136 nm diameter pillar – single slip

The before and after compression dark-field TEM images of a 136 nm diameter Cu(1 0 0) pillar using the $g_{002}$ diffraction vector are shown in Fig. 1 along with the measured load–displacement data. The total deformation was 200 nm at a displacement rate of 1 nm s$^{-1}$. The bright and dark rings of contrast along the perimeter of the sample are thickness fringes, but most of the mottled contrast in the pillar comes from defects [23]. The pillar shows some reduction of the FIB-induced surface defects during loading, the so-called mechanical annealing effect [10]. The deformation by dislocation slip is confined to the sample top and, despite the symmetrical slip orientation, limited to only one set of slip planes. From the in situ movie it is seen that plasticity starts at the very sample top, at the interface between the specimen and the flat punch, and continues on parallel slip planes with increasing distance from the interface with ongoing compression. The pronounced load drops shown in the load–displacement data (Fig. 1c) correlate to the activation of new parallel slip planes. The inclined contrast features in the deformed specimen are Moiré fringes due to a slight crystal rotation between the slipped and un-slipped part of the pillar.

2.2.2. In situ compression test of a 208 nm diameter pillar – alternating slip

Dark-field TEM images of a medium-sized 208 nm diameter Cu(1 0 0) sample before and after compression using the $g_{002}$ imaging condition are presented in Fig. 2a and b, respectively. Similar to the previous example, this specimen was compressed by 200 nm at a displacement rate of 1 nm s$^{-1}$. However, in this case alternating slip on two inclined sets of slip systems was observed, starting again at the pillar top and progressing further down the sample height with ongoing compression, as the stresses near the interface decreased due to the evolving mushroom shape. Note that there is also indication of a slight amount of slip close to the bottom of the sample in Fig. 2b. The small contact load followed by a total load drop at the beginning of the test (Fig. 2c) was caused by unwanted contact due to surface roughness of the diamond tip.
2.2.3. In situ compression test of a 535 nm diameter pillar – multiple slip

The before and after compression dark-field images of a 535 nm diameter Cu(1 0 0) sample are shown in Fig. 3a, b using a $g_{002}$ imaging condition. Because of the thickness of the sample, only the edges are electron transparent. The specimen was compressed by 400 nm with a displacement...
rate of 2 nm s\(^{-1}\). Dislocation activity was observed on multiple slip systems and led to significant barreling of the sample, comparable to what would be expected for a macroscopic Cu(1 0 0) single crystal. Notably, the load-displacement data in Fig. 3c display numerous load drops, but they are less pronounced in terms of the relative amount of load reduction compared to the smaller samples shown in Figs. 1c and 2c.

2.3. Evaluation of yield point and true stresses

As already noted in the introduction, there are recent publications questioning the extent to which the quotation of offset stresses is instrumental in order to discuss the origins of the observed size effects in small-scale testing [17,21,24]. The rationale behind this is the fact that at some percent plastic strain different hardening mechanisms (e.g. source truncation, exhaustion hardening, and forest hardening) can contribute to the measured stress. It was suggested to use micro-Laue to determine what was termed the Laue yield [21]. This, however, requires use of synchrotron facilities and cannot be routinely accomplished in a standard laboratory.

Here we suggest a simple procedure to determine the sample yield point by making use of an in situ TEM setup. We define yield as the point at which the first set of dislocations extend across the entire sample. The procedure is shown in Fig. 4 and explained in the following. Focusing to the first few nm of the load-displacement data (Fig. 4b), the first dislocation burst is identified. This gives the yield load \( F_Y \) and the yield displacement \( u_Y \). The critical diameter \( d_c \) (actual contact diameter or smallest sample diameter after slip step formation) is tracked during the test by analysis of individual video frames (Fig. 4c) and fit to an empirical equation to determine the diameter at yield, \( d_Y \), or any given displacement (Fig. 4d). Assuming that the out-of-plane deformation is well behaved and the sample cross-section can be reasonably described by a circle, this allows the calculation of true stresses in general and the yield stress in particular.

3. Results

The following section will first present the size-dependent yield strength of Cu(1 0 0) pillars and correlate this to the observed deformation morphology. Next the influence of sample size and plastic strain on the hardening will be shown in comparison to macroscopic values. Last, we provide a direct and quantitative in situ observation of a hardening mechanism operating in the nanometer regime.

3.1. Size-dependent yield strength and deformation morphology

In Fig. 5, we plot the yield stress from 24 separate Cu(1 0 0) samples compressed in situ in the TEM. Error bars were derived by an error propagation calculation from the experimental noise without contact during unloading for each experiment. Compared to earlier suggestions based on the experimental noise during a holding period under load [7], this is clearly a more conservative worst-case
approach, as the noise in contact is lower [25]. Thus, it is worth noticing that the error bars are still well within the symbol size for dimensions larger than $\frac{74}{150}$ nm. Clear size dependency of the yield strength is apparent, increasing from $\frac{74}{1243}$ MPa for the 1700 nm specimen to 1243 MPa for the 90 nm pillar. The different deformation behaviors identified in the previous section (single slip, alternating slip, multiple slip) are color coded and the inset images identify the individual pillars described in Section 2.2. In general, samples with diameters smaller than $\frac{180}{400}$ nm deform in the quasi-single-slip mode described in Section 2.2.1, while specimens with diameters above $\frac{400}{180}$ nm exhibit a bulk-like multiple slip deformation accompanied by significant barreling as shown in Section 2.2.3. The alternating slip deformation introduced in Section 2.2.2 was observed for the intermediate diameter range. Note that for completeness we included three tests in Fig. 5 that were run under load control (indicated by a red outline). No influence of the loading mode on the deformation morphology was observed within this limited data set.

3.2. Size and plastic strain dependence of hardening at a nanometer scale

In order to investigate the size dependency of the hardening behavior in small dimensions, we have evaluated the true stress, $\sigma_\varepsilon$, at 1%, 3%, 5%, 10%, and 20% plastic strain. Subsequently, the hardening, $\theta$, at the given plastic strains, $\varepsilon$, was calculated with respect to the yield stress, $\sigma_Y$, and...

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Fig. 4. Example showing the determination of the yield point, understood as the first bunch of dislocations extending across the sample, from the in situ data. Corresponding positions are indicated by a red arrow. (a) Measured load–displacement data. (b) Load, $F_Y$, and displacement, $u_Y$, at yield, respectively, are derived from the first discrete dislocation burst event. (c) The minimal sample diameter is tracked during the whole test. (d) Fitting the data from (c) with an empirical equation allows determining the contact diameter at any displacement, which is essential to evaluate true stresses.

Fig. 5. Yield stress of Cu(100) compression samples ranging in diameter from 90 nm to 1700 nm. Besides an established size effect on strength, a change in the deformation morphology as described in detail in Section 2.2 is shown, indicated by differently colored symbols and color bars indicating the regimes. The insets identify the individual pillars presented in Section 2.2.
normalized with the bulk modulus, $G$, to facilitate comparison to bulk hardening data by applying:

$$\theta = \frac{\sigma - \sigma_y}{\varepsilon} \cdot \frac{m^2}{G} \quad (1)$$

A Schmid factor, $m_{SG}$, of $m_{100} = 0.408$ and a bulk shear modulus of $G = 47$ GPa were used. The normalized hardening data are presented in Fig. 6a as a function of the pillar diameter and in b dependent on the plastic strain, where an additional color coding was used to indicate the sample diameter. The deformation regimes observed earlier (Figs. 1–3) are indicated in Fig. 6a as well. Changes of the sample diameter during compression were in both cases taken into account. Generally, higher scatter is observed at lower strain values (Fig. 6b). Moreover, the highest hardening rates are found in the smallest samples, especially for plastic strains below 3%. Notably, hardening decreases with strain, reaching values in the range of stage II bulk strain-hardening [26,27] for strains around 5–10% and even lower for higher strains. A few samples show negative values at low strains (for example a green and a blue symbol at $\sim$250 nm diameter) in Fig. 6a. This is due to large load drops shortly after yield from which the sample has not fully recovered at the given strain.

3.3. In situ observation of a nanoscale hardening mechanism

The shutdown of a dislocation source by dislocation interaction, leading to hardening in the sub-micron regime, is depicted in Fig. 7 using still images extracted from a dark-field video recorded during in situ compression of a 169 nm diameter Cu(1 0 0) sample (a–e) along with the measured load–displacement data (f). The $g_{002}$ diffraction condition and traces of active glide planes are shown in the inset in (a). The duration of operation of the individual sources is indicated by color bars in (f).

Initial deformation is to some extent governed by a partial loss of near surface FIB defects [23] escaping from the specimen during loading, a phenomenon termed “mechanical annealing” [10], resulting in a reduction of the dislocation density. Subsequently, a source or several adjacent sources situated close to the interface between pillar and flat punch near the upper left corner is/are operative on a (111) plane, but cease to operate due to an unknown event (b). Thereafter, another dislocation source situated close to the first one but on a (TT1) glide plane becomes activated at higher stresses (c). This operation is shut-down by interaction with a pre-existing dislocation, upon which this defect partly disappears (compare (c) and (d)). As seen in (f), an increase in the external stress is required to enable deformation by mobilization of dislocations situated closer to the sample base. At maximum load in (f) a new source is activated close to the upper right corner of the sample on a (1T1) plane, and a large slip event leads to a loss of contact between flat punch and pillar, as seen in (e) and the full unload in (f). This observation directly confirms the presence of dislocation-interaction-mediated exhaustion-hardening processes as observed by Rao et al. [19] for Ni and Senger et al. [28] for Al using 3D-DDD simulations. It should be noted that the above crystallographic analysis of the slip traces necessarily comes from the two-dimensional projection of individual frames during the in situ experiments, thus we have indexed the planes as best we can but cannot verify the exact inclination angle of the slip systems during the in situ test.

4. Discussion

In recent years, many reports on different scaling exponents in small-scale mechanics were published, see for example Refs. [1–3], and attempts were made to identify dislocation mechanisms based on such exponents. As a matter of fact, the scaling $\sigma \sim d^{-n}$ caused by common dislocation
The yield stresses of the larger compression samples reported here correlate well to published data for single-slip-oriented Cu(1 0 0) micro-tensile specimens [7] and extend the data towards smaller dimensions. A comparison between yield stress in the current study and 10% flow stresses reported in Ref. [7] is in this case meaningful because of the absence of hardening in the micro-tensile samples in this strain range. The scaling exponents were $n = 0.47 \pm 0.08$ in the case of the micro-tensile tests ranging from 0.5\(\mu\)m to 8\(\mu\)m [22], $n = 1.05 \pm 0.10$ for the present nano-compression tests, and $n = 1.04 \pm 0.08$ if the data from nano-compression and micro-tension are fitted together, as the power law fit is dominated by the small sample dimensions. One might consider that differences between these exponents could still arise due to the different testing setups used in the two studies, variations in the dislocation density of the single crystals [30], or the different data range, since fabrication parameters and data evaluation were comparable and the lateral stiffness low in both cases [22]. Also the difference in Schmid factors with $m_{100} = 0.408$ and $m_{136} = 0.422$ is negligible considering the experimental scatter typical for such experiments [15].

Possibly the scaling exponent changes with strain due to a hardening contribution. Evaluations at yield have been reported by Ng and Ngan [13] with $n = 0.92 \pm 0.06$ for Al(3 1 5) pillars ranging from 0.7\(\mu\)m to 6.5\(\mu\)m diameter and at 1% engineering strain by Dimiduk et al. [15] for Ni(2 6 9) with $n = 0.64$ for untapered samples ranging in diameter from 1\(\mu\)m to 22.2\(\mu\)m.

A recent comprehensive review by Kraft et al. [3] provided a comparison of data published on Cu from different
groups using various techniques. The authors conclude that different mechanisms are operative at three different size regimes, namely for dimensions \(>1\ \mu m\), for an intermediate sub-micron regime, and for dimensions \(<100\ nm\). The same three regimes and possible mechanisms have been independently proposed earlier by Kiener et al. [31]. These independent findings render a single exponent insufficient to describe the whole data and question the use of scaling exponents. Instead, we anticipate that further understanding of small-scale mechanics requires identification of the mechanisms, for example by analyzing the deformation morphology and resolving the dislocation structures and mechanisms as in the current study.

4.2. Source-limited deformation morphology

Before discussing the deformation morphology, some critical comments regarding the used geometry should be made. From the deformed samples shown in Figs. 1–3 it is seen that deformation always starts at the sample top. The inevitable taper and top rounding when fabricating pillars using annular milling results in higher stresses at the sample top. This is further amplified by the friction between diamond and specimen top, leading to a multi-axial stress state which promotes dislocation nucleation [18]. Other important factors are misalignment and lateral stiffness of the system. The main effect of misalignment is a reduction of the linear loading slope [32,33], which does not affect our data evaluation of the yield strength. Moreover, we observe close agreement between loading and unloading slope (see Figs. 1–3, 4 and 7), indicating good alignment. A high lateral stiffness affects the measured hardening rates, as free deformation is limited and hardening sets in immediately [7,22,26]. As stated earlier, our rather high aspect ratios \(>5:1\) minimize lateral constraints and the yield stress data fit well to published micro-tensile tests [7,22].

The deformation morphologies observed for the different sample sizes suggest that the deformation behavior is source-limited. Smaller samples, albeit being oriented for multiple slip, deform like single-slip-oriented crystals due to the limited number of dislocation sources in the sample volumes. The larger the samples, the more closely the observed deformation behavior comes to what is known from bulk Cu with a (1 0 0) orientation. This source-limited situation might become even more pronounced due to the taper of the pillars, as this creates a stress gradient over the sample height, thereby confining the actual deforming volume further [10,18].

An evaluation of scaling exponents for the present data subset that deformed in single slip \((d < 180\ nm, \ blue\ symbols\ in\ Fig.\ 5)\) gives \(n = 1.31 \pm 0.17\) for the yield stress, while the result for samples that deformed under multiple slip \((d > 400\ nm, \ black\ symbols\ in\ Fig.\ 5)\) yields \(n = 0.89 \pm 0.20\). While the exponents of these subsets do differ with respect to their statistical certainty, indicating a possible change in deformation mechanism, an interpretation attempt without in situ observation would seem rather daring.

While source-limited deformation explains the change in deformation morphology, it is interesting to note the rather low scatter in the experiments (Fig. 5) compared to yield data published for larger samples (e.g. [1,4,13,15,24]), as one might expect even higher scatter for smaller samples that deform under single slip. For example, it was shown by Bei et al. [24] that the observed scatter is increasing with fewer dislocations present in the sample, until the material becomes defect-free and deforms at the ideal strength [34,35]. While the probability of finding grown-in dislocations in our small samples is low, they initially contain a rather high density of FIB defects [10,23] and surface ledges [36] that can act as dislocation sources, given the stress levels of several hundred MPa to well above 1 GPa (Fig. 5). Such sources could not be activated in micro-scale FIB-fabricated Cu samples or in initially defect-free Cu whiskers during flow after yield, deforming at typical stresses in the 10–100 MPa range [2,3].

Other factors contributing to scatter are unwanted hardening contributions to the yield stress (see the variations in Fig. 6) or experimental inaccuracies. However, by the data evaluation procedure introduced above (Fig. 4) we eliminated hardening contributions and experimental errors from the instrumentation are of minor concern (see the error bars in Fig. 5).

4.3. Nanoscale hardening

As previously mentioned, besides the common notion of a size effect on the yield or flow stress in small dimensions [1–4], also the observation of (strain-)hardening was noted in several publications. Using the term strain-hardening in this context is debatable, as it refers to the well-established macroscopic strain-hardening concept of (fcc) metals, which is governed by strain-driven dislocation storage [37]. Micro- and sub-micro-scale post mortem TEM investigations [38] as well as computational studies [19,20] demonstrated that there is a breakdown of this Taylor hardening behavior. In small samples, due to limited slip distances and scarcity of dislocation interactions [9], not necessarily representative volume elements with bulk-like dislocation storage rates exist. We therefore use the neutral term “hardening” or quote the terms used by other authors when referring to their work.

Volkert and Lilleodden [17] examined Au and evaluated nominal strain-hardening between 5% and 15% plastic strain. They reported a hardening scaling exponent of 1.07 relating the strain-hardening rate, \(\theta_{5/15}\), and the undeformed pillar diameter at half height, \(d_{\text{mid}}\), \(\theta_{5/15} \sim d_{\text{mid}}^{-1.07}\). Frick et al. [16] investigated Ni(1 1 1) and reported strain-hardening rates between 3% and 10% strain in conjunction with a hardening scaling exponent of 1.0 with respect to the undeformed pillar top diameter, \(d_{\text{top}}\), \(\theta_{3/10} \sim d_{\text{top}}^{-1.0}\). For comparison, evaluation of the data in this study yields exponents of \(1.27 \pm 0.48\) following the procedure of...
Volkert and Lilleodden [17] and 1.34 ± 0.32 using the parameters applied by Frick et al. [16]. However, for better comparison to well-established bulk hardening data [26,27], we prefer to normalize our data with the bulk shear modulus.

We observe rather high hardening well above the bulk stage II range at low plastic strains (Fig. 6) caused by the shutdown of an operating source and the necessity to form or activate a new one in the limited volume, in most cases at higher stresses (Fig. 7). At low plastic strains the individual sources tend to operate only a few times before they cease to operate, a phenomenon that was also observed in 3D-DDD simulations using naturally formed dislocation structures instead of fixed pinning points [39]. This can be caused by changes in the stress distribution [8], reaction with other dislocations (Fig. 7 and Refs. [19,28,39]), movement of the pinning point [39,40], or the dissolution of the pinning point by unzipping [39]. As mentioned, in the current case the taper of the samples leads to a pronounced stress gradient during initial deformation. Therefore, the initial yield and hardening response is controlled by the deforming top portion of the pillar. With ongoing deformation, a larger sample volume is probed and the stress state becomes more uniform. The formation of more stable dislocation pinning points requires more complicated, thus less probable, dislocation reactions, which presumably can occur with increasing likelihood at higher strains, since larger volumes are probed. The observation of rather low hardening or even softening at higher strain values supports this view, indicating the formation of more stable dislocation sources, which are then capable of sustaining larger amounts of plastic deformation [29]. From these observations, exhaustion hardening [19] can be seen as the transition regime between yield and sustained plasticity by the formation of stable spiral sources [8].

While the hardening was determined as a function of strain and size in this work, also an evaluation of the dislocation densities as a function of strain as performed ex situ by Norfleet et al. [38] for micro-scale Ni specimens would be desirable. However, due to the geometric limitations of such pillar tests (e.g. taper, top rounding) and the resulting gradients [18], the dislocation densities in such sub-micro-scale samples are inhomogeneous and concentrated near the sample top [10]. Therefore, we did not attempt such an evaluation in the present study. It is, however, worth mentioning that Oh et al. [8] performed an in situ TEM straining experiment of a 450 nm thick single crystal Al wire with constant cross-section. They observed a constant dislocation density of \( \sim 5 \times 10^{13} \text{ m}^{-2} \) up to strains of 160%, while Norfleet et al. [38] report a moderate reduction in dislocation density from initial \( \sim 5 \times 10^{13} \text{ m}^{-2} \) to \( \sim 3 \times 10^{13} \text{ m}^{-2} \) after \( \sim 45\% \) strain for 2 \( \mu \text{m} \) diameter Ni pillars. The strain rate of \( 1 \times 10^{-4} \text{ s}^{-1} \) was in both cases lower than in the present study.

Finally, an interesting observation emerges from comparison of the deformation morphology indicated by the color bars in Fig. 5 and Fig. 6a to the measured strain-dependent hardening as shown in Fig. 6b. From the shape of the deformed samples one could expect stage I hardening for the small samples exhibiting single slip, and stage II for the larger alternating and multiple slip specimens. However, this is not reflected in the strain-dependent hardening data in Fig. 6b, where on the contrary the small single-slip samples exhibit high stage II hardening values in the range of 5–10% strain and even much higher at lower strains. This indicates, as elaborated before for the yield stress scaling, that hardening rates alone provide no insight into the underlying mechanism.

5. Conclusion

Our results demonstrate that in situ experiments provide the required information to understand small-scale dislocation mechanisms, while scaling exponents linking stresses and sample dimensions are interesting mainly from an engineering standpoint. Furthermore, it is important to realize that, besides yield, also hardening in small dimensions is primarily governed by a limitation of available dislocation sources, and not solely by the interaction of slip systems as in the macroscopic world. This is clearly revealed by the transition from multiple slip to single slip for the (100) oriented Cu samples with decreasing pillar diameter. Moreover, the hardening contribution to the observed size effect on flow strength is most pronounced at low strains and gets reduced with increasing plastic strain, as hardening is highest for small samples and low strains and reduces with size and strain.

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