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Contrasting the nature of plastic fluctuations in small-sized systems of BCC and FCC materials

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Abstract

It is now well established that, upon decreasing system sizes down to a few μm or below, the nature of plasticity of metallic materials is changing. Two important features of this small-sizes plasticity are two size effects, which can be summed up as "smaller is stronger" and "smaller is wilder", this last observation meaning that the jerkiness of plastic deformation becomes prominent at small enough system sizes. In FCC and HCP materials, this is now rather well understood within the framework of *obstacle-controlled* plasticity, from the key role of a scaling ratio between the system size L and an internal scale l mainly dictated by dislocation patterning in pure materials, or by the nature of extrinsic disorder in alloys. The situation is more complex in BCC materials, for which screw dislocation motion becomes *lattice-controlled*, i.e. is thermally activated, below a transition temperature T_a. Therefore, in small-sized BCC systems, temperature, size and strain-rate effects combine to give rise to a complex landscape. We show, from an analysis of the literature as well as micropillar compression tests on Molybdenum performed with different sample sizes, under different temperatures and different applied strain-rates, that (i) near or above T_a, the plasticity of pure BCC metals is athermal and obstacle-controlled, much like at bulk scales, therefore mimicking that of pure FCC metals; (ii) below T_a and for sample sizes larger than ∼1 μm, BCC plasticity becomes lattice-controlled, this damping dislocation avalanches and thus reducing wildness; but (iii) for very small systems, still below T_a, the role of screw dislocations on plasticity vanishes, i.e. is no more lattice-controlled, opening again the door for wild plastic fluctuations and jerkiness.

Keywords: Plastic intermittency, Micropillars, BCC materials, Size effects

Introduction

With the growing interest towards manufacturing and using small technological devices with typical sizes L of the order of a few μm or below, it was realized over the last decades that several aspects of material engineering, well established at large scales, have to be reconsidered in case of miniaturization. This is particularly true for the plastic behavior of metals, characterized by emergent features specific to such small system sizes. The first discovered was a "smaller is stronger" size effect (Uchic et al. 2004), generally described from an empirical relation of the form $\tau_y/\mu \sim (L/b)^{-\alpha}$, with τ_y a (plastic) yield shear stress, μ the shear modulus, b the Burgers' vector, and α an empirical exponent that can vary with the crystalline structure, the system geometry (e.g. micropillars,

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nanoparticles, nanowires,..), or the testing procedure, but is generally close or smaller than 1 (Greer and De Hosson 2011). In case of nanoparticles with L in the range of a few hundreds of nm, this yield stress can even approach the theoretical strength of the material, i.e. $\tau_y \simeq \mu/10$ (Sharma et al. 2018). The classical interpretation of this size effect on strength is related to an increasing role of free surfaces in small systems, through a dislocation source truncation mechanism (Parthasarathy et al. 2007), or dislocation starvation at even smaller sizes (Greer and Nix 2006). However, molecular dynamics (MD) simulations of the deformation of very small pillars (diameter of 36 nm) of Molybdenum, a BCC material, at a temperature below its athermal temperature T_a, suggested that dislocation starvation is unlikely in BCC materials as the result of a self-multiplication mechanism coming from the combined effects of the image stress and dislocation core structure (Weinberger and Cai 2008). This mechanism was not observed for FCC materials (gold and aluminum), pointing out a first important difference in the sub-μm plasticity of BCC vs FCC materials.

Such size effect might appear beneficial at first glance for nanoscale engineering. It is however accompanied by another "smaller is wilder" size effect (Weiss et al. 2015, 2021), meaning that the jerkiness of plastic deformation becomes prominent upon decreasing the system size (Dimiduk et al. 2006; Papanikolaou et al. 2017) (see below for a more formal definition of "wildness"). This takes the form of either intermittent abrupt stress drops (under strain-control conditions) or strain bursts (under load-control) of various sizes on stress-strain (SS) curves. These sudden, large and unpredictable strain jumps represent a detrimental effect for the structural stability of micro- to nano-components and the control of plastic-forming processes at those scales (Csikor et al. 2007; Liu et al. 2022). A thorough analysis of plastic fluctuations and of the underlying physical mechanisms is therefore a key question for nanoscale engineering. Qualitatively, they can be present for all types of metallic crystalline structures (HCP, FCC, BCC) (Brinckmann et al. 2008; Ispánovity et al. 2022), although the exact nature of plastic fluctuations, their distribution of sizes, and the scale range over which they occur depend on the material, and can be partly controlled by alloying (Zhang et al. 2017). Here, we will focus on a comparison between pure FCC and BCC metals in terms of plastic intermittency and dislocations avalanches. Owing to the specific nature of plasticity in BCC materials (Po et al. 2016), characterized by a strong dependence on temperature, such comparison allows to explore the role of thermally activated processes on dislocation motion in micro- and nano-components, thereby on the nature of plastic fluctuations. As it will become apparent below, our understanding of this problem largely rests on recent works of Pr Ghoniem and colleagues about screw dislocation mobility in BCC materials (Po et al. 2016) and its role on plastic avalanches in small-sized systems (Cui et al. 2016, 2020).

The nature of plastic fluctuations and its controlling factors (in FCC and HCP crystals)

Despite the seminal study of Becker and Orowan in 1932 (Becker and Orowan 1932), the fluctuating nature of crystal plasticity and the importance of correlated and fast motions of dislocations, called dislocation avalanches, was essentially overlooked for many years, mainly because these phenomena are generally not detectable on the macroscopic behavior and SS-curves at bulk scales (except for some specific cases related to complex interactions between dislocations and the diffusion of solutes towards their cores (Lebyodkin et al. 1995)). At these large scales, however, plastic fluctuations can sometimes be detected from their acoustic signature. This is particularly the case for some HCP single crystals such as ice, Cadmium or Zinc (Miguel et al. 2001; Richeton et al. 2006; Weiss 2019), characterized by a very strong plastic anisotropy with most of dislocation motion occurring along the basal planes of the hexagonal structure, this minimizing the role of short-range interactions between dislocations and therefore precluding the formation of stable dislocation patterns. In these materials, acoustic emission (AE) revealed a so-called *wild* dynamics consisting of intermittent events, power-law distributed in size (or energy) (Miguel et al. 2001), clustered in both space (Weiss and Marsan 2003) and time (Weiss and Miguel 2004), i.e. arguing for a scale-free critical character of plasticity at odds with the smooth, classical vision of plastic flow. Instead, AE during plastic deformation of large ($\gg \mu$m) FCC crystals mainly consists of a continuous "noise" resulting from the cumulative effect of small and uncorrelated dislocation motions, distinctive of a *mild* plasticity (Weiss et al. 2015, 2019). If AE measurements argued in some cases for the coexistence of wild and mild fluctuations during plastic deformation (Weiss et al. 2015), this was later confirmed in compression tests on micro- to nano-pillars of FCC materials (Zhang et al. 2017), for which plastic fluctuations are directly measurable on SS-curves (Note that a strong correlation was recently shown between stress drops on the SS-curves and AE bursts during compression of micro-pillars of Zinc (Ispánovity et al. 2022)). Overall, it was shown that the statistics of plastic jump sizes X (in nm, or unit of strain if normalized) follow the generic distribution (Weiss et al. 2015, 2021):

$$P(X) = \frac{X_0^{\kappa-1}}{\Gamma(\kappa-1)X^{\kappa}} e^{-X_0/X} \tag{1}$$

$$P(>X) = 1 - \frac{\Gamma(\kappa-1, \frac{X_0}{X})}{\Gamma(\kappa-1)}, \tag{2}$$

where $\Gamma(a,x)$ is the incomplete gamma function. This expresses a power-law tail ($P(X) \sim X^{-\kappa}$ for $X \gg X_0$) representing wild fluctuations, while the exponential term, which dominates at small sizes ($X \ll X_0$), accounts for mild plasticity. We can then define the wildness W as the fraction of total plastic strain dissipated through wild avalanches. In this framework, the power-law exponent κ is non-universal, instead a universal relation between κ and W emerges, with an exponent κ increasing with increasing mildness (see Weiss et al. (2021) and below).

If the distribution (1) appears generic, the wildness W varies with the crystal symmetry, the degree of alloying, i.e. on the amount of disorder, and the system size L (Weiss et al. 2021). This last point expresses a "smaller is wilder" size effect. For FCC materials and alloys, as well as HCP materials deforming preferentially through basal slip, i.e. for situations where lattice friction plays a minor role on dislocation motion, which is therefore athermal, these different controlling factors on wildness can be unified from a ratio of external *vs* internal length scales, $R = L/l$, where the internal length scale l can be generically defined as $l = \mu b/\tau_{pin}$, and τ_{pin} represents the effective pinning strength of the obstacles (of any type) impeding dislocation motion. In pure FCC metals, the main contribution to τ_{pin} comes from the pinning strength of forest dislocations, i.e. $\tau_{pin} \simeq \tau_f = \mu b \rho_f^{1/2}$, where ρ_f is the forest dislocation density. In these materials, dislocation patterns emerge spontaneously with plastic deformation, with a characteristic scale l_p given by the "similitude principle", $l_p = k\frac{\mu b}{\tau_{pin}} = kl$, and the dimensionless constant $k \simeq 7.5$ empirically obtained from experimental data (Sauzay and Kubin 2011). This internal length scale l_p appears as a natural mean free path for dislocation motion. One can then understand the key role of the scale ratio R as follows: For system sizes $L \gg l_p$, i.e. $R \gg 1$, dislocation avalanches can hardly propagate beyond l_p, i.e. the plastic response, resulting from the cumulative effect of numerous "confined" motions, appears smooth at the system scale. On the reverse, for small systems ($L < l_p$), dislocation avalanches are not frustrated and become prominent on SS-curves. This simple argument is also consistent with a wild plasticity observed over an extended range of external sizes L in case of HCP materials deforming through basal slip, as dislocation patterns do not emerge in this case and the dynamics is controlled by long-ranged elastic interactions between dislocations, promoting wild fluctuations (avalanches) (Weiss 2019). Note, however, that this argument is likely too simple in case of *very* small systems (L <100 nm), when surface effects can play a significant role on dislocation nucleation and multiplication (Weinberger and Cai 2008; Song et al. 2019). If the image stresses are assumed generally to strengthen the escape of dislocations in small-sized systems, therefore reinforcing dislocation starvation in FCC materials, this seems to be no longer true for BCC materials below their athermal temperature T_a (Cui et al. 2016). In this case, the image stresses appear to instead promote dislocation self-multiplication. These image stress effects and their dependence on system size should therefore be introduced in the modeling of the plasticity of small systems, e.g. through a surface-nucleation and/or a self-nucleation mechanism (Nicola et al. 2007; Papanikolaou et al. 2017; Song and Papanikolaou 2019), in particular when considering the evolution of dislocation densities, a point which was not considered in the simple argumentation above.

In case of alloys, the pinning strength is dominated by the average effect of extrinsic obstacles, such as solutes or precipitates, but the framework described above holds (Zhang et al. 2017). Expressed in other words, the internal scale l corresponds to a distance at which elastic interactions between dislocations, scaling as $\mu b/l$, is comparable to the dislocation/obstacles (including forest dislocations) interaction stress τ_{pin}. This type of plasticity can therefore be considered as being *obstacle-controlled* (Kubin 2013).

To conclude this section, we can mention that, in case of wild dynamics, the scale-free character of plasticity appears as a signature of criticality, although the exact nature of this criticality has been debated. Friedman et al. (2012) argued that scale-free dynamics results from a tuned (depinning) criticality for both FCC and BCC *pure* crystals. This was later challenged by Ispanovity et al. (2014) who showed that in these pure systems, self-organization gives rise to a so-called extended criticality (not tuned). However, in the presence of strong enough extrinsic obstacles, the nature of criticality can switch to a depinning scaling picture (Ovaska et al. 2015). We finally note that these different scaling behaviors cannot be distinguished on the basis of wildness only (Zhang et al. 2020).

The question we will raise now is the following: Can we extend the above reasoning to BCC materials, for which lattice friction plays an important role on dislocation motion, at least at bulk scales ?

Lattice-controlled *vs* obstacle-controlled plasticity in BCC materials

The plasticity of BCC materials differs significantly from that of FCC metals, owing to the non-degenerated core configuration of screw dislocations (Rodney et al. 2017). We recall below these main specific features, before to show their consequences in terms of plastic fluctuations in small-sized systems.

At least for large-sized systems, the plasticity of BCC metals below a material-dependent athermal transition temperature T_a is controlled by 1/2<111> screw dislocations. As the result of their core structure, the Peierls stress τ_p (i.e. at 0°K) of these dislocations is very large, of the order of $5.10^{-3}- 10^{-2}\mu$ (Po et al. 2016; Kubin 2013), meaning that screw dislocations will be immobile for a resolved shear stress below τ_p. At finite $T < T_a$ and resolved stresses below τ_p, the lattice energy barriers can be overcome by thermal activation, leading to a temperature-dependent lattice friction $\tau_l(T)$ and a *slow* motion occurring through the thermally-activated nucleation of kink-pairs and their subsequent migration. This gives rise to a *lattice-controlled* plastic flow (Kubin 2013). Note, however, that for very small-sized systems, with L of the order of a few tens of nm, MD simulations suggest that this kink-pair mechanism is no longer the dominant factor of screw dislocation motion, which is instead primarily controlled by single-kink nucleation from the surface (Weinberger and Cai 2008). Unfortunately, experimental observations are scarce at those very small scales. On the reverse, above T_a, $\tau_l(T)$ almost vanishes, at least becomes negligible compared to the athermal interactions between dislocations, or with other defects. Much like for FCC metals, this leads to an *obstacle-controlled* plasticity. Various temperature and strain-rate dependencies, absent in FCC materials, are signatures of the lattice-controlled plasticity of BCC materials at low temperatures. One of the most discussed is an increase of the yield stress τ_y upon decreasing the temperature below T_a, while the behavior is athermal above this transition temperature (e.g. Argon and Maloof (1966); Werner (1987)). Another signature of the thermally activated dislocation motion is an increase of τ_y with increasing strain-rate, still for $T < T_a$ (Khan and Liang 1999), quantified from the strain-rate sensitivity $m = \partial \ln\tau_y/\partial \ln\dot{\varepsilon}$. In case of obstacle-controlled plasticity, such as for BCC crystals above T_a, or FCC crystals, the strain-rate sensitivity is therefore expected to vanish. As shown below, this is consistent with observations, as long as the applied strain-rate is not too large, and/or dislocation densities not too small. However, for extreme strain-rates ($\dot{\varepsilon} \gg 1s^{-1}$), a strain-rate sensitivity emerges even in FCC materials, e.g. as the result of phonon drag (Fan et al. 2021). The conditions considered here do not belong to this regime.

If dislocation patterns spontaneously emerge from mutual short-range interactions between dislocations in deformed FCC metals, this is no longer true for BCC materials below a certain value of the applied strain, which itself increases with decreasing temperature, meaning that such patterns are in fact absent at low enough temperatures. This is once again related to the temperature dependence of lattice friction, and to the limited ability of screw dislocations to cross-slip at those temperatures (Keh and Weissmann 1963; Kubin 2013).

In summary, above a material-dependent athermal transition temperature T_a, the obstacle-controlled plasticity of large-sized BCC crystals resembles that of FCC crystals: no strain-rate or temperature dependence of the yield stress, formation of well-defined dislocation patterns and cells that will impact the propagation of dislocation avalanches. Instead, below T_a, the motion of *individual* screw dislocations becomes lattice-controlled, delaying or suppressing the formation of dislocation patterns and giving rise to specific temperature- and strain-rate dependencies. The questions now are: Do these specific features persist in small-sized systems ? What is the role of lattice-controlled mechanisms on plasticity at those scales, and particularly on the nature and statistics of plastic fluctuations ?

The nature of plastic fluctuations in BCC small-sized crystals

Several studies were performed on the plasticity of BCC micro- to nano-crystals, either experimentally (Brinckmann et al. 2008; Zaiser et al. 2008; Kim et al. 2010; Friedman et al. 2012; Rizzardi et al. 2022) or from numerical simulations (MD (Alcalá et al. 2020; Weinberger and Cai 2008) or DDD (Cui et al. 2020; Po et al. 2016; Cui et al. 2016; Aragon et al. 2021; Rizzardi et al. 2022)). We already mentioned the work of Weinberger and Cai (2008) that showed that, below T_a , and for very small system sizes, non-trivial surface effects could promote dislocation self-multiplication. Different studies that considered the role of temperature on small-sized BCC samples mainly focused on strength (e.g. Abad et al. (2016); Aragon et al. (2021); Kim et al. (2010)), without paying too much attention to the nature of plastic fluctuations. For these systems, temperature- and size-effects combine to complicate the picture. At high temperatures ($T \geq T_a$), the size effect on yield stress is pronounced, with an exponent α in the range 0.7-1, i.e. similar to that of FCC crystals, while α is much smaller below T_a (Abad et al. 2016; Brinckmann et al. 2008; Kim et al. 2010). In addition, for "large" sample sizes ($L > 1\,\mu\text{m}$), the yield stress significantly increases with decreasing temperature, much like for the same BCC materials at bulk scales (Abad et al. 2016; Rizzardi et al. 2022). However, this temperature effect, typical of BCC metals, disappears for very small system sizes (Cui et al. 2016). In addition, for $T < T_a$, the strain-rate sensitivity is significant ($m \geq 0.05$) for "large" L, but becomes negligible for very small systems ($L \ll 1\,\mu\text{m}$) (Huang et al. 2015).

Altogether, these observations suggest that:

(i) The plasticity of BCC crystals larger than $\sim 1\,\mu\text{m}$ is similar to that observed at bulk scales, i.e. is obstacle-controlled above T_a and lattice-controlled below this athermal transition temperature.

(ii) For very small-sized systems, the signatures of lattice-controlled plasticity seem to disappear, even at low T.

Is point (ii) related to a reduced role of screw dislocations on the plasticity of BCC materials in small-sized systems ? How this, and points (i) and (ii), would translate in terms of plastic fluctuations ?

Several previous works that examined plastic fluctuations in small-sized BCC crystals discussed the role of screw dislocations on avalanche dynamics, but did not converge towards a definitive conclusion. Compressing micro- to nano-pillars of various sizes of Mo at room temperature, i.e. well below $T_a \simeq 190-230°C$ for this material, and under different applied stress-rates, Zaiser et al. (2008) reported jerky SS-curves and power-law distributed (i.e. wild) strain bursts, apparently independently of the system size or the stress-rate, and interpreted this as a minor role of screw dislocations in avalanche dynamics at low T. In contrast, from a comparison of the internal time-structure of dislocation avalanches (the avalanche shape) in Au (FCC) and Nb (BCC) micropillars at room temperature, Sparks and Maaß (2018) argued that the slower decay of avalanches in Nb is a signature of lattice-controlled screw dislocation motion. For the same material, Rizzardi et al. (2022) reported a transition from wild (power-law distributed) plastic fluctuations above T_a ($\simeq 20-50°C$ for Nb) to mild ones at temperatures much below T_a, for a fixed micropillar size $L=2\mu m$. Finally, from a DDD simulation study of the compression of micropillars of tungsten, Cui et al. (2020) concluded that, for 2-μm pillars, strain bursts were dominated by the slow screw dislocation behavior while, for smaller samples (500 nm) the rapid motion of non-screw dislocations was the dominant process under low applied strain-rates. However, at such small L, the role of screw dislocations seems to increase under high strain-rates. This might appear also somehow in contradiction with MD simulations of the plasticity of Mo *very* small pillars (36 nm of diameter), which seems controlled by surface nucleation and pure screw dislocation motion (Weinberger and Cai 2008). This illustrates the overall complexity of size effects on BCC plasticity.

To try to further decipher the combined roles of temperature, system size, and strain-rate on plastic fluctuations in BCC materials, we recently performed a statistical analysis of plastic fluctuations in compressed micropillars of Mo over a range of both pillar sizes (500 nm$\leq L \leq$3500 nm), temperature ($25°C \leq T \leq 200°C$) and strain-rates ($2\times10^{-4}s^{-1} \leq \dot{\varepsilon} \leq 2\times10^{-2}s^{-1}$) (Zhang et al. 2023). The experimental details on the experiments and their physical interpretation have been given elsewhere (Zhang et al. 2023). Our goal here is to summarize the main results of this work, present some additional data and details to further compare the nature of plastic fluctuations in this BCC material to what is known for FCC materials, and contextualize them in light of the literature discussed above.

We compare first, on Fig. 1, the size effects on SS-curves and on the distributions of displacement jumps X for compressed micropillars of pure Mo at a temperature T=200° C$\simeq T_a$ in one hand, and of pure Al at 25°C on the other hand. These micropillar SS-curves are compared as well to the macroscopic behavior of polycrystalline samples of the same materials under tension. If this macroscopic behavior is smooth ("mild") in both materials, the micropillar SS-curves exhibit an increasing jerkiness upon decreasing the sample size L, with obvious plastic instabilities even for $L=6\,\mu$m (Al sample). In Al, all distributions $P(>X)$ are well fitted by Eq. (2). They are dominated by a power law tail with a lower bound X_0 in the range 0.6-1 nm (the detection threshold is 0.3 nm). Consequently, most of plastic deformation occurs through dislocation avalanches of sizes $X>X_0$, and the wildness W is large, close to 0.9 (Zhang et al. 2017; Weiss et al. 2021). In this material, mildness becomes significant only for $L=6\,\mu$m ($W<0.4$), which is compatible with a internal length scale l_p associated to dislocation patterning of the order of 10 μm (Sauzay and Kubin 2011; Zhang et al. 2017). In Mo at 200°C, the distribution $P(>X)$ observed for a system size $L=3.5\,\mu$m is still dominated by the power law tail,

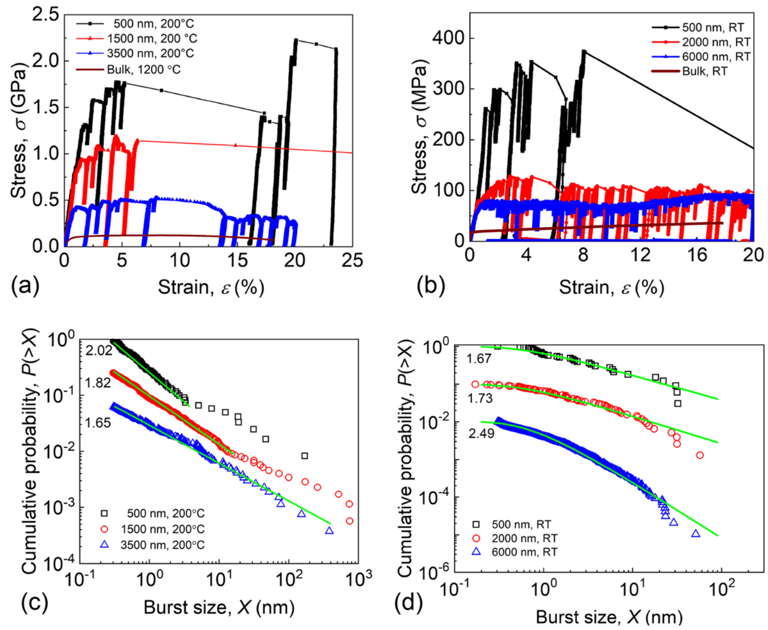

Fig. 1 Comparison of SS-curves (top) and cumulative probability distributions of displacement jumps X (bottom) for Mo pillars of different sizes L compressed at 200°C and $\dot{\varepsilon} = 2 \times 10^{-3} s^{-1}$(left) and Al pillars compressed at 25°C and $\dot{\varepsilon} = 2 \times 10^{-4} s^{-1}$(right). The small numbers indicate the corresponding best-fitted exponents κ of Eq. (2)

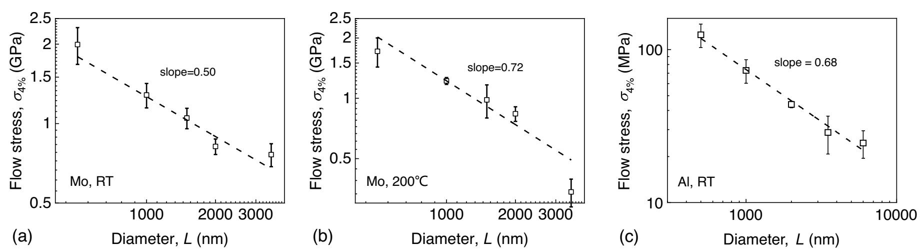

Fig. 2 Comparison of size effects on the yield strength τ_y of micropillars estimated at 4% of plastic strain for **a** pure Mo at 25°C, i.e. well below T_a, **b** pure Mo at $T \simeq T_a$ (200°C), and **c** pure Al at 25°C

associated to a wildness close to 0.9, but one can expect a transition to a mild behavior at larger scales as the result of dislocation patterning, which is also observed in BCC materials above T_a (Kubin 2013), and in agreement with a mild behavior at bulk scales. At even smaller L, the tails of the distributions are characterized as well by few outliers, or "dragon-kings", a signature of super-criticality (Sornette and Ouillon 2012). Each of these giant strain bursts, corresponding to a system-spanning avalanche, accounts individually for more than 10% of plastic strain. They are reminiscent of similar events observed during the plastic deformation of FCC nanoparticles (Sharma et al. 2018; Mordehai et al. 2011), have been also reproduced from a minimal automaton model of crystal plasticity (Zhang et al. 2020), but never observed so far to our knowledge in FCC micropillars. A more thorough discussion about this super-criticality in BCC plasticity can be found elsewhere (Zhang et al. 2023). Here, we simply summarize that the plasticity of BCC pure materials at high temperatures $T \geq T_a$ is very similar to that of FCC materials, i.e. is *obstacle-controlled*: dominated by wild fluctuations for $L < l_p$, transitioning towards

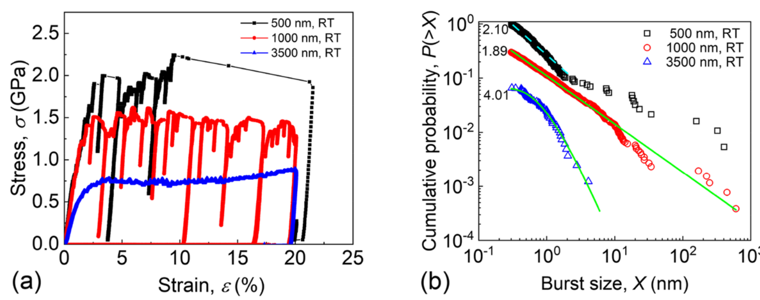

Fig. 3 SS-curves (left) and cumulative probability distributions of displacement jumps X (right) for Mo pillars of different sizes L compressed at 25°C, i.e. well below T_a. The small numbers shown on the right figure indicate the corresponding best-fitted exponents κ of Eq. (2)

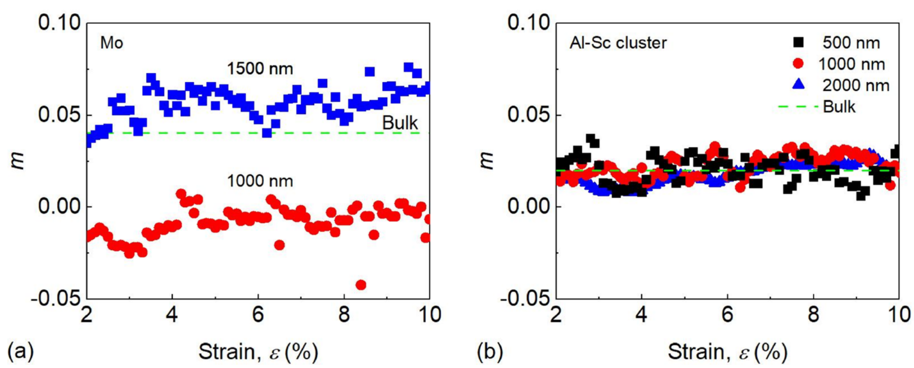

Fig. 4 Strain-rate sensitivity, m, as a function of strain, obtained from compression tests performed under applied strain-rates ranging from $2 \times 10^{-4}\,s^{-1}$ to $2 \times 10^{-2}\,s^{-1}$. Mo pillars with L=1.5 µm (upper left) show a positive m=0.05, signature of lattice controlled plasticity, while strain-rate sensitivity is absent in the same material for smaller sample sizes (L= 1 µm; lower left), or for pillars of an Al-Sc alloy, whatever their sizes (right). In each case, the data shown were obtained from an averaging over several samples (5 to 8)

a mild behavior at larger scales ($L > l_p$) as the result of the emergence of dislocation patterning. The similarity extends to the size effect on strength, which is significant in both cases with a relatively large exponent $\alpha \simeq 0.7$, whereas $\alpha \simeq 0.5$ is smaller for the BCC material below T_a (Fig. 2). This last observation is consistent with previous works (Brinckmann et al. 2008; Abad et al. 2016).

As expected above, the situation at low temperatures ($T < T_a$) is more complex. The size effect on SS-curves and distributions of plastic fluctuations is pronounced (Fig. 3). For large pillars ($L > 1\,\mu m$), as observed previously (Rizzardi et al. 2022), SS-curves are smooth and the distributions are clearly not power laws: well fitted by Eq. (2), the effect of the lower cut-off X_0 is pronounced and the corresponding wildness W decreases down to $\sim 1\%$ for $L = 3.5\,\mu m$. This mild character of plasticity, compared to what is observed for the same sizes L at 200°C (see Fig. 1), can be attributed to a *lattice-controlled* mechanism, i.e. to the thermally-activated motion of screw dislocations, much like what is known at bulk scales (Kubin 2013; Po et al. 2016). Several signatures of such mechanism can be mentioned. The first one is, as observed at bulk scales (Khan and Liang 1999), a positive strain-rate sensitivity of the yield stress, $m > 0$. To illustrate this, Fig. 4 contrasts a positive $m \simeq 0.05$ for 1500 nm Mo micropillars at room temperature, with (i) a significantly smaller value ($m = 0.02$) for micropillars of an Al-Sc alloy (FCC), and (ii) an absence of strain-rate sensitivity ($m = 0$) for 1000 nm Mo micropillars at the same temperature. We note that the m-values for 1.5 µm Mo samples and for Al-Sc samples (whatever their size) are consistent with those obtained at bulk scales for the same materials. On the reverse, the disappearance of strain-rate sensitivity for smaller Mo pillars (point (ii)) is a first indication of a fading of lattice-controlled plasticity in BCC materials at very small scales, discussed in more details below and in agreement with results obtained on Tungsten (Srivastava et al. 2021).

A second symptom of thermally-activated motion of screw dislocations is a scaling between the average maximum (or peak) displacement velocity during an avalanche, $\langle \dot{X}_p \rangle$ (in nm/s), and the corresponding average size, $\langle X \rangle$ (in nm), which can be expressed as $\langle \dot{X}_p \rangle \sim \langle X \rangle^{\gamma}$. Such scaling has been mentioned in some previous works. Compiling data from various micropillars of sizes ranging from ~ 200 nm to ~ 6 µm, Zaiser et al. (2008) reported a positive correlation between $\dot{X}_p$ and X, but did not find a clear signature of scaling, maybe as a result of mixing up data from various sample sizes. Focusing on a unique pillar size (L=2 µm), Sparks and Maaß (2018) observed a $\langle \dot{X}_p \rangle \sim \langle X \rangle^{\gamma}$ scaling for both a BCC material (Nb) with γ=0.44 and a FCC material (Au) with γ=0.66. The smaller peak velocities (for a given avalanche size X) as well as the slower increase of $\langle \dot{X}_p \rangle$ with increasing $\langle X \rangle$ (smaller γ) in the BCC material was attributed to the slow motion of screw dislocations, damping avalanche propagation (Sparks and Maass 2018). Our results are essentially consistent with this scenario, showing for Mo a dependence of γ on both system size L and temperature T (Zhang et al. 2023): the peak velocity $\langle \dot{X}_p \rangle$ is almost insensitive to the avalanche size $\langle X \rangle$ at room temperature for large pillars (3.5 µm), i.e. γ=0.10, while much larger exponents, in the range 1.0-1.3 are observed at higher T and/or smaller sizes L, much like for Al (Fig. 5). Note that, in these last cases, the estimation of $\dot{X}_p$ is biased by the finite sampling frequency (400 Hz in our experiments), meaning that these γ-values represent a lower bound for the true exponents. Overall, a positive correlation is observed between the exponent γ and the wildness W (Zhang et al. 2023). So, low γ and W are associated, for BCC micropillars, to a lattice-controlled plasticity while large values are signatures of an obstacle-controlled plasticity.

Finally, another sign of the role of thermally activated motion of screw dislocations is presented on Fig. 6, which shows the triggering of a plastic avalanche during an

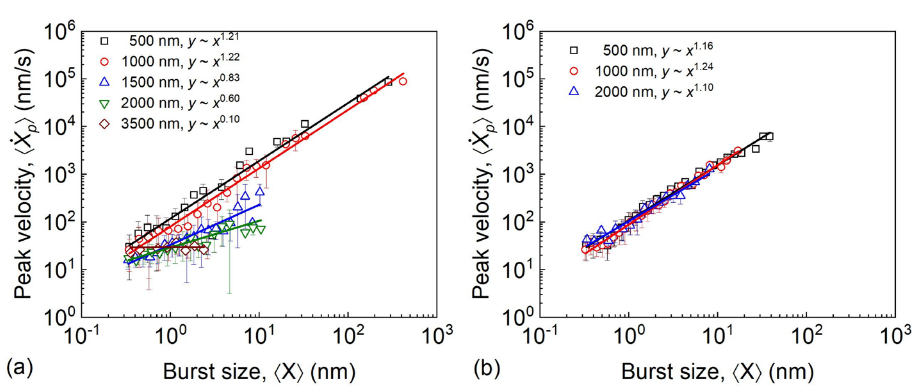

Fig. 5 Scaling between the average peak velocity $\langle \dot{X}_p \rangle$ of the avalanche and the corresponding average size $\langle X \rangle$. **a** For Mo micropillars of different sizes L tested at 25°C; **b** For Al micropillars of different sizes L tested at 25°C

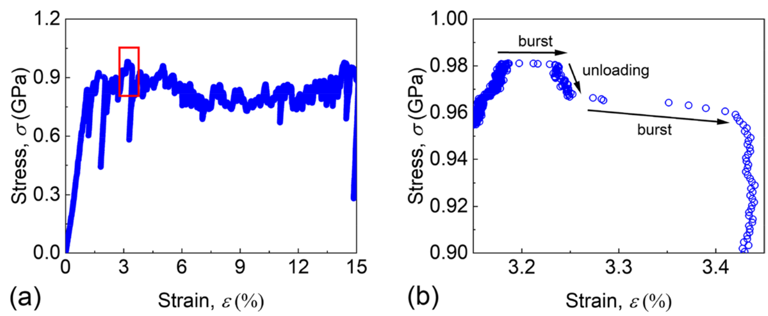

Fig. 6 Triggering of a dislocation avalanche during the unloading stage following a previous one, for a 1.5 μm Mo micropillar at 25°C. **a** The entire SS-curve, zoomed in (**b**) to show such event. This signature of thermally activated triggering was neither observed for Mo pillars with sizes L <1 μm and/or T=200°C, nor for Al pillars, whatever their sizes

unloading stage resulting from a previous instability. This would be hardly reconcilable with an obstacle-controlled plasticity, and was actually observed neither in Al samples, nor in Mo micropillars at 200°C and/or with L <1 μm, but is compatible with a thermally activated triggering.

At this point, we can temporarily conclude that (i) at high temperatures ($T \geq T_a$), the plasticity of small-sized BCC samples is strongly similar to that of FCC materials, with size effects on strength and wildness that can be understood in a context of obstacle-controlled plasticity, while (ii) for large enough (L >1 μm) samples tested below T_a, plasticity becomes lattice-controlled, much like at bulk scales. In this last case, the mildness, which increases with decreasing temperature (Rizzardi et al. 2022; Zhang et al. 2023), does not result from the pinning effect of obstacles (such as forest dislocations in pure metals, or other obstacles in alloys (Zhang et al. 2017)) but from the damping of the thermally-activated motion of screw dislocations. In our understanding, this damping does not allow a strong acceleration of collective dislocation dynamics during avalanches. We note that this interpretation slightly differs from that of Rizzardi et al. (2022) who argued that screw dislocation motion remains athermal during avalanches, even below T_a, considering that these avalanches are triggered by local stress concentrations strong enough to overcome lattice friction. However, in Zhang et al. (2023), we argued, on the basis of the results of DDD simulations (Csikor and Groma 2004), that this is unlikely.

The next question is: What happens at very small scales (L ≤1 μm) in BCC materials at low temperatures ? Indeed, at those scales, strain-rate sensitivity disappears, see Fig. 4, suggesting a fading of lattice-controlled plasticity that remains to be confirmed and explained.

Besides an absence of strain-rate sensitivity, the plasticity of sub-μm Mo samples below T_a closely mimics that of the same material at high temperatures, or of FCC materials: Jerky SS-curves are associated to critical (power law) or super-critical distributions of jump sizes X (see Fig. 7), the average peak displacement velocity during an avalanche, $\langle \dot{X}_p \rangle$, is large, as well as the associated scaling exponent γ (Fig. 5), whereas triggering during an unloading stage, such as depicted on Fig. 6 for a larger sample (L=1.5 μm), is not observed. All of this suggests that the role of (slow) screw dislocation motion vanishes in such very small systems. In line with the work of Cui et al. (2016), we proposed to explain this from the nature of single-arm dislocation sources (SAS) in small-sized

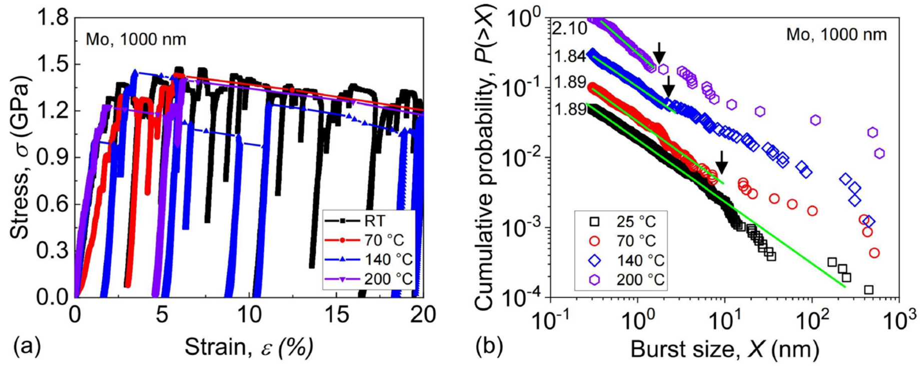

Fig. 7 SS-curves (**a**) and corresponding cumulative probability distributions of displacement jumps X (**b**) for 1 μm Mo pillars compressed at different temperatures. An absence of temperature-sensitivity of the yield stress τ_y is already apparent in (**a**). On (**b**), the small numbers indicate the corresponding best-fitted exponents κ of Eq. ((2)), while the small arrows show the transition from power-law distributed events to dragon-kings

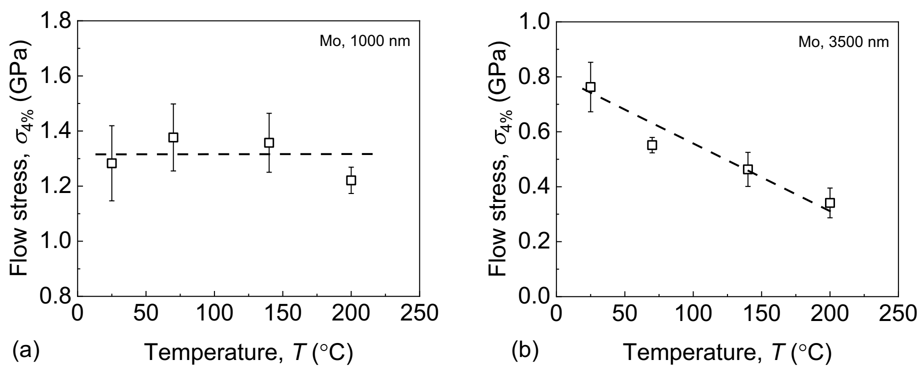

Fig. 8 Temperature sensitivity of the yield stress τ_y estimated at 4% of plastic strain for **a** 1 μm Mo pillars (no temperature dependence) and **b** 3.5 μm Mo pillars (τ_y increases with decreasing temperature, much like at bulk scales)

BCC systems (Zhang et al. 2023): TEM observations in Fe revealed that the operation of such SAS involves, at low temperatures, a slow screw segment connected to a curved, non-screw, and therefore strongly mobile part (Caillard 2010). In $L > 1\,\mu m$ systems, the non-screw segments control the slowness of these SAS, damp dislocation avalanches, i.e. strongly reduces the jerkiness of the lattice-controlled plasticity, and induces a positive strain-rate sensitivity. However, upon reducing further the system size L below $\sim 1\,\mu m$, these screw segments disappear (Zhang et al. 2023), leading to a vanishing fraction of screw dislocations (Cui et al. 2016), opening once again the door for a jerky behavior and wild fluctuations (Figs. 3 and 7), accompanied as well by a strain-rate- (Fig. 4) and a temperature- (Fig. 8) *in*sensitivity of the flow stress. This last observation is in full agreement with the previous numerical work of Cui et al. (2016).

Summary and conclusions

It is now well established that the plasticity of metals and alloys is dramatically modified upon decreasing the system size near or below a few µm. Two prominent features of this small-sizes plasticity are an increasing yield/flow stress, unfortunately counterbalanced by ubiquitous plastic instabilities (avalanches) and consequently a likely larger variability of the strength (e.g. Parthasarathy et al. (2007); Papanikolaou et al. (2017)). In case of obstacle-controlled plasticity, such as in FCC and HCP materials as well as their alloys, the statistics of these plastic fluctuations and their size-dependencies are now rather well understood (Weiss et al. 2015; Zhang et al. 2017; Weiss et al. 2021). In particular, they are well fitted by eqs. (1) and (2), from which it is possible to deduce a relationship between the wildness W (the fraction of plastic strain accommodated through wild fluctuations), and the power-law exponent κ:

$$W(\kappa) = 1 - \frac{\Gamma(\kappa - 1, \ln(1/K))}{\Gamma(\kappa - 1)}, \tag{3}$$

where $K = e^{-X_0/X_{min}}$ and X_{min} a value, close to X_0, above which fluctuations can be considered as being power-law distributed. Equation (3) accounts for the fact that large W are associated to low exponents κ, and is the signature of an underlying universal physics, at least for obstacle-controlled plasticity. Indeed, in this case, a link between the exponent κ (so, the wildness W) and a ratio of external L vs internal l length scales has been proposed from a stochastic mean-field model, $\kappa - 1 \sim L/l = R$, where the internal scale l is inversely proportional to the effective pinning strength τ_{pin} of the obstacles (of any type) impeding dislocation motion. It is important to remark, however, that a constant nucleation rate, independent of the dislocation density itself, was so far considered in this stochastic mean-field framework for dislocation density evolution. I.e., possible self-nucleation processes (Papanikolaou et al. 2017; Song and Papanikolaou 2019), which seems to play a significant role in BCC plasticity for very small-sized systems (Weinberger and Cai 2008), are ignored.

In this context, the situation for BCC materials is expected to be more complex, as the plasticity of many of these materials is know be become lattice-controlled below T_a. Experimental works (Brinckmann et al. 2008; Zaiser et al. 2008; Kim et al. 2010; Abad et al. 2016), as well theoretical and numerical ones (e.g. Alcalá et al. (2020); Aragon et al. (2021)) including some key contributions of Pr. Ghoniem et al. (2016; 2016; 2020), explored this problem. This is likely to rapidly become of great practical interest, as BCC metals are involved in the fields of micro- to nano-technologies (see e.g. Ha et al. (2004) in case of Mo for gate technology). In the present work, we complemented these former works with recent results obtained on the size-, temperature- and strain-rate dependencies of plastic fluctuations in compressed Mo micropillars, fully detailed elsewhere (Zhang et al. 2023). Our main conclusions are as follows:

(i) Near or above T_a, the plasticity of pure BCC materials is athermal and obstacle-controlled, strongly mimicking that of pure FCC materials. Consequently, a strong smaller-is-wilder size effect occurs, with a jerky behavior for small ratios of length scales $R = L/l$ and a progressive transition towards a smoother behavior at larger system sizes. In this case, what is recalled just above for FCC materials is expected to hold. Indeed, Mo data at 200°C are fully consistent with Eq. (3) (Zhang et al. 2023).

(ii) Below T_a and for system sizes above $\sim 1\ \mu$m, the BCC plasticity becomes lattice-controlled, i.e. is thermally activated, much like at bulk scales for the same T. This gives rise to a temperature- as well as strain-rate-sensitivities of the yield stress. In addition, this thermally-activated motion of screw dislocations damps dislocation avalanches, leading to a much milder behavior than at higher temperatures for the same system sizes. In this situation, the arguments developed for obstacle-controlled plasticity and recalled above are not expected to hold. Indeed, the main barrier to screw dislocation motion becomes the lattice friction $\tau_l(T)$. However, identifying this lattice friction as a "pinning strength" τ_{pin} to estimate a transition ratio R, as initially proposed in Louchet (2006), Weiss and Louchet (2006) and recently taken (Rizzardi et al. 2022), might be misleading, since τ_l depends as well on screw dislocation velocity, i.e. will vary during the accelerating and decelerating stages of dislocation avalanches (Zhang et al. 2023). If the above definition of the scaling ratio R does not hold, we surprisingly observed that the statistics of plastic fluctuations of Mo below T_a are still consistent with Eq. (3), extending further the universal character of this relationship (Zhang et al. 2023).

(iii) Still below T_a, but for small systems ($100nm < L < 1\ \mu$m), the role of screw dislocation segments on the operation of SAS in particular, and on plasticity in general, vanishes. Consequently, plasticity is no more lattice-controlled, strain-rate- as well as temperature-sensitivities of the yield stress disappear and, at such very small scales, plastic fluctuations become wild again, much like for FCC materials at those scales.

(iv) For *very* low systems ($L < 100nm$), MD simulations indicate instead a plasticity controlled by surface-effects and a self-nucleation mechanism of pure screw dislocations (Weinberger and Cai 2008). These types of self-nucleation processes were not considered in the modeling approach used to derive Eqs. (1) to (3) (Weiss et al. 2015; Zhang et al. 2017; Weiss et al. 2021), although it could be introduced in future works from a dislocation density-dependent nucleation rate. Such self-nucleation mechanisms were considered in some numerical models of small-scale plasticity (Papanikolaou et al. 2017; Song and Papanikolaou 2019). This illustrates the complexity of size effects on BCC plasticity, and calls for further studies, in particular experimental, at those scales.

All of this raises interesting additional questions, left for future work. First, an analysis of plasticity and dislocation avalanches in *alloyed* BCC materials would be worth doing, e.g. to check if this can tame jerkiness at high temperatures and/or for sample sizes below $\sim 1\ \mu$m, as it does in case of FCC materials (Zhang et al. 2017). Second, on a more theoretical side, some HCP materials, such as Titanium or Zirconium, mainly deform through the lattice-controlled motion of screw dislocations on prismatic planes (Kubin 2013), leading to a temperature dependence of the yield stress (Naka et al. 1988). For those materials, at odds with what is happening for HCP materials mainly deforming on the basal planes and which are paradigmatic examples of wild plasticity (Ice, Zn, Cd,..) (Richeton et al. 2006; Weiss 2019), we could expect a scenario similar to what is detailed above for BCC materials.

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Authors' contributions
J.W. and G.L. designed research. P.Z. conducted the micro-pillar compression experiments. P.C. performed the bulk Mo experiments. P.Z. and J.W. analyzed the experimental data. J.W. wrote the paper.

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Availability of data and materials
Data and materials can be obtained from the authors upon reasonable request.

Declarations

Competing interests
The authors declare no competing interests.

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