MATERIALS SCIENCE

Deformation and fracture mechanisms of nanotwinned metals

Zesheng You[1,2] and Lei Lu[1,*]

Nanotwinned (NT) metal represents a unique hierarchical nanostructured material, with a large number of highly organized nanoscale twins embedded within micro- or submicro-sized grains. Twin boundaries (TBs) are effective barriers blocking dislocation motion and thus substantially strengthen metals, as effectively as general high-angle grain boundaries (GBs) [1]. More importantly, due to their high symmetry and extremely low excess energy, TBs can also generate and store plenty of mobile dislocations, making compatible plastic deformation with macroscopic hardening [2]. NT metals, with high thermal and mechanical stabilities, exhibit ultrahigh strength and good tensile ductility [3], enhanced cyclic performance [4], high electrical conductivity [5] and superior resistance to electro-migration [6]. Their superior mechanical, physical and chemical properties have drawn worldwide research interests over the last decade and NT metals have been recognized as a remarkable breakthrough in the area of metallic structured materials.

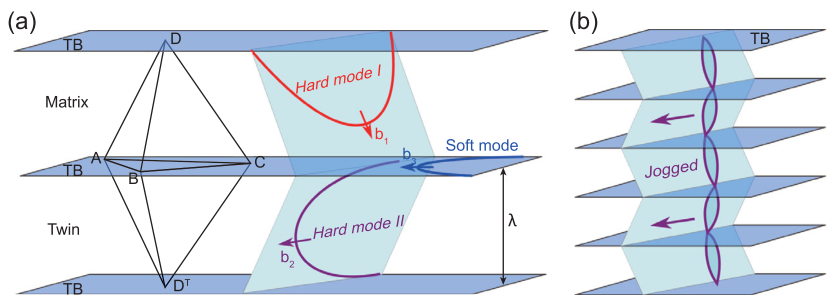

Figure 1. Schematic illustration of possible deformation mechanisms in face-centered cubic NT metals. (**a**) Three different dislocation modes depending on the orientation relationship between the active slip system and TB: dislocation pile-up and slip transfer (hard mode I); threading dislocations in between TBs (hard mode II); twinning partials gliding at TBs (soft mode) [8]. (**b**) Jogged dislocation mechanism in NT metals as twin thickness is reduced to a few nanometers [9].

Plenty of experimental, computational and theoretical studies have been conducted to investigate the plastic deformation mechanisms of NT metals. It is now well established that the novel mechanical properties of NT metals arise from the dislocation–TB interactions that dominate the plastic deformation as TB spacing is refined to the nanometer scale.

Taking face-centered cubic (fcc) NT metals as an example, the possible dislocation–TB interactions can be classified into three different categories, depending on the orientation relationship between active slip system (slip plane and slip direction) and TB, illustrated as a pair of Thompson tetrahedra corresponding to the matrix and twinned crystals shown in Fig. 1a. Dislocations may either pile up against and slip transfer across TBs (hard mode I, with Burgers vector DA, DB, DC or D^{T}A, D^{T}B, D^{T}C on slip planes inclined to TB) or slip in between the NT lamellar channels like threading dislocations in confined-layer slip mode (hard mode II, with Burgers vector AB, BC or CA on inclined slip planes) [7]. Alternatively, twinning partials are activated under large shear stresses along TBs and slip without strong obstruction (soft mode, with Burgers vector AB, BC or CA on TBs) [8].

Recently, molecular dynamic (MD) simulations revealed a transition from threading dislocation to jogged dislocation as the twin thickness is extremely small, i.e. a few nanometers [9]. This jogged dislocation mode, operative either if loading direction is parallel to TBs [9] or under high stresses adjacent to crack tip [10], involves collective movements of linked extended dislocations that span multiple TBs.

The dominant deformation mode controls the mechanical properties, such as strength, ductility, work hardening and strain rate sensitivity. Both internal microstructural parameters (grain size, twin spacing and orientation) and external loading conditions affect the dominant deformation mode. For instance, strong anisotropies in flow strength and strain hardening were detected when a preferentially oriented NT Cu was deformed in various directions with respect to TBs [8].

Inspecting the damage tolerance, such as fracture and fatigue resistance, of NT metals is essential for their practical applications. Compared with the in-depth studies on the plastic deformation, the damage processes of NT metals remain largely unexplored. In nanocrystalline metals, the high-density incoherent interfaces are potential nucleation sites and propagation paths for cracks, leading to dramatically reduced fracture resistance. Contrarily, the low-energy TBs are expected to be much more resistant to direct void nucleation, because the built stress concentration can be plastically relaxed by aforementioned dislocation–TB interactions.

Besides crack nucleation, the presence of NT lamellae also enhances its propagation resistance by various crack–TB interactions, as shown by *in-situ* transmission electron microscope (TEM) fracture tests and MD simulations of thin film samples [11–13]. As a crack approaches a TB, it becomes substantially blunted by emitting a large number of dislocations from the crack tip. These dislocations then accumulate along the TB and resist further crack extension [13]. As shown in Fig. 2, zig-zag crack extension arising from periodic deflections of the crack path by TBs [11,12] and thin nanoscale twins acting as crack-bridging ligaments [13] clearly demonstrate the toughening mechanism associated with the nanoscale twins.

Although qualitatively valid, there is still doubt as to whether the toughening mechanisms revealed in thin films can carry over to bulk metals. Investigations are currently scarce regarding the plain-strain fracture toughness and damage mechanisms of bulk NT metals. The main reason lies in the difficulty in preparing NT samples sufficiently large for conventional fracture tests. In bulk form, the NT samples fail in ductile mode with coarse deep dimples, and higher twin density leads to enhanced fracture resistance [14]. Recently, a contactless video crack opening displacement gauging system has been designed to enable accurate determination of the intrinsic fracture toughness of nanostructured materials using relatively small specimens [15]. This opens a route to further determine the influence of microstructural parameters, such as grain size, twin thickness and TB orientation, on the intrinsic fracture resistance in bulk form.

Despite the considerable progresses made in revealing the interaction mechanisms between dislocations and TBs, there are some unsolved issues related to the NT materials. For instance, understandings of the interactions between dislocation nucleation and GB structure and between cracks and NTs are still in their infancy. It is also challenging at this time to prepare bulk NT samples and to scrutinize the detailed damage mechanisms in three dimensions. In addition, since most of the current understanding comes from studies of fcc metals, the development in other NT materials, such

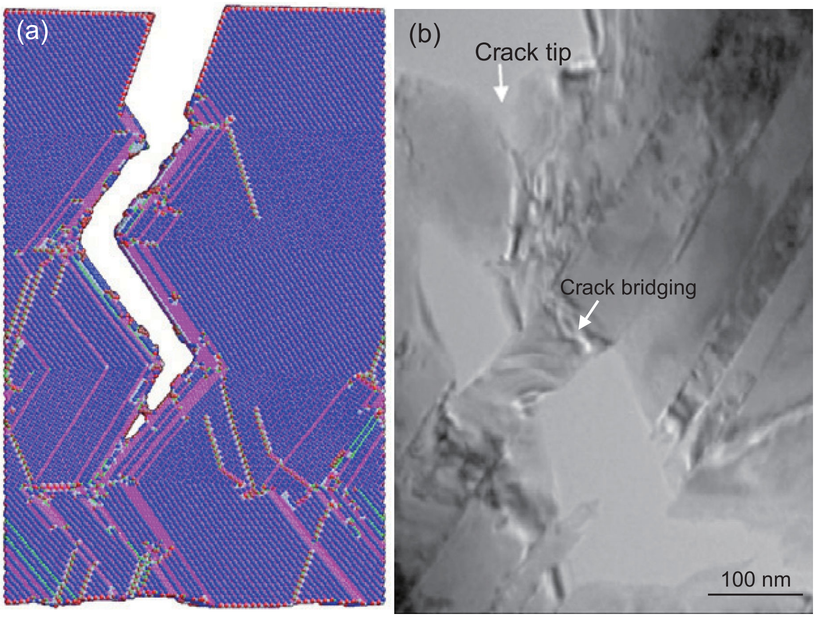

Figure 2. Crack growth processes in NT thin films. (**a**) MD snapshots demonstrating the zig-zag crack growth mode involving dislocation-mediated local thinning in twin/matrix lamellae, adapted with permission from Ref. [12], Elsevier. (**b**) A twin segment near the crack tip acting as ductile crack bridging, adapted with permission from Ref. [13], Elsevier.

as hexagonal closed-packed or body-centered cubic metals, ceramics and composite materials, remains an important direction for future research. Further quantifying correlations between microstructures and deformation/damage behavior are essential for optimizing the global combination of strength and fracture toughness, and for extending the engineering applications of the hierarchical NT materials.

FUNDING

This work was supported by the National Natural Science Foundation of China (51420105001, 51371171, 51471172 and 51401211) and the Key Research Program of Frontier Sciences, Chinese Academy of Sciences.

Zesheng You[1,2] and Lei Lu[1,*]
[1]Shenyang National Laboratory for Materials Science, Institute of Metal Research, Chinese Academy of Sciences, China
[2]Herbert Gleiter Institute of Nanoscience, Nanjing University of Science and Technology, China
*__Corresponding author.__ E-mail: llu@imr.ac.cn

REFERENCES

1. Lu K. *Nat Rev Mater* 2016; **1**: 16019.
2. Lu K, Lu L and Suresh S. *Science* 2009; **324**: 349–52.
3. Shen YF, Lu L and Lu QH *et al. Scr Mater* 2005; **52**: 989–94.
4. Pan QS, Lu QH and Lu L. *Acta Mater* 2013; **61**: 1383–93.
5. Lu L, Shen Y and Chen X *et al. Science* 2004; **304**: 422–6.
6. Chen K-C, Wu W-W and Liao C-N *et al. Science* 2008; **321**: 1066–9.
7. You ZS, Lu L and Lu K. *Acta Mater* 2011; **59**: 6927–37.
8. You Z, Li X and Gui L *et al. Acta Mater* 2013; **61**: 217–27.
9. Zhou H, Li X and Qu S *et al. Nano Lett* 2014; **14**: 5075–80.
10. Zhou H and Gao H. *J Appl Mech* 2015; **82**: 071015.
11. Shan ZW, Lu L and Minor AM *et al. JOM* 2008; **60**: 71–4.
12. Zeng Z, Li X and Lu L *et al. Acta Mater* 2015; **98**: 313–7.
13. Kim S-W, Li X and Gao H *et al. Acta Mater* 2012; **60**: 2959–72.
14. Qin EW, Lu L and Tao NR *et al. Acta Mater* 2009; **57**: 6215–25.
15. Luo SS, You ZS and Lu L. *Scr Mater* 2017; **133**: 1–4.

National Science Review
4: 519–521, 2017
doi: 10.1093/nsr/nwx043
Advance access publication 8 April 2017