techniques will be a promising route to teach us the lessons needed to improve our engineering skills. ❐

Christine Semmrich and Andreas R. Bausch are at the Lehrstuhl für Biophysik-E27, Technische Universität München, Garching, Germany.
e-mail: abausch@ph.tum.de

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CRYSTALLINE MATERIALS

Twin behaviour and size

For a Ti alloy single crystal, the stress required for deformation twinning increases dramatically as the size of the crystal decreases, until at submicrometre sizes, deformation occurs solely by dislocation motion.

Oliver Kraft

Danny de Vito would be happy to hear that for twin formation the paradigm 'smaller is stronger' still prevails. Twinning is a deformation process in crystals defined as the collective shearing of one portion of the crystal with respect to the rest. However, compared with plasticity based on dislocation glide, twinning has been scarcely studied in the deformation of metals. This is because twinning occurs (when slip by dislocation glide is suppressed) mostly for a few hexagonal-close-packed metals such as Ti, or at very low temperatures. Thus, twinning in metals has been considered of rather small technical importance. Furthermore, twinning is difficult to study as it occurs as a collective mechanism, and its origin and evolution in space and time remain puzzling. Writing in *Nature*, Yu *et al.*[1] now describe size effects on the deformation of small, micrometre- and submicrometre-sized Ti alloy crystals and, in doing so, provide new insights and questions relating to the deformation twinning process.

It has long been known that size effects have a prominent role in the deformation and strength of metals. In the 1950s, it was observed that grain size has a strong impact on the strength of pure metals and alloys. Also, it was demonstrated impressively that specimen size is important and that small whiskers, presumably dislocation-free, can have very high strength. Over the decades, such size effects have been observed for many samples and testing conditions, including metal thin films and, most recently, for small single crystals in microcompression tests[2–4]. Although not all details are understood, it is now recognized that these size effects relate to the confinement of the motion of dislocations and the availability of dislocations and dislocation sources[5]. Moreover, it has been shown that nanoscale growth twins can lead to high-strength materials because the twins confine dislocation motion[6]. This observation, however, is not to be confused with twinning as a deformation mechanism, which is discussed here.

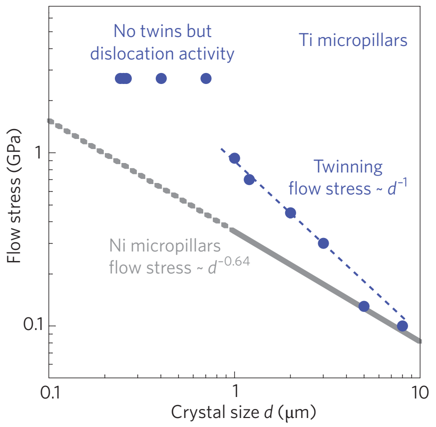

Figure 1 | Mechanical data as a function of the crystal size (d) for Ti alloy micropillars[1] compared with the typical strength values for Ni micropillars. The trend line is taken from ref. 3 in which data for the 1–10 μm region are presented. In Ti, the size effect for twinning is more pronounced with an exponent of −1 compared with −0.6 for many fcc metals[4,5].

Intuitively, one might imagine that there is no size effect for twinning of a small single crystal, which can freely deform. It could be argued that a smaller crystal twins more easily because a smaller volume needs to be sheared compared with a large crystal. However, Yu *et al.* show that the opposite is true: the smaller the crystal, the greater the stress required for deformation twinning (Fig. 1). For submicrometre-sized crystals, the flow stress is found to be constant and of the order of the theoretical strength of Ti. Using *in situ* transmission electron microscopy, it is well demonstrated that deformation is governed by dislocation motion. Compared with results from similar experiments on face-centred-cubic (fcc) metals (such as Ni, as shown in Fig. 1), it becomes clear that for twinning, the size effect is stronger than for dislocation plasticity. Therefore, the limit of theoretical strength, at which apparently homogeneous dislocation occurs, is reached for larger crystals.

The dramatic increase in flow stress with a decreasing size of single crystal is explained using a 'stimulated slip' model (Fig. 2). The model incorporates a pole, for example a screw dislocation, perpendicular to the slip plane, which acts as a promoter that is responsible for twin nucleation. The main idea is that in a large crystal, with enough dislocations, there is always a suitable promoter dislocation present; twinning occurs when the stress is high enough to drive a partial dislocation on the twinning plane that wraps around the pole, leading to a collective, stimulated-slip phenomenon. This implies that there is a certain critical crystal size, below which the initial dislocation density is not large enough to have a twinning promoter. Instead, in such a small crystal, dislocation plasticity with strain hardening occurs. The dislocation density increases until a twinning promoter is available and a twin is formed, corresponding to a large burst in the stress–strain curve. Indeed, for a critical crystal size and below, the crystal is only deformed by dislocations at a stress level near the theoretical strength of the

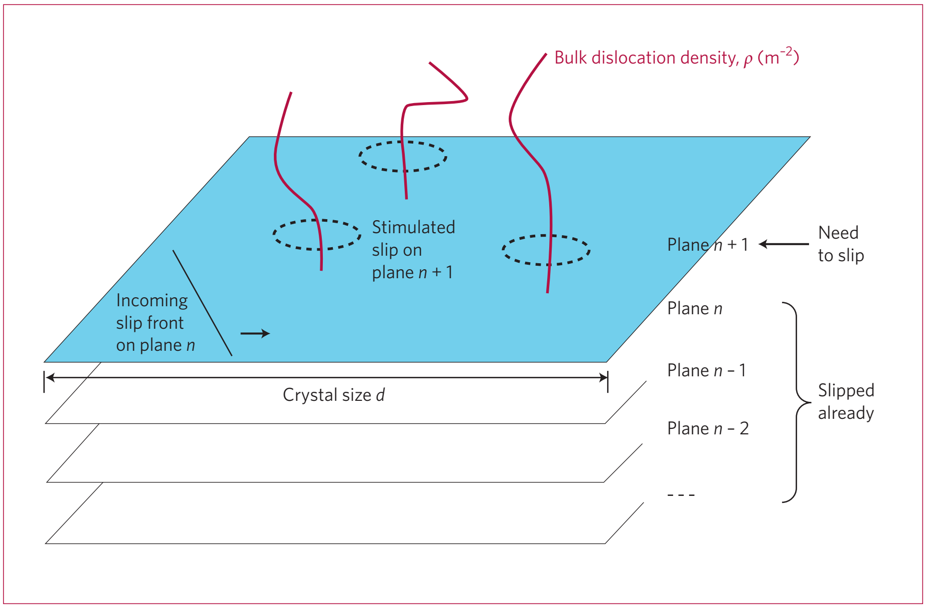

Figure 2 | Schematic of the stimulated-slip model[1]. The formation of the twin occurs by thickening of the twin. This comes about when an incoming slip front hits bulk screw dislocations (red lines) that act as promoters. Such promoters are thought to be only a small subset of the bulk dislocation population and the basis for the proposed stimulated-slip model.

metal. Yu *et al.* support this argument by proposing a simple quantitative model, which agrees with their data, which assumes that not all bulk dislocations can act as a promoter but that only a small subset of about 1% can. This implies that the proposed model is a statistical one. As a result, measured strength data should scatter, to some extent, because the first or the hundredth dislocation in the crystal may act as the twinning promoter. Moreover, the scatter should increase with decreasing crystal size. This is not seen in the published data, possibly because of the small number of samples tested. This is an issue certainly worthy of further study.

The most important technical implication of the observed twinning behaviour is that the critical size is about 1 μm, at which this Ti alloy, and presumably other hexagonal-close-packed metals, deform at the limit of their theoretical strength. This size is about one order of magnitude larger than the one found for fcc or base-centred-cubic metals where theoretical strength is observed when presumably no dislocations are present[7]. Based on this, it could be argued that 1-μm-thick Ti or Mg fibres with the correct texture constitute the ultimate high-strength lightweight metallic material. Such fibres would be easier to produce and manufacture into components compared with metallic nanowires with similar strength. ❐

Oliver Kraft is at the Karlsruhe Institute of Technology, Karlsruhe, Germany.
e-mail: oliver.kraft@kit.edu

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ELECTRON MICROSCOPY

New views of catalysts

Developments in electron microscopy are generating more realistic views of catalysts, allowing optimization of their structure to improve their performance.

Chris Kiely

Improvements to catalysts that even slightly increase their activity, selectivity for products or lifetime can have a huge impact on the economic viability of an industrial-scale chemical process. Recent advances in electron microscope instrumentation and techniques are allowing researchers to obtain unprecedented views of real, as opposed to model, catalyst materials. Crucial new information concerning the identity of the catalytically active species or the mechanism of deactivation can then be fed back into the catalyst design process to produce more effective and stable catalysts. Presentations at the 1st International Symposium on Advanced Electron Microscopy for Catalysis and Energy Storage Materials, held on 17–19 January 2010 at the Fritz-Haber Institute in Berlin, provided a fascinating snapshot of the latest developments in catalyst microscopy. Aberration-corrected microscopes, cryo-tomography and specialty heating stages are shedding new light on the nanoworld of catalysts and are beginning to facilitate improvements in catalyst design.

The sophisticated lens systems designed to correct aberrations in transmission electron microscopes (TEMs) essentially come in two varieties. First, there are instruments with 'probe correctors' that are situated between the electron source and specimen, which allow the formation of high-intensity, subangstrom-sized electron probes and result in greatly improved imaging resolution and higher-sensitivity compositional analysis (Fig. 1). Second, there are those instruments with 'image correctors' placed after the sample that give clearer atomic-resolution lattice images, with fewer image artefacts, of surfaces and interfaces, than are obtainable in conventional high-resolution electron microscopes. Both types of