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A Dislocation Perspective on Strength and Toughness in Ceramics

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ABSTRACT

Ceramics are widely known to be brittle. Yet in recent years, there has been a growing interest in the research domain of strengthening or toughening ceramics with dislocations, a fundamental crystal defect that has been perceived as uncommon or irrelevant in ceramics. With new methodologies being developed for mechanical deformation, fabrication, and processing, dislocations are increasingly spotted or purposely engineered in ceramics for the technological potential they may hold. This article aims to provide an overview of the historical development of this topic concerning dislocation-impacted strength and toughness, covering three major research waves over the last nine decades. This historical retrospective provides valuable insights and useful guidance for understanding the currently ongoing research endeavors in using dislocations for potentially stronger and tougher ceramics, as well as for balancing the mechanical and functional properties. A few outstanding issues concerning dislocation engineering, work hardening, dislocation fatigue, dislocation/defects interactions, and external fields affecting the strength and toughness of ceramics are discussed. Although there have not yet been mature engineering applications for dislocations in ceramics, with dislocation being a fundamental crystal defect and more readily present in ceramics, the increasing proofs-of-concept continue to add weight to its roles played in advanced ceramics engineering.

1 | Introduction

Tackling the conflict between strength and toughness in engineering structural materials has been an ongoing endeavor [1, 2]. An outstanding challenge remains in achieving strong and tough ceramics, which are infamous for being brittle and exhibiting little plasticity at ambient conditions. Oxide ceramics that can be plastically deformed in bulk compression at room temperature (in single-crystal form) have often been tagged as “surprising” [3] or “unexpected” [4]. Due to the brittleness caused by the strong and directional bonding, ceramics have much lower resistance to crack propagation, i.e., fracture toughness (about 0.1–20 $\text{MPa}\cdot\text{m}^{1/2}$) in comparison to metals (about 20–200 $\text{MPa}\cdot\text{m}^{1/2}$). Their brittle nature limits the engineering applications of ceramics as structural/load-bearing materials. Current strategies for toughening structural ceramics can mainly be categorized into process zone toughening (including transformation toughening, ferroelastic toughening, and microcracking toughening), bridging zone toughening, and crack deflection [1, 2].

Apart from these research lines and engineering applications, dislocations in ceramics have experienced a rather bumpy journey, as briefly discussed by the current author [5]. Recent years have witnessed a renewed interest in dislocations in advanced ceramics, owing to the technical potential that dislocations may hold [6]. Contrasting with the previous research waves on dislocation-mediated strength and toughness in structural ceramics, the most recent studies have placed more emphasis on the functional properties. Along this path, a new trend is emerging in fusing these two aspects, namely, by examining the synergy as well as the potential competition/trade-off in using dislocations to modulate both mechanical and physical properties [6, 7].

This work will focus on the mechanical properties, aiming to provide an overview of the role played by dislocations in strength and toughness in oxide ceramics. For those interested in dislocation-tuned functional properties in advanced oxides, the readers may refer to [8–11]. Given the fact that dislocation plasticity as well as dislocation-mediated strength and toughness are predominantly investigated in metallic materials, a few fundamental concepts are often directly borrowed from metals to apply to ceramics. This appears to have created some gaps and will first be explained and discussed (Section 2). This addresses the terms of "ductility," "strength", "toughness," and "fracture toughness," and the discrepancies in the use of these terms across the two material systems for tensile/compressive testing, small-scale/bulk testing, as well as at room/high temperatures will be discussed. The discussion will also extend to independent slip systems and the impact of grain boundaries (GBs), as well as point defects. Then, in the following Section 3, a few historical highlights of dislocations in ceramics on strength and toughness are first gathered, aiming to offer a broader view of these research outcomes. This line of historical evolution is useful to revisit defect engineering in ceramics, which has so far been primarily focused on point defect and planar defect engineering, by reflecting the ongoing efforts following the historical highlights. In Sections 4 and 5, the correlation and discrepancy between toughness and fracture toughness will be discussed. A few outstanding issues toward dislocation engineering in ceramics will be presented in Section 6 for this re-emerging topic. The final objective is to stimulate the discussions on the possibility and feasibility of using dislocations to simultaneously enhance the strength and toughness of some ceramics for advanced engineering applications, which may eventually pave the road for the integration of both mechanical and functional properties [7].

2 | A Few Crevices

There is an increasing tendency to abuse the term *"ductile/ductility"* for plastically deformable ceramics tested under compression. Meanwhile, in the framework of dislocation-mediated plasticity, yield strength should be included in addition to the conventional fracture strength in brittle ceramics. Additionally, the terms "toughness" and "fracture toughness" are often misused. Other confusions or inconsistencies occur regarding the boundary conditions often not being specified, whether in small-scale testing or bulk testing, so is the temperature. For the sake of building a common ground, it is helpful to begin by clarifying these few critical fundamental concepts and defining the boundary conditions, before we discuss the content in the following Sections 3–5.

For **ductility** in ceramics, the author refers to Sprackling in his monograph on the plastic deformation of simple ionic crystals [12]: "*Ductility is difficult to define in any precise quantitative manner, but a working rule often adopted is to describe a sample deformed in tension as showing ductility if there is a measurable decrease in cross-section at fracture.*" Accepting this definition, as a consensus in both textbooks and in application, rules out the many claims of "ductile ceramics" popping out recently in the literature. Most of these false claims in the literature appear to be caused simply by the misuse of the term, without referring to or being aware of the definition "in tension." It may also be the case that, since it was first used in metals subjected to tensile testing, discussions and usage of this term at a later stage are by default for tensile tests, but have not been specified or have been simply neglected when transferring the terms to ceramics. This confusion has now accumulated in the domain of ceramics that can be plastically deformed at ambient conditions. The question now is: even if a ceramic piece can be plastically deformed in tension, can we label it as truly ductile? This comes to the latter part of Sprackling's definition regarding the "measurable decrease in cross-section at fracture." In this sense, most ceramics, even if plastically deformable (as predominantly in compression), fail to meet the criterion for being labeled as ductile at room temperature or low temperature. Truly ductile ceramics are available at high enough temperatures, where drawing/pulling in tension has been proven feasible in the case of superplasticity of nanograined ceramics [13–15]. Nevertheless, the most exciting yet frustrating pursuit appears to lie in room-temperature "ductile ceramics," with a few exceptions to be discussed later in detail.

Strength in ceramics primarily refers to *fracture strength* (in compression, tension, and bending for flexural strength [16], see also Figure 1). When dislocation plasticity is considered, regardless of the temperature, length scales, or microstructure, *yield strength* should be included when fracture does not occur prematurely. It should be specified here that the yield strength for plastically deformable ceramics is predominantly given in compression. The definition of yield strength, as originally defined in tension for metals (Figure 1b), may be adopted for ceramics under compression for convenience. Namely, the yield strength can be defined with a 0.2% offset as with metals [16] (note that a 0.1% offset was also previously used as in [12]). Other, less popular definitions, such as using the proportional limit or the lower yield point in the case of yield drop [17, 18], are also available, as discussed by Sprackling [12].

As illustrated later in Figure 1, for metals widely used in structural applications, the yield stress is usually lower than the (tensile) fracture strength (with a ratio of about 1.1–3 [16]), where dislocations cause yielding but also contribute to increasing the fracture toughness. In contrast, for most ceramics subjected to bulk tests, the (tensile) fracture strength is much lower than the (compressive) fracture strength (with a ratio of about 10–15 [16]), as well as the yield strength (provided that the materials can yield before fracture). For ceramics to exhibit appreciable plasticity, it is required that the yield strength be lower than the fracture strength. This is achievable depending on the temperature, length scale, and microstructure of the materials, which will be discussed in the later sections.

Toughness and fracture toughness are often used interchangeably in ceramics literature. **Toughness** is defined as a material's capability to absorb energy and plastically deform without fracturing. One definition is the amount of energy per unit volume that a material can absorb before fracture (shaded area under the stress–strain curve in Figure 1a, note that here the schematic is for tension). High toughness requires both high yield strength

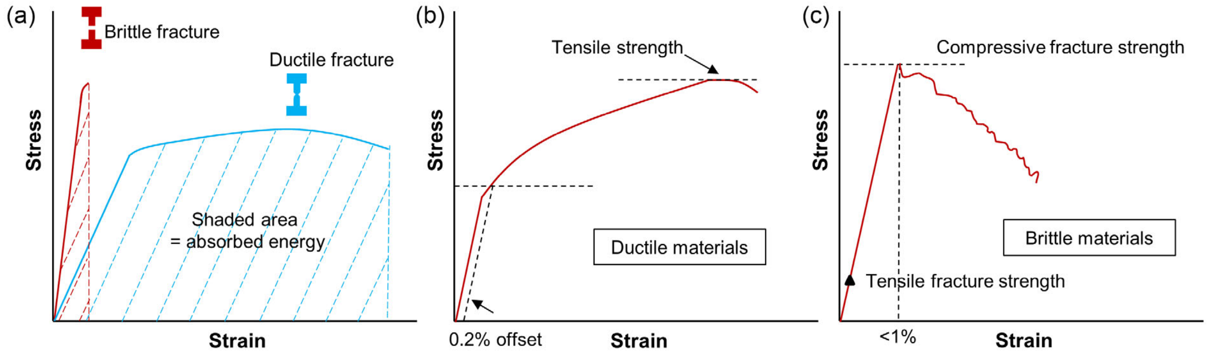


FIGURE 1 | The shaded area underneath the stress–strain curves in (a) indicates the toughness, i.e., absorbed energy. Representative stress–strain curves for (b) ductile metals in tension, and (c) brittle ceramics in tension and/or compression. Fracture toughness, i.e., the resistance to crack propagation, is not included in this figure. Note that for brittle ceramics, the fracture strength depends strongly on the loading mode, with tensile fracture strength much lower (about 1/10–1/15) than the compressive fracture strength. Yield strength for plastically deformable ceramics is not directly depicted here but resembles that in (b).

and ductility. For the sake of discussing plastically deformable ceramics, it might be convenient to refer to good plasticity (as mostly tested in compression) instead of ductility. The Charpy and Izod notched impact strength tests are typical ASTM tests used to determine the toughness of metals. **Fracture toughness**, on the other hand, refers to the resistance to crack propagation. A prerequisite here is the existence of a crack. ASTM testing standards for determining the fracture toughness of ceramics can be found in [19]. Note that the unit for fracture toughness is either MPa·$m^{1/2}$ (stress intensity factor *K*-criterion) or J/m^2 (energy release rate *G*-criterion), which is different from the unit of toughness (J/m^3).

All these four mechanical properties are sensitive to structure, size, and temperature [16]. By manipulating the microstructures, testing temperature, or often by simply decreasing the tested volume from macroscale to nano-/microscale, the values for these properties can be very different. These aspects concerning the size effect, temperature, and microstructure (GBs, dislocations, and point defect densities) will be discussed later.

3 | A Brief Historical Retrospective

The author has previously presented a brief overview [5] discussing the historical research lines for dislocations in ceramics. Here, more details are included, with a focus on the impact of dislocations on strength and toughness. The following Sections 3.1–3.3 focus more on strength, while an independent Section 4 is devoted to discussing toughness and fracture toughness influenced by dislocations.

To begin with, dislocations in ceramics have enjoyed a long yet less-known history. Looking over the last century's research in this research domain, three major research waves (Figure 2) can roughly be identified since the conceptualization of edge dislocation independently by Taylor [25, 26], Orowan [27], and Polanyi [28] in 1934 to explain the large discrepancy between the measured yield strength and the theoretical upper bound in metals [29]. Nevertheless, predating the conceptualization of dislocations, bulk plastic deformation in minerals [30], with a particular focus on rock salt, was studied in the 1920s [31]. Interestingly enough, room-temperature tensile tests were also successfully performed on some of these samples. Given the fact that these experimental studies are phenomenological, they are excluded from Figure 2 as well as from the following discussions.

3.1 | The 1st Wave: Yield Strength in Bulk Simple Ionic Crystals at Room Temperature

Early research on dislocations in ceramics (primarily on single crystals) appeared to be motivated by confirming the dislocation theories developed in the 1930s, with the materials and experimental techniques just becoming available at that time [32, 33]. It was no surprise that crystal plasticity, particularly in these conventionally-believed-to-be brittle materials, was fascinating also from a purely curiosity-driven perspective. Another advantage of using these simple ionic crystals is that the limited number of independent slip systems at room temperature, as well as the lower density of dislocations, makes it easier to interpret the experimental observations, in which the chemical etching technique has played [33] and is still playing a critical role.

Room-temperature bulk compression of simple ionic crystals such as NaCl, KCl, KBr, and LiF. have been extensively studied within the framework of dislocation theory since the 1950s, extending to the 1980s. These crystals exhibit surprisingly low yield strength, falling between several MPa and tens of MPa at room temperature. Sprackling summarized the studies in 1976 [12], followed by another short review that appeared later by P. Haasen in 1985 [34], summarizing the interaction between dislocations and point defects in ionic crystals.

In LiF crystals, a series of classic and elegant studies on dislocation multiplication, motion, and nucleation was carried out by Gilman & Johnston [35], who used the chemical etching method to directly visualize the dislocation etch pits (Figure 3). This dislocation etch pit method has played a significant role in understanding the dislocation behaviors [29], such as the edge/screw dislocation velocity [32], dislocation multiplication mechanisms [21], as well as the importance of dislocation sources for plasticity [37]. Although the resolution is limited, it can provide a large overview for statistical analysis, as well as local identification

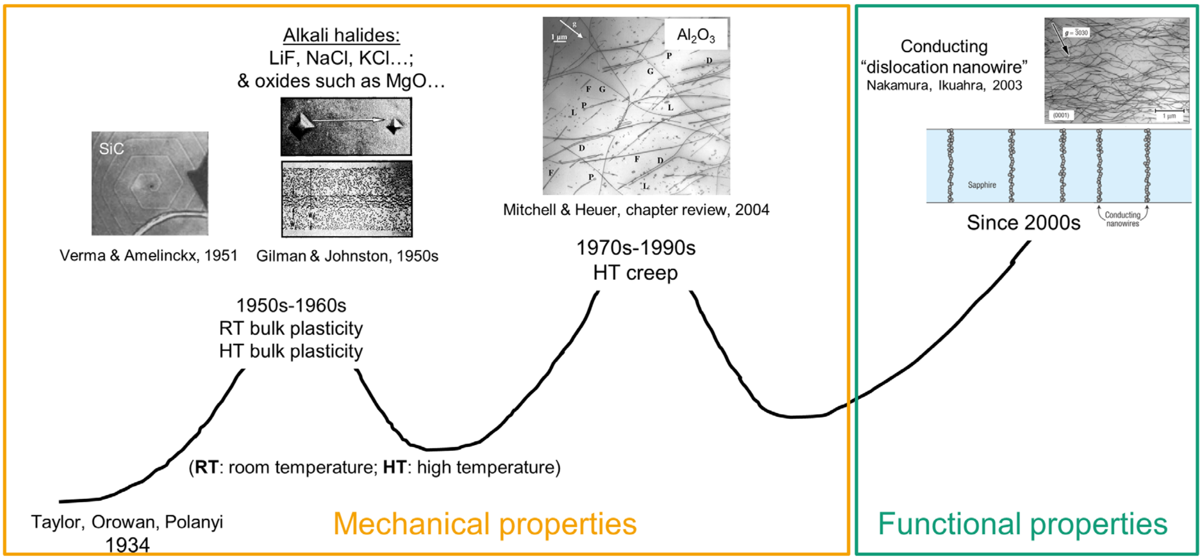


FIGURE 2 | Three major research waves on dislocations in ceramics since the conceptualization of edge dislocation in 1934. The first two waves on the mechanical properties primarily focused on bulk deformation at both room temperature (RT) and high temperature (HT). The third wave has much more weight on the functional properties and small-scale testing at nano-/microscales. Note that the earlier research on plastic deformation of rock salts predating 1934 is not included. Figure adapted from Reference [5], with permission from the American Ceramic Society. The micrographs are adapted from Reference [20] with permission, copyright of Springer Nature; Reference [21] with permission, copyright of AIP Publishing, and Reference [22] with permission, copyright of Springer Nature. For completeness, the continuous research endeavors on dislocations in semiconductors (with some overlap in oxides) can be found in the monograph by Holt and Yacobi [23], and the review by Figelski [24].

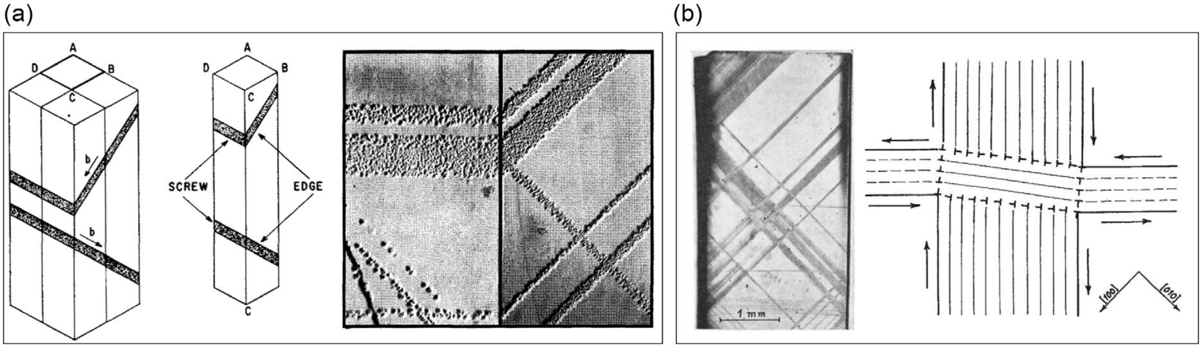


FIGURE 3 | (a) Room-temperature uniaxial bulk compression of LiF crystals, with the etch pit studies revealing the slip bands and dislocation types (edge/screw). Image reproduced from Reference [21] with permission, copyright of AIP publishing. (b) Plastically deformed MgO crystals under uniaxial bulk compression at room temperature (left), with the schematic drawing (right) illustrating the formation of the kink bands due to intersecting slip bands. The formation of kink bands can eventually lead to local crystal rotation and crack formation. Image adapted from References [36] with permission, copyright of Springer Nature.

of the dislocation type and 3D distribution of the dislocations. This can be highly complementary to the electron microscope techniques for understanding dislocations in new materials. A prerequisite, however, is to find a suitable chemical solution for etching.

For simple oxides with ionic bonding, MgO stands out as a prototypical material (Figure 3), which has arguably been the most extensively studied simple oxide for dislocations so far. In the 1950s–1960s, Stokes [38–40], Argon and Orowan [36, 41], and Pask [42–44], among many others, pioneered the dislocation research in MgO, primarily at room temperature in bulk deformation, including tensile tests [39]. To date, MgO still finds great interest in dislocation plasticity for its fundamental role in, e.g., geology [45] and simulation [46]. A comprehensive review for dislocations in MgO can be found in Reference [47]. More bulk plasticity studies on other simple oxides, such as SrO and NiO, at room/cryo-temperatures can be found in the most recent materials list compiled by Frisch et al. [48].

Due to the ionic bonding in these materials, the charged dislocations, originally addressed in the seminal work of J. Eshelby et al. [49], have attracted considerable attention since the 1950s. So far, the most comprehensive summary on charged dislocations in alkali halides has remained the review by Whitworth [50], which appeared in 1975. Later, Y. Osip'yan et al. in 1986 [51] summarized the interaction between electric charges and mobile dislocations in II–VI group semiconductors, with a focus on the *photoplastic effect* (where dislocation plasticity is affected by light illumination, first reported on CdS crystals by Osip'yan and

Savchenko in 1968 [52], see Figure 4a), and the *electroplastic effect* (where dislocation plasticity is influenced by an electric field, see Figure 4b, by Kataoka et al. [53] in 1975). More recently, there have been renewed studies of photoplasticity [54–56] and electroplasticity [57].

This raises the question of why these ionic crystals exhibit such low lattice friction stress that allows for easy glide of dislocations once they are mechanically generated. First-principles calculations in AgCl and NaCl have provided new insights [58, 59]. The local electronic structure around the dislocation cores, as well as the bonding interactions between atomic orbitals, have been examined. A much smaller excess energy was found due to the small dislocation core energy for the screw dislocations in AgCl compared to NaCl. Simulations suggest that the dislocations in these simple ionic crystals can be formed more easily. In both cases, the dislocations show an extended core (undissociated), which is the same case for simple oxides such as MgO [47].

3.2 | The 2nd Wave: High-Temperature Dislocations in Structural Ceramics

By the 1970s, likely due to the requirements of aerospace engineering for using high-temperature structural ceramics and the readily accessible high-temperature deformation setup [30], the research focus on dislocations in ceramics shifted to high-temperature deformation, where plastic yielding behavior and dislocation creep became the main focus [30, 60, 61]. The representative materials investigated were Al_2O_3, ZrO_2, TiO_2, spinel ($MgAl_2O_4$), and forsterite (Mg_2SiO_4) [30]. During this period, the transmission electron microscope (TEM) became widely available, providing more direct evidence and revealing the complexity of the dislocation structure at the nanoscale. A typical example is the TEM observation of the climb dissociation of dislocations in sapphire (α-Al_2O_3) tested at 1400°C [62].

Realizing the importance of dislocation plasticity in structural ceramics at high temperatures (e.g., above $0.5T_m$, T_m being the melting temperature), and the fact that "*in this* (*temperature*) *regime*, *interaction of point defects with dislocations is especially significant*," Mitchell et al. [63] assembled an overview entitled "Interaction Between Point Defects and Dislocations in Oxides." An excellent example illustrating the importance of dislocation–point defect interaction is the temperature-dependent yield strength of single-crystal TiO_2, reported by Ashbee and Smallman [64, 65], which exhibited a double up-and-down trend in the critical resolved shear stress as a function of temperature (see Figure 5b). This also extends to the ongoing investigations of the impact of defect chemistry (point defect engineering) on room-temperature plasticity in reduced and doped oxides, which will be discussed in greater detail later in the 3rd wave.

Another case study was the temperature-dependent plasticity in yttria-stabilized zirconia (YSZ), systematically carried out by Heuer's group, as assessed by high-temperature Vickers indentation and chemical etching (see Figure 6). An interesting technique of sequential polishing and etching was developed to provide an in-depth dislocation distribution in YSZ, as presented by Holmes et al. [68]. A similar methodology has also been used recently in MgO [69] and $SrTiO_3$ [70] for 3D reconstruction of dislocation structures underneath the indents.

Possibly the most widely studied structural oxide ceramic for high-temperature deformation is alumina, owing to its technical relevance, relatively easy access to crystals, and the precise control of defects/impurities. In a concise and elegant study, Heuer et al. [17] investigated the yield point of sapphire via high-temperature (1500°C) tensile tests. They compared the different crystal conditions (as received, reground, air/vacuum annealed, flame polished) and their impact on the yield drop (see Figure 7) and concluded that the initial mobile dislocation source density, instead of the pinning of dislocations by impurities (Cottrell mechanism), controls the lower yield point. Such a yield point can be increased by carefully polishing the surface using flame polishing or annealing. Later, in 2017, Nakamura et al. [71] proposed that the recombination of climb-dissociated partial dislocations back onto the basal glide plane can be another origin of the yield drop behavior in sapphire deformed via basal slip. Again, the Cottrell mechanism was ruled out for the yield drop, as well as the Johnston–Gilman model for dislocation multiplication.

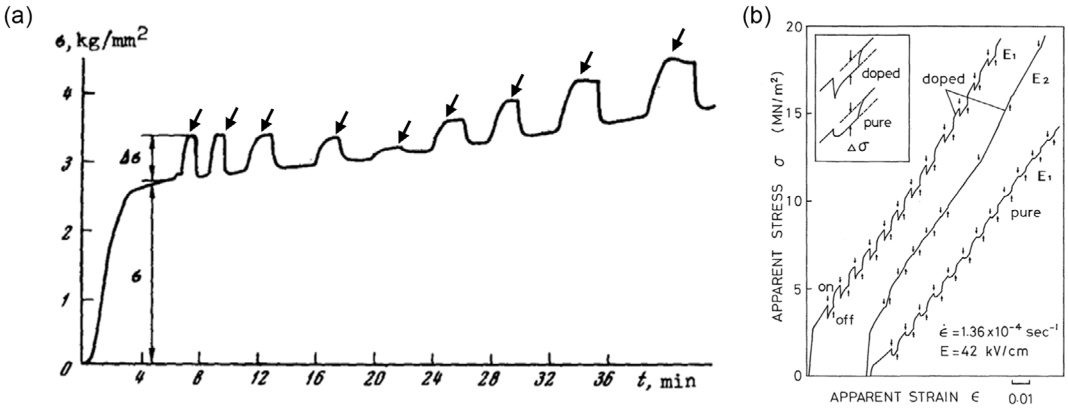


FIGURE 4 | (a) Impact of light on the flow stress (note the increase under light illumination, indicated by the arrows) in CdS crystals tested at 75°C. The term photoplasticity was coined for such an effect. Image adapted from Reference [52]. (b) Fluctuation in the flow stress in KCl (for both doped and pure crystals) in accordance with the application and cut off (indicated by the arrows) of the electric field, for the phenomenon termed electroplasticity. The arrows indicate the onset and cut-off of the electric field. Image reproduced from Reference [53], copyright of IOP Publishing.

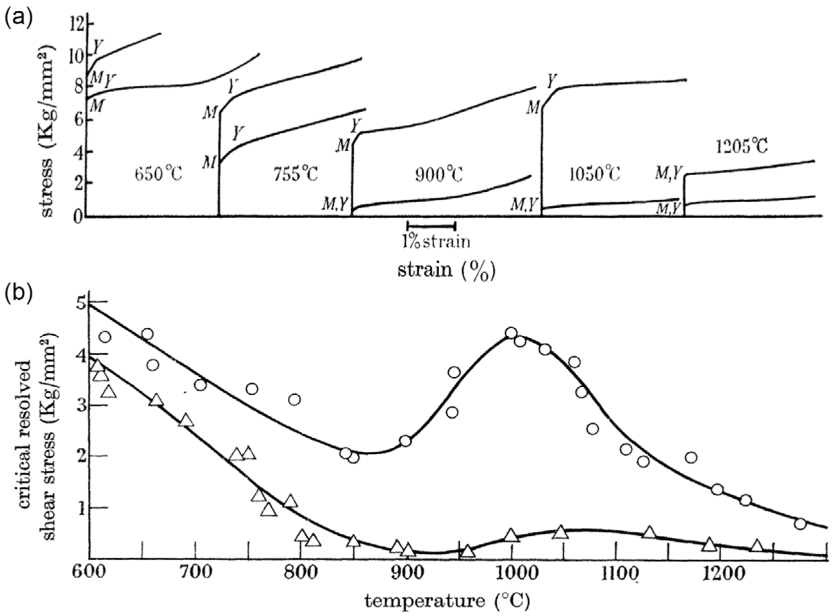


FIGURE 5 | (a) Stress–strain curves for plastically deformed TiO_2 bulk crystals at various temperatures. The lower curves refer to the tests performed at 760 mmHg pressure, and the upper curves at 10^{-5} mmHg pressure. (b) Variation of the critically resolved shear stress for the (−101) [66] slip system as a function of temperature, with the upper data points (circles) corresponding to tests performed at 10^{-5} mmHg pressure, and the lower data points (triangles) at 760 mmHg pressure. The original pressure conditions are kept here for consistency. Images adapted from Reference [64] with permission, copyright of the Royal Society.

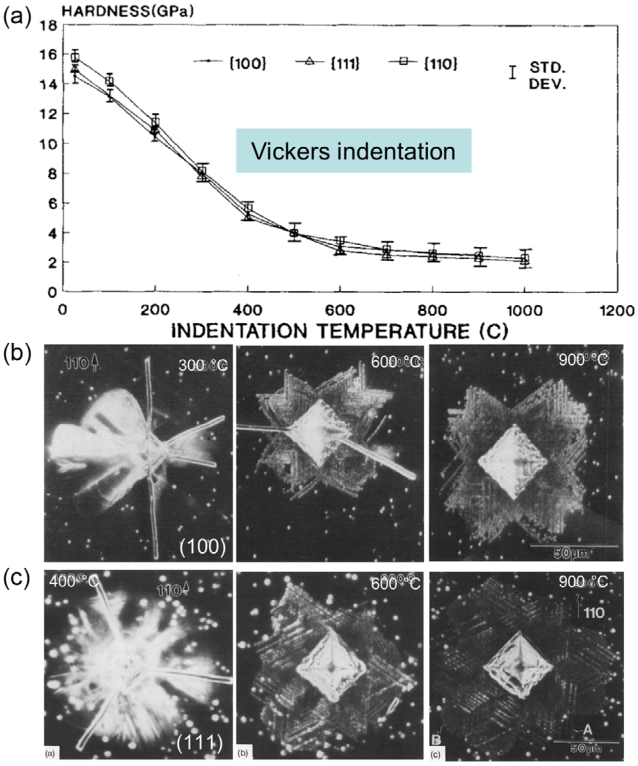


FIGURE 6 | (a) Temperature-dependent Vickers hardness for 9.5 mol%-Y_2O_3-fully-stabilized zirconia; (b) etch pits studies at selected temperatures on the (100) surface; and (c) etch pit studies at selected temperatures on the (111) surface. Images adapted from Reference [67] with permission, copyright The American Ceramic Society.

3.3 | The 3rd Wave: Dislocation Technology in Functional Ceramics

Contrasting with the previous research interests with a major focus on mechanical behavior, the latest research efforts are fueled by a new aspect of functionalizing dislocations in advanced ceramics, with mechanical properties running in parallel [6, 7]. This research wave witnessed a more aware and intentional approach to dislocation engineering. To make use of dislocations for functionality tuning, the bottleneck in dislocation engineering must first be overcome (see also Section 6). For bulk samples, the main method to generate controllable dislocations is through high-temperature deformation. As will be showcased later, this current research wave has a more versatile feature of various research topics.

3.3.1 | High-Temperature Dislocation Engineering

In 1983, W. Shockley speculated on the concept of "using dislocations in (semiconductor) crystals as microwiring" [72], which could be pictured as a conductive tube (along the dislocation core) filled with metallic elements. The experiment was realized 20 years later by Nakamura et al. [22], who successfully deformed bulk single-crystal sapphire at 1400°C and then diffused Ti along the dislocations to achieve a ~10^{13} times higher conductivity compared to the pristine insulating sapphire. Based on these proofs-of-concept results, the (dislocation) nanowire design for "dislocation technology" was proposed by Nakamura and Ikuhara et al. [9, 22]. This methodology was established on Al_2O_3, involving high-temperature uniaxial bulk compression for dislocation engineering and subsequent slicing of the samples containing dislocations. Later, it was extended to YSZ [73], followed by using bicrystal

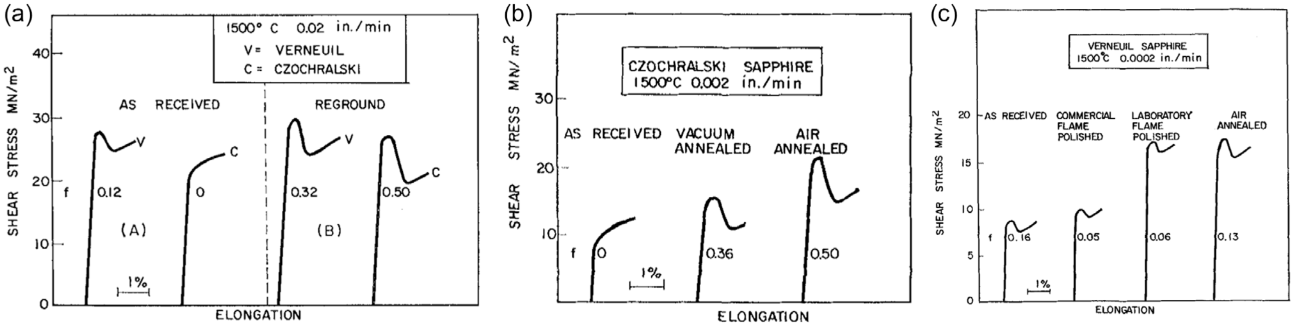


FIGURE 7 | Yield point of bulk sapphire in high-temperature tensile tests. Two types of crystals were used, grown by the Verneuil (V) method and the Czochralski (C) method, respectively. (a–c) Different crystal conditions (as received, reground, air/vacuum annealed, flame polished) before the mechanical loading were investigated to understand the impact of the initial density of mobile dislocation sources on the yield strength. Note that only the as-received Czochralski-grown crystals (with a higher dislocation source density) do not show a drop in the yielding behavior. Images adopted from Reference [17] with permission, copyright The American Ceramic Society.

$LiNbO_3$ [74], for evaluating the electrical conductivity enhanced by dislocations. Worth noting is that, without decorating the dislocations with metal elements, as in Reference [22], the increase in electrical conductivity by dislocations appears marginal. In Rödel's group, high-temperature deformation of bulk crystals further finds its playground in $BaTiO_3$ [11] for tuning the electromechanical properties via dislocation-domain wall interaction, enhancing the photocatalytic efficiency [10], and in TiO_2 [75] for regulating the electrical conductivity, where the concept of mechanical imprinting and mechanical doping (or dislocation self-doping, contrasting with conventional doping strategies using point defects) were coined and validated.

These high-temperature tests also provided new insights into dislocation-tuned strength. For instance, Nakamura et al. [76] compared the deformed single-crystal alumina at 1400°C with that first deformed at 1400°C and then continued at 1200°C (see Figure 8), and found an increased yield point during the 2nd stage of deformation. The slip traces and dislocation structures were also altered distinctly. This finding is in line with a later example of first performing a high-temperature test followed by room-temperature deformation [77], where a large increase in the plastic strain at room temperature was achieved, with yet still a very high yield strength, as will be further discussed later.

3.3.2 | Room-Temperature Plasticity in Oxides: Bulk Versus Small-Scale Tests

A. Bulk Deformation at Room Temperature

In 2001, Brunner et al. [3] first reported the bulk plasticity in single-crystal $SrTiO_3$ under uniaxial compression at room temperature, with a yield strength of ~120 MPa along the <110> axis. This surprising finding sparked new interest in understanding the origins of this room-temperature dislocation plasticity. Given that $SrTiO_3$ is one of the most adopted model perovskite oxides for versatile functionality studies in electronics and solid-state physics, the bulk plasticity found in $SrTiO_3$ has further boosted the exploration of using dislocations for probing dislocation-tuned electrical conductivity [78], thermal conductivity [79, 80], photoconductivity [81], superconductivity [82], and flexoelectricity [83], to name a few.

In 2016, Mark et al. [4] reported the temperature-dependent plastic deformation in single-crystal $KNbO_3$, including room-temperature

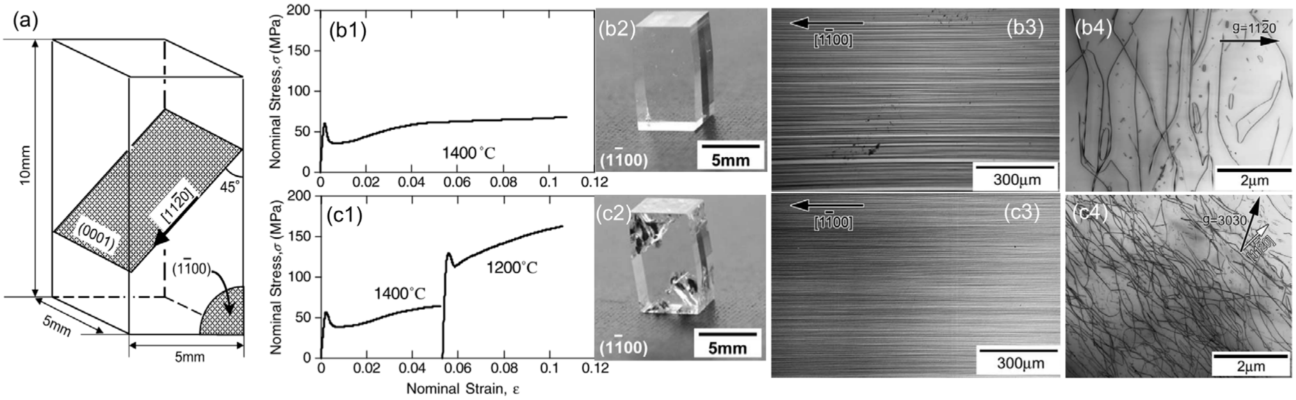


FIGURE 8 | High-temperature compression of single-crystal Al_2O_3: (a) Schematic of the crystal orientation, with the easy glide basal plane is highlighted in gray; (b1–b4) deformation at 1400°C, with stress–strain curve, crystal shape, surface slip traces, and dislocation structures presented; and (c1–4) deformation first at 1400°C, followed by deformation at 1200°C: where the strain–strain curve exhibit more pronounced hardening, the crystal deformed with much finer surface slip traces and higher-density dislocations which are also more uniformly distributed. Images adapted from Reference [76] with permission, copyright Elsevier Ltd.

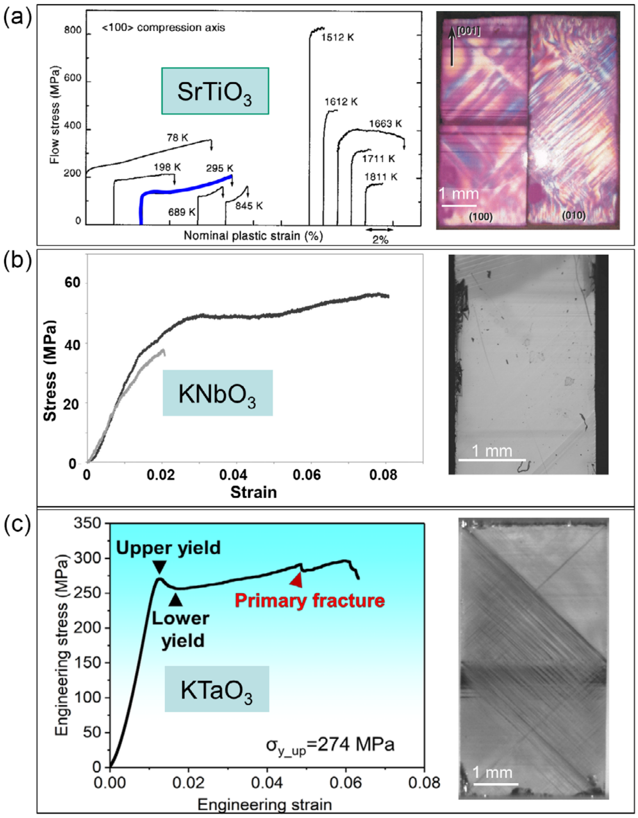


FIGURE 9 | Room-temperature bulk plasticity under uniaxial compression for single-crystal perovskite oxides: (a) $SrTiO_3$ (blue curve for room-temperature test); (b) $KNbO_3$; and (c) $KTaO_3$. For each subfigure, the representative stress–strain curves are presented on the left, with the right images showcasing the samples' surfaces after plastic deformation, where slip traces are visible. Images reproduced from Reference [3] with permission, copyright The American Ceramic Society, Reference [4] with permission, copyright Elsevier Ltd., and Reference [18], under Creative Commons CC-BY License.

bulk plasticity. Motivated by the dislocation-tuned functionalities studies, Höfling et al. [84] experimentally validated the bulk plasticity in $KNbO_3$ and further investigated the dislocation and domain structure at various length scales using different dislocation imaging techniques, including electron channeling contrast imaging and TEM. Down this line, the author and coworkers [18] reported the room-temperature bulk plasticity in $KTaO_3$ in 2024, in parallel with Greven's group [85].

At the time of writing, in the literature, there are only three perovskite oxides ($SrTiO_3$ [3, 86, 87], $KNbO_3$ [4, 84], and $KTaO_3$ [18, 85, 88]) reported that exhibit room-temperature plasticity under bulk compression (see also Figure 9). Spotting one of the main bottlenecks as being limited to the plastically deformable materials, the author's group has attempted to summarize ceramics that exhibit room-temperature bulk plasticity, and 44 ceramic compounds have been listed [48]. Clearly, the number is limited. Nevertheless, it may not be that pessimistic: there appears to have been little effort so far in actively searching for more oxides that exhibit bulk plasticity at room or low temperatures. A feasible starting point can be looking at the solid solution for $KNbO_3$ and $KTaO_3$, namely, $KTa_xNb_{1-x}O_3$ (x is in the range from 0 to 1), which has been experimentally confirmed by the current author's group. Finding more room-temperature plastically deformable perovskite oxides can be foreseen to bring major advantages for fast, efficient, and cost-effective dislocation engineering compared to the energy-intensive, high-temperature deformation.

B. Small-Scale Deformation

Due to the limited number of bulk plastically deformable ceramics at room temperature, the relevant studies on various materials have been predominantly focused on small-scale testing, including nanoindentation [89–93] (see Figure 10), nano-/micropillar compression [94–97] (see Figure 11), and in situ

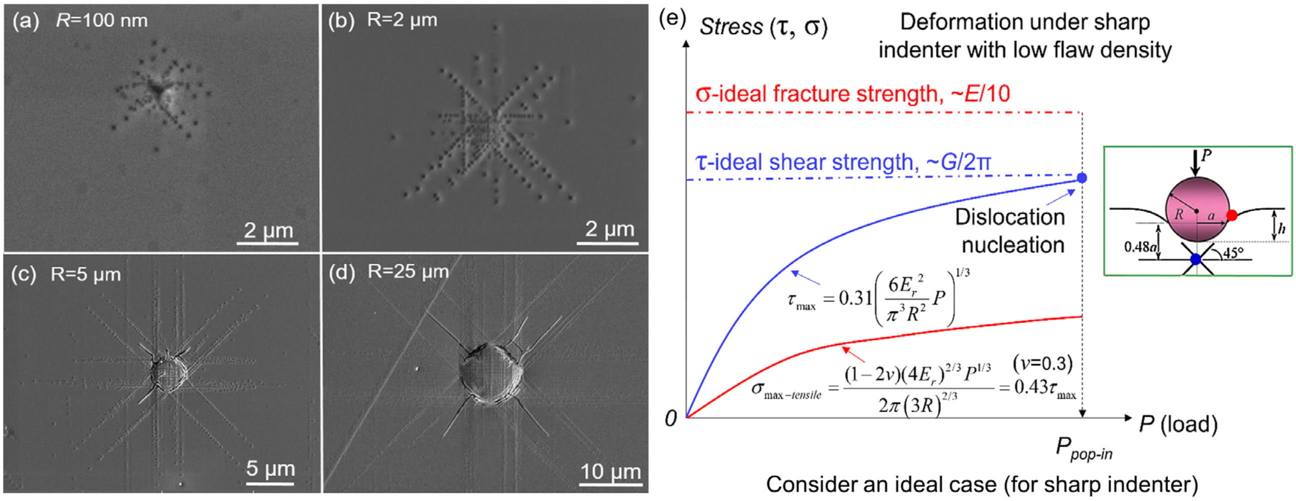


FIGURE 10 | Size effect for the competition between incipient plasticity and fracture during pop-in events in nanoindentation tests: (a–d) transition from purely dislocation plasticity to concomitant cracking and plasticity with increasing tip radius and (e) for a sharp indenter in an ideal case, the Hertzian contact theory predicts that the applied maximum shear stress (for plasticity) under the indenter tip is always larger than the tensile stress (for fracture), reaching the ideal shear strength first to generate dislocations. Images reproduced from Reference [92], under Creative Commons CC-BY License.

tensile testing in the TEM [99, 100]. To avoid the complex stress field underneath the indenter tip, small-scale uniaxial testing has been favored for understanding the dislocation mechanisms, which has been made available thanks to the development of FIB (focused ion beam) and in situ testing setup in the scanning electron microscope (SEM) and TEM for directly visualizing the dislocation behavior.

The small deformed volumes (typically ranging from hundreds of nm^3 to tens of μm^3) significantly minimize the number of pre-existing flaws. The additional geometrical constraints (in pillar compression) as well as the locally high compressive hydrostatic stress (in nanoindentation) favor plastic flow over crack initiation and suppress crack propagation. In contrast, conventionally sintered bulk ceramic samples or grown single crystals have a much lower density of grown-in dislocations (e.g., $10^{10}/m^2$ or lower, depending on the crystal growth techniques). These samples can be considered as dislocation-free or dislocation-scarce. For dislocation plasticity to occur in these samples, the bottleneck is shifted from dislocation motion to dislocation nucleation [101], which has to be effectively operative (to also facilitate dislocation multiplication [102]) before fracture occurs. For a dislocation-free volume to be plastically deformed, homogeneous dislocation nucleation needs to take place, which is energetically unfavorable and requires stresses approaching the theoretical shear strength of $G/2\pi$ (G is the shear modulus) [66]. This stress level is often two orders of magnitude higher than the actual fracture strength of most ceramics. The lower fracture strength in bulk ceramic samples is often correlated with the pre-existing flaws, such as pores and GBs from sintering, and (surface) microcracks from the machining of the samples, which, in the small-scale testing, are significantly minimized or intentionally avoided.

As briefly discussed above, by overcoming the barrier for dislocation nucleation, namely, by engineering pre-existing dislocations into the samples by increasing the dislocation sources, dislocation plasticity can be significantly enhanced. This has been achieved for both single crystals and polycrystalline oxides. For single crystals, the author and coworkers have successfully applied the concept of "mechanically seeded dislocation" [103] in single-crystal $SrTiO_3$ [104], MgO [105], $KTaO_3$ [106], etc., to significantly improve the plasticity in compression. The term "mechanical seeding" [103] was coined due to the fact that these dislocations are introduced by mechanical deformation. Although these compression tests were performed in microscale samples, mesoscale testing using the cyclic Brinell indentation/scratching method suggests that this concept can be extended to hundreds of micrometers, with tunable dislocation densities over 4–5 orders of magnitude. This further allows for tuning the yield strength too, as will be presented with more detail in the next Section 3.3.3. In principle, improved bulk compression plasticity can be expected, provided the dislocations are successfully engineered into the whole sample volume, which is still being attempted. Another open question is, with these mechanically seeded dislocations, can tensile plasticity be realized to finally confirm the "ductility" of these materials by circumventing the bottleneck of dislocation nucleation?

For polycrystalline samples, the tests are primarily limited to microsized pillars, where the samples are FIB-milled from flash-sintered samples. Flash sintering is a relatively new method to achieve fast densification of ceramic green bodies. Due to the highly nonequilibrium process, various types of crystal defects, including dislocations, stacking faults, and a high concentration of point defects (vacancies), can be generated, in addition to traditional flaws such as voids, pores, and GBs in conventionally sintered samples. Although a comprehensive mechanistic understanding of the generation of dislocations is under investigation [107], it does not compromise the promising outcome of improved compressive plasticity in micropillar compressions in various oxide ceramics [108], as showcased in Figure 12.

To rule out the complexity of other defects in polycrystalline samples and focus on the impact of dislocations improving the plasticity, high-temperature preloading of single-crystal oxides was performed [77] (see Figure 13, on single-crystal TiO_2). It is worth noting that the yield strength is still rather high (higher than 4 GPa in Figure 13) after mechanically seeding the micropillar

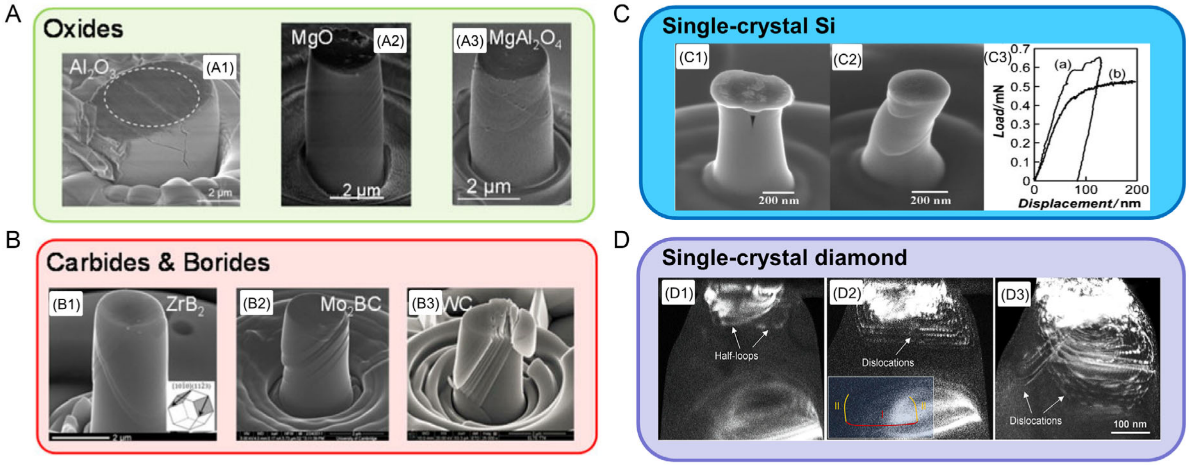


FIGURE 11 | Micro-/nanopillar compression of various brittle materials tested at room temperature, showing appreciable plasticity: (a,b) Oxides, carbides, & borides. Images adapted from Reference [94], under Creative Commons Attribution 4.0 International License. (c) Single-crystal Si with the size-dependent "brittle to ductile" transition. Image reproduced from Reference [96] with permission, copyright 2009 Wiley. (d) Single-crystal diamond. Images reproduced from Reference [98] with permission, copyright Elsevier Inc.

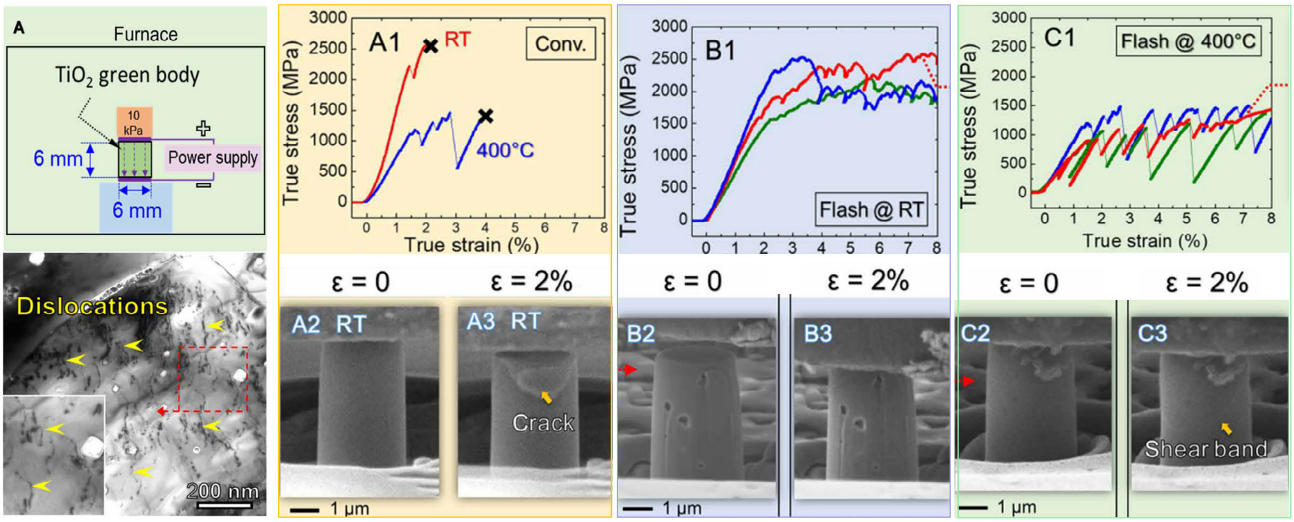


FIGURE 12 | (A) Flash-sintering generated a high density of defects that facilitate the plasticity in micropillar compression, showcased in polycrystalline TiO_2. (A1–C3) In comparison to a conventionally sintered sample that fractured almost immediately after the elastic limit, flash-sintered samples display even room-temperature plastic deformation up to about 10% total strain without fracture. Images adapted from Reference [109], under Creative Commons CC-BY-NC license. Copyright The American Association for the Advancement of Science.

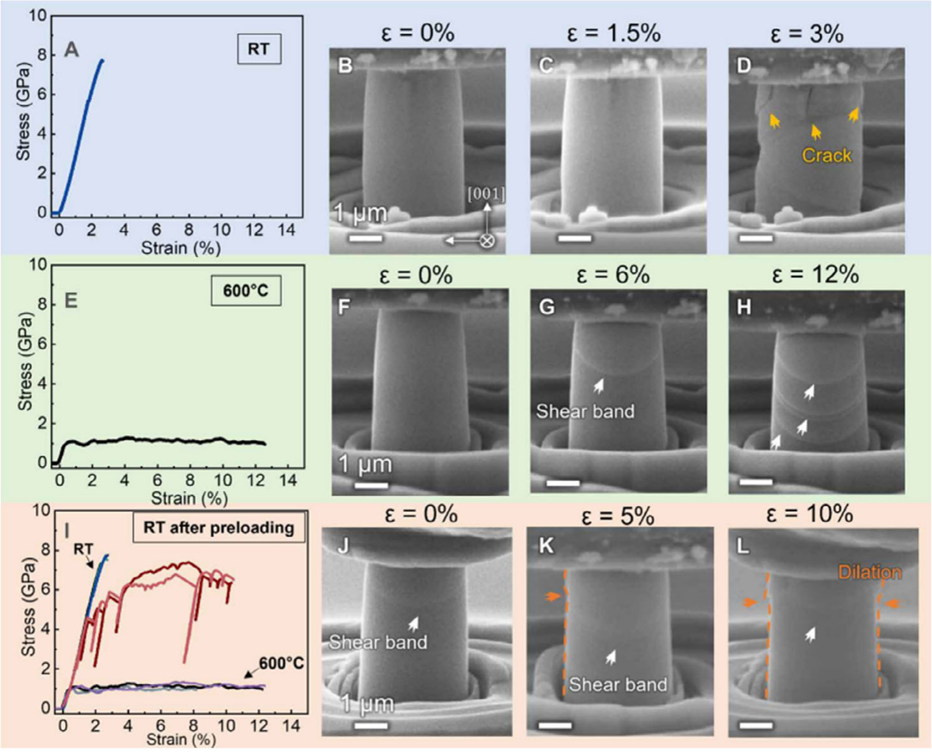


FIGURE 13 | Preloading single-crystal TiO_2 at elevated temperature, aided by thermal activation, to generate dislocations as sources for the following room-temperature deformation. This high-temperature preloading approach can effectively increase the plastic deformability at room temperature on a small scale. Images reproduced from Reference [77], under Creative Commons CC-BY-NC license. Copyright The American Association for the Advancement of Science.

with preloading-induced dislocations. This value is much higher than the ~1 GPa at 600°C for the 1^{st} round of pillar compression. This difference in the yield strength is more evident for single-crystal alumina [77], where the yield strength after dislocation seeding is almost the same as the fracture strength before seeding, suggesting that the high-temperature dislocations still encounter a very high lattice friction stress at room temperature. Moreover, the size effect also plays an important role here in producing the very high strength values in micropillar compression, as compared to the critical resolved shear stress of ~40–50 MPa in bulk deformation of TiO_2, see Figure 5 by Ashbee et al. [64] and ~40 MPa at the lower yield point in single-crystal alumina (see Figure 8 by Nakamura et al. [76]). Besides the size effect, the comparison here needs further consideration of the GBs in the flash-sintered polycrystalline samples (as do the micropillars, although the number of GBs may not be very high). Scaling up to bulk samples may find its challenge in shifting the dominating defects from dislocations/stacking faults to GBs and pores as stress concentrators in polycrystalline samples, as will be discussed later in Sections 4 and 5.

In short, conventional high-temperature sintering renders ceramics virtually dislocation-free/scarce. Homogeneous dislocation nucleation in a perfect crystal lattice requires approximately the theoretical shear strength (a fraction of the shear modulus). It was also calculated that nucleating the first dislocation from a grain boundary can be extremely difficult, requiring stresses in the range of 20 GPa for $SrTiO_3$ (at room temperature) [110]. Hence, as a starting point, plastic yield in ceramics is limited by dislocation nucleation. In fact, it is strikingly easy to move dislocations in many ceramics even at room temperature [48], provided they are nucleated or generated via synthesis. The underlying mechanisms for such low Peierls stress are still being investigated. This, however, does not affect the dislocation engineering for achieving plastic yield in some ceramics.

3.3.3 | Strength Tuning Via Dislocation Engineering

In Section 3.3.2, abundant experimental evidence has been provided, including flash sintering and high-temperature predeformation, to showcase that circumventing dislocation nucleation can effectively lower the stress for the onset of plastic yield and increase the plastic strain. At room temperature, the same design concept applies as showcased in $SrTiO_3$ crystals via mechanically seeded dislocations (in contrast to the grown-in dislocations that do not belong to the active slip systems), where near-surface dislocations were generated via grinding on sandpapers. A stark contrast in the yield behavior with/without preinduced dislocations is illustrated in Figure 14. This work [103] addressed only two extreme cases, i.e., dislocation-scarce and high-density dislocations up to about $10^{14}/m^2$. Nevertheless, these results provided feasible proof-of-concept for the hypothesis that, if the dislocation densities can be further tuned (Figure 15), a V–shaped curve for the density-dependent strength should be achievable (Figures 14c and 15i). This was later validated experimentally for the first time in ceramics such as $SrTiO_3$ [104] and MgO [105] via micropillar compression tests.

This V–shaped curve capturing the yield strength as a function of dislocation density has been previously reported only in metals,

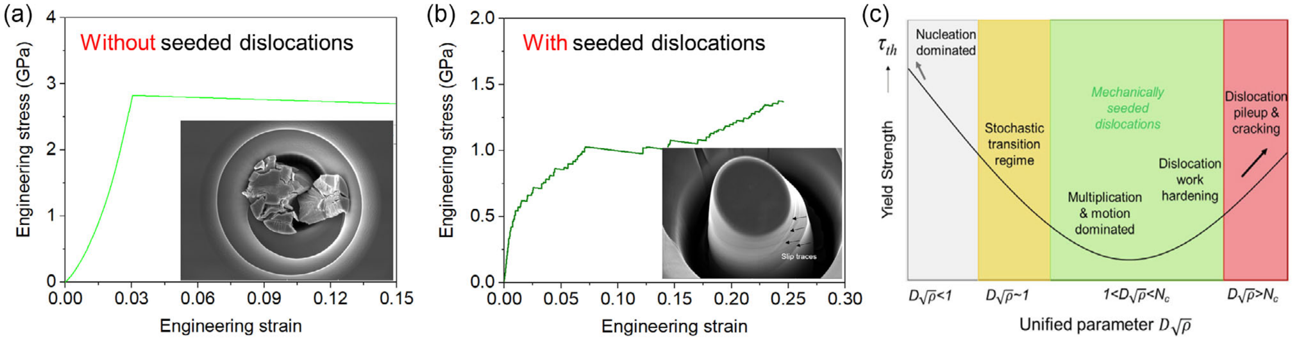


FIGURE 14 | Impact of mechanically seeded dislocations for effective dislocation sources engineering, validated using micropillar compression in $SrTiO_3$ at room temperature: (a) Brittle failure with high fracture strength without seeded dislocations; (b) plastic flow with large compressive strain with lower yield strength with seeded dislocations; and (c) hypothetical V–shaped curve of the yield strength as a function of dislocation density in the given strained volume, where a unified parameter using the product of the characteristic dimension of the pillar (D) and the square root of dislocation density (ρ) is used to estimate the number of dislocations in the strained volume. Images reproduced from Reference [103] with permission. Copyright Elsevier Ltd.

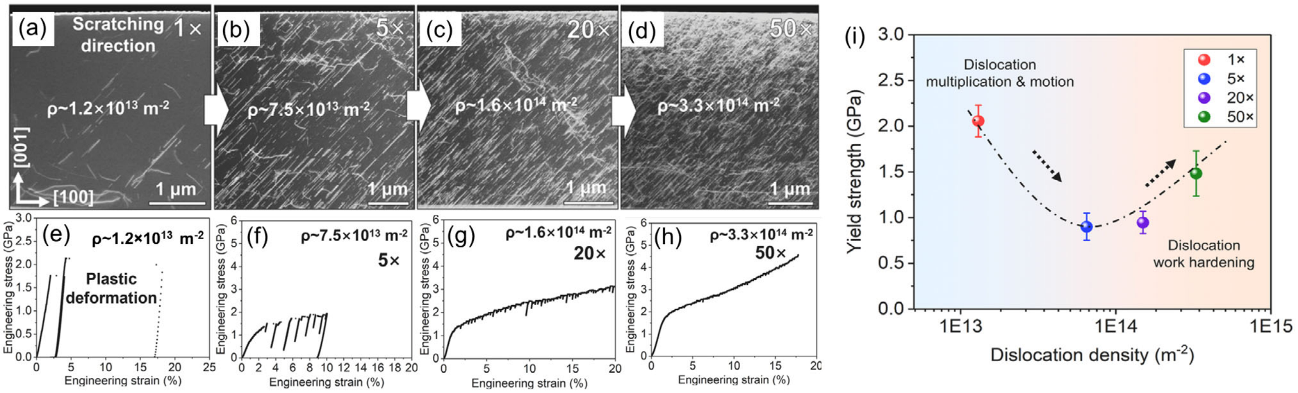


FIGURE 15 | (a–d) Tunable dislocation densities in single-crystal $SrTiO_3$ by cyclic scratching using a Brinell indenter at room temperature; (e–h) corresponding engineering stress–strain curves in micropillar compression tests with different preseeded dislocation densities; and (i) yield strength as a function of the dislocation density, exhibiting a V–shaped curve, suggesting different dislocation mechanisms upon plastic yield. Images adapted from Referene [104], under Creative Commons CC-BY license.

with early experimental work on single-crystal copper (with a much lower dislocation density, though) dating back to the 1970s (see, e.g., Hildebrand [111]), and recently discussed in simulation, including comparison with compiled experimental data from the 1960s to 1980s [112]. The V-shaped trend in yield strength reflects the dislocation-controlled transition from dislocation-source strengthening to forest-dominated strengthening [112]. In other words, as illustrated in Figure 14c, the high strength on the left side is controlled by dislocation nucleation, with the lower strength in the middle is dominated by easy glide and multiplication of dislocations. With a higher dislocation density on the right side, work hardening sets in to increase the strength. Worth noting is that, with few or no dislocations in the strained volume in ceramics, the yield strength may well be replaced by fracture strength when the size exceeds a critical value, as captured by the size-dependent "brittle-to-ductile" transition (BDT) in many brittle solids [96, 113–116] (see also Figure 11, for example). Note here that the term "ductile" in the BDT refers to plasticity under compression due to flaw/defects elimination due to the size reduction (unlike the temperature-dependent BDT as in BCC metals such as tungsten [117]), and appears to be a "consensus" in the small-scale mechanics community when describing the BDT behavior at small scales, and it should not be confused with the ductility as defined in Section 2 in the conventional sense.

Unlike the BDT induced by reducing the size (see, e.g., Figure 11), where the dislocation density in the mother sample remains the same except that there will be possibly no dislocations in the strained volume when the sample is reduced sufficiently small, the BDT illustrated in Figures 14 and 15 is controlled by engineering effective dislocation sources, as in Figures 12 and 13. One major difference between ceramics and metals is that, as shown in Figure 14c, when the dislocation density becomes very low (to the extreme left) or very high (toward the extreme right), brittle failure will set in. Hence, a full spectrum of dislocation-density tuning shall eventually lead to a complete brittle-ductile-brittle transition (BDBT instead of just a single BDT), as most recently validated experimentally in $KTaO_3$ [106].

3.4 | Correlation Between the Three Research Waves

Looking at the three research waves, it is clear that the first wave is more curiosity-driven and focused almost purely on fundamental questions such as dislocation mobility, dislocation multiplication mechanisms, eventually dislocation sources (nucleation), and their interaction with point defects and GBs.

The second wave is leaning more into high-temperature engineering applications, with new fundamental insights gained on dislocation structure, stability, and climb at elevated temperatures.

The first two research waves have clearly paved the road for the current third wave, which is gaining increasing research momentum. As discussed in the last few sections, the fundamental mechanisms and engineering approaches are being fully exploited now in order to effectively and efficiently engineer dislocations for the potential realization of dislocation technology and the integration of both mechanical and functional properties wherever applicable. Particularly for the third research wave, new toolboxes such as small-scale testing, advanced characterization techniques, extension of the materials toolbox, as well as the steady and fast development in computational modeling and eventually data-driven [118] and machine-learning-based approaches [119], are bringing strong driving forces for the topic.

4 | Dislocation-Mediated Toughness and Fracture Toughness

In Section 2, we clarified the concepts for toughness and fracture toughness. For the majority of ceramics, the fracture toughness lies in the range of $\sim$0.1–15 MPa·m$^{1/2}$, which is 2–3 orders of magnitude lower than the values in ductile metals. Various toughening mechanisms, including process-zone toughening, crack bridging, and crack deflection, are available in ceramics. The question is: what is truly a tough ceramic? In ceramics, the concept of toughness appears to be related to a high resistance to catastrophic failure, which naturally includes a high resistance to crack propagation (fracture toughness). Therefore, ceramics that have higher fracture toughness are often regarded as tougher since the plastic deformation is mostly irrelevant, particularly at room temperature.

In this sense, for monolithic ceramic materials (fiber-reinforced ceramic composites and nature-inspired nacre-mortar ceramics are not included in this work for the discussion), we can consider the main material system of partially stabilized zirconia (PSZ) to be tough (the so-called transformation-toughened ceramics [120, 121]), with an appreciable fracture toughness up to $\sim$20 MPa·m$^{1/2}$, while the bulk fracture strength can be higher than 2 GPa. But it is not possible to achieve the highest values for both at the same time [120, 121] due to the conflict between strength and (fracture) toughness [1]. It is frustrating to see that there are no other prominent examples of ceramics that can be considered tough in bulk.

The recent work by Porz et al. [122] sets a new path for dislocation-toughened ceramics (Figure 16a,b), with examples demonstrated on single-crystal $SrTiO_3$ in the near-surface region. With pre-engineered surface dislocations via grinding and polishing, an almost doubled crack-tip fracture toughness was observed. Despite the still rather low absolute value ($\sim$1.6 MPa·m$^{1/2}$) after dislocation-toughening, it was further observed that the cracks can be completely suppressed below a threshold load (10g in Vickers indentation in [122]). This has paved the road for the later studies by the same group to showcase the dislocation toughening concept in other oxides such as $KNbO_3$ [123] and MgO [124, 125]. A most striking observation was on MgO [125], with the dislocation density increased up to $\sim 10^{15}/m^2$, complete crack suppression was also observed at a higher load (100 g) during Vickers indentation. The high compressive stress, as well as the pre-engineered high-density dislocations, are found to synergistically increase the damage tolerance, namely, resistance to crack formation. For high-temperature tests with Vickers indentation on regions with pre-engineered dislocations, a few previous reports by Moon et al. stated that it is effective in the surface regime [126–128], with a 2–4 fold increase of the Vickers indentation fracture toughness.

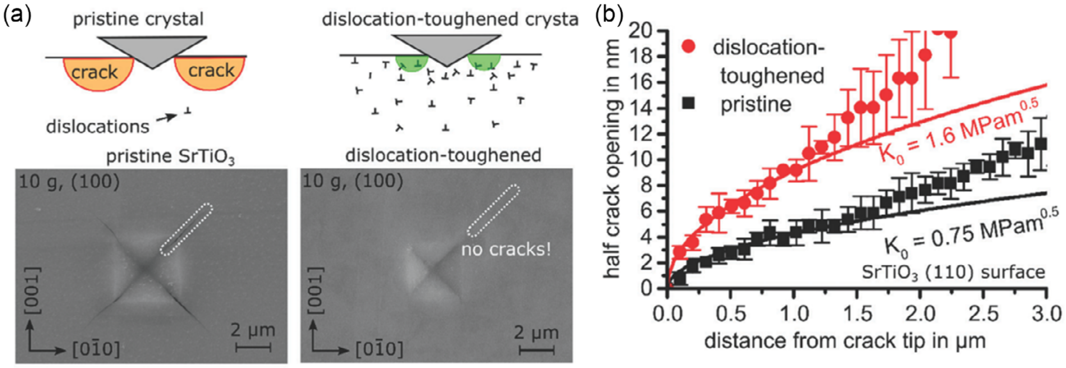


FIGURE 16 | (a) Grinding-induced surface dislocations can completely suppress crack formation under point loading (Vickers indentation) under a threshold load. (b) Crack-tip toughness increased by about twofold using the crack-tip displacement opening method. Images reproduced from Reference [122], under Creative Commons CC-BY License.

In these studies discussed above, Vickers indentation tests were unanimously used for evaluating the toughening effect, by examining either the crack-tip fracture toughness using crack-tip opening displacement [122, 129] or the indentation fracture resistance [125] via the indentation crack length [123, 125, 130, 131]. These authors have clearly noted that these are not standard testing methods [19] for measuring the fracture toughness, mainly limited by the small sample and plastic zone size due to the lack of effective dislocation engineering in bulk samples (see Section 6 later). The use of indentation methods leads to the complication of mixing crack initiation and propagation; meanwhile, it also plays the trick that high local hydrostatic stress, as well as the decaying applied stress intensity factor, helps to confine the crack after generation. Despite the ramifications of using Vickers indentation to evaluate the fracture toughness [132], it still provides useful semiquantitative information in understanding the impact of dislocations on the fracture behavior of these materials.

Since the indentation method does not address clean crack propagation, Preuß et al. adopted the double cleavage drilled compression test [124] to directly capture the impact of dislocation-rich zones impeding the crack propagation, corroborated by phase field modelling coupled with crystal plasticity. Other methods, such as notched microcantilever fracture tests [133], can eventually be an optimal approach to evaluate the dislocation-tuned fracture toughness in these samples that are pre-engineered with dislocations. The notched microcantilever method has the capability of examining pure crack propagation, as well as staying in the micro-/mesoscale volume to have uniformly distributed dislocations.

To date, there have been extensive works discussing the crack tip–dislocation interaction in brittle and semibrittle materials. It has been widely accepted that the very low fracture toughness in these materials is due to the lack of spontaneous dislocation emission at the crack tip before crack propagation [134, 135] (which is readily occurring in FCC metals for a large contribution of plastic work, accompanied by significant crack tip blunting to reduce the stress intensity factor). In the plastically deformable ionic crystals (or sometimes termed as semibrittle materials in the mechanics literature), spontaneous crack-tip dislocation emission at room temperature has indeed been observed in, e.g., LiF [136, 137], MgO [138–141], NaCl [142], and KCl [143, 144]. Models for crack-tip shielding [145–147] (or antishielding) depending on the dislocation configuration at the crack tip have been developed to explain the increase in the crack-tip fracture toughness. Nevertheless, only a limited fracture toughness increase was observed.

In short, the enduring pursuit of truly *ductile and tough* ceramics in bulk at room temperature has been tempting and, meanwhile, frustrating. Tremendous efforts have been made to achieve tougher ceramics, whereas the most effective attempts are almost exclusively extrinsic by crack deflection or grain bridging with some of the wisdom gained from nature (such as the nacre-mortar structures [1, 148–152]) or via intrinsic toughening based on, e.g., transformation toughening. The idea of using dislocations has been greatly discouraged as it "would require changing the bond strength—not a feasible undertaking" as discussed by Ritchie [1]. Hence, intrinsically *ductile and tough* ceramics using dislocations at room temperature have, unfortunately and painfully, never been achieved.

Nevertheless, with the new possibility of preseeding of dislocations before the crack propagation [7], new research endeavors down this path have been carried out, and the previous models and approaches are being revisited. Besides the lack of active slip systems at room temperature for these plastically deformable crystals, new insights have been gained [153] regarding the preferential spatial order of dislocations on the shielding/antishielding mechanisms, work hardening, and low dislocation velocity caused by the pre-engineered, high-density dislocations at the crack-tip, as well as assisted crack-tip emission at elevated temperatures with additionally activated slip systems. Therefore, combining the dislocation seeding and elevated-temperature deformation can be a new path to explore.

On another note, with the preseeding dislocations showcased in Figures 12 and 13 at elevated temperature and in Figures 14 and 15 at room temperature, the microscale testing results showed a significantly increased area below the stress–strain curves in dislocation-scarce samples. Provided that compression is acceptable for this type of toughness evaluation for ceramics, these dislocation-seeded materials shall be considered as much tougher than their pristine states. Given that scaling up to bulk samples still faces great challenges (see Sections 5 and 6), the dislocation-seeding approach may prove more useful for miniaturized components or devices in the near future.

5 | Discrepancy Between Micro- and Macroscale Tests

In Section 3, the discrepancy in yield strength between micro- and macroscale tests has been discussed in the framework of "size effect" in strength by the impact of defects (mainly dislocations) population in various deformation volumes and under geometry constraints. The weakest-link theory has been adopted by Phani et al. to address the stochasticity in the yield strength using the dislocation distribution in metals [154]. A comprehensive review by Gao et al. [155] has addressed the strength statistics of single crystals (and metallic glasses) with small stressed volumes by considering the impact of pre-existing defects. The weakest-link theory is also successful in explaining the fracture strength in brittle ceramics, where pre-existing flaws (microcracks, pores) mainly act as stress concentrators for decreasing the fracture strength as well as for causing the Weibull distribution.

A more critical discrepancy arises in recent studies concerning the strength and toughness of ceramics with dislocations. Many reports have reported a significant increase in plasticity, an increase in toughness at small scales, and improved damage tolerance. However, when extending to bulk, these enhancements either become marginal or even bring a negative impact.

This can be understood as follows:

1. Lack of an effective approach for uniform dislocation engineering at the bulk scale, which is now readily available in micro/mesoscale.
2. Change in the dominating weakest link across length scale, which has a direct consequence: the focus of the problem (the deformation process) shifts from crack initiation (microscale: strength) to crack propagation (macroscale: fracture toughness), as illustrated in Figure 17.

Contrasting with that strength is scale-dependent, fracture toughness in most monolithic brittle ceramics is almost scale-invariant (except for those that exhibit *R*-curve behavior and another interesting observation in Si by an in situ notched cantilever bending test [156]). This discrepancy appears to mainly stem from the misuse of the different concepts (Section 2) as well as the different testing protocols (not being able to follow the ASTM standard [19] due to, e.g., limited sample size and geometry).

6 | Toward Engineering: A Few Outstanding Issues

The current research aims to achieve dislocation technology in ceramics. For potential dislocation-based devices, a few attempts have been made in thin films [157, 158]. We may posit that the future dislocation-based devices will find applications in a variety of fields, including thermal, electrical, light, magnetic, and mechanical loading technologies. For instance, the principle noted by Shockley [72, 159] may enable the use of dislocations in ceramics for emerging quantum/superconducting studies [82], based on the assumption that broken/dangling bonds in crystals with a diamond structure may lead to a 1D band of edge states. These edge states, as in Shockley's words, "*may be partially filled, thus causing each dislocation to become a 1D degenerate-electron-gas conductor.*" This is probably the earliest idea of using dislocations as quantum wires [157, 158] or as demonstrated in Ti-decorated dislocations in Al_2O_3 as conductive nanowires [22]. Other possible application scenarios may be for water splitting [10] or ferroelectric devices [160].

Overall, this aligns with the ongoing research lines covering: (i) dislocation-mediated mechanical properties (strength and toughness), (ii) dislocation-tuned functional properties, and (iii) eventually achieving a balanced combination of both in addressing the potential synergy or competition/trade-off between mechanical and functional properties. To this end, one major challenge remains for effective dislocation engineering as well as the stability/evolution of the dislocations (while interacting with other defects) under various external fields. In what follows, we discuss these aspects.

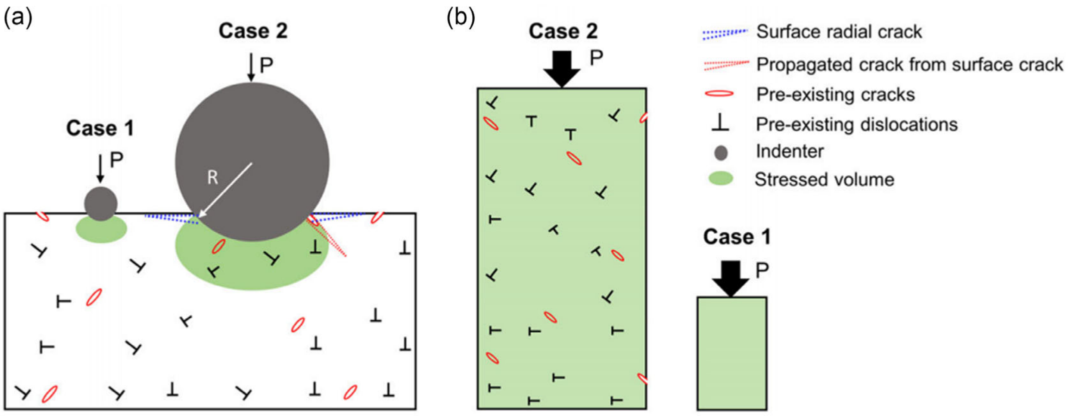


FIGURE 17 | Schematic illustration of the dominating defects in (a) spherical indentation and (b) uniaxial compression. Note the competition between dislocations and pre-existing cracks for different strained volumes (colored in green). The testing volume is reflected in the size-dependent competition between plasticity (small scale, dislocation nucleation and multiplication) and cracking (large scale, crack propagation). Image reproduced from Reference [7], under Creative Commons CC-BY License.

6.1 | Effective Dislocation Engineering

The major bottleneck remains how to effectively and efficiently engineer dislocations into ceramics, especially in bulk polycrystalline samples. The author previously summarized the various routes available to generate dislocations [7]. New approaches have emerged over the past few years. For the sake of keeping up with the state of the art, an updated overview is presented here. Overall, the methods can be approximately grouped into deformation, processing, fabrication, and irradiation, as outlined in Figure 18. To avoid a lengthy discussion, irradiation will not be covered here, and the other three aspects will be discussed with an emphasis on their relation to dislocation mechanics (nucleation, multiplication, and motion).

6.1.1 | Mechanical Deformation

The essence of mechanically inducing dislocations into ceramics can be understood by the three fundamental factors for dislocation mechanics, namely, dislocation nucleation, multiplication, and motion [35, 103]. It is of interest to briefly compare these three factors in metals and plastically deformable ceramics. As discussed in the foregoing sections, in conventionally high-temperature sintered ceramics, dislocations are scarce in their presence, with a density several orders of magnitude lower compared to that in metals. This stems not only from the very different bonding types, but also from the fundamentally different processing routes for these two classes of materials. For most ceramic materials, dislocation nucleation is a primary bottleneck to overcome before desirable plasticity can be achieved, provided that the lattice friction is sufficiently low, as in the case of alkali halides or simple oxides. In a perfect crystal without flaws, for both metals and ceramics, the onset of plasticity requires homogeneous dislocation nucleation, which is energetically expensive at room temperature. This leads to a shear stress level that approaches the theoretical limit $G/2\pi$. The experimentally determined maximum shear stress for dislocation nucleation, dominantly by nanoindentation pop-in study [161, 162], for a wide range of metals and ceramics can be found in References [103, 163].

Once dislocations are nucleated or present (as pre-existing ones), effective dislocation multiplication mechanisms mainly involve Frank–Read sources and cross-slip [57]. In the case of FCC metals, the role of dislocation cross-slip and glissile junctions has been discussed in the framework of computational simulation [164, 165]. In ceramics with simple rock-salt structure, dislocation cross-slip has been reported to be an effective multiplication mechanism in, e.g., LiF [21, 35] and MgO [166]. However, in perovskite oxide with a more complex structure, such as $SrTiO_3$, it remains an open question if cross-slip is effectively operative at room temperature for the profuse generation of dislocations [103]. Drawing inspiration from metals but bearing in mind the differences, such as the impact of oxygen vacancies or vacancy clusters, as well as the role played by debris in ceramics, may be a feasible route. Nevertheless, extensive simulation efforts will be indispensable. Such a simulation toolbox, unfortunately, is still severely lacking in the current research endeavors for understanding dislocation multiplication in plastically deformable ceramics. Last but not least, contrasting the well-understood dislocation mobility in most simple metals, the underlying mechanism for the easy glide of dislocations in bulk plastically deformable ceramics such as $SrTiO_3$, $KTaO_3$, etc., remains a fundamental question to be addressed.

For experimental testing, in addition to the aforementioned uniaxial compression (in bulk and for small samples), indentation (nano-/micro-/mesoscale tests), and scratching, other methods such as high-temperature rolling [167], sandblasting or shot peening with a revisited focus on dislocations [168, 169], and grinding [103, 122, 170, 171] have been adopted for dislocation engineering. Conventional bulk tests at high temperature can be rather tedious and energy-intensive, yet it remains to date the only practical approach to mechanically engineer dislocations into bulk polycrystalline samples.

Worth mentioning is that, for sandblasting or shot peening, the past focus on ceramics was mainly attributing the increased strength and toughness to the high residual compressive stress. Recent studies have started to pay more attention to dislocations

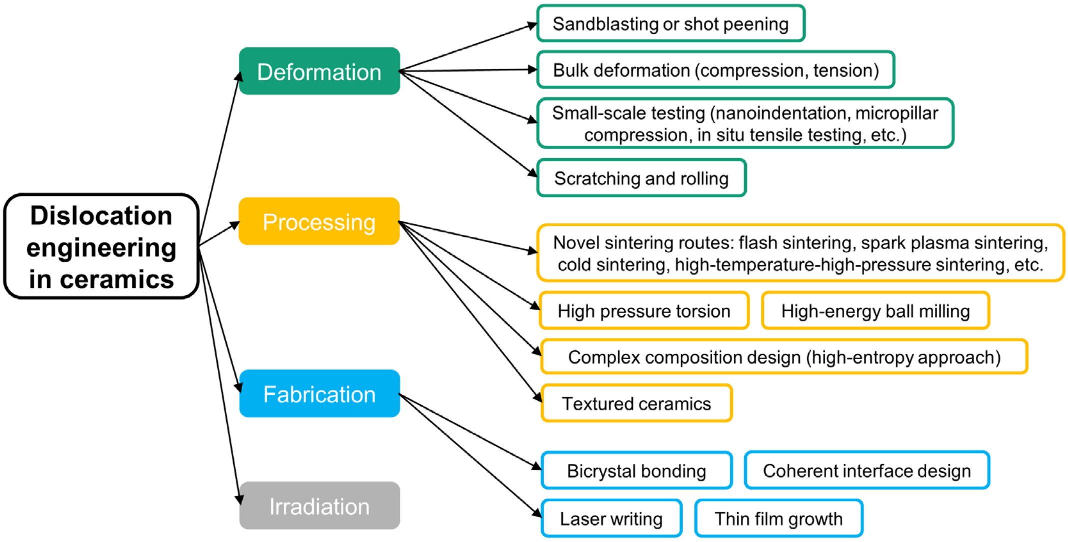


FIGURE 18 | Dislocation engineering routes for ceramics. The experimental approaches are roughly categorized into deformation, processing, fabrication, and irradiation. Updated based on the toolbox summarized in Reference [7].

and their structure, which was possibly left out unintentionally in the past in ceramics. Down this line, a pertinent question remains how to effectively suppress crack formation during shot peening or sandblasting, as the materials, size, and speed of the particles can significantly affect the outcome (being plastically deformed or inducing cracks, similar to, e.g., Figure 10). This question can be simplified by using high-strain-rate indentation (HSRI) [172, 173] equipped with spherical indenters with different tip radii [92, 162] to highlight the deformation mechanisms. Worth noting is that nanoindentation techniques, now capable of high-speed screening, may eventually be used for programming desired arrays of indents for plastic zone/dislocation engineering, namely, for nanopatterning [93, 174]. It has the potential to become a very powerful technique.

6.1.2 | Processing

At first sight, processing/synthesis may be deemed irrelevant for connecting dislocations with the strength and toughness of ceramics. By taking a closer look, recent studies using high-energy ball milling [175, 176] to prepare the powders have shown evidence of dislocations retained in the sintered samples. From a mechanical point of view, this is not surprising since the high local shear force exerted on the particles during milling can easily reach the plastic deformation threshold for dislocations to be generated. An extreme case has been shown in very hard alumina particles enriched with dislocations after milling [177]. The question then becomes: how can we use mechanics, by understanding the strength and toughness of particles, to regulate dislocation generation in powders? This may spark research interest in plastic deformation of ceramic particles under compression (with hydrostatic pressure) as well as under shear. Then, coupled with novel sintering techniques, particularly cold sintering taking place at low temperatures, it may be possible to achieve bulk samples that are rich in dislocations [178]. Cold sintering is particularly of interest here due to the reason that conventional high-temperature sintering will likely annihilate the dislocations in the powders during grain growth.

Here, we get back to flash sintering and examine the strength that might play a role in the sintering process. As discussed by Rheinheimer et al. [107] using the two-particle model, during flash sintering, the local stress required to move dislocations can be lower than 20 MPa (due to surface tension and locally elevated temperature), which is argued to be sufficient to generate dislocations in, e.g., $SrTiO_3$. One may find this argument useful in explaining the dislocations generated in other oxides during flash sintering, such as TiO_2, which also exhibits a very low yield stress at sufficiently high temperatures (∼20 MPa at 900°C [64]). Abundant dislocations have been observed in flash-sintered TiO_2 [108]. It is worth noting that the same principles shall be applicable for the dislocation mechanics analysis in the other field-assisted sintering techniques (FAST), such as spark plasma sintering (SPS) [179] and ultrafast high-temperature sintering (UHS) [180], where dislocations are also being observed [181, 182].

A more extreme case will be the high-temperature high-pressure (HTHP) sintering that has been recently reported to generate high-density dislocations in oxides [183, 184] and borides [185]. A dislocation-based model has been adopted by the current author to explain the dislocation generation in HTHP sintering. As the particle gets smaller and the applied pressure gets higher, local shear stress can reach the critical value for dislocation nucleation [184], which is complementary to Ashby's deformation maps [16]. In short, studying dislocation-based strength in various ceramics (in the form of particles) will bring valuable contributions to the mechanistic understanding of dislocation generation in these new sintering techniques [107, 186].

Last but not least, other processing methods such as high-pressure torsion [187], high-entropy design [188], and extrusion [189] have also been shown to generate high-density dislocations, yet remain difficult to control and tune the density and distribution.

6.1.3 | Fabrication

Fabrication of bicrystals via thermal diffusion bonding [190] has often been adopted for fundamental studies, including dislocation core characterization [191, 192], electrical conductivity, and mechanical properties in the sense of dislocation-GB interactions. Other techniques, including laser writing/patterning [193], can generate dislocations in the near-surface region of oxide crystals. Although amorphization can be triggered due to the high power and local rapid heating/cooling, this technique might hold potential for dislocation patterning (writing and rewriting) for encryption [193]. Down the line of intentional structural design, coherent interface construction for metal/ceramic [99] and ceramic/ceramic [194] has been achieved for improved plasticity and strength. Other fabrication techniques, such as texturing ceramics [195] and (microscale) additive manufacturing [196], may also be worth exploring.

6.2 | Mechano-Functional Integration and Trade-Off

Balancing the mechanical and functional properties is desirable for engineering applications, and has been receiving increasing attention [197–199]. Successful engineering of dislocations into ceramics promises potential integration of both mechanical and functional properties [6, 7]. This proposition has been validated very recently by examining the thermoelectric studies, where dislocations were reviewed [200] and discussed to play a critical role in increasing both the mechanical stability and the thermoelectric figure of merit. The thermal electric figure of merit, zT, is a function of $S^2\sigma T/(\kappa_{ele}+\kappa_{lat})$, with S being the Seebeck coefficient, σ electrical conductivity, T absolute temperature, κ_{ele} electronic thermal conductivity, and κ_{lat} lattice thermal conductivity. High-density dislocations have been introduced to greatly suppress the phonon transport by scattering the mid-frequency phonons (while high-frequency phonons and low-frequency phonons are scattered by point defects and boundaries, respectively), which leads to further reduction in the lattice thermal conductivity κ_{lat}. With the introduced dislocations not compromising the electrical conductivity σ, the significant decrease in the lattice thermal conductivity is therefore beneficial for increasing zT.

On the other hand, the potential competition between the functional and mechanical aspects should also be taken into consideration. From the previous discussion on dislocation-tuned strength (Section 3.3.3), we noticed a decrease and then an increase in strength as the dislocation density increases (the V–shaped curve), namely, the strength has a non-monotonous dependence on dislocation density. However, in the case of thermal conductivity, a monotonous decrease with increasing dislocation density has been predicted and observed. This suggests a possible trade-off between dislocation-regulated mechanical and functional properties, which has recently been experimentally observed and discussed on single-crystal $KTaO_3$ with a tunable dislocation density up to $\sim 10^{15}/m^2$ [106].

6.3 | Work Hardening in Plastically Deformable Ceramics

Work hardening, or strain hardening, refers to a material's strength increase during plastic deformation. It primarily occurs in metallic materials under tension. Work hardening is "a hard problem," as noted by Cottrell in the opening commentary chapter for *Dislocations in Solids* with a whole volume dedicated to this topic [201]. By definition, this sets apart the *ductile* metals from *brittle* ceramics. Therefore, work hardening only becomes relevant for ceramics exhibiting good dislocation plasticity.

Here, we confine the discussion to compression tests, although it should be more properly evaluated in tensile tests. A typical example has been previously discussed in the case of micropillar compression in, e.g., $SrTiO_3$ and MgO in Section 3, in which the dislocation density becomes tunable by cyclic scratching. Tested in compression, the stress–strain curves exhibit a clear increase in apparent work hardening. Work hardening clearly exists in these materials, especially when the dislocation density becomes high. This not only leads to the V–shaped curve in these oxides in micropillar compression, but also raises new questions, such as how dislocation–dislocation interaction, dislocation-point defects interaction, and dislocation-grain boundary interactions contribute to work hardening in plastically deformable ceramics.

In simple ionic crystals, the hardening behavior has been extensively studied in terms of temperature, strain rate, impurity (point defects), crystal orientation, etc., during the first research wave [12]. Models such as Taylor hardening, debris hardening, and the three-stage hardening (which is established in rock salt and fluorite structure single crystals deforming predominantly via single glide) were discussed back then. A more recent example was presented by Yang et al. [202], who reported a three-stage work hardening in $SrTiO_3$ (note the test was performed under uniaxial bulk compression, see Figure 19).

At small scales, one interesting example is the effect of preinduced dislocations (via first higher-load indentation) on the Taylor hardening and indentation size effect on CaF_2 crystals in a subsequent lower-load indentation inside the plastic zone [90]. The experimental observations were captured satisfactorily by the molecular dynamics simulations.

Given the discussion in Section 3.3, work hardening is relevant for contributing to the potential synergy of mechano-functional integration. Yet, as the materials become more work-hardened, the (indentation) hardness increases [90, 203] at the potential cost of losing plasticity and eventually leading to crack formation, as exemplified in single-crystal $SrTiO_3$ [104], MgO [105], and $KTaO_3$ [106] via micropillar compression.

6.4 | Is It Too Early to Consider Dislocation Fatigue in Plastically Deformable Ceramics?

We first refer to Suresh [204] regarding the definition of ***fatigue***, which distinguishes between ceramics and metals. First, for metals, the term *fatigue* is widely accepted for describing the deformation and failure under cyclic loading conditions. While in ceramics, the particular expression *cyclic fatigue* is used for cyclic loading conditions, contrasting *static fatigue* that refers to stable cracking under sustained loads in an embrittling environment [204]. Second, as pointed out by Suresh [204], earlier extensive research on fatigue in metallic materials was intimately related to dislocation slip and motion, which has led to the traditional perception that dislocation motion and plastic slip are necessary conditions for fatigue. This underscored the role of kinematically irreversible cyclic events such as microcracking, martensitic transformation, interfacial sliding, or creep in promoting permanent damage and crack nucleation, as in the case of brittle ceramics. This also explains the bumpy history of fatigue of

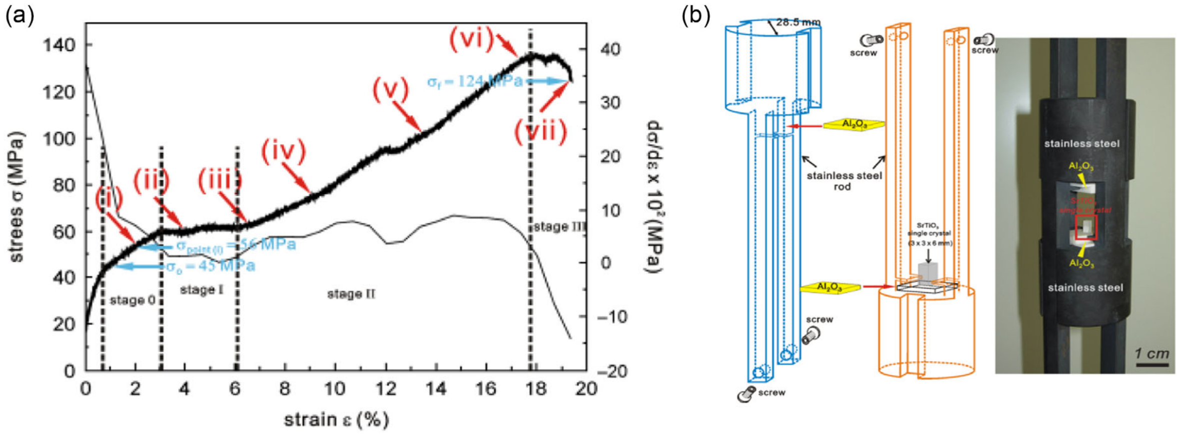


FIGURE 19 | (a) Work hardening (in bulk compression, not in tension) for single-crystal $SrTiO_3$ at room temperature. (b) The testing setup. Images reproduced and adapted from Reference [202] with permission. Copyright The American Ceramic Society.

ceramics, debating if it exists or not (see, e.g., Evans and Fuller [205] proposed the absence of fatigue in porcelain in 1974; which was contradicted later in PSZ by Dauskardt, Yu, and Ritchie [206], who attributed to PSZ a similar crack propagation behavior as metals with cyclic crack growth). Nowadays, the consensus of cyclic fatigue in ceramics refers mainly to the degradation of the toughening mechanisms. For the discussion here related to dislocations in ceramics, we use the term "dislocation fatigue" to avoid confusion.

Fatigue deformation characteristics related to dislocations in simple ionic crystals (e.g., AgCl, NaCl, LiF, and MgO) were extensively investigated in the first wave of dislocation research, as also summarized by Suresh [204]. For instance, Forsyth (1957) [207] observed in cyclically strained AgCl that intrusions and extrusions formed on the sample surface, similar to the surface roughness developed in fatigued FCC metals, and found that the crevices eventually developed into fatigue cracks. A long search for similar ductile metal fatigue modes in these ionic crystals was conducted in the earlier investigations, for instance, by McEvily [208], Alden [209], Subramanium & Washburn [210, 211], and Argon & Godrick [212]. A simple experimental setup was developed for room-temperature cyclic bending of thin slabs of these crystals (see example in Figure 20). However, no dislocation substructure comparable to those found in fatigued FCC metals was observed. Contrasting these room-temperature efforts, Argon & Godrick [212] showed that elevated-temperature fatigue deformation of LiF crystals exhibited similar characteristics as in the room-temperature cyclically deformed FCC metals.

A more dedicated push-pull loading setup was adopted by Majumdar & Burns [213–215] in the early 1980s to study the dislocation fatigue in LiF crystals at room temperature and elevated temperatures. In the latter case, dislocation bands with alternate regions of high/low dislocation density were observed, which may indicate dynamic recovery. At higher temperatures (300°C), ladder-like dislocation structures appear to exist within the PSBs (persistent slip bands) and resemble more closely but are still distinctive from FCC metals. Later, Majumdar & Burns [216] also investigated the dislocation structures using TEM on fatigued MgO crystals deformed by push–pull loading at elevated temperatures. They observed that dense dislocation bundles were developed, similar to the vein structure evolving from early-stage fatigue in FCC metals.

Interestingly, both LiF and MgO showed, after cyclic loading, extensive strain hardening. It was concluded that dislocation movement in these materials also leaves behind microstructural damage (e.g., *debris*, or in Gilman & Johnston's definition, edge dislocation dipoles in the case of LiF [21, 217]; or vacancy clusters as in the case of oxides such as $SrTiO_3$ [122]), causing irreversibility of dislocation movements. Yet, fatigued ionic crystals at room temperature do not fail in typical fatigue as in FCC metals because cross-slip is difficult [208], resulting in easier crack formation due to dislocation pile-up. Later studies on repeatedly loaded MgO with an indenter [218] revealed crack initiation by dislocation pileup.

In short, as briefly summarized by Suresh [204], the fatigue studies in simple ionic crystals suggested that: (i) the lack of FCC metal fatigue characteristics in these crystals tested at room temperature may be due to the limited possibility for cross slip; (ii) when cross slip is favored by the combination of temperature and strain rate, similar dislocation fatigue characteristics such as PSB and cellular structures can be observed; and (iii) further pileup of dislocations can lead to crack nucleation.

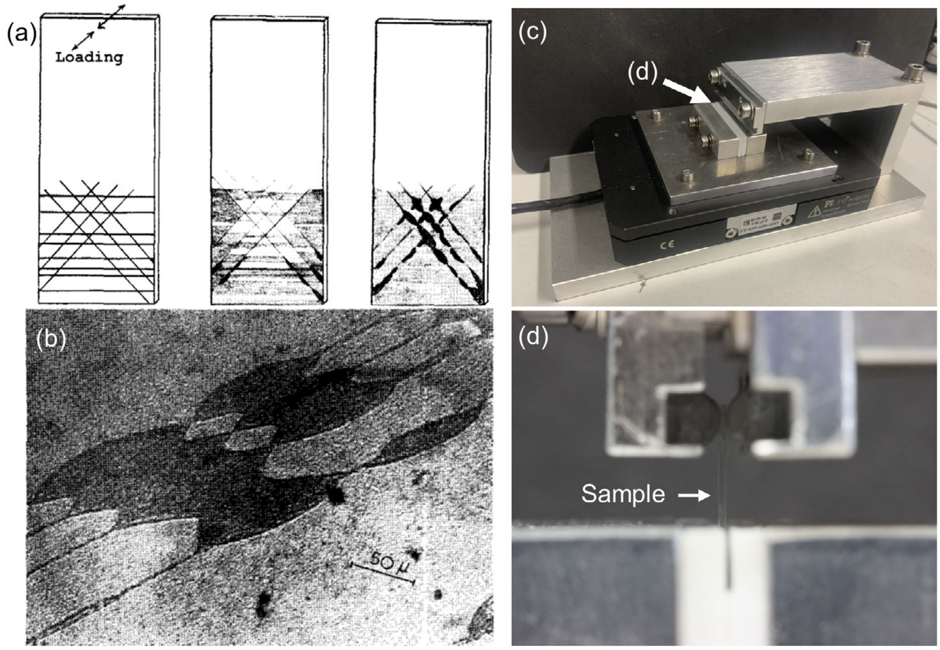


FIGURE 20 | (a,b) Cyclic bending fatigue tests were performed at room temperature for single-crystal MgO thin slabs. Note the "patches" of slip bands (dark region) revealed by the chemical etching in (b), with images reproduced and adapted from Reference [210] with permission. Copyright AIP Publishing. (c,d) A modern version of a cyclic bending testing stage based on a PI actuator, with sample mounting showcased in (d). Image courtesy of D. Isaia and T. Woelke from TU Darmstadt.

The author would like to point out that, in addition to cross slip at elevated temperatures, high temperatures also facilitate activating the additional slip systems, as oxide ceramics do have different slip systems activated at various temperatures, with an excellent example on $SrTiO_3$ [219]. This switch-on/off of the active slip systems at different temperatures [220] and its impact on the dislocation fatigue remains an open question, which is currently being undertaken by the author's group.

Now, back to the ongoing third research wave, we anticipate that, provided dislocation-based devices were to be produced, where the service conditions involve cyclic loading (either purely mechanical, or subjected to thermal cycles, electric fields, etc.), dislocation fatigue in these ceramic materials will become relevant from both engineering and scientific aspects.

6.5 | Defects Interactions and Impact of External Stimuli

6.5.1 | Defects Interactions

Crystal defects can be briefly categorized by dimension into point defects (0D), line defects (1D), planar defects (2D), and 3D precipitates or secondary phases (Figure 21). In plastically deforming polycrystalline materials, the interactions among these defects are most relevant to be examined. There is a wide variety of possible interactions, all of which can be broken down to be within the framework of dislocation nucleation, motion, and multiplication that control material strength and toughness.

The impact of GBs on the yield strength of ceramics, particularly at room temperature, can be complicated due to cracking. Although there are new studies for room-temperature dislocation–GB interaction without cracking at room temperature at both nanoscale [221–223] and mesoscale [224, 225], these studies often deal with only one or several GBs in the deformation volume, rendering ineffective use of the Hall–Petch relation for meaningful analysis. Therefore, dislocation–GB interaction will not be covered here.

Here, we highlight the interactions between dislocations and point defects by considering the impact of defect chemistry. As discussed before [7, 220], the valence and concentration of the charged point defects can modify the radius of the space charge (Debye–Hückel radius) around the dislocation core. Therefore, tuning the defect chemistry [226–229] (for example, via acceptor or donor doping) is expected to modify the dislocation mobility, hence the strength and toughness. An outstanding example can be found in NaCl (with divalent impurities), investigated by Eshelby et al. [49]. They observed a temperature-dependent fluctuation of the yield stress (σ_Y) in such materials and attributed it to the temperature-dependent charge characters of the dislocations. It was proposed that at lower temperatures, the sweeping of charged point defects by the mobile dislocations results in a higher σ_Y. With the temperature increasing to the isoelectric value (T_{iso}), yield strength σ_Y drops to a lower value as the dislocations become charge-neutral. The charge can be further reversed to raise σ_Y as the temperature further increases [49]. Finally, at sufficiently high temperature, thermal activation facilitates breaking the charge cloud drag of the dislocations and leads to a drop of σ_Y again [230]. Further studies in this direction are available in other alkali halides [50].

In oxide ceramics, the impact of defect chemistry and nonstoichiometry on the yield strength was briefly mentioned in the foregoing Section 3.2. The temperature-dependent bulk compression of nonstoichiometric rutile TiO_2 by Ashbee et al. [65] revealed an analogous variation of σ_Y (double maxima and double minima, resembling those in impure NaCl discussed above), with no such

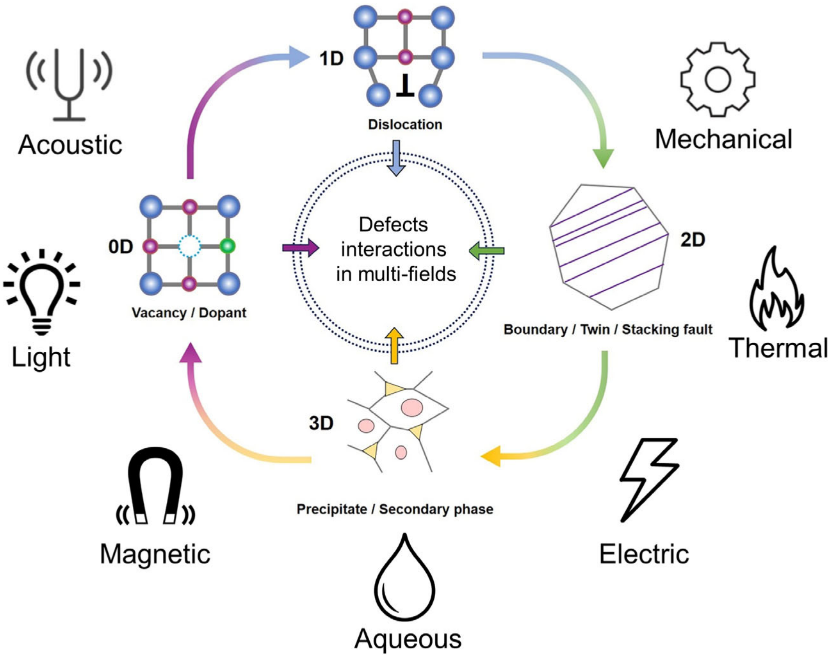


FIGURE 21 | Multidefect interactions (image in the center, courtesy of S. Gao), and their coupling with multiple fields as external stimuli.

effect found in the stoichiometric samples. Recently, Nakamura et al. [231] performed room-temperature bulk compression of $SrTiO_3$ crystals with different Sr/Ti ratios, and observed a ∼15% lower σ_Y and ∼60% higher fracture strain in the Sr/Ti = 1.04 samples, in comparison with the Sr/Ti = 1.00 samples. The underlying mechanisms were later revealed by Fang et al. [232] using nanoindentation pop-in and creep tests (see Figure 22), assisted by impedance spectroscopy. They identified the dual role of oxygen vacancies in facilitating dislocation nucleation (lower pop-in load) while impeding dislocation motion after dislocations are nucleated (lower indentation creep rate and smaller dislocation spacing while gliding). In addition, the pre-existing dislocation densities also play a role in reversing the incipient plasticity from nanoscale to microscale. Having noticed the potential coupling and competition of the dislocation densities as well as the oxygen vacancy concentration changes in the previous experiment, Okafor et al. [233] developed a coupled electromigration-nanoindentation method to enable the tuning of oxygen vacancies without altering the other defect types and concentrations (Figure 23). Their experimental findings, supported by molecular dynamics simulation, confirmed the earlier conclusion in [232].

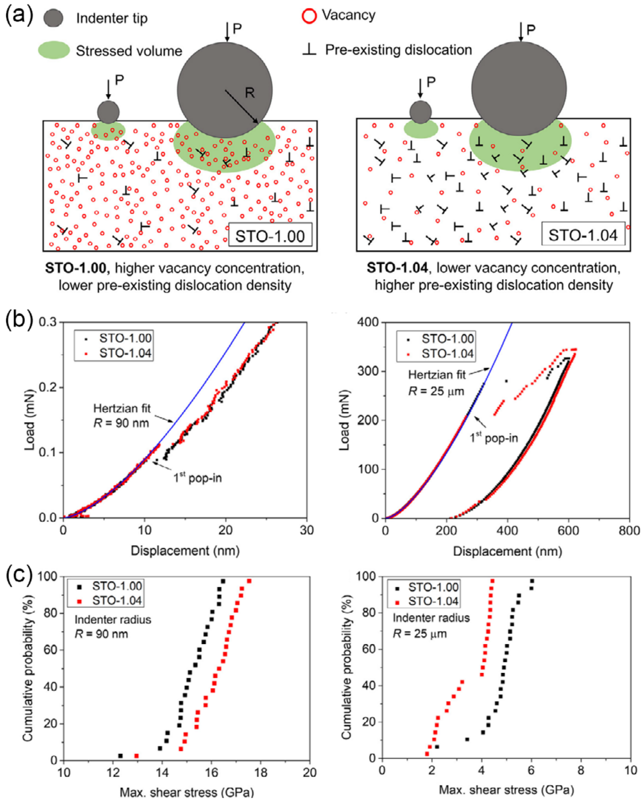


FIGURE 22 | Impact of defect chemistry and dislocation density on the reversed incipient strength at nanoscale and microscale: (a) Schematic illustration of the samples of $SrTiO_3$ crystals with two ratios of Sr/Ti (1.00 and 1.04) in the starting powders for crystal growth, which leads to higher vacancy concentration in STO-1.00 but higher dislocation density in STO-1.04; (b) representative load–displacement curves showcasing the reserved pop-in loads at nanoscale and microscale for two different crystals; and (c) statistical distribution of the maximum shear stress for two crystals at nanoscale and microscales, note the reversed trend. Images reproduced and adapted from Reference [232] with permission. Copyright Elsevier Ltd.

For high-temperature studies, dislocation climb becomes relevant. By examining the interaction of vacancies with a charged edge dislocation core, Hirel et al. [234] simulated dislocation climb using atomic-scale simulation in perovskite $MgSiO_3$. The results concluded that an electrostatic force dominates the dislocation–vacancy interaction, and the dislocations favor a nonstoichiometric, oxygen-deficient configuration associated with a positive charge. This chemoplasticity effect [7] centered around charged dislocations deserves more attention, posing a challenge in experimental design for new functional oxides development [235] as well as in computational simulation for understanding the charged dislocation cores [236].

6.5.2 | Influence of External Stimuli

The above discussions have highlighted the charged dislocation cores in ceramics with ionic and/or covalent bonding [50, 220, 237]. The charged dislocation cores, when subjected to external physical stimuli (e.g., electric field or light illumination, see also Figure 21), can be tuned to alter the dislocation mobility, hence different mechanical responses in strength and toughness are expected. For instance, it has been recently shown that single-crystal ZnS reveals extraordinary bulk plasticity when subjected to complete darkness [56] while it quickly fractures at a higher yield strength under light illumination (Figure 24). The most recently developed experimental methodology of *photoindentation* [54], which couples nanoindentation with precise control of light illumination, reveals at the nanoscale that the dislocation nucleation is marginally affected by light, while dislocation mobility in ZnS is significantly reduced when shining UV light with a wavelength that is close to the bandgap. This photoindentation approach has found its strength in studying photoplasticity in various materials, including ZnO [238, 239], GaN [240], and $SrTiO_3$ [241] at small scales. Very often, bulk crystals are either too expensive or not available.

The impact of light extends beyond strength to also affect the crack-tip fracture toughness, as showcased by Zhu et al. [242] in single-crystal ZnS with the crack-tip opening displacement method. The crack tip opening was examined for cracks generated using Vickers indentation under UV light and complete darkness (see Figure 25). An almost doubled crack-tip fracture toughness was calculated for deformation in complete darkness, compared to that under UV light.

Charged dislocation cores also offer rich opportunities for tuning ionic [181], electrical [243], and thermal [176] conductivities. These new research topics cannot be fully explored without a comprehensive understanding of the mechanics of charged dislocations in ceramics. The multifield coupling problem will be necessary for investigation, potentially opening exciting new research domains for dislocation-based ceramics' stability and degradation.

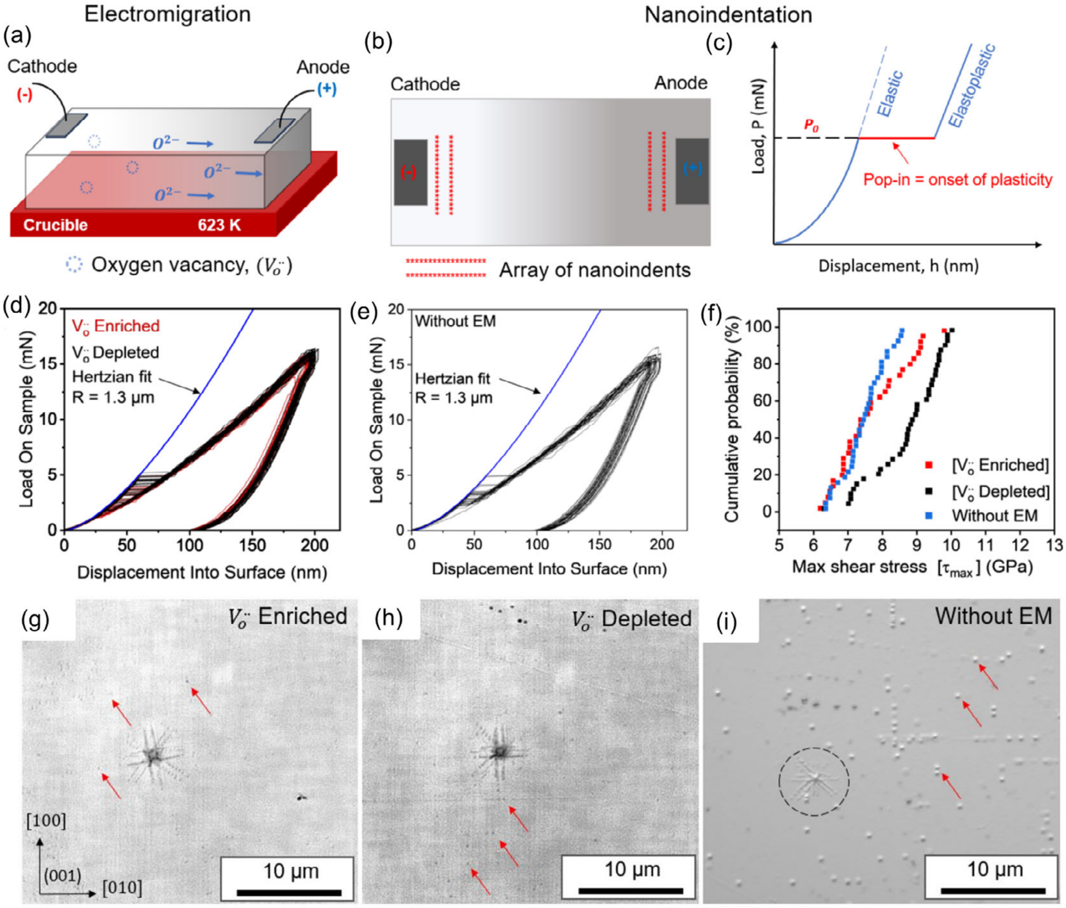
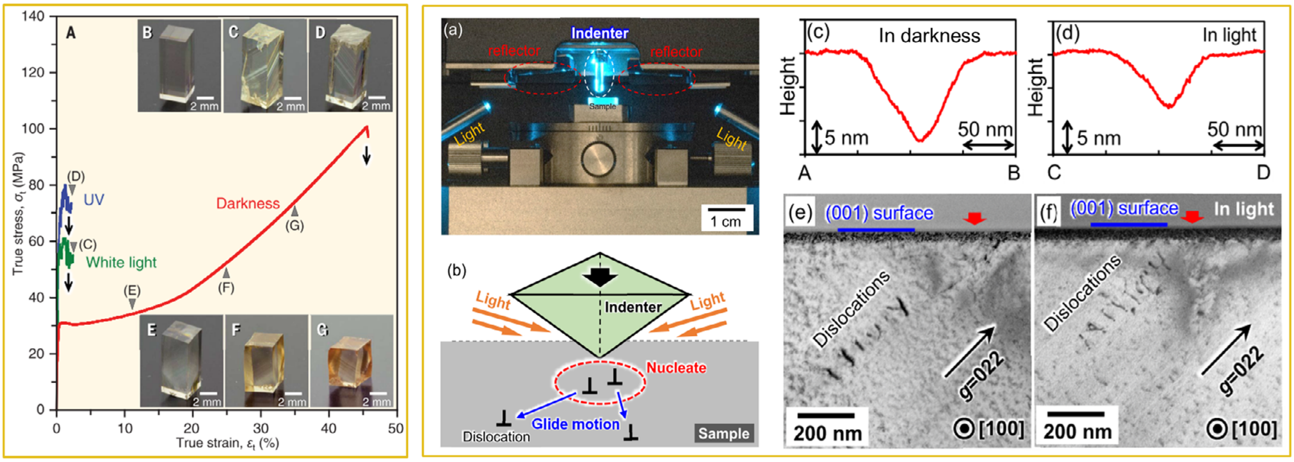


FIGURE 23 | Coupled electromigration–nanoindentation study on dislocation nucleation in single-crystal $SrTiO_3$: (a–c) Electromigration experimental design and placing of nanoindents near the two electrodes; (d–f) load-displacement curves with pop-in events as well as the cumulative probability of the maximum shear strength corresponding to the pop-in events; and (g–i) etch pits analysis of the indents in oxygen vacancy rich and depleted regions, in comparison with reference samples without going through the electromigration treatment. Image reproduced from Reference [233], under Creative Commons CC-BY-NC-ND License.



Bulk compression

Nanoindentation coupled with light illumination (*Photoindentation*)

FIGURE 24 | (Left, a) Bulk strength switching in ZnS crystals under light and darkness. Image reproduced from Reference [56] with permission. Copyright The American Association for the Advancement; (Right, a–f) newly developed *photoindentation* (nanoindentation coupled with precise control of light illumination in terms of wavelength and density) for studying dislocation nucleation and motion coupled with postmortem TEM characterization, demonstrated on single-crystal ZnS. Images adapted from Reference [54] with permission. Copyright American Chemical Society.

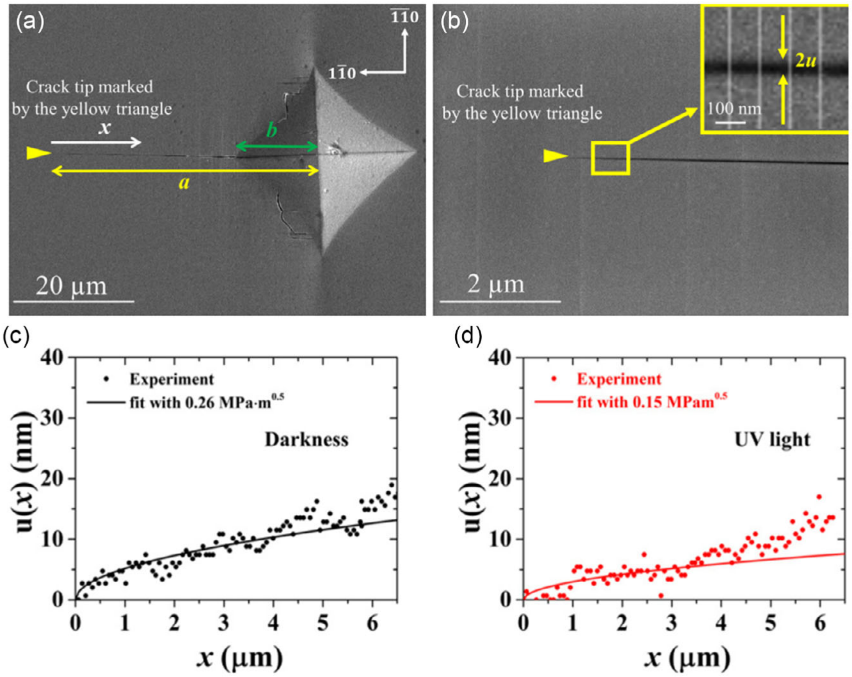


FIGURE 25 | (a,b) Crack tip toughness switching in (c) darkness and (d) under UV light on single-crystal ZnS. The crack-tip opening displacement method was adopted to evaluate the crack-tip fracture toughness. Image reproduced from Reference [242], under Creative Commons CC-BY License.

7 | Conclusion

Despite the several research waves on dislocations in ceramics over the past 90 years, the renewed research interest in this field is gaining growing momentum for engineering versatile properties of advanced ceramics. In addition to dislocation-tuned functional properties, multiple approaches and efforts are in action for dislocation-based mechanical properties, aiming to take the challenge to strengthen and toughen ceramics by examining dislocation nucleation, multiplication, and motion across the length and time scales, as well as their interactions with point defects, GBs, and external fields, leveraging the charged features of dislocations in ceramics. By defining the length scale, temperature, microstructure, and deformation methods, dislocation plasticity in ceramics can be achieved at room temperature, with crack initiation and propagation effectively suppressed.

Looking back on the historical development and looking forward to the challenges of balancing mechanical and functional properties tuned by dislocations, there appear to be exciting times ahead for dislocations in ceramics. The rich knowledge on dislocations in simple ionic crystals and high-temperature deformation in the past is invaluable in guiding the current research wave to avoid completely “reinventing the wheel.” Ongoing endeavors in achieving efficient and effective dislocation-engineering in ceramics are laying the ground: new materials are being discovered, new problems are being defined, new proofs-of-concept are being presented, and new knowledge will be gained. With dislocations being a fundamental defect in crystalline solids and now being increasingly visible in ceramics thanks to the new deformation methods and novel sintering approaches, the wheel may just become stronger and tougher to finally make dislocations in ceramics roll into truly meaningful engineering applications.

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Conflicts of interest

The author declare no conflicts of interest.

Data Availability Statement

No new data was generated in this review.

Statement

The author declares that no Artificial Intelligence Generated Content (AIGC) tools, such as ChatGPT and others based on large language models (LLMs), were used in developing any portion of this manuscript.

References

1. R. O. Ritchie, “The Conflicts between Strength and Toughness,” *Nature Materials* 10, no. 11 (2011): 817–822.

2. O. Gavalda-Diaz, E. Saiz, J. Chevalier, and F. Bouville, “Toughening of Ceramics and Ceramic Composites through Microstructure Engineering: A Review,” *International Materials Reviews* 70, no. 1 (2025): 3–30.

3. D. Brunner, S. Taeri-Baghbadrani, W. Sigle, and M. Rühle, “Surprising Results of a Study on the Plasticity in Strontium Titanate,” *Journal of the American Ceramic Society* 84, no. 5 (2001): 1161–1163.

4. A. F. Mark, M. Castillo-Rodriguez, and W. Sigle, “Unexpected Plasticity of Potassium Niobate during Compression between Room Temperature and 900°C,” *Journal of the European Ceramic Society* 36, no. 11 (2016): 2781–2793.

5. X. Fang, “A Brief History of Dislocations in Ceramics: From Steinsalz to Quantum Wires,” *ACerS Bulletin* 104, no. 5 (2025): 32–36.

6. X. Fang, A. Nakamura, and J. Rödel, “Deform to Perform: Dislocation-Tuned Properties of Ceramics,” *ACerS Bulletin* 102, no. 5 (2023): 24–29.

7. X. Fang, “Mechanical Tailoring of Dislocations in Ceramics at Room Temperature: A Perspective,” *Journal of the American Ceramic Society* 107, no. 3 (2024): 1425–1447.

8. M. D. Armstrong, K. W. Lan, Y. Guo, and N. H. Perry, “Dislocation-Mediated Conductivity in Oxides: Progress, Challenges, and Opportunities,” *ACS Nano* 15 (2021): 9211–9221.

9. Y. Ikuhara, “Nanowire Design by Dislocation Technology,” *Progress in Materials Science* 54, no. 6 (2009): 770–791.

10. Y. Zhang, K. Feng, M. Song, et al., “Dislocation-Engineered Piezocatalytic Water Splitting in Single-Crystal $BaTiO_3$,” *Energy & Environmental Science* 18 (2025): 602–612.

11. M. Höfling, X. Zhou, L. M. Riemer, et al., “Control of Polarization in Bulk Ferroelectrics by Mechanical Dislocation Imprint,” *Science* 372 (2021): 961–964.

12. M. T. Sprackling, *The Plastic Deformation of Simple Ionic Crystals* (Academic Press Inc., 1976).

13. F. Wakai, S. Sakaguchi, and Y. Matsuno, “Superplasticity of Yttria-Stabilized Tetragonal ZrO_2 Polycrsytals,” *Advanced Ceramic Materials* 1, no. 3 (1986): 259–264.

14. F. Wakai, Y. Kodama, S. Sakaguchi, N. Murayama, K. Izaki, and K. Niihara, “A Superplastic Covalent Crystal Composite,” *Nature* 344 (1990): 421–423.

15. T. Hiraga, T. Miyazaki, M. Tasaka, and H. Yoshida, “Mantle Superplasticity and Its Self-Made Demise,” *Nature* 468, no. 7327 (2010): 1091–1094.

16. M. F. Ashby, *Materials Selection in Mechanical Design*, 4th ed. (Butterworth-Heinemann, Elsevier, 2011).

17. R. F. Firestone and A. H. Heuer, “Yield Point of Sapphire,” *Journal of the American Ceramic Society* 56, no. 3 (1973): 136–139.

18. X. Fang, J. Zhang, A. Frisch, et al., “Room-Temperature Bulk Plasticity and Tunable Dislocation Densities in $KTaO_3$,” *Journal of the American Ceramic Society* 107 (2024): 7054–7061.

19. A. International, ASTM C1421−16, *Standard Test Methods for Determination of Fracture Toughness of Advanced Ceramics at Ambient Temperature* (ASTM International, 2016).

20. A. R. Verma and S. Amelinckx, “Spiral Growth on Carborundum Crystal Faces,” *Nature* 167 (1951): 939.

21. W. G. Johnston and J. J. Gilman, “Dislocation Multiplication in Lithium Fluoride Crystals,” *Journal of Applied Physics* 31, no. 4 (1960): 632–643.

22. A. Nakamura, K. Matsunaga, J. Tohma, T. Yamamoto, and Y. Ikuhara, “Conducting Nanowires in Insulating Ceramics,” *Nature Materials* 2, no. 7 (2003): 453–456.

23. D. B. Holt and B. G. Yacobi, *Extended Defects in Semiconductors: Electronic Properties, Device Effects and Structures* (Cambridge University Press, 2007).

24. T. Figielski, “Dislocations as Electrically Active Centres in Semiconductors—Half a Century from the Discovery,” *Journal of Physics: Condensed Matter* 14 (2002): 12665–12672.

25. G. I. Taylor, “The Mechanism of Plastic Deformation of Crystals. Part II.—Comparison with Observations,” *Proceedings of the Royal Society A: Mathematical, Physical and Engineering Sciences* 145 (1934): 388–404.

26. G. I. Taylor, “The Mechanism of Plastic Deformation of Crystals. Part I.—Theoretical,” *Proceedings of the Royal Society A: Mathematical Physical and Engineering Sciences* 145 (1934): 362–387.

27. E. Orowan, ”Zur Kristallplastizitgt. I. Tieftemperaturplastizität und Beckersche Formel, ”*Zeitschrift für Physik* 89 (1934): 605–613.

28. M. Polanyi, “Über Eine Art Gitterstörung, Die Einen Kristall Plastisch Machen könnte,” *Zeitschrift für Physik* 89 (1934): 660–664.

29. D. Hull and D. J. Bacon, *Introduction to Dislocations*, 5th ed. (Elsevier, 2011).

30. T. Mitchell, A. Heuer, E. by, F. R. N. Nabarro, and J. P. Hirth, “Chapter 68 - Dislocations and Mechanical Properties of Ceramics,“ *Dislocations in Solids*, eds. F. R. N. Nabarro, J. P. Hirth (Elsevier, 2004), vol. 12, 339–402.

31. W. Ewald and M. Polanyi, “Plastizität Und Festigkeit Von Steinsalz Unter Wasser,” *Zeitschrift für Physik* 28 (1924): 29–50.

32. W. G. Johnston and J. J. Gilman, “Dislocation Velocities, Dislocation Densities, and Plastic Flow in Lithium Fluoride Crystals,” *Journal of Applied Physics* 30, no. 2 (1959): 129–144.

33. S. Amelinckx, *The Direct Observation of Dislocations_Supplement 6 of Solid State Physics* (Academic Press, 1964).

34. P. Haasen, “Dislocations and the Plasticity of Ionic Crystals,” *Materials Science and Technology* 1, no. 12 (2013): 1013–1024.

35. J. J. Gilman and W. G. Johnston, “Dislocations in Lithium Fluoride Crystals,” *Solid State Physics* 13 (1962): 147–222.

36. A. Argon and E. Orowan, “Lattice Rotation at Slip Band Intersections,” *Nature* 192 (1961): 447–448.

37. J. J. Gilman, “Dislocation Sources in Crystals,” *Journal of Applied Physics* 30 (1584): 1959.

38. R. J. Stokes and C. H. Li, ”Dislocations and the Strength of Polycrystalline Ceramics,” *Materials Science Research* (Springer US, 1963), 133–157.

39. R. J. Stokes and C. H. Li, “Dislocations and the Tensile Strength of Magnesium Oxide,” *Journal of the American Ceramic Society* 46, no. 9 (1963): 423–434.

40. R. J. Stokes, T. L. Johnston, and C. H. Li, “Effect of Slip Distribution on the Fracture Behaviour of Magnesium Oxide Single Crystals,” *Philosophical Magazine* 6, no. 61 (1961): 9–24.

41. A. S. Argon and E. Orowan, “Plastic Deformation in MgO Single Crystals,” *Philosophical Magazine* 9, no. 102 (1964): 1003–1021.

42. C. O. Hulse, S. M. Copley, and J. A. Pask, “Effect of Crystal Orientation on Plastic Deformation of Magnesium Oxide,” *Journal of the American Ceramic Society* 46, no. 7 (1963): 317–323.

43. A. E. Gorum, W. J. Luhman, and J. A. Pask, “Effect of Impurities and Heat-Treatment on Ductility of MgO,” *Journal of the American Ceramic Society* 43, no. 5 (1960): 241–246.

44. A. E. Gorum, E. R. Parker, and J. A. Pask, “Effect of Surface Conditions on Room-Temperature Ductility of Ionic Crystals,” *Journal of the American Ceramic Society* 41, no. 5 (1958): 161–164.

45. P. Cordier, J. Amodeo, and P. Carrez, “Modelling the Rheology of MgO Under Earth’s Mantle Pressure, Temperature and Strain Rates,” *Nature* 481, no. 7380 (2012): 177–180.

46. J. Amodeo, P. Carrez, B. Devincre, and P. Cordier, “Multiscale Modelling of MgO Plasticity,” *Acta Materialia* 59, no. 6 (2011): 2291–2301.

47. J. Amodeo, S. Merkel, C. Tromas, et al., ”Dislocations and Plastic Deformation in MgO Crystals: A Review,” *Crystals* 8, no. 6 (2018): 240–292.

48. A. Frisch, C. Okafor, O. Preuß, et al., “Room-Temperature Dislocation Plasticity in Ceramics: Methods, Materials, and Mechanisms,” *Journal of the American Ceramic Society* 108 (2025): e20575.

49. J. D. Eshelby, C. W. A. Newey, P. L. Pratt, and A. B. Lidiard, “Charged Dislocations and the Strength of Ionic Crystals,” *Philosophical Magazine* 3, no. 25 (1958): 75–89.

50. R. W. Whitworth, “Charged Dislocations in Ionic Crystals,” *Advances in Physics* 24, no. 2 (1975): 203–304.

51. Y. A. Osip’yan, V. F. Petrenko, A. V. Zaretskii ˇ, and R. W. Whitworth, “Properties of II–VI Semiconductors Associated with Moving Dislocations,” *Advances in Physics* 35, no. 2 (1986): 115–188.

52. Y. A. Osip’yan and I. B. Savchenko, “Experimental Observation of the Influence of Light on Plastic Deformation of Cadium Sulfide,” *JETP Letters* 7 (1968): 100.

53. T. Kataoka, M. Sakamoto, and T. Yamada, “Influence of the Electric Field on Flow Stress in KCI Crystals Containing Ca++ Impurity,” *Japanese Journal of Applied Physics* 14 (1975): 1609–1610.

54. A. Nakamura, X. Fang, A. Matsubara, et al., “Photoindentation: A New Route to Understanding Dislocation Behavior in Light,” *Nano Letters* 21 (2021): 1962–1967.

55. K. Matsunaga, S. Hoshino, M. Ukita, Y. Oshima, T. Yokoi, and A. Nakamura, “Carrier-Trapping Induced Reconstruction of Partial-Dislocation Cores Responsible for Light-Illumination Controlled Plasticity in an Inorganic Semiconductor,” *Acta Materialia* 195 (2020): 645–653.

56. Y. Oshima, A. Nakamura, and K. Matsunaga, “Extraordinary Plasticity of an Inorganic Semiconductor in Darkness,” *Science* 360 (2018): 772–774.

57. O. Guillon, C. Elsässer, O. Gutfleisch, et al., “Manipulation of Matter by Electric and Magnetic Fields: Toward Novel Synthesis and Processing Routes of Inorganic Materials,” *Materials Today* 21, no. 5 (2018): 527–536.

58. M. Ukita, A. Nakamura, T. Yokoi, and K. Matsunaga, “Electronic and Atomic Structures of Edge and Screw Dislocations in Rock Salt Structured Ionic Crystals,” *Philosophical Magazine* 98, no. 24 (2018): 2189–2204.

59. A. Nakamura, M. Ukita, N. Shimoda, Y. Furushima, K. Toyoura, and K. Matsunaga, “First-Principles Calculations on Slip System Activation in the Rock Salt Structure: Electronic Origin of Ductility in Silver Chloride,” *Philosophical Magazine* 97, no. 16 (2017): 1281–1310.

60. T. Bretheau, J. Castaing, J. Rabier, and P. Veyssière, “Summary Dislocation Motion and High Temperature Plasticity of Binary and Ternary Oxides,” *Advances in Physics* 28, no. 6 (2006): 829–834.

61. T. E. Mitchell, K. P. D. Lagerlöf, and A. H. Heuer, “Dislocations in Ceramics,” *Materials Science and Technology* 1, no. 11 (2013): 944–949.

62. T. E. Mitchell, B. J. Pletka, D. S. Phillips, and A. H. Heuer, “Climb Dissociation of Dislocations in Sapphire (α-Al_2O_3),” *Philosophical Magazine* 34, no. 3 (1976): 441–451.

63. T. E. Mitchell, L. W. Hobbs, A. H. Heuer, J. Castaing, J. Cadoz, and J. PhIlibert, “Interaction between Point Defects and Dislocations in Oxides,” *Acta Metallurgica* 27 (1979): 1677–1691.

64. K. H. G. Ashbee and R. E. Smallman, “The Plastic Deformation of Titanium Dioxide Single Crystals, Proceedings of the Royal Society A: Mathematical,” *Physical and Engineering Sciences* 274, no. 1357 (1962): 195–205.

65. K. H. G. Ashbee and R. E. Smallman, “The Fracture of Titanium Dioxide Single Crystals with Particular Reference to Non-Stoiehiometry,” *Physica Status Solidi (B)* 4 (1964): 289–298.

66. D. Lorenz, A. Zeckzer, U. Hilpert, P. Grau, H. Johansen, and H. S. Leipner, “Pop-in Effect as Homogeneous Nucleation of Dislocations during Nanoindentation,” *Physical Review B* 67, no. 17 (2003): 172101.

67. G. N. Morscher, P. Pirouz, and A. H. Heuer, “Temperature Dependence of Hardness in Yttria-Stabilized Zirconia Single Crystals,” *Journal of the American Ceramic Society* 74 (1991): 491–500.

68. D. Holmes, A. H. Heuer, and P. Pirouz, “Dislocation Structures around Vickers Indents in 9.4 mol% Y_2O_3-Stabilized Cubic ZrO_2 Single Crystals,” *Philosophical Magazine A* 67, no. 2 (1993): 325–342.

69. Y. Gaillard, C. Tromas, and J. Woirgard, “Quantitative Analysis of Dislocation Pile-Ups Nucleated during Nanoindentation in MgO,” *Acta Materialia* 54, no. 5 (2006): 1409–1417.

70. F. Javaid, A. Stukowski, and K. Durst, “3D Dislocation Structure Evolution in Strontium Titanate: Spherical Indentation Experiments and MD Simulations,” *Journal of the American Ceramic Society* 100, no. 3 (2017): 1134–1145.

71. A. Nakamura, K. Matsunaga, E. Tochigi, N. Shibata, Y. Ikuhara, and K. P. D. Lagerlöf, “Another Origin of Yield Drop Behavior in Sapphire Deformed via Basal Slip: Recombination of Climb-Dissociated Partial Dislocations,” *Scripta Materialia* 138 (2017): 109–113.

72. W. Shockley, “Do Dislocations Hold Technological Promise?,” *Solid State Technology* 26, no. 1 (1983): 75–78.

73. K. Otsuka, A. Kuwabara, A. Nakamura, T. Yamamoto, K. Matsunaga, and Y. Ikuhara, “Dislocation-Enhanced Ionic Conductivity of Yttria-Stabilized Zirconia,” *Applied Physics Letters* 82, no. 6 (2003): 877–879.

74. Y. Furushima, A. Nakamura, E. Tochigi, Y. Ikuhara, K. Toyoura, and K. Matsunaga, “Dislocation Structures and Electrical Conduction Properties of Low Angle Tilt Grain Boundaries in $LiNbO_3$,” *Journal of Applied Physics* 120, no. 14 (2016): 142107.

75. Q. K. Muhammad, L. Porz, A. Nakamura, et al., “Donor and Acceptor-Like Self-Doping by Mechanically Induced Dislocations in Bulk TiO_2,” *Nano Energy* 85 (2021): 105944.

76. A. Nakamura, K. P. D. Lagerlöf, K. Matsunaga, J. Tohma, T. Yamamoto, and Y. Ikuhara, “Control of Dislocation Configuration in Sapphire,” *Acta Materialia* 53, no. 2 (2005): 455–462.

77. C. Shen, J. Li, T. Niu, et al., “Achieving Room Temperature Plasticity in Brittle Ceramics through Elevated Temperature Preloading,” *Science Advances* 10 (2024): eadj4079.

78. L. Porz, T. Fromling, A. Nakamura, et al., “Conceptual Framework for Dislocation-Modified Conductivity in Oxide Ceramics Deconvoluting Mesoscopic Structure, Core, and Space Charge Exemplified for $SrTiO_3$,” *ACS Nano* 15 (2021): 9355–9367.

79. J. Ding, J. Zhang, J. Dong, et al., “Effective Reduction in Thermal Conductivity by High-Density Dislocations in $SrTiO_3$,” *Applied Physics Letters* 126, no. 25 (2025): 253301.

80. M. Johanning, L. Porz, J. Dong, A. Nakamura, J.-F. Li, and J. Rödel, “Influence of Dislocations on Thermal Conductivity of Strontium Titanate,” *Applied Physics Letters* 117, no. 2 (2020): 021902.

81. M. Soleimany, T. Frömling, J. Rödel, and M. Alexe, “Dislocation-Induced Local and Global Photoconductivity Enhancement and Mechanisms in Iron-Doped $SrTiO_3$,” *Advanced Functional Materials* 35, no. 13 (2025): 2417952.

82. S. Hameed, D. Pelc, Z. W. Anderson, et al., “Enhanced Superconductivity and Ferroelectric Quantum Criticality in Plastically Deformed Strontium Titanate,” *Nature Materials* 21 (2022): 54–61.

83. P. Gao, S. Yang, R. Ishikawa, et al., “Atomic-Scale Measurement of Flexoelectric Polarization at $SrTiO_3$ Dislocations,” *Physical Review Letters* 120, no. 26 (2018): 267601.

84. M. Hoefling, M. Trapp, L. Porz, et al., “Large Plastic Deformability of Bulk Ferroelectric $KNbO_3$ Single Crystals,” *Journal of the European Ceramic Society* 41 (2021): 4098–4107.

85. I. Khayr, S. Hameed, J. Budić, et al., “Structural Properties of Plastically Deformed $SrTiO_3$ and $KTaO_3$,” *Physical Review Materials* 8, no. 12 (2024): 124404.

86. D. Brunner, “Low-Temperature Plasticity and Flow-Stress Behaviour of Strontium Titanate Single Crystals,” *Acta Materialia* 54, no. 19 (2006): 4999–5011.

87. P. Gumbsch, S. Taeri-Baghbadrani, D. Brunner, W. Sigle, and M. Ruhle, “Plasticity and an Inverse Brittle-to-Ductile Transition in Strontium Titanate,” *Physical Review Letters* 87, no. 8 (2001): 085505.

88. P. Hirel, F. J. K. Yewou, J. Zhang, W. Lu, X. Fang, and P. Carrez, “Dislocations and Plasticity of $KTaO_3$ Perovskite Modeled with an Interatomic Potential,” *Physical Review Materials* 9, no. 5 (2025): 053605.

89. F. Javaid, E. Bruder, and K. Durst, “Indentation Size Effect and Dislocation Structure Evolution in (001) Oriented $SrTiO_3$ Berkovich Indentations: HR-EBSD and Etch-Pit Analysis,” *Acta Materialia* 139 (2017): 1–10.

90. M. A. Lodes, A. Hartmaier, M. Göken, and K. Durst, “Influence of Dislocation Density on the Pop-in Behavior and Indentation Size Effect in CaF_2 Single Crystals: Experiments and Molecular Dynamics Simulations,” *Acta Materialia* 59, no. 11 (2011): 4264–4273.

91. P. Sadrabadi, K. Durst, and M. Göken, “Study on the Indentation Size Effect in CaF_2: Dislocation Structure and Hardness,” *Acta Materialia* 57, no. 4 (2009): 1281–1289.

92. X. Fang, H. Bishara, K. Ding, et al., “Nanoindentation Pop-in in Oxides at Room Temperature: Dislocation Activation or Crack Formation?,” *Journal of the American Ceramic Society* 104 (2021): 4728–4741.

93. X. Fang, A. Clausner, A. M. Hodge, and M. Sebastiani, “Modern Strategies in Classical Fields of Nanoindentation: Semiconductors, Ceramics, and Thin Films,” *MRS Bulletin* 50, no. 6 (2025): 726–734.

94. S. Korte-Kerzel, “Microcompression of Brittle and Anisotropic Crystals: Recent Advances and Current Challenges in Studying Plasticity in Hard Materials,” *MRS Communications* 7, no. 2 (2017): 109–120.

95. R. Soler, J. M. Wheeler, H.-J. Chang, et al., “Understanding Size Effects on the Strength of Single Crystals through High-Temperature Micropillar Compression,” *Acta Materialia* 81 (2014): 50–57.

96. F. Östlund, K. Rzepiejewska-Malyska, K. Leifer, et al., “Brittle-to-Ductile Transition in Uniaxial Compression of Silicon Pillars at Room Temperature,” *Advanced Functional Materials* 19, no. 15 (2009): 2439–2444.

97. J. Michler, K. Wasmer, S. Meier, F. Östlund, and K. Leifer, “Plastic Deformation of Gallium Arsenide Micropillars Under Uniaxial Compression at Room Temperature,” *Applied Physics Letters* 90 (2007): 043123.

98. A. Nie, Y. Bu, J. Huang, et al., “Direct Observation of Room-Temperature Dislocation Plasticity in Diamond,” *Matter* 2 (2020): 1222–1232.

99. L. R. Dong, J. Zhang, Y. Z. Li, et al., “Borrowed Dislocations for Ductility in Ceramics,” *Science* 385 (2024): 422–427.

100. H. Idrissi, C. Bollinger, F. Boioli, D. Schryvers, and P. Cordier, “Low-Temperature Plasticity of Olivine Revisited with In Situ TEM Nanomechanical Testing,” *Science Advances* 2 (2016): e1501671.

101. I. Issa, J. Amodeo, J. Réthoré, et al., “In Situ Investigation of MgO Nanocube Deformation at Room Temperature,” *Acta Materialia* 86 (2015): 295–304.

102. I. Issa, L. Joly-Pottuz, J. Amodeo, et al., “From Dislocation Nucleation to Dislocation Multiplication in Ceramic Nanoparticle,” *Materials Research Letters* 9, no. 6 (2021): 278–283.

103. X. Fang, W. Lu, J. Zhang, et al., “Harvesting Room-Temperature Plasticity in Ceramics by Mechanically Seeded Dislocations,” *Materials Today* 82 (2025): 81–91.

104. J. Zhang, X. Fang, and W. Lu, “Impact of Dislocation Densities on the Microscale Strength of Single-Crystal Strontium Titanate,” *Acta Materialia* 291 (2025): 121004.

105. J. Zhang, Z. Li, Y. Zhang, et al., “Scale-Bridging Dislocation Plasticity in MgO at Room Temperature,” *International Journal of Plasticity* 195 (2025): 104533.

106. J. Zhang, W. Lu, and X. Fang, “Tunable Dislocations Overcome Mechano-Functional Tradeoff in Perovskite Oxides,” preprint, arXiv: 2602.20121, 2026.

107. W. Rheinheimer, X. L. Phuah, L. Porz, M. Scherer, J. Cho, and H. Wang, “The Impact of Flash Sintering on Densification and Plasticity of Strontium Titanate: High Heating Rates, Dislocation Nucleation and Plastic Flow,” *Journal of the European Ceramic Society* 43 (2023): 3524–3537.

108. X. L. Phuah, J. Cho, T. Tsakalakos, A. K. Mukherjee, H. Wang, and X. Zhang, “Defects in Flash-Sintered Ceramics and Their Effects on Mechanical Properties,” *MRS Bulletin* 46 (2021): 44–51.

109. J. Li, H. Wang, and X. Zhang, “Nanoscale Stacking Fault–assisted Room Temperature Plasticity in Flash-Sintered TiO_2,” *Science Advances* 5 (2019): eaaw5519.

110. S. F. Yang, N. Zhang, and Y. P. Chen, “Concurrent Atomistic-Continuum Simulation of Polycrystalline Strontium Titanate,” *Philos Mag* 95, no. 24 (2015): 2697–2716.

111. H. Hildebrand, “The Effect of the Initial Dislocation Density on Dislocation Multiplication and Work-Hardening Characteristics of Copper Single Crystals,” *Physica Status Solidi (a)* 12, no. 1 (1972): 239–249.

112. J. A. El-Awady, “Unravelling the Physics of Size-Dependent Dislocation-Mediated Plasticity,” *Nature Communications* 6 (2015): 5926.

113. F. Östlund, P. R. Howie, R. Ghisleni, et al., “Ductile–brittle Transition in Micropillar Compression of GaAs at Room Temperature,” *Philosophical Magazine* 91, no. 7–9 (2011): 1190–1199.

114. W. W. Gerberich, K. M. Schmalbach, Y. Chen, E. Hintsala, and N. A. Mara, “Quantifying Physical Parameters to Predict Brittle/ Ductile Behavior,” *Materials Science and Engineering: A* 808 (2021): 140899.

115. E. Hintsala, C. Teresi, and W. W. Gerberich, “Linking Nanoscales and Dislocation Shielding to the Ductile–Brittle Transition of Silicon,” *Metallurgical and Materials Transactions A* 47, no. 12 (2016): 5839–5844.

116. W. W. Gerberich, J. Michler, W. M. Mook, et al., “Scale Effects for Strength, Ductility, and Toughness in “Brittle” Materials,” *Journal of Materials Research* 24, no. 3 (2009): 898–906.

117. P. Gumbsch, J. Riedle, A. Hartmaier, and H. F. Fischmeister, “Controlling Factors for the Brittle-to-Ductile Transition in Tungsten Single Crystals,” *Science* 282 (1998): 1293–1295.

118. I. Słodczyk, A. Frisch, X. Fang, I. Gitman, and F. Liu, “A Data-Driven Quest for Room-Temperature Bulk Plastically Deformable Ceramics,” *Journal of the American Ceramic Society* 109, no. 2 (2026): e70577.

119. S. Zhang, Y. Li, S. Suzuki, A. Nakamura, and S. Ogata, “Neural Network Potential for Dislocation Plasticity in Ceramics,” *Npj Computational Materials* 10, no. 1 (2024): 266.

120. D. B. Marshall and R. H. J. Hannink, Ceramics: Transformation Toughening, Encyclopedia of Materials: Technical Ceramics and Glasses, 1 (2021): 818-822.

121. D. J. Green, R. H. J. Hannink, and M. V. Swain, Transformation Toughening of Ceramics, (CRC Press, 1989).

122. L. Porz, A. J. Klomp, X. Fang, et al., "Dislocation-Toughened Ceramics," *Materials Horizons* 8 (2021): 1528–1537.

123. O. Preuß, E. Bruder, W. Lu, et al., "Dislocation Toughening in Single-Crystal $KNbO_3$," *Journal of the American Ceramic Society* 106 (2023): 4371–4381.

124. O. Preuß, Z. Li, E. Bruder, et al., "Stable Crack Propagation in Dislocation-Engineered Oxide Visualized by Double Cleavage Drilled Compression Test," *Journal of the European Ceramic Society* 46 (2026): 117905.

125. O. Preuß, E. Bruder, J. Zhang, W. Lu, J. Rödel, and X. Fang, "Damage-Tolerant Oxides by Imprint of an Ultra-High Dislocation Density," *Journal of the European Ceramic Society* 45, no. 2 (2025): 116969.

126. W.-J. Moon and H. Saka, "Toughening of a Brittle Material by Means of Dislocation Subboundaries," *Philosophical Magazine Letters* 80, no. 7 (2000): 461–466.

127. W.-J. Moon, T. Ito, S. Uchimura, and H. Saka, "Toughening of Ceramics by Dislocation Sub-Boundaries," *Materials Science and Engineering: A* 387–389 (2004): 837–839.

128. H. Saka, D. Goto, and W. J. Moon, "Dislocations in Plastically Deformed Apatite," *Journal of Materials Science* 43, no. 9 (2008): 3234–3239.

129. J. Seidel and J. Rödel, "Measurement of Crack Tip Toughness in Alumina as a Function of Grain Size," *Journal of the American Ceramic Society* 80, no. 2 (1997): 433–438.

130. M. N. Salem, K. Ding, J. Rödel, and X. Fang, "Thermally Enhanced Dislocation Density Improves Both Hardness and Fracture Toughness in Single-Crystal $SrTiO_3$," *Journal of the American Ceramic Society* 106 (2023): 1344–1355.

131. G. R. Anstis, P. Chantikul, B. R. Lawn, and D. B. Marshall, "A Critical Evaluation of Indentation Techniques for Measuring Fracture Toughness: I, Direct Crack Measurements," *Journal of the American Ceramic Society* 64, no. 9 (1981): 533–538.

132. G. D. Quinn and R. C. Bradt, "On the Vickers Indentation on Fracture Toughness," *Journal of the American Ceramic Society* 90, no. 3 (2007): 673–680.

133. J. Ast, M. Ghidelli, K. Durst, M. Göken, M. Sebastiani, and A. M. Korsunsky, "A Reviewof Experimental Approaches to Fracture Toughness Evaluation at Themicro-Scale," *Materials and Design* 173 (2019): 107762.

134. J. R. Rice, "Dislocation Nucleation from a Crack Tip: An Analysis Based on the Peierls Concept," *Journal of the Mechanics and Physics of Solids* 40, no. 2 (1992): 239–271.

135. J. R. Rice and R. Thomson, "Ductile versus Brittle Behaviour of Crystals," *The Philosophical Magazine: A Journal of Theoretical Experimental and Applied Physics* 29, no. 1 (2006): 73–97.

136. S. J. Burns, "Crack Tip Dislocation Nucleation Observation in Bulk Specimens," *Scripta Metallurgica* 20 (1986): 1489–1494.

137. K. Y. Chia and S. J. Burns, "Crack Tip Deformation in LiF Single Crystals," *Scripta Metallurgica* 18 (1984): 467–472.

138. B. S. Majumdar and S. J. Burns, "Crack Tip Dislocations in MgO," *Scripta Metallurgica* 14 (1980): 653–656.

139. F. Appel, U. Messerschmidt, and M. Kuna, "Crack Propagation in MgO during In-Situ Deformation in the High-Voltage Electron Microscope," *Physica Status Solidi* 55 (1979): 529.

140. K. Higashida, N. Narita, S. Asano, and R. Onodera,"Dislocation Emission from a Crack Tip in MgO Thin Crystals," *Materials Science and Engineering A* 285 (2000): 111–121.

141. K. Higashida, S. Sadamatsu, and M. Tanaka, "High-Voltage Electron Microscopy Analyses for Crack-Tip Dislocations and Their Shielding Effect on Fracture Toughness in MgO and Si Crystals," *Journal of the American Ceramic Society* 108 (2025): e20378.

142. N. Narita, K. Higashida, and S. Kitano, "Dislocation Distribution around a Crack Tip and the Fracture Toughness in NaCl Crystals," *Scripta Materialia* 21 (1987): 1275–1278.

143. K. Higashida, N. Narita, and K. Matsunaga, "Effect of Dislocation Activities on Crack Extension in Ionic Crystals," *Materials Science and Engineering A* 176 (1994): 147–153.

144. K. Higashida and N. Narita, "Crack Tip Plasticity in Ionic Crystals with the NaCl-Type Structure," *Materials Transactions* 42 (2001): 33–40.

145. B. S. Majumdar and S. J. Burns, "A Griflith Crack Shielded by a Dislocation Pile-up," *International Journal of Fracture* 21 (1983): 229–240.

146. B. S. Majumdar and S. J. Burns, "Crack Tip Shielding-an Elastic Theory of Dislocations and Dislocation Arrays near a Sharp Crack," *Acta Metallurgica* 29 (1981): 579–588.

147. N. Narita, K. Higashida, T. Torii, and S. Miyaki, "Crack-Tip Shielding by Dislocations and Fracture Toughness in NaCl Crystals," *Materials Transactions JIM* 30, no. 11 (1989): 895–907.

148. A. G. Evans, "Perspective on the Development of High-Toughness Ceramics," *Journal of the American Ceramic Society* 73, no. 2 (1990): 187–206.

149. A. Wat, C. Ferraro, X. Deng, et al., "Bioinspired Nacre-Like Alumina with a Metallic Nickel Compliant Phase Fabricated by Spark-Plasma Sintering," *Small* 15 (2019): e1900573.

150. E. Munch, M. E. Launey, D. H. Alsem, E. Saiz, A. P. Tomsia, and R. O. Ritchie, "Tough, Bio-Inspired Hybrid Materials," *Science* 322 (2008): 1516–1520.

151. S. Deville, E. Saiz, R. K. Nalla, and A. P. Tomsia, "Freezing as a Path to Build Complex Composites," *Science* 311 (2006): 515–518.

152. U. G. Wegst, H. Bai, E. Saiz, A. P. Tomsia, and R. O. Ritchie, "Bioinspired Structural Materials," *Nature Materials* 14, no. 1 (2015): 23–36.

153. O. Preuß, *Dislocation-Based Damage Tolerance and Toughening in Ceramics* (PhD Thesis, Department of Materials and Earth Sciences, Technical University of Darmstadt, 2025).

154. P. S. Phani, K. E. Johanns, E. P. George, and G. M. Pharr, "A Simple Stochastic Model for Yielding in Specimens with Limited Number of Dislocations," *Acta Materialia* 61, no. 7 (2013): 2489–2499.

155. Y. Gao and H. Bei, "Strength Statistics of Single Crystals and Metallic Glasses Under Small Stressed Volumes," *Progress in Materials Science* 82 (2016): 118–150.

156. I. Issa, C. Gammer, S. Kolitsch, et al., "In-Situ TEM Investigation of Toughening in Silicon at Small Scales," *Materials Today* 48 (2021): 29–37.

157. C. L. Reynolds, J. G. Reynolds, A. Crespo, J. K. Gillespie, K. D. Chabak, and R. F. Davis, "Dislocations as Quantum Wires: Buffer Leakage in AlGaN/GaN Heterostructures," *Journal of Materials Research* 28, no. 13 (2013): 1687–1691.

158. M. Reiche, M. Kittler, H. Uebensee, E. Pippel, A. Haehnel, and S. Birner, "Electronic Properties of Dislocations," *Applied Physics A* 122, no. 4 (2016): 389.

159. W. Shockley, "Dislocations and Edge States in the Diamond Crystal Structure," *Physical Review* 91 (1953): 228.

160. F. Zhuo, X. Zhou, S. Gao, et al., "Anisotropic Dislocation-Domain Wall Interactions in Ferroelectrics," *Nature Communications* 13 (2022): 6676.

161. W. W. Gerberich, J. C. Nelson, E. T. Lilleodden, P. Anderson, and J. T. Wyrobek, “Indentation Induced Dislocation Nucleation: The Initial Yield Point,” *Acta Materialia* 44, no. 9 (1996): 3585–3598.

162. J. R. Morris, H. Bei, G. M. Pharr, and E. P. George, “Size Effects and Stochastic Behavior of Nanoindentation Pop In,” *Physical Review Letters* 106, no. 16 (2011): 165502.

163. H. Oguri, Y. Li, A. I. Osaka, et al., ”Atomic Step–Terrace Ordering Enables Unprecedentedly Low Pop-in Stress Scatter in GaN (0001),” *Small* early view (2026). e12390.

164. M. Sudmanns, M. Stricker, D. Weygand, T. Hochrainer, and K. Schulz, “Dislocation Multiplication by Cross-Slip and Glissile Reaction in a Dislocation Based Continuum Formulation of Crystal Plasticity,” *Journal of the Mechanics and Physics of Solids* 132 (2019): 103695.

165. M. Stricker and D. Weygand, “Dislocation Multiplication Mechanisms – Glissile Junctions and Their Role on the Plastic Deformation at the Microscale,” *Acta Materialia* 99 (2015): 130–139.

166. U. Messerschmidt, *Dislocation Dynamics During Plastic Deformation* (Springer, 2010).

167. Z. Chen, X. Liu, and Y. Li, “Enhancing the Fracture Toughness and Hardness of Sapphire by High-Temperature Rolling,” *Journal of the American Ceramic Society* 107, no. 5 (2024): 2793–2799.

168. W. Pfeiffer and J. Wenzel, “Shot Peening of Brittle Materials – Status and Outlook,” *Materials Science Forum* 638–642 (2010): 799–804.

169. W. Pfeiffer and T. Frey, “Strengthening of Ceramics by Shot Peening,” *Journal of the European Ceramic Society* 26, no. 13 (2006): 2639–2645.

170. R. Wang, Y. Zhu, and S. M. Shapiro, “Structural Defects and the Origin of the Second Length Scale in $SrTiO_3$,” *Physical Review Letters* 80, no. 11 (1998): 2370–2073.

171. L. Jin, X. Guo, and C. L. Jia, “TEM Study of <110>-Type 35.26° Dislocations Specially Induced by Polishing of $SrTiO_3$ Single Crystals,” *Ultramicroscopy* 134 (2013): 77–85.

172. P. S. Phani, B. L. Hackett, C. C. Walker, W. C. Oliver, and G. M. Pharr, “High Strain Rate Nanoindentation Testing: Recent Advancements, Challenges and Opportunities,” *Current Opinion in Solid State and Materials Science* 27, no. 1 (2023): 101054.

173. Y. Zhang, J. Dong, B. L. Hackett, S. J. Kotze, K. Y. Xie, and G. M. Pharr, “Plastic Deformation of Ceramics via Quasistatic and High Strain Rate Nanoindentation,” *Journal of the American Ceramic Society* 109, no. 1 (2025): e70391.

174. H. Bishara, H. Tsybenko, S. Nandy, et al., “Dislocation-Enhanced Electrical Conductivity in Rutile TiO_2 Accessed by Room-Temperature Nanoindentation,” *Scripta Materialia* 212 (2022): 114543.

175. J. Li, Y. Pan, C. Wu, F. Sun, and T. Wei, “Processing of Advanced Thermoelectric Materials,” *Science China Technological Sciences* 60, no. 9 (2017): 1347–1364.

176. Y. Pan, U. Aydemir, J. A. Grovogui, et al., “Melt-Centrifuged $(Bi, Sb)_2Te_3$: Engineering Microstructure toward High Thermoelectric Efficiency,” *Advanced Materials* 30, no. 34 (2018): e1802016.

177. B. K. Sarkar, “The Effect of Ball Milling on Alumina Powders,” *Transactions of the Indian Ceramic Society* 34, no. 1 (2014): 8–12.

178. K. Yasui and K. Hamamoto, “Possibility of Production of All-Dislocation-Ceramics by Cold Sintering,” *Journal of Materials Research* 41 (2026): 839–850.

179. O. Guillon, J. Gonzalez-Julian, B. Dargatz, et al., “Field-Assisted Sintering Technology/Spark Plasma Sintering: Mechanisms, Materials, and Technology Developments,” *Advanced Engineering Materials* 16, no. 7 (2014): 830–849.

180. C. Wang, W. Ping, Q. Bai, et al., “A General Method to Synthesize and Sinter Bulk Ceramics in Seconds,” *Science* 368 (2020): 521–526.

181. K. K. Adepalli, M. Kelsch, R. Merkle, and J. Maier, “Enhanced Ionic Conductivity in Polycrystalline TiO_2 by “one-Dimensional Doping,” *Physical Chemistry Chemical Physics* 16, no. 10 (2014): 4942–4951.

182. T. P. Mishra, S. Wang, C. Lenser, et al., “Ultra-Fast High-Temperature Sintering of Strontium Titanate,” *Acta Materialia* 231 (2022): 117918.

183. S. Cottrino, T. Gaudisson, T. Douillard, et al., “Effect of High Pressure on Microstructure and Mechanical and Optical Properties of Nano-Structured $MgAl_2O_4$ Spinel Fabricated by High Pressure Spark Plasma Sintering,” *Journal of the European Ceramic Society* 45, no. 15 (2025): 117579.

184. Z. Yin, H. Zhang, Y. Wang, et al., “Ultrahigh-Pressure Structural Modification in BiCuSeO Ceramics: Dense Dislocations and Exceptional Thermoelectric Performance,” *Advanced Energy Materials* 15 (2025): 2403174.

185. H. Xu, W. Ji, W. Guo, et al., “Enhanced Mechanical Properties and Oxidation Resistance of Zirconium Diboride Ceramics via Grain-Refining and Dislocation Regulation,” *Advanced Science* 9, no. 6 (2022): e2104532.

186. K. Tsuji, Z. Fan, S. H. Bang, S. Dursun, S. Trolier-McKinstry, and C. A. Randall, “Cold Sintering of the Ceramic Potassium Sodium Niobate, $(K_{0.5}Na_{0.5})NbO_3$, and Influences on Piezoelectric Properties,” *Journal of the European Ceramic Society* 42 (2022): 105–111.

187. K. Edalati, A. Bachmaier, V. A. Beloshenko, et al., “Nanomaterials by Severe Plastic Deformation: Review of Historical Developments and Recent Advances,” *Materials Research Letters* 10, no. 4 (2022): 163–256.

188. Y. Han, X. Liu, Q. Zhang, et al., “Ultra-Dense Dislocations Stabilized in High Entropy Oxide Ceramics,” *Nature Communications* 13, no. 1 (2022): 2871.

189. M. Han, C. Dong, C. Yao, et al., ”Asymmetric Stress Engineering of Dense Dislocations in Brittle Superconductors for Strong Vortex Pinning,” *Advanced Materials* 37 (2025): e13265.

190. Y. Ikuhara, H. Nishimura, A. Nakamura, K. Matsunaga, T. Yamamoto, and K. P. D. Lagerloef, “Dislocation Structures of Low-Angle and Near-Sigma 3 Grain Boundaries in Alumina Bicrystals,” *Journal of the American Ceramic Society* 86, no. 4 (2003): 595–602.

191. P. Gao, R. Ishikawa, B. Feng, A. Kumamoto, N. Shibata, and Y. Ikuhara, “Atomic-Scale Structure Relaxation, Chemistry and Charge Distribution of Dislocation Cores in $SrTiO_3$,” *Ultramicroscopy* 184 (2018): 217–224.

192. Z. Zhang, W. Sigle, W. Kurtz, and M. Rühle, “Electronic and Atomic Structure of a Dissociated Dislocation in $SrTiO_3$,” *Physical Review B* 66, no. 21 (2002): 214112.

193. R. Ma, B. Zhang, G. Xu, et al., “Reversible Writing of High-Density Dislocations with Multidimensional Controllability in PMN-PT Crystal,” *Nature Communications* 16, no. 1 (2025): 5966.

194. Y. Aoki, H. Masuda, E. Tochigi, and H. Yoshida, “Overcoming the Intrinsic Brittleness of High-Strength Al_2O_3-$GdAlO_3$ Ceramics through Refined Eutectic Microstructure,” *Nature Communications* 15, no. 1 (2024): 8700.

195. G. L. Messing, S. Poterala, Y. Chang, et al., “Texture-Engineered Ceramics—Property Enhancements through Crystallographic Tailoring,” *Journal of Materials Research* 32, no. 17 (2017): 3219–3241.

196. N. Rohbeck, M. Watroba, C. Gunderson, et al., “Microscale Additively Manufactured 3D Metal-Ceramic Nanocomposites with Improved Strength and Thermal Stability,” *Additive Manufacturing* 111 (2025): 104957.

197. R. Yang, T. T. Zate, S. Mojumder, et al., “Strong Lead-Free Bioinspired Piezoceramics for Durable Energy Transducers,” Preprint on arXiv: 2508.15382, 2025.

198. Z. Tao, P. Zheng, G. Dong, and M. Liu, “The Mechanical Behaviors in Oxides: Beyond Brittleness,” *Advanced Materials* (2026): e15834.

199. L. Dou, B. Yang, X. Ye, et al., “Flexible High-Entropy Functional Ceramics,” *Nature Communications* 16, no. 1 (2025): 5915.

200. B. Ge, Y. Yu, and C. Zhou, “Dislocation-Mediated Thermoelectric Performance and Mechanical Behavior,” *Advanced Engineering Materials* 27, no. 16 (2025): 2500553.

201. F. R. N. Nabarro and M. S. Duesbery, *Dislocations in Solids* (Work Hardening, 2002), vol. 11.

202. K.-H. Yang, N.-J. Ho, and H.-Y. Lu, “Plastic Deformation of <001> Single-Crystal $SrTiO_3$ by Compression at Room Temperature,” *Journal of the American Ceramic Society* 94, no. 9 (2011): 3104–3111.

203. J. Zhang, O. Preuß, X. Fang, and W. Lu, “Nanoindentation Crack Suppression and Hardness Increase in $SrTiO_3$ by Dislocation Engineering,” *JOM* 77 (2025): 3503–3512.

204. S. Suresh, *Fatigue of Materials*, 2nd ed. (Cambridge University Press, 1998).

205. A. G. Evans and E. R. Fuller, “Crack Propagation in Ceramic Materials Under Cyclic Loading Conditions,” *Metallurgical Transactions* 5 (1974): 27–33.

206. R. H. Dauskardt, W. Yu, and R. O. Ritchie, “Fatigue Crack Propagation in Transformation-Toughened Zirconia Ceramic,” *Journal of the American Ceramic Society* 70, no. 10 (2005): C-248–C-252.

207. P. J. E. Forsyth, “Slip-Band Damage and Extrusion,” *Proceedings of the Royal Society A* 242 (1957): 198–202.

208. A. J. McEvily Jr and T. L. Johnston, “The Role of Cross-Slip in Brittle Fracture and Fatigue,” *International Journal of Fracture Mechanics* 3, no. 1 (1967): 45–74.

209. T. H. Alden, “Dynamic Recovery and Fatigue in Some Brittle Crystals,” *Acta Metallurgica* 11 (1963): 791–794.

210. K. N. Subramanian and J. Washburn, “Fatigue Deformation of Magnesium Oxide,” *Journal of Applied Physics* 34, no. 11 (1963): 3394–3397.

211. K. N. Subramanian, “Cyclic Stressing in Crystals of Rock-Salt Structure (Review),” *Physica Status Solidi* 29, no. 9 (1968): 9–37.

212. A. S. Argon and J. A. Godrick, “Fatigue in Lithium Fluride Single Crystals at Elevated Temperatures,” *Czechoslovak Journal of Physics B* 19 (1969): 341–342.

213. B. S. Majumdar and S. J. Burns, “Push-Pull Fatigue of LiF at Elevated Temperatures—I. Mechanical Properties,” *Acta Metallurgica* 30 (1982): 1743–1750.

214. B. S. Majumdar and S. J. Burns, “Push-Pull Fatigue of LiF at Elevated Temperatures—II. Microstructures,” *Acta Metallurgica* 30 (1982): 1751–1760.

215. B. S. Majumdar and S. J. Burns, “Fatigue Hardening of LiF Single Crystals in Push-Pull Cyclic Deformation,” *Acta Metallurgica* 29 (1981): 425–436.

216. B. S. Majumdar and S. J. Burns, “TEM Study of the Recovery Process of Cyclically Deformed MgO Single Crystals,” *Journal of Materials Science* 22 (1987): 1157–1162.

217. J. J. Gilman, “Debris Mechanism of Strain-Hardening,” *Journal of Applied Physics* 33, no. 9 (1962): 2703–2709.

218. M.-O. Guillou, J. L. Henshall, and R. M. Hooper, “Indentation Cyclic Fatigue of Single-Crystal Magnesium Oxide,” *Journal of the American Ceramic Society* 76, no. 7 (1993): 1832–1836.

219. S. Taeri, D. Brunner, W. Sigle, and M. Rühle, “Deformation Behaviour of Strontium Titanate between Room Temperature and 1800 K Under Ambient Pressure,” *International Journal of Materials Research* 95, no. 6 (2022): 433–446.

220. P. M. Anderson, J. P. Hirth, and J. Lothe, *Theory of Dislocations*, 3rd ed. (Cambridge University Press), 2017.

221. S. Kondo, T. Mitsuma, N. Shibata, and Y. Ikuhara, “Direct Observation of Individual Dislocation Interaction Processes with Grain Boundary,” *Science Advances* 2 (2016): e1501926.

222. R. Nakamura, H. Masuda, and H. Yoshida, “Nanoindentation Responses near Single Grain Boundaries in Oxide Ceramics,” *Journal of the American Ceramic Society* 106, no. 3 (2023): 2061–2072.

223. F. Janus, R. Kinoshita, R. Li, et al., “A Nanoindentation Study on the Room-Temperature Plasticity in Titanium Dioxide Bicrystals,” *Journal of the American Ceramic Society* 108 (2025): e20593.

224. C. Okafor, K. Ding, O. Preuß, N. Khansur, W. Rheinheimer, and X. Fang, “Near-Surface Plastic Deformation in Polycrystalline $SrTiO_3$ via Room-Temperature Cyclic Brinell Indentation,” *Journal of the American Ceramic Society* 107 (2024): 6715–6728.

225. K. Ding, A. Nakamura, P. Cordier, and X. Fang, “Dislocation Interaction with a Tilt Low-Angle Grain Boundary in Bi-Crystal $SrTiO_3$,” *Journal of the American Ceramic Society* 109, no. 1 (2026): e70492.

226. J. Maier, “Defect Chemistry : Composition, Transport, and Reactions in the Solid State; Part I: Thermodynamic,” *Angewandte Chemie International Edition* 32, no. 3 (1993): 313–456.

227. J. Maier, “Defect Chemistry : Composition, Transport, and Reactions in the Solid State; Part II: Kinetics,” *Angewandte Chemie* 32 (1993): 528–542.

228. R. A. De Souza, “Oxygen Diffusion in $SrTiO_3$ and Related Perovskite Oxides,” *Advanced Functional Materials* 25, no. 40 (2015): 6326–6342.

229. D. Marrocchelli, L. Sun, and B. Yildiz, “Dislocations in $SrTiO_3$: Easy to Reduce but Not so Fast for Oxygen Transport,” *Journal of the American Chemical Society* 137, no. 14 (2015): 4735–4748.

230. J. S. Koehler, D. Langreth, and B. von Turkovich, “Charged Dislocations in Ionic Crystals,” *Physical Review* 128, no. 2 (1962): 573–580.

231. A. Nakamura, K. Yasufuku, Y. Furushima, K. Toyoura, K. Lagerlöf, and K. Matsunaga, “Room-Temperature Plastic Deformation of Strontium Titanate Crystals Grown from Different Chemical Compositions,” *Crystals* 7, no. 11 (2017): 351.

232. X. Fang, K. Ding, S. Janocha, et al., “Nanoscale to Microscale Reversal in Room-Temperature Plasticity in $SrTiO_3$ by Tuning Defect Concentration,” *Scripta Materialia* 188 (2020): 228–232.

233. C. Okafor, A. Sayyadi-Shahraki, S. Bruns, et al., “Coupled Electromigration–nanoindentation Study on Dislocation Nucleation in $SrTiO_3$,” *Journal of the American Ceramic Society* 108 (2025): e70015.

234. P. Hirel, P. Carrez, E. Clouet, and P. Cordier, “The Electric Charge and Climb of Edge Dislocations in Perovskite Oxides: The Case of High-Pressure $MgSiO_3$ Bridgmanite,” *Acta Materialia* 106 (2016): 313–321.

235. Z. Zhang, B. Yang, L. Chen, Z. Zhang, and J. Guo, “Oxygen Vacancy-Induced Strengthening and Toughening in $(K,Na)NbO_3$-Based Piezoceramics Revealed via Nanoindentation,” *Nature Communications* 16, no. 1 (2025): 7015.

236. P. Hirel, P. Cordier, and P. Carrez, “$\langle 110\rangle\{1^{\rightarrow}10\}$ Edge Dislocations in Strontium Titanate: Charged vs Neutral, Glide vs Climb,” *Acta Materialia* 285 (2025): 120636.

237. K. Szot, C. Rodenbücher, G. Bihlmayer, et al., “Influence of Dislocations in Transition Metal Oxides on Selected Physical and Chemical Properties,” *Crystals* 8, no. 6 (2018): 241–317.

238. H. Oguri, Y. Li, X. Fang, and A. Nakamura, “Photoplastic Anisotropy in Nanoindentation of Wurtzite ZnO Single Crystals,” *Applied Physics Letters* 126, no. 11 (2025): 114101.

239. Y. Li, X. Fang, E. Tochigi, et al., “Shedding New Light on the Dislocation-Mediated Plasticity in Wurtzite ZnO Single Crystals by Photoindentation,” *Journal of Materials Science & Technology* 156 (2023): 206–216.

240. R. Kinoshita, Y. Li, H. Oguri, et al., “Effect of Light Irradiation on Indentation-Induced Dislocation Behavior in GaN Oriented for Basal Slip,” *Journal of the American Ceramic Society* 108, no. 9 (2025): e20663.

241. N. Nammahachak, M. Sriondee, T. Eknapakul, et al., “Dynamic Change in Light-Regulated Hardness in $SrTiO_3$,” *Applied Physics Letters* 127, no. 17 (2025).

242. T. Zhu, K. Ding, Y. Oshima, et al., “Switching the Fracture Toughness of Single-Crystal ZnS Using Light Irradiation,” *Applied Physics Letters* 118, no. 15 (2021): 154103.

243. K. K. Adepalli, M. Kelsch, R. Merkle, and J. Maier, “Influence of Line Defects on the Electrical Properties of Single Crystal TiO_2,” *Advanced Functional Materials* 23, no. 14 (2013): 1798–1806.