Solute adsorption, disconnections, and grain growth: Potentially critical aspects of microstructural evolution

Rachel Marder, W. Craig Carter, and Wayne D. Kaplan*

Grain growth, governed by the mobility and migration mechanisms of grain boundaries, is a fundamental process shaping the microstructure and properties of polycrystalline materials. This article explores the critical interplay between solute adsorption, disconnections, and complexion transitions in determining grain-boundary behavior. Traditional models based on curvature-driven motion are insufficient, as experimental and computational studies reveal complex mechanisms, including disconnection-mediated migration and transitions in structure or interface excess concentration. Solute atoms not only exert classical drag effects, but can also catalyze or suppress mobility by altering activation energies for disconnection formation or modifying boundary structures. The concept of grain-boundary complexions provides a thermodynamic framework for understanding structural and chemical transitions at interfaces, with direct implications for mobility. Special attention is given to ionic systems, where charge effects and electric fields further influence migration. Collectively, these insights highlight new strategies for tailoring grain growth to engineer advanced materials.

The impact of grain growth

The engineering importance of grain size and shape

One of the most fundamental aspects of materials science and engineering is the concept and control of grain size in polycrystalline materials. In fact, one could argue that the emergence of materials engineering as a distinct discipline was largely driven by the recognition of how microstructure influences materials properties.

Among the various features of a microstructure, grain size is particularly significant. The effects of grain size on materials properties were frequently measured in the 19th century.[1] Its impact on mechanical and functional properties continues to be extensively documented in metallic systems, semiconductors, and ceramics. One of the most well-known outcomes of grain size control is described by the Hall–Petch relationship, which shows that reducing grain size leads to increased strength in metallic systems.[2] This strengthening is typically attributed to the role of grain boundaries in impeding dislocation motion.[3,4]

Beyond plastic deformation, grain boundaries also play a central role in determining the fracture toughness and wear resistance of brittle materials; properties that are of key engineering relevance.[5,6] Equally important is the effect of grain size on the functional properties of polycrystalline materials.[7] In such cases, charge-carrier scattering and electromagnetic wave scattering at the perturbed crystal interfaces, the two-dimensional (2D) grain boundaries, significantly affect electrical or optical properties.

Given the engineering importance of grain size in polycrystalline materials, controlling the average grain size (and distribution of grain sizes) has become a central goal in the development of advanced engineering materials. The mechanisms used to limit grain size vary depending on the nature of the material system. In classical metallic systems, recrystallization of heavily deformed alloys can refine the grain size and increase the yield strength.[8]

Another approach to limiting grain growth is to restrict grain-boundary motion by pinning the boundaries with secondary phases.[9–11] This method is effective, but it can only be applied if the presence of these secondary phases is compatible with the overall material design.

An alternative method that does not require secondary phases involves the solute drag phenomenon.[12] In solute drag, segregating solute atoms accumulate at grain boundaries; the accumulation was believed to lead to a reduction in

Rachel Marder, Department of Materials Science and Engineering, Technion–Israel Institute of Technology, Haifa, Israel; rachelma@technion.ac.il
W. Craig Carter, Department of Materials Science and Engineering, MIT–Massachusetts Institute of Technology, Cambridge, USA; ccarter@mit.edu
Wayne D. Kaplan, Department of Materials Science and Engineering, Technion–Israel Institute of Technology, Haifa, Israel; kaplan@technion.ac.il
*Corresponding author
doi:10.1557/s43577-025-01025-3

© The Author(s) 2026

grain-boundary mobility. However, it seems plausible that solutes could also catalyze rearrangement in the boundary region and result in increased mobility. In general, there will be excess solute concentration in the boundary's vicinity; if those solutes are in equilibrium with those in the bulk, then the interfacial region's concentration will differ from the bulk. This is because the activity (or equivalently, the chemical potential) of any species must be the same everywhere at equilibrium, and the boundary vicinity has a different enthalpy and entropy density.[13] Thus, there is a driving force for solutes to segregate to (or from) a grain boundary until its concentration reaches its equilibrium value and the boundary's excess energy is minimal. The values of interface excess concentration are limited by the values of chemical potential in the bulk; within a phase-diagram's single-phase region, chemical potentials—and thus interface concentrations—can vary continuously. Within two-phase regions, the chemical potentials are pinned by the nucleating phase and so grain-boundary excess concentrations will be the same for each boundary. (Strictly speaking, they would be the same for boundaries with the same misorientation and grain-boundary plane.) Beyond the solubility limit, secondary phase nucleation (typically) occurs preferentially at boundaries.

The core idea of solute drag is that a migrating grain boundary influences the concentration profile of solutes. If solutes are attracted to a migrating grain boundary, then solutes behind the grain boundary must have a sufficiently large drift velocity to keep up; otherwise, the boundary energy would increase. If solutes desorb from the boundary, then those solutes ahead of the advancing boundary must themselves advance, otherwise they would be absorbed by the boundary. Unless the solutes diffuse with a sufficient drift velocity in the direction of motion, the boundary would sweep up the solutes and thus have a larger adsorption that it would have at equilibrium, and the boundary energy would increase. In either case, the energy increase diminishes the driving force creating a drag. As a result, the grain boundary's movement is limited by the diffusion rate of the solute cloud. This phenomenon plays a critical role in the structural evolution of engineering materials. However, this simplified view likely underestimates the complexity of the segregation effects because it ignores the effect of solutes on migration mechanisms.

Conceptual understanding of migration mechanisms historically derives from a heuristic of thin grain boundaries in which the transition from one orientation to another takes place over a few atoms. In addition to a bias that derives from early models for pure metals, thin boundaries are easier to image in a transmission electron microscope (TEM) and easier to study with numerical models. Boundaries are certainly more complex in materials with larger, multiatom, unit cells, and significantly less ordered (nearly amorphous) grain boundaries are observed in ceramics.[14–16] Identifying a boundary as amorphous is prone to misinterpretation; while there are certainly cases where the grain boundaries are far less ordered than the adjacent crystals, and there can even be a gradient in order across the boundary, we do not mean to imply that for thicker grain boundaries complete liquid-like disorder is achieved.[17] What is meant by "order" in this case? Suppose a diffraction pattern could be attained with enough spatial resolution that there would be an identifiable change as the probe traverses perpendicularly to the grain-boundary region. In the abutting grains close to the grain-boundary region, the diffraction pattern would have sharp reflections according to the grain's orientation. As the probe enters the grain-boundary region, the spots would broaden and the pattern would evolve toward—but not completely to—that which is typical for scattering from amorphous materials.[18] Upon passing toward the second crystal, those diffuse features would collapse into reflections indicating the structure and orientation of the second crystal. The peak broadening could be used to construct an order parameter to characterize the boundary.

Driving force for grain growth

The reduction in the excess Gibbs free energy during grain growth is conceptually straightforward. As grains grow, the total grain-boundary area decreases. Since grain boundaries are two-dimensional defects that carry excess energy, this decrease in boundary area leads to a corresponding reduction in the system's free energy.

Because this driving force is the same as for soap froth, it was natural to draw an analogy between grain growth in a polycrystalline solid and bubble growth in a foam.[2] In such a model, the movement of grain boundaries is thought to be governed by their curvature, with the growth direction determined by the boundary's shape.[7] The velocity ($\vec{v}$) of grain-boundary motion is then assumed to be proportional to the local rate of interfacial energy decrease with respect to the volume (V) of material transferred between the two grains $\frac{\delta\gamma\vec{A}}{\delta V} = \gamma\vec{\kappa}$ (if γ is unchanged by the motion) where $\vec{\kappa}$ is the local mean curvature vector (which is normal to the interface and directed toward the center of curvature), γ is the grain-boundary excess energy, and A is the area of the grain boundary. This simple model is that $\vec{v} = M\gamma\vec{\kappa}$ where the proportionality constant M is called the grain-boundary mobility. Following this, models for grain growth were developed.[19,20] It should be recognized that, here, mobility is a proportionality factor between velocity and curvature and is given in the context of a simple model. Below, mobility refers to processes that facilitate motion and not to grain-boundary energy—however, motion and energy are certainly coupled to the boundary's atomic structure.

Bubble growth in a foam does not capture the full complexity of the atomistic structure of boundaries between crystals:[21,22] coarsening in foams occurs by atomic diffusion through an intermediate phase, whereas grain boundaries move by local reorientation of atoms. The physics governing grain-boundary migration in crystalline solids is more complex and cannot be fully captured by models derived from fluid foams.

The complexity of grain-boundary motion has recently been demonstrated by measuring the velocity of very large samples of individual boundaries as a function of time in

polycrystals.[23–26] These experimental and simulation studies demonstrated that there is no experimental correlation between curvature and velocity of grain boundaries. This of course does not mean that eradication of the grain-boundary area and its excess energy does not result in a reduction of energy of the system, but curvature is insufficient to determine a grain boundary's driving force and velocity. Furthermore, the behavior of grain growth in all materials should not be inferred from experiments on a single material: there may be similarities in simple metals and similarities in complex oxides, but they are unlikely to be similar to each other. In any case, the mechanism by which grain boundaries move must be included.

Grain-boundary migration mechanisms: Disconnections versus order–disorder

Grain growth occurs when atoms change their crystallographic orientation from that of the shrinking grain to that of the growing grain. Mott hypothesized that grain growth occurs by a process akin to a phase transition ("…the crystal grows at each step by the disordering or melting of a group of *n* atoms … then takes up the structure of the growing crystal"[27]). This is plausible in boundaries with a thick quasicrystalline/quasi-amorphous internal structure. Gleiter developed a model for thin grain boundaries that can be represented as a 2D crystallographic defect: the grain boundary is displaced by the motion of steps (i.e., the termination of a lattice plane in the vicinity of a grain boundary). The coordinated movement of these steps enables atoms to migrate from one grain to the other, thereby facilitating grain growth.[28]

A grain boundary is a crystalline defect, but it is not a 2D defect. While some TEM images show boundary structures—especially high-symmetry boundaries—that are very sharp, other quantities (e.g., excess enthalpy, excess solute atmospheres, and excess entropy) extend farther than what might be inferred from a boundary's structure. Quantities such as the interface normal, $\hat{n}$, and curvature, $\vec{\kappa}$, are associated with a 2D mathematical surface; but, as Gibbs showed, these can be associated with any two-dimensional surface which is co-parallel to the boundary region. Gibbs also recognized that surfaces have a finite extension into the abutting phases: the 2D reduction was a mathematical convenience. Cahn developed a surface thermodynamics that is equivalent to that of Gibbs, but treats the boundary as a layer.[13] It is not uncommon to use "2.5-dimensional defect" as a designation for grain boundaries.

This initial concept of moving steps within the grain-boundary plane was later expanded by Hirth and Pond, who identified linear defects within the grain boundary's 2D periodic structure that can exhibit step character, dislocation character, or both. They referred to these defects as disconnections and proposed that grain-boundary motion is governed by the formation of new disconnections and their subsequent movement along the grain-boundary plane.[29]

The proposition that disconnections facilitate grain-boundary motion is natural for planar thin metallic boundaries or grain boundaries that have 2D periodicity (see **Figure** 1). For thicker boundaries in which limited periodicity exists in the boundary's proximity, Mott's hypothesis is plausible: motion must be achieved by a transition from one crystallographic orientation (on the retreating side of the grain boundary) to a non-crystallographic material between the two orientations and then subsequent transition to the orientation of the growing grain (see Figure 1c). It is plausible that the two transitions are associated with independent step motion on the quasicrystalline/quasi-amorphous surfaces (by analogy to crystallization from a melt). The motion of these steps is likely to become more correlated for thinner quasi-amorphous regions. Continuing the analogy to interphase boundaries; understanding of the role of solutes in Stranski–Krastanov growth is applicable to thick boundary motion as well.[30] There are parallels between boundary motion and thin-film growth:[31] the Frank–van der Merwe epitaxial growth-mode (where ledges grow laterally on the surface of the growing crystal) and its coupling to stress are akin to the disconnection mode of grain-boundary motion. The Stranski–Krastanov and Volmer–Weber modes (where there is normal attachment from an amorphous phase) are akin to the processes that occur for boundaries with a quasi-amorphous character.

The advent of disconnections led to important studies investigating shear coupling between grain-boundary motion and shear strain.[32,33] Detailed electron microscopy, both *ex situ* and *in situ*, and simulation tools have since been used to characterize the role of disconnections in grain growth and in phase transformations.[34–45] Understanding the activation energy required to nucleate disconnections is just as important as the mechanism itself. Simulations have shown that these activation energies can be highly anisotropic, meaning that grain growth is predominantly governed by specific types of disconnections at any given temperature.[40] As the temperature rises, this anisotropy may diminish, enabling other types of disconnections to become active. This behavior resembles the temperature-dependent activation of different dislocation glide systems in bulk materials.

If the types of disconnections formed can be influenced by the temperature via an activation energy, then it is reasonable to assume that disconnection formation may be affected by segregating solutes, as demonstrated schematically in Figure 1. Consequently, it is reasonable to expect that absorbing solute atoms may also affect the activation energy required to form disconnections and/or their mobility in the grain-boundary plane. This introduces specific mechanisms into the broader concept of solute–drag, highlighting how adsorbing solutes might alter the kinetics of boundary motion. Although this remains poorly understood at this time, a deeper understanding of the interactions between solutes and disconnections could provide valuable insight into engineering microstructural evolution. It is plausible that adsorbing solutes could increase or decrease the activation energy (i.e., they passivate grain growth or catalyze it), and the effect

depends on boundary misorientation and the particular solute/host combination.

If boundary migration occurs by the order/quasi-disorder/order process, it is plausible that solutes that lower the glass-transition temperature would reduce the activation energy for grain-boundary motion.

Influencing disconnections so as to influence grain growth

Understanding how adsorbing solutes influence the formation and motion of disconnections requires a more complete understanding of the 2D structures present at thin grain boundaries. There is a long history of theoretical and experimental work aimed at describing and predicting these structures, most of which has relied on geometric models derived from O-lattice theory and the coincident site lattice theory.[46–49] However, more recent studies have shown that geometric models do not reliably predict which grain-boundary structures will form, because they do not include the finer aspects of the local "2D" structure and chemical bonding.[50,51] These studies also reveal that grain-boundary structures can vary with changes in temperature, pressure, and chemical composition.[52] The behavior is analogous to bulk phase transformations and basically identical to the concept of surface reconstruction, which has been an important topic in the physics of surfaces and catalysis.[53] Moreover, at higher temperatures, fluctuations in the boundary structure and position confound the applications of 2D periodic models—especially for those boundaries with larger excess entropies.

Fundamental theories addressing the energetic contributions of structural order, chemical composition, electrostatics, and crystallographic orientation at grain boundaries (and surfaces more generally) have been rigorously developed within the framework of diffuse interface models.[54–57] This approach allows surfaces and interfaces to be treated analogously to bulk phases, in the sense that they can undergo first-order or higher-order transformations in response to changes in thermodynamic conditions (see **Figure** 2).

However, two important caveats must be considered when drawing such an analogy. First, surfaces and interfaces are not bulk phases; their existence depends on coexistence with neighboring bulk phases. Interfaces do not possess a Gibbs–Duhem relation that is independent of that of the existing phase(s): An interface does not place further restrictions on equilibrium changes in the intensive thermodynamic variables. Moreover, an interface's thermodynamic properties are excess layer quantities that refer to their abutting phase(s) and they must use those phases as a reference for the proper identification of layer quantities.[13] Second, grain boundaries can only be in local thermodynamic equilibrium; polycrystals will always have a free energy in excess of a single crystal's. There can be many grain-boundary structures with nearly the same excess free energy and a boundary's structure and position will undergo equilibrium fluctuations. Moreover, the boundary structure can be a function of the driving force and applied fields. These effects may have profound implications for processing: Metastable grain-boundary structures can provide rapid transport (grain-boundary diffusion) and decrease the time required to densify materials during sintering.[58]

Despite these limitations, the concept of first-order or higher-order transitions resulting in reversible changes in the structure and chemistry of grain boundaries may be critical for understanding grain-boundary-dependent properties.[59] Grain growth kinetics are certainly influenced by such transitions. Consequently, attention has been directed toward the potential role of first-order transitions in grain growth behavior. This focus led to the adoption of a simplified version of diffuse interface theory, in which the term complexion was introduced to describe the lowest free energy state of a grain boundary (and to disambiguate boundary transitions from bulk phase transitions that could occur in their vicinity), and attempts were made to map complexion transitions onto changes in grain-boundary mobility.[60–63] Fundamentally, there is no contradiction between the concept of complexions and transitions at grain boundaries involving solute adsorption, as defined by Gibbs.[11] However, the mechanism by which transitions from one complexion to another influence grain-boundary mobility

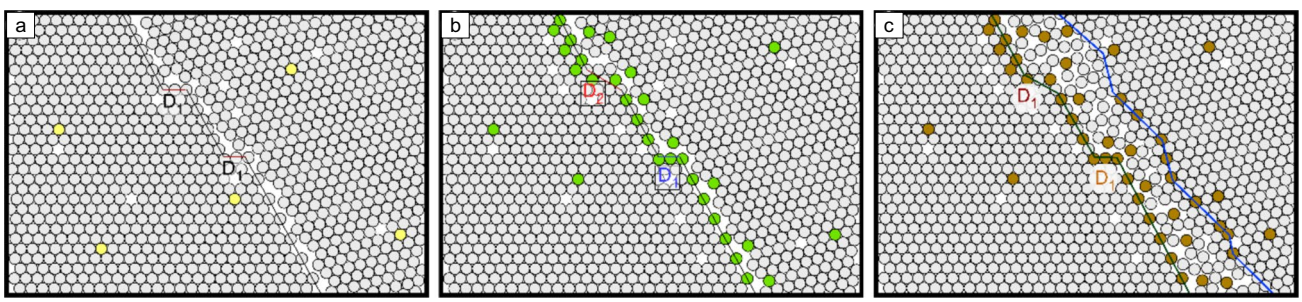

Figure 1. Schematic drawing of a general high-angle grain boundary between two crystals, containing solute atoms. In (a), there is no detectable adsorption of the solute atoms to the grain boundary, which contains two disconnections (D_1). In (b), solute adsorption is easily detected, and in addition to reducing the grain-boundary excess energy, the step plane of one of the disconnections has changed (D_2). In (c), the grain boundary is diffuse, and it is not clear if disconnection motion or the Mott mechanism of step dissolution and redeposition facilitates grain-boundary motion.[27]

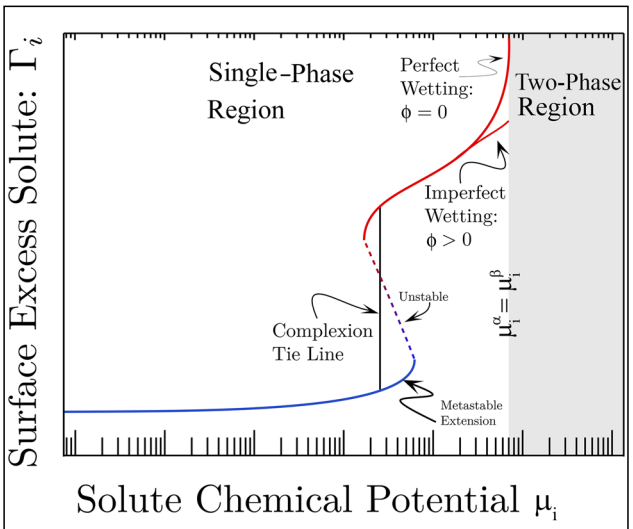

Figure 2. Schematic illustration of an adsorption isotherm (blue, black, and red line) for a grain boundary in a single-phase polycrystalline material (white background): the surface excess concentration of species *i* is plotted as a function of its chemical potential for compositions that lie in the system's phase diagram. As μ_i is increased, Γ_i increases for the complexion associated with the blue curve. A complexion coexistence line (tie line, black) connects the low μ_i stable complexion to another complexion (red) that is stable at higher μ_i. The jump $\Delta\Gamma_i$ is associated with the first-order transition. There are metastable extensions for each complexion where (presumably) a nucleation barrier exists for the complexion transition. The metastable extensions terminate at values of Γ_i that are unconditionally unstable. This is the same construction as a chemical potential as it traverses a spinodal in composition. As μ_i approaches its two-phase coexistence value (gray background region) ($\mu_i^\alpha = \mu_i^\beta$), a second phase would wet the grain boundary: when the wetting angle is small ($\phi_{contact} \approx 0$), Γ_i becomes very large at the two-phase boundary.

remains unclear and is still an open question. The only way to investigate this is by directly correlating the mechanism by which grain boundaries move; whether through disconnections or atomic ordering onto steps (the Mott mechanism), with transitions in the structural or chemical state of the boundary. This remains an important and active direction for future research.

Grain-boundary structure and chemistry and their influence on mobility

Dillon et al. classified grain boundaries in alumina and identified a correlation between grain-boundary width and mobility.[60] They constructed an Arrhenius plot of mobility versus 1/T showing that thicker boundaries had higher mobilities. Remarkably, the mobilities' activation energies were nearly the same—suggesting that the activation processes may be the same for all the observed boundaries. This is consistent with the Mott-type order/quasi-disorder/order two-step mechanism for boundary migration: The activation energy is that required for the transformation of an atom group between order and quasi-disorder. However, the Arrhenius intercepts are offset (i.e., the *y*-intercepts differ) which suggests that the activation entropies increase with thickness, and this is consistent with a hypothesis that the number of activated configurations increases with thickness: There are more ways to transfer order between the advancing and receding interfaces in thick boundaries (see **Figure** 3).[64]

If there is a correlation between mobility and structure, and if there is a structural change in some of the migrating boundaries, grain growth kinetics should change at those temperatures and solute chemical potentials where such structural transitions occur. In general, changes in the system's intensive variables (P, μ_i, T, etc.) can produce first-order transitions among boundaries with different excess extensive values ([V],[N_i],[S]) and structure. The various boundaries can be called complexions.[54–60] The values of the intensive variables at which transitions between complexions will depend on boundary misorientation and normal vector.

This suggests a scenario for how mobility might evolve with composition (via changes in the chemical potentials). Whether boundary migration occurs by disconnections or a Mott-like process, solutes in the boundary's vicinity can affect the migration mechanism and its thermal activation. Whether the solute increases the mobility or not is illustrated in **Figure** 4. The red curve is for the case where the solute catalyzes the boundary migration process (the curve commences for single complexion in equilibrium with a composition in a single-phase region of the bulk material's phase diagram). A transition to a new complexion at a different composition will change the mobility, but that change could either be an increase or a decrease. Subsequent to the formation of a new complexion, a solute's effect on mobility may differ—in a different structure a solute could either catalyze or poison the migration process. This is illustrated by the bifurcations in Figure 4 (the illustration should be interpreted that—subsequent to the transition—the behavior is either poisoning or catalyzing and not both). If there is a solubility limit, then a second phase can appear at the boundary and Zener drag will play an additional role that is external to that boundary's internal migration mechanisms.

For moving boundaries with solutes, there are two time scales: one associated with the migration step and the other with the solute diffusion step. If solutes would adsorb to the boundary, but the migration step is faster than the diffusion step, then the boundary can break away from its atmosphere and migrate with a nonequilibrium solute excess composition. This is one of the modes in Cahn's solute drag model. However, Cahn's model does not include the catalyzing/poisoning effects previously described. If breakaway results in the loss of a solute that would catalyze the boundary, then the boundary would slow down, and the breakaway transition would be forestalled.

Grain-boundary migration in ionic materials

The character of grain boundaries is more complex in ionic materials. The arrangement of anions and cations in the unit cell will certainly limit the number of crystallographic planes that abut a grain boundary, and the step characteristics will reflect this—the size of disconnections will be larger, and their thermal activations will have larger energies. For boundaries that are thin enough to have an identifiable 2D crystallography,

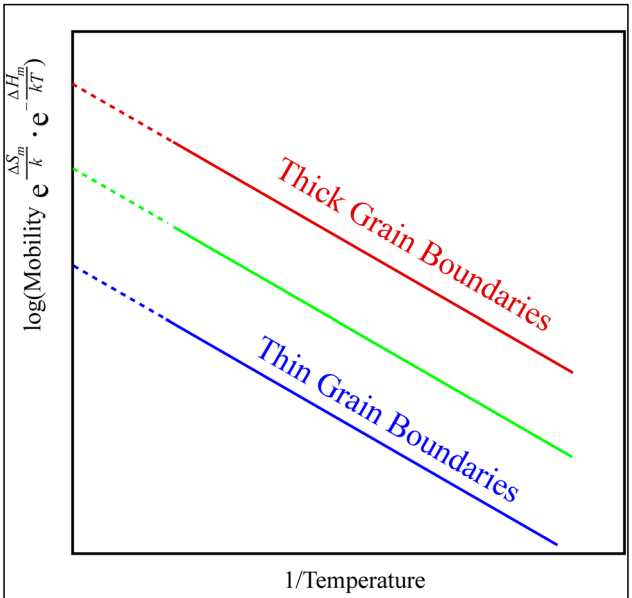

Figure 3. Schematic drawing of the mobility versus temperature for grain boundaries with different thicknesses implied by results of Dillon et al.[60] Similar values for the activation energy (related to the slope $\frac{-\Delta H_m}{kT}$) suggest that the migration mechanism is the same. Whereas, the intercept at $T = \infty$ (i.e., $\frac{\Delta S_m}{k}$) suggests a larger multiplicity of activated configurations for thicker grain boundaries.

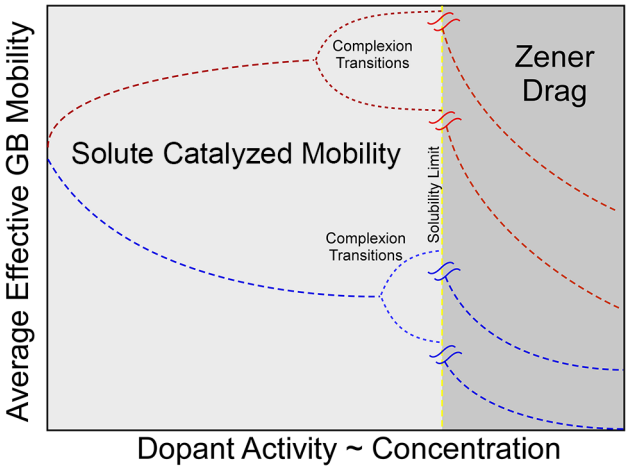

Figure 4. Schematic illustration of how complexion transitions may affect the average grain-boundary (GB) mobility in a polycrystalline material. Depending on the species, solute adsorption can increase (dashed red line) the GB mobility or decrease (dashed blue line). Complexion transitions would result in gradual changes in the mobility, because not every GB will undergo a complexion transition at the same concentration, nor will the magnitudes and signs of the grains' mobility change: There will be a distribution of mobilities. At concentrations beyond the solubility limit, adsorption will likely be dominated by Zener drag.

it is plausible that a Kröger–Vink defect thermodynamics could be constructed for disconnections and an analogy to a Pourbaix diagram for boundary migration mechanisms result. This may be a direction for future useful research.

For boundaries that have a quasi-amorphous interior, the foundations of the Kröger–Vink method disappear; there is no site conservation and the formal charge at an atom's location is vague. Nevertheless, it is plausible that charged crystalline defects within the grains will interact with the quasi-amorphous grain-boundary interior and this would affect migration. The development of a new theoretical framework may shed insights.

In any case, the grain region may be charge polarized, and this would create an asymmetry with respect to the direction of an applied electric field $\vec{E}$. Grain evolution in ionic polycrystals would depend on the orientation of applied field. Moreover, an electrical double layer could develop at the boundaries exterior that would create an additional activation step for a solute to adsorb. For thicker boundaries, double layers could develop within the quasi-amorphous layer that would affect character of the migration configurations and its interaction with an applied $\vec{E}$.

Summary

Grain growth is a central process in the evolution of polycrystalline materials, with grain-boundary mobility directly influencing mechanical and functional properties. Traditional models assume that grain-boundary motion is driven solely by curvature and energy reduction, but recent studies demonstrate that migration mechanisms are more complex. Two primary pathways are emphasized: migration via disconnections, which are step-like defects enabling atomic transfer; and order–disorder transitions, where boundaries with quasi-amorphous structures rearrange as atoms change crystallographic orientation.

The role of solute atoms is shown to be more complicated than the classical solute–drag model suggests. Adsorbed solutes can impede, accelerate, or redirect boundary motion by modifying activation energies or altering boundary structures. This interplay is framed within the concept of grain-boundary complexions, which describes structural and chemical transitions at interfaces analogous to phase transformations. Such transitions strongly influence mobility and, consequently, grain growth kinetics. A deeper understanding of these mechanisms should open new pathways for tailoring microstructures and designing advanced materials with controlled grain growth.

Author contributions

All authors whose names appear on the submission: (1) made substantial contributions to the conception and design of the work; (2) participated in the initial draft and its revision; and (3) approved the version to be published. All three authors participated in the concept of the review, contributed to the identification of important literature which should be included, and drafted and critically revised the work.

Funding

Open access funding provided by Technion - Israel Institute of Technology. W.D.K. and R.M. acknowledge partial support from the Israel Science Foundation (ISF 544/22).

Competing interests

The authors have no conflicts of interest or competing interests regarding this manuscript.

Open Access

This article is licensed under a Creative Commons Attribution 4.0 International License, which permits use, sharing, adaptation, distribution and reproduction in any medium or format, as long as you give appropriate credit to the original author(s) and the source, provide a link to the Creative Commons licence, and indicate if changes were made. The images or other third party material in this article are included in the article's Creative Commons licence, unless indicated otherwise in a credit line to the material. If material is not included in the article's Creative Commons licence and your intended use is not permitted by statutory regulation or exceeds the permitted use, you will need to obtain permission directly from the copyright holder. To view a copy of this licence, visit http://creativecommons.org/licenses/by/4.0/.

References

1. K.T. Aust, *Metall. Trans.* **1**, 2055 (1970). https://doi.org/10.1007/BF02642827
2. C.S. Smith, *Metallogr. Microstruct. Anal.* **4**, 543 (2015). https://doi.org/10.1007/s13632-015-0241-1
3. Z. Shen, R.H. Wagoner, W.A.T. Clark, *Acta Metall.* **36**, 3231 (1988). https://doi.org/10.1016/0001-6160(88)90058-2
4. S. Kondo, T. Mitsuma, N. Shibata, Y. Ikuhara, *Sci. Adv.* **2**, e1501926 (2016). https://doi.org/10.1126/sciadv.1501926
5. J.J. Kruzic, R.M. Cannon, R.O. Ritchie, *J. Am. Ceram. Soc.* **88**, 2236 (2005). https://doi.org/10.1111/j.1551-2916.2005.00434.x
6. N.A. Yahya, R.I. Todd, *J. Eur. Ceram. Soc.* **32**, 4003 (2012). https://doi.org/10.1016/j.jeurceramsoc.2012.07.003
7. B. Poudel, Q. Hao, Y. Ma, Y. Lan, A. Minnich, B. Yu, X. Yan, D. Wang, A. Muto, D. Vashaee, X. Chen, J. Liu, M.S. Dresselhaus, G. Chen, Z. Ren, *Science* **320**, 634 (2008). https://doi.org/10.1126/science.1156446
8. A. Rollett, F.J. Humphreys, G.S. Rohrer, M. Hatherly, *Recrystallization and Related Annealing Phenomena*, 2nd edn. (Elsevier, Oxford, 2004). https://doi.org/10.1016/B978-0-08-044164-1.X5000-2
9. C.S. Smith, *Trans. Metall. Soc. AIME* **175**, 15 (1948). https://doi.org/10.1007/s11663-010-9364-6
10. C.H. Hsueh, A.G. Evans, R.L. Coble, *Acta Metall.* **30**, 1269 (1982). https://doi.org/10.1016/0001-6160(82)90145-6
11. E. Nes, N. Ryum, O. Hunderi, *Acta Metall.* **33**, 11 (1985). https://doi.org/10.1016/0001-6160(85)90214-7
12. J.W. Cahn, *Acta Metall.* **10**, 789 (1962). https://doi.org/10.1016/0001-6160(62)90092-5
13. J.W. Cahn, "Thermodynamics of Solid and Fluid Surfaces," in *Interfacial Segregation*, ed. by W.C. Johnson, J.M. Blakely (ASM International, Metals Park, 1979), pp. 3–23
14. D.R. Clarke, *J. Am. Ceram. Soc.* **70**, 15 (1987). https://doi.org/10.1111/j.1151-2916.1987.tb04846.x
15. I. Tanaka, H.J. Kleebe, M.K. Cinibulk, J. Bruley, D.R. Clarke, M. Ruhle, *J. Am. Ceram. Soc.* **77**, 911 (1994). https://doi.org/10.1111/j.1151-2916.1994.tb07246.x
16. A. Subramaniam, C.T. Koch, R.M. Cannon, M. Ruhle, *Mater. Sci. Eng. A* **422**, 3 (2006). https://doi.org/10.1016/j.msea.2006.01.004
17. M. Baram, S.H. Garofalini, W.D. Kaplan, *Acta Mater.* **59**, 5710 (2011). https://doi.org/10.1016/j.actamat.2011.05.046
18. M. Gandman, Y. Kauffmann, C.T. Koch, W.D. Kaplan, *Phys. Rev. Lett.* **110**, 086106 (2013). https://doi.org/10.1103/PhysRevLett.110.086106
19. W.W. Mullins, *J. Appl. Phys.* **27**, 900 (1956). https://doi.org/10.1063/1.1722511
20. L.A. Barrales Mora, G. Gottstein, L.S. Shvindlerman, *Acta Mater.* **56**, 5915 (2008). https://doi.org/10.1016/j.actamat.2008.08.006
21. A.P. Sutton, *Int. Met. Rev.* **29**, 377 (1984). https://doi.org/10.1179/imtr.1984.29.1.377
22. J.E. Taylor, *Acta Metall. Mater.* **40**, 1475 (1992). https://doi.org/10.1016/0956-7151(92)90091-R
23. A. Bhattacharya, Y.-F. Shen, C.M. Hefferan, S.F. Li, J. Lind, R.M. Suter, C.E. Krill, G.S. Rohrer, *Science* **374**, 189 (2021). https://doi.org/10.1126/science.abj3210
24. M. Lyu, Z. Xu, G.S. Rohrer, E.A. Holm, *Scr. Mater.* **256**, 116429 (2025). https://doi.org/10.1016/j.scriptamat.2024.116429
25. V. Muralikrishnan, Z. Xu, G.S. Rohrer, A.R. Krause, *J. Am. Ceram. Soc.* **108**, e20319 (2024). https://doi.org/10.1111/jace.20319
26. C. Qiu, D.J. Srolovitz, G.S. Rohrer, J. Han, M. Salvalaglio, *Proc. Natl. Acad. Sci. U.S.A.* **122**, e2500707122 (2025). https://doi.org/10.1073/pnas.2500707122
27. N.F. Mott, *Proc. Phys. Soc.* **60**, 391 (1948). https://doi.org/10.1088/0959-5309/60/4/309
28. H. Gleiter, *Acta Metall.* **17**, 565 (1969). https://doi.org/10.1111/jace.20319
29. J.P. Hirth, R.C. Pond, *Acta Mater.* **44**, 4749 (1996). https://doi.org/10.1111/jace.20319
30. I.N. Stranski, L. Krastanow, *Monatsh. Chem.* **71**, 351 (1937). https://doi.org/10.1007/BF01798103
31. H. Liu:, "Growth Kinetics of Thin Film Epitaxy," in *21st Century Surface Science-A Handbook*, edited by P.V. Pham, P. Goel, S. Kumar, K. Yadav (IntechOpen, London, 2020), chap. 1, pp. 3–26. https://doi.org/10.5772/intechopen.87891
32. A. Rajabzadeh, M. Legros, N. Combe, F. Mompiou, D.A. Molodov, *Philos. Mag.* **93**, 1299 (2013). https://doi.org/10.1080/14786435.2012.760760
33. N. Combe, F. Mompiou, M. Legros, *Phys. Rev. Mater.* **3**, 060601 (2019). https://doi.org/10.1103/PhysRevMaterials.3.060601
34. T. Radetic, C. Ophus, D.L. Olmsted, M. Asta, U. Dahmen, *Acta Mater.* **60**, 7051 (2012). https://doi.org/10.1016/j.actamat.2012.09.012
35. M.L. Bowers, C. Ophus, A. Gautam, F. Lançon, U. Dahmen, *Phys. Rev. Lett.* **116**, 106102 (2016). https://doi.org/10.1103/PhysRevLett.116.106102
36. H. Sternlicht, W. Rheinheimer, M. Hoffmann, W.D. Kaplan, *J. Mater. Sci.* **51**, 467 (2016). https://doi.org/10.1007/s10853-015-9058-1
37. H. Sternlicht, W. Rheinheimer, R.E. Dunin-Borkowski, M.J. Hoffmann, W.D. Kaplan, *J. Mater. Sci.* **54**, 3694 (2019). https://doi.org/10.1007/s10853-018-3096-4
38. H. Sternlicht, W. Rheinheimer, J. Kim, E. Liberti, A.I. Kirkland, M.J. Hoffmann, W.D. Kaplan, *J. Mater. Sci.* **54**, 3710 (2019). https://doi.org/10.1007/s10853-018-3095-5
39. H. Sternlicht, W. Rheinheimer, A. Mehlmann, A. Rothschild, M.J. Hoffmann, W.D. Kaplan, *Scr. Mater.* **188**, 206 (2020). https://doi.org/10.1016/j.scriptamat.2020.07.015
40. J. Han, S.L. Thomas, D.J. Srolovitz, *Prog. Mater. Sci.* **98**, 386 (2018). https://doi.org/10.1016/j.pmatsci.2018.05.004
41. C. Wei, S.L. Thomas, J. Han, D.J. Srolovitz, Y. Xiang, *J. Mech. Phys. Solids* **133**, 103731 (2019). https://doi.org/10.1016/j.jmps.2019.103731
42. K. Chen, J. Han, D.J. Srolovitz, *Acta Mater.* **194**, 412 (2020). https://doi.org/10.1016/j.actamat.2020.04.057
43. L. Wang, Y. Zhang, Z. Zeng, H. Zhou, J. He, P. Liu, M. Chen, J. Han, D.J. Srolovitz, J. Teng, Y. Guo, G. Yang, D. Kong, E. Ma, Y. Hu, B. Yin, X. Huang, Z. Zhang, T. Zhu, X. Han, *Science* **375**, 1261 (2022). https://doi.org/10.1126/science.abm2612
44. G. Wei, H. Xie, J.-P. Du, T. He, G. Lu, S. Ogata, *Acta Mater.* **280**, 120325 (2024). https://doi.org/10.1016/j.actamat.2024.120325
45. Y. Chen, J. Han, H. Deng, G. Cao, Z. Zhang, Q. Zhu, H. Zhou, D.J. Srolovitz, J. Wang, *Acta Mater.* **268**, 119717 (2024). https://doi.org/10.1016/j.actamat.2024.119717
46. D.G. Brandon, *Acta Metall.* **14**, 1479 (1966). https://doi.org/10.1016/0001-6160(66)90168-4
47. W. Bollmann, *Surf. Sci.* **31**, 1 (1972). https://doi.org/10.1016/0039-6028(72)90250-6
48. R.W. Balluffi, A. Brokman, A.H. King, *Acta Metall.* **30**, 1453 (1982). https://doi.org/10.1016/0001-6160(82)90166-3
49. H. Grimmer, R. Bonnet, S. Lartigue, L. Priester, *Philos. Mag. A* **61**, 493 (1990). https://doi.org/10.1080/01418619008231929
50. G.S. Rohrer, E.A. Holm, A.D. Rollett, S.M. Foiles, J. Li, D.L. Olmsted, *Acta Mater.* **58**, 5063 (2010). https://doi.org/10.1016/j.actamat.2010.05.042
51. T. Brink, L. Langenohl, H. Bishara, G. Dehm, *Phys. Rev. B* **107**, 054103 (2023). https://doi.org/10.1103/PhysRevB.107.054103
52. S. Zhang, E. Isotta, R. Bueno Villoro, C. Scheu, *MRS Bull.* **51**(2) (2026). https://doi.org/10.1557/s43577-025-01041-3
53. G.A. Somorjai, *Annu. Rev. Phys. Chem.* **45**, 721 (1994). https://doi.org/10.1146/annurev.pc.45.100194.003445
54. C.M. Bishop, R.M. Cannon, W.C. Carter, *Acta Mater.* **53**, 4755 (2005). https://doi.org/10.1016/j.actamat.2005.07.008
55. M. Tang, W.C. Carter, R.M. Cannon, *Phys. Rev. B* **73**, 024102 (2006). https://doi.org/10.1103/PhysRevB.73.024102
56. M. Tang, W.C. Carter, R.M. Cannon, *J. Mater. Sci.* **41**, 7691 (2006). https://doi.org/10.1007/s10853-006-0608-4
57. M. Tang, W.C. Carter, R.M. Cannon, *Phys. Rev. Lett.* **97**, 075502 (2006). https://doi.org/10.1103/PhysRevLett.97.075502

58. Z. Guo, R.I. Todd, *Acta Mater.* **282**, 120471 (2025). https://doi.org/10.1016/j.actamat.2024.120471
59. S. Calderon, W.-Y. Chen, M.S. Islam, S.S. Tripathy, M.P. de Boer, E.C. Dickey, G.S. Rohrer, *Scr. Mater.* **269**, 116902 (2025). https://doi.org/10.1016/j.scriptamat.2025.116902
60. S.J. Dillon, M. Tang, W.C. Carter, M.P. Harmer, *Acta Mater.* **55**, 6208 (2007). https://doi.org/10.1016/j.actamat.2007.07.029
61. S.J. Dillon, M.P. Harmer, *J. Am. Ceram. Soc.* **91**, 2304 (2008). https://doi.org/10.1111/j.1551-2916.2008.02454.x
62. W. Kaplan, D. Chatain, P. Wynblatt, W.C. Carter, *J. Mater. Sci.* **48**, 5681 (2013). https://doi.org/10.1007/s10853-013-7462-y
63. P.R. Cantwell, M. Tang, S.J. Dillon, J. Luo, G.S. Rohrer, M.P. Harmer, *Acta Mater.* **62**, 1 (2014). https://doi.org/10.1016/j.actamat.2013.07.037
64. D. Turnbull, *Trans. AIME* **191**, 661 (1951). https://doi.org/10.1007/BF03397362

Publisher's note Springer Nature remains neutral with regard to jurisdictional claims in published maps and institutional affiliations.



Rachel Marder is a research scientist in the Department of Materials Science and Engineering at the Technion–Israel Institute of Technology. She received her PhD degree in materials science from the Technion in 2015. Her research focuses on ceramics, with emphasis on sintering, grain growth, and microstructure evolution, particularly the influence of dopants on microstructural development. She has expertise in advanced sintering methods, including spark plasma sintering, and in electron microscopy for characterizing ceramics. In addition to her research, she lectures on ceramic materials to undergraduate students. Marder can be reached by email at rachelma@technion.ac.il.



W. Craig Carter is the Toyota Professor of Materials Science at the Massachusetts Institute of Technology. His research interests include theory and computational modeling of materials. He has worked on many different subfields of materials science, including microstructural evolution, interface thermodynamics, mechanics and reliability, and kinetics. He is interested in innovations in teaching materials science and has several teaching awards. Carter can be reached by email at ccarter@mit.edu.



Wayne D. Kaplan is the Karl Stoll Chair in Advanced Materials at the Technion–Israel Institute of Technology, where he has been a faculty member since 1995. At the Technion, he has served as Dean, Vice President for Research, and Vice President for External Relations and Resource Development. His research focuses on interface science, particularly wetting, solute adsorption, and grain-boundary mobility in ceramics, where he employs advanced electron microscopy and associated techniques. Kaplan can be reached by email at kaplan@technion.ac.il.