Viewpoint article

Unusual dislocation behavior in high-entropy alloys

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This Viewpoint article articulates ten unusual features associated with dislocations and their slip processes in high-entropy alloys (HEAs). Attention will be directed towards the composition undulation inherent in these highly concentrated solutions, even when the alloy is a single-phase and nominally random solution. This inhomogeneity is often exacerbated by lattice distortion and local chemical order arising from the interaction among multiple principal elements. As a consequence, dislocation motion faces a rugged atomic and energy landscape, presenting new twists in defect energetics and properties, as well as cocktail strengthening as a new form of solid solution hardening beyond the conventional mechanism. Our systematic account from this standpoint may help answer the continual query in the community as to what is special about these HEAs that makes them different from traditional solid solutions.

There has seen an ongoing rush worldwide in the research on "high-entropy alloys" (HEAs) and their derivatives (e.g., medium-entropy alloys, MEAs) [1-4]. These HEAs are primarily solid solutions, often based on simple face-centered cubic (FCC) [1] or body-centered cubic (BCC) [4] crystal structures. However, unlike conventional solutions that are normally based on a single host metal, the HEAs are composed of multiple principal elements and hence also termed complex concentrated alloys [3]. Papers subscribing to this new paradigm of metallurgy are being published in large numbers each year [2-4], mostly about data collection on new compositions in the previously unexplored middle section of the phase diagram in multicomponent systems. This leaves a glaring hole in the HEA literature: it remains unclear as to what is special that sets these HEAs apart from elemental metals and traditional (relatively dilute) solid solutions. With HEAs, are we gaining materials science insight beyond prior knowledge about conventional metals? For example, broadly speaking the fundamental plastic deformation behavior of HEAs seem similar to those of conventional FCC and BCC metals [5]. There is a pressing need, therefore, for a different way of thinking at another level, to uncover noteworthy differences. This is the purpose of this Viewpoint commentary.

We take an up-to-date look at this issue, and offer a point of view from the angle of dislocation mechanisms that govern mechanical properties. Dislocations are the elementary defects that carry plasticity. Their behaviors are well understood in traditional metals. In what follows, we highlight ten unfamiliar features in multi-principal-element alloys, some of which may help enrich current textbook

understanding of dislocation behavior in general. The ten points we discuss include not only root causes but also foreseeable consequences (to exemplify new behavior). The list is not meant to be exhaustive, but rather a starting point to stimulate future studies.

Concentration inhomogeneity inherent in HEA solutions

To set the stage for our discussion, we begin by emphasizing a salient feature characteristic of HEAs: these solid solutions are highly concentrated by definition, and possess unprecedented degree of compositional inhomogeneity. The traditional solid solutions, based on a host FCC metal for example, can be approximated as a dilute solution: the overwhelming majority of the lattice sites are uniformly occupied by identical atoms of the host metal. In contrast, in a HEA with multiple principal elements, spatial composition variation becomes appreciable and the norm. This concentration undulation can sometimes come from inadequate homogenization of the alloy during heat treatment at elevated temperatures. But it can also simply arise from statistical fluctuation (e.g., scaling with the square root of the concentration which is now rather high). Note that here we are talking about a nominally random solution, one that is commonly regarded as homogeneous.

Even when we leave out such fluctuations in our discussion, there are also other intrinsic factors that exacerbate the inhomogeneity in a HEA. The concentration undulation, including chemical ordering or segregation, can be enhanced by the heat of mixing (ΔH_{mix}) as the driving force. This ΔH_{mix} can come from the atomic size mismatch (such as the case in a Ag-Cu solid solution [6]) and electronic interactions;

the two effects are often intertwined, e.g., the chemical effect from electronic interaction can be accompanied by a change in the apparent atomic size due to electron transfer [7]. In general, there will be local preference (or avoidance) for bonding between/among certain nearest-neighbor species, together with strain relaxation to accommodate lattice distortion due to atomic-size mismatch. In a highly concentrated HEA, these complex interactions often necessitate inhomogeneous distribution of the elements. Even for the pioneering HEA, the single-phase FCC CrMnFeCoNi known as the Cantor alloy [1], atomic-resolution maps [8] reveal that Cr-, Co-, Mn- and Ni-rich atomic columns line up, and Fe columns tend to aggregate, and the Ni-poor region is filled by more Fe and Co atomic columns rather than Cr and Mn. These local groups of atomic columns are suggestive of local chemical order (LCO) as suggested in recent modelling studies [9-12], even though heat treatment was performed under conditions that are projected to produce a random solid solution. Very recently, the predicted LCO was directly verified in a presumably random NiCoCr MEA, using high resolution and energy-filtered transmission electron microscopy [13]. So HEAs/MEAs are at best nominally random and homogeneous. Furthermore, compositional inhomogeneity becomes much more pronounced and easier to detect, when substitution is made with some species considerably different in atomic size and electronegativity. For example, pronounced concentration variation was reported when Mn in the Cantor alloy is replaced by Pd (or 12% Al is added in place of Mn, or 5% W into a NiCoCr MEA) [8]. As shown in Fig. 1, composition profiling in CrFeCoNiPd reveals concentration peaks and valleys with modulations 1 to 3 nanometer in wavelength, rather than flat

concentration profiles measured for a truly random solution. Such a correlated undulation over nanometer length scale [8] (see Fig. 1) obviously contains LCOs [12-16] on short-range (nearest neighbors) and medium-range (< ~2 nm) scale, even though from simple X-ray diffraction the HEA may be perceived as an ideal solution with no chemical order/segregation. In any event, local environment variability reigns at ultrafine length scales in HEAs.

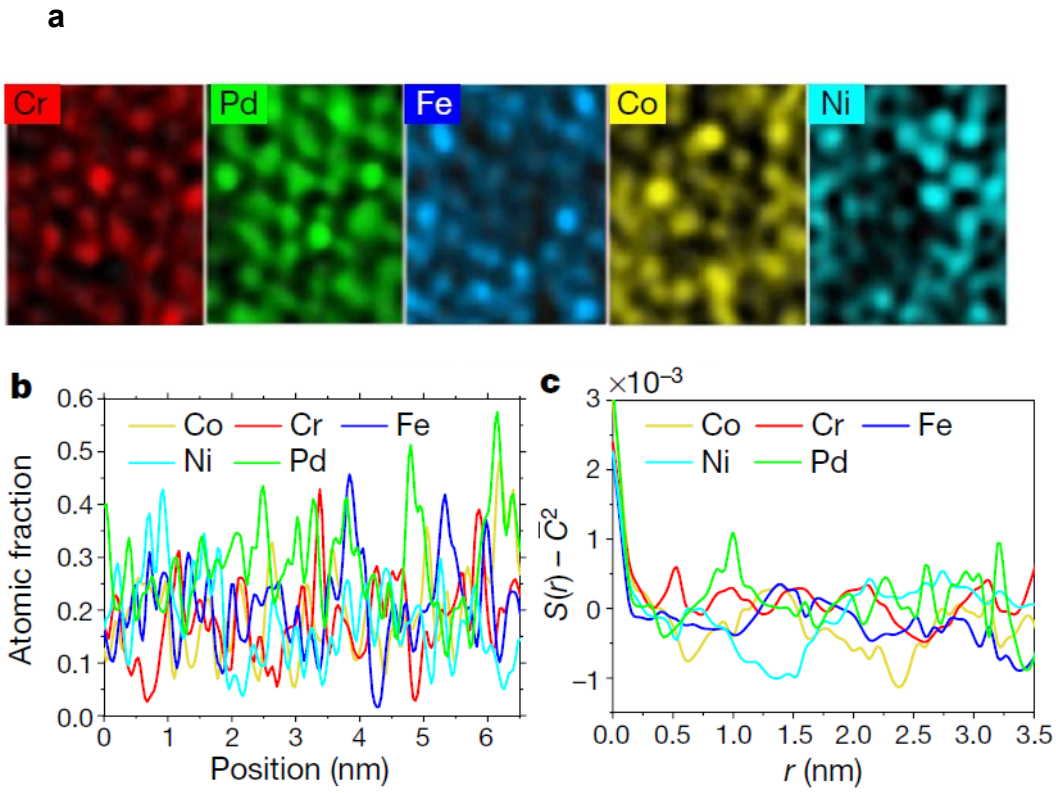

Figure 1 Composition inhomogeneity in a CrFeCoNiPd HEA. **a**, local region showing that atomic columns poor in Ni and Fe concentration are filled by the other constituent elements. **b**, Across a larger area, line profile of atomic fraction of each constituent element shows peaks up in the 40% to 58% range with valleys (~2%) nearby. The concentration waves are indicated by **c**, where pair correlation function $S(r)$ of each of the five elements shows correlation peaks at concentration wavelength (repeat length) r of 1 to 3 nm. Note that the HEA was heat treated at 1150°C for 1 h, a condition not particularly conducive to pronounced development of order/segregation. Yet the element distribution is clearly inhomogeneous with nanometer scale undulations, very different from the flat profiles measured for random solutions. For details, see **Ref. 8**, from which this figure is adapted.

As a result, moving dislocations will confront a choppy sea of inhomogeneities, in lieu of the smooth sailing in an otherwise single-phase crystal. Putting it another way, the inhomogeneity imposes a rugged landscape, in terms of the atomic arrangements on the otherwise uniform lattice and the minimum energy paths for dislocation motion. This ushers in at least the following ten complications that influence the dislocation behavior and mechanical properties.

Excess energy associated with dislocation debris

First off, we start with the **stacking fault** to demonstrate the new twists with defects in HEAs. For conventional metals, when a partial dislocation traverses the crystal lattice, it leaves in its wake a faulted stacking sequence, characterized by a stacking fault energy (SFE). The SFE normally has a fixed magnitude regardless how the alloy is processed. For the dislocation core, the dissociation distance (width of the stacking fault region between the partial dislocations) does not vary much among dislocations with similar orientation. In contrast, a given HEA has a spatially variable SFE. This is because, as shown in the example of Fig. 1, locally the concentration of each species can deviate significantly away from the sample average, and change from one location to another, modifying the local value of the SFE. The large variability of SFE is reflected by the drastic change in dissociation distance, along a given dislocation line, or for dislocations of the same character in a given sample. The first observation in this regard, made for the Cantor alloy, led Smith *et al*. to the notion of "local SFE" [17], i.e., SFE becomes a spatially local property in HEAs. A similar example is displayed in Fig. 2 (left panel) for a Al0.1CoCrFeNi HEA: the dislocation dissociation

distance varies from one segment to another, indicating the fluctuation of SFE from location to location [18]. In modeling, e.g., simulation result [12] displayed in the middle panel in Fig. 2, the local SFE spans a wide range and its distribution depends on how the HEA is processed. The global (sample-average) SFE changes along with the shift of the entire distribution, when the HEA model is annealed at different temperatures that leaves different degree/extent of inhomogeneity in the alloy, from a random solution to one that contains considerable LCOs [11,12].

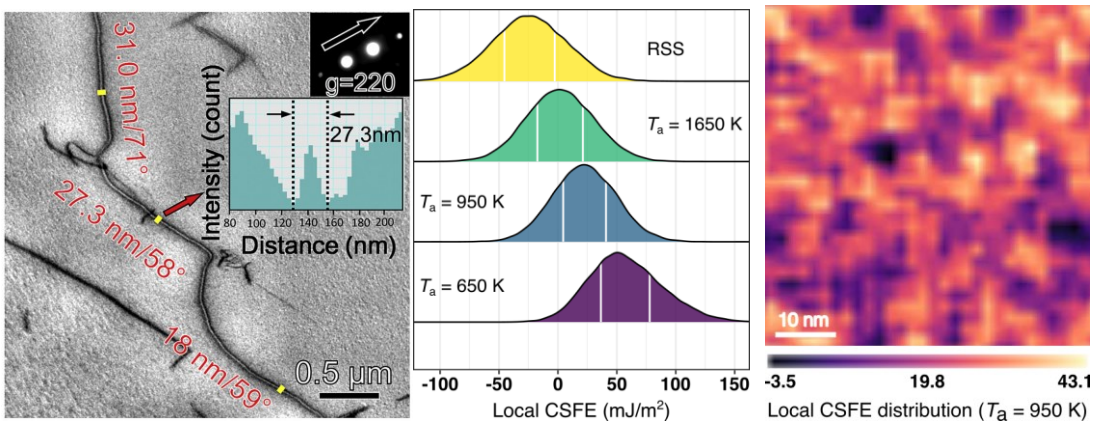

Fig. 2 Variable stack fault energy. **Left panel**: Bright-field transmission electron microscope image of dislocations in the deformed Al0.1CoCrFeNi HEA. The dislocation core width of the dissociated dislocation, as measured using the intensity profile in the lower inset, changes with dislocation character angles along the curved dislocation line [18]. The upper inset diffraction pattern shows the diffraction vector of the two-beam image condition. **Middle panel**: The wide range of local SFE values, in a ternary FCC HEA model mimicking NiCoCr [12]. The distribution profile shifts to the right and the sample-average SFE increases, because of the increasing degree and extent of chemical inhomogeneity inside the alloy, including LCO, with decreasing annealing temperature, *T*a. The data were obtained using molecular dynamics simulations for a FCC MEA model employing embedded-atom-method (EAM) interatomic potentials. The potentials were developed by fitting to NiCoCr data [12], but are only empirical and ignored magnetic effects. As such, the model should be taken more like a general A-B-C ternary system that has chemical interactions and atomic sizes similar to those in concentrated solutions, meant to illustrate the expected trends in HEAs. **Right panel**: Spatial distribution of local SFE value in a sample with *T*a = 950 K, showing obvious variation (see color bar for scale) on nanometer scale.

Second, we look at deformation **twinning**, which involves partial dislocations as well. Recall that in a conventional FCC metal or solution, the widening of an already nucleated deformation twin does not cost additional twin boundary (TB) energy. The layer-by-layer twin thickening is mediated only by the glide of a twinning partial dislocation at the already-present TB, moving the latter by one atomic layer (the next (111) plane) at a time, while the lattice on either side of the TB (in both the twin and the matrix) remains perfect FCC. The situation in a HEA is different. Now the passage of a twinning partial alters the relaxed favorable local environment of the atomic species in the layers next to the TB. In other words, the favorable local environment (sometimes with LCO [12,13]) is ruined (see further discussion in the next paragraph for the passage of a full dislocation). This incurs an energy penalty for every single atomic layer incorporated into the widening twin, anytime the twin thickens [12]. As a consequence, in HEAs nucleated twins would be difficult to grow, which may be related to the experimental observation that all deformation twins stay on nanometer scale without thickening into wide lamellae [19]. The same can also be said for another partial-dislocation mediated process, the FCC to HCP martensitic transformation [12,19], where a partial dislocation runs on every other (111) plane.

Third, continuing along this line of thought, even a **full dislocation** passing through the HEA can leave a higher-energy configuration. This is also different from a conventional metal, where the passage of a full dislocation simply leaves behind a perfect lattice. As discussed above, in HEAs composition inhomogeneities are inherent

from the mixing of concentrated species into energetically favored local atomic environments. These relaxed local arrangements/compositions are altered upon dislocation shear (the example of a twinning partial dislocation slipping at the TB has already been discussed in the preceding paragraph). This can be better appreciated by looking at an extreme case when strong LCOs are developed: chemically segregated domains can form (even though they may be as small as sub-nanometer in dimensions) [12]. In this case, when an extended full dislocation cuts through, the leading partial creates a stacking fault, but the trailing partial eliminating the fault still fails to restore the original local atomic arrangements. Instead, it would create a local, and diffuse, anti-phase boundary, which carries energy [12,13]. This local scenario is akin to an ordered intermetallic, even though the HEA has not established a long-range sublattice.

Dislocation slip modes

The **fourth** point we note here is that the stress-driven dislocation shear on a given slip plane, if repeated a number of times, would homogenize the atomic species towards complete randomness and uniformity. All local regions tend towards the sample-average composition. Across this entire slip plane the favorable configurations (and inhomogeneity-induced hardening associated with these arrangements, to be discussed later) are no longer left, bringing about "glide plane softening" [20] for ensuing moving dislocations. As shown in Fig. 3, dislocation glide in the CoCrFeMnNi HEA is strongly localized on a distinct set of {111}-type planes [21]. Such "planar slip" is also observed very recently in NiCoCr MEA [13]. The planar slip mode is consistent with the expectation known from other types of alloys that short-range order or

segregation/clustering [20, 22], together with the low SFE [23,24] and the high friction stress [25] (both are characteristic of many HEAs), promote **planar dislocation slip**. This may topologically confine and localize plastic strain onto a single or limited slip planes with a dislocation pile-up. Interestingly, the local heterogeneities in HEAs, when intentionally made pronounced to effectively block and divert dislocations, have been found to also give way to cross-slip, transforming the planar slip into wavy slip, i.e., three-dimensional arrangement via multiple slip systems [26]. One example is the CrFeCoNiPd HEA in Fig. 1. As seen in Fig. 3, dislocation motion on the primary slip plane is sluggish, indicative of considerable lattice friction in this HEA [8]. This can be attributed to the trapping effects of pronounced composition inhomogeneity, causing sustained dislocation pile-up. A number of dislocations in the pile-up, and eventually all of them, are eventually forced to cross-slip, Fig. 3 [8]. A similar scenario of slip mode change is also seen in the BCC $(TiZrHfNb)_{98}O_2$ HEA [27]. The authors intentionally introduced 2% oxygen solutes, to form numerous nanometer-scale (O, Zr, Ti)-rich atomic complexes, which arise because these three elements have high chemical affinity conducive to LCOs. Some examples of strong pinning of dislocations due to these ordered oxygen complexes are shown in Fig. 3. Again, the nano-spaced heterogeneities resulted in double cross-slip, such that the dislocation configuration is characterized by wavy slip (Fig. 3), in lieu of the co-planar arrays in the base HEA without oxygen [27]. The implications of such massive cross-slip on strain hardening will be discussed later (the tenth point).

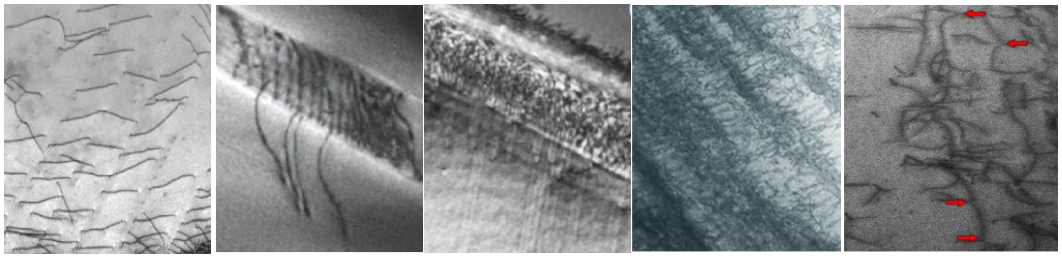

Fig. 3 Dislocation slip modes in HEAs. Displayed in the five panels are, from left to right: Planar slip of ½(110)-type dislocations on{111}-type planes in a CoCrFeMnNi HEA at 273 K [21]; The leading dislocation on a slip plane in the in CrFeCoNiPd HEA experienced high resistance, sustaining a dislocation pile-up [8]; Along this dislocation pile-up cross-slip happens almost everywhere [8]; In a deformed $(TiZrHfNb)_{98}O_2$ HEA, dislocations show dipolar wall with wavy substructures caused by cross-slip of screw dislocations [27]; The ordered oxygen complexes suppress dislocation motion at the pinning points - examples are marked by the red arrows [27]. These five panels are cropped from figures published in the references cited above. See these original references [5,21,27] for the scale bar and more detailed legends and captions about each image.

Many other interesting dislocation features emerge, originating from the defect energetics discussed above. We illustrate a few as examples, starting with our **fifth point**: in the presence of the chemical complexity and abundant inhomogeneity, the moving dislocation line tends to adopt a wavy morphology, different from that in traditional metals. Here we compare the behavior with the familiar Cu in Fig. 4 (bottom panel). In this reference elemental FCC metal the dislocation line is long and straight, and moves easily as a whole from one Peierls valley to the next. In contrast, strong **waviness** down to nanometer scale is seen along the dislocation line in HEAs, as shown for a simulated (NiCoCr) random solution (Fig. 4 top panel). The roughness of the dislocation profile further increases when the solution is made to contain more LCO (Fig. 4 middle panel). That the dislocation line tends to be wavy and rough is not a

surprise, given our discussion above that the spatial distribution of multiple principal elements is prone to fluctuation/undulation in their concentrations, even if the HEA is nominally a random solution. When there are considerable LCOs and local lattice misfits, the spatial undulation will be further accentuated on ultrafine scale. As a result, the dislocation line propagates across a rugged terrain, with some segments resting in energetically favorable fluctuations and others struggling to escape from energetically-unfavorable fluctuations. Under applied stresses, the forward motion of the dislocation is therefore non-uniform and associated with the forward burst (stick-slip) of local segments one at a time. This is important, considering that it is the dislocation movement that controls the strength of an alloy, as discussed next.

Activated process governing dislocation motion and strength

It naturally follows from the picture above that the abundant inhomogeneity in a HEA solution sets up numerous short-distance obstacles tripping and trapping the gliding dislocations [4, 28, 29]. This brings us to our **sixth** point: an unusual "detrapping from the heterogeneity" process is required for dislocation motion. This results in a new slip mode: stick-slip dislocation forward-glide [12], one **nanoscale segment detrapping** (NSD) at a time. This mode and the resultant dislocation morphology is illustrated in Figure 4, where the locally swept areas associated with two typical intermittent NSD events are marked in red [12], for both the random and the LCO-riddled MEA. The corresponding activation volume is only tens of b^3, where b is the Burgers vector, in agreement with those measured in recent experiments on HEAs

[4,30,31]. This contrasts with the much larger activation volume (hundreds up to thousands of b^3, and correspondingly low strain rate sensitivity) in normal metals such as Cu used in Fig. 4 (bottom panel) as a comparison.

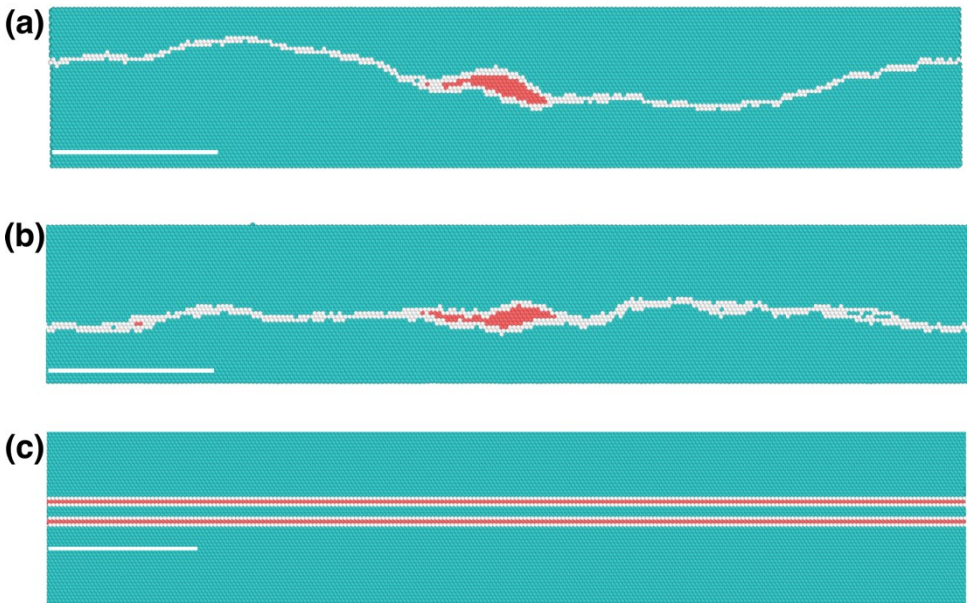

Fig. 4 Dislocation motion via the new nanoscale segment detrapping (NSD) mechanism. The molecular dynamics simulation was performed at 300 K using a NiCoCr model, see Ref. [12]. **(a)** In this model MEA random solution, the leading partial dislocation (white color) adopts a wavy line morphology, and moves upward in a stick-slip manner under a shear stress of 50 MPa. **(b)** The nanoscale roughness increases after the model MEA was aged at 950 K to induce more LCOs, and the dislocation line now moves under 300 MPa. In both panels, highlighted in red are swept areas, between two neighboring snapshots when the dislocation settles down briefly. These red areas correspond to the activation volume involved in the activated detrapping of these nanoscale segments. **(c)** For Cu, in contrast, the screw dislocation lines (dissociated into a leading-trailing partial dislocation pair as colored in white) are straight. The leading partial and trailing partial glide upward together smoothly and uniformly under a low constant shear stress of ~10 MPa. Highlighted in red are the swept areas, showing a much larger activation volume compared with the MEA. All these snapshots were obtained by energy minimization to remove thermal noise. The (white) scale bar is 10 nm.

Seventh, there is an energy cost associated with this stress-assisted, thermally activated NSD as the rate-limiting process. In other words, the new slip mode entails

an elevated resistance to dislocation motion. The NSD entails a sizable **barrier** corresponding to the energy expense (on the order of a fraction of one eV [12]) for dislodging the local dislocation segment trapped in a favorable potential energy fluctuation. This contrasts with traditional FCC metals, where the barriers in between Peierls valleys are very low (~10^{-2} eV [32]) and uniform. This NSD energy barrier is also higher than that associated with conventional solid solution hardening, where each individual solute atom interacts separately with the gliding dislocation through elastic strain fields. The collective solution hardening via NSD naturally necessitates a higher applied stress (see caption of Fig. 4) to move the dislocation. The macroscopic mechanical strength is increased as a result. We coin a new term here, "**cocktail hardening"**, to highlight this new form of hardening in highly concentrated HEA solutions. It is distinctly different from the conventional solid solution hardening mechanism, where the dislocation line interacts through elastic strain field interactions with individual solute atoms that are clearly defined and dilutely separated. This key difference is practically important. Indeed, generally the yield strength of a HEA is higher than the constituent metals and conventional solution hardening. This "cocktail strengthening" in HEAs, and in particular its tunability via tailoring the degree and extent of inhomogeneity, is discussed further as Point #9 later.

The **eighth** point we make extrapolates the insight above to **BCC HEAs** (the examples discussed above are mainly for FCC HEAs). BCC has a non-planar dislocation core structure [33] and no stacking faults, but the inevitable composition fluctuation/inhomogeneity, and therefore the NSD mechanism as well, is also expected

in BCC HEAs. The high yield stress of the BCC HEAs was in fact already linked to the effect of short-distance obstacles on the gliding dislocations by Rao *et al*. [33] in atomistic simulations in a BCC multi-component concentrated solid solution. They attribute the short-distance obstacles in this random solution to the local concentration fluctuations, which impact the dislocation core configuration and dislocation mobility. We project that the double-kink nucleation mechanism, commonly known to be the rate-limiting step for screw dislocation motion in BCC metals, would no longer be the dominant activated process controlling strength for BCC HEAs [34,35]. Rather, screw dislocations adopt a kinked structure as the minimum total energy configuration, followed by lateral kink glide [34]. As such, screw dislocation moves via Peierls-type mechanism much like the FCC case in Fig. 4. The NSD mechanism can take over to become the activation process that governs dislocation mobility, presenting an activation barrier of similar magnitude for both screw and edge dislocations. As a result, edge dislocations would be slowed down to glide with a speed similar to that of screw dislocations. This is very different from conventional BCC metals and alloys, where edge dislocations glide easily over low barriers, orders of magnitude faster than screws. In plastically deforming BCC-HEA samples we should then observe edge and screw dislocations with similar mobility and both present in post mortem images (rather than long screw dislocations alone). The lattice resistance to edge dislocation motion should therefore also contribute to strength, in a capacity comparable to that of screw dislocations. This was implemented in a recent strength model [34], treating both types of dislocations using the same formulation.

Tweaking the dislocation behavior to influence macroscopic properties

We reiterate that a conventional solution based on a host metal would only offer a fixed set of mechanical properties. This holds even for a HEA, if it is a truly homogeneous solution with fixed composition. But a HEA at a given overall composition can be processed to reach enhanced composition inhomogeneity, especially if LCOs and atomic-strain-induced undulations are developed. This opens a "high-entropy" playground, where dislocations in the lattice can be made to face different energy landscapes and thus slip in different modes, leading to new property opportunities. Experimental examples of tuning properties via purposely increased chemical and topological complexity in single-phase HEAs have emerged recently, as discussed next.

Let us start with the capacity to strengthen. We discussed earlier that the resistance to dislocation motion from collective trapping in the "cocktail solution" is fundamentally higher than traditional solid solution hardening, due to a higher energy barrier (see Point #7). The resultant "cocktail strengthening", for which the height of the NSD barrier can be purposely **tuned to tailor strength**, offers a knob to turn as the cocktail mixture can be made inhomogeneous to various degrees. This is the **ninth** point we make here. Specifically, the strength can be intentionally elevated by increasing the misfit volume and lattice distortion [29], or selecting constituent elements to tailor the atomic-level pressure and electron charge transfer [36], or purposely tuning composition to increase inhomogeneity [8,27]. For example, very recent efforts used

substitution of suitable element [8] by replacing Mn in the Cantor alloy with Pd to markedly increase the difference in electronegativity and atomic size, or employed doping to take advantage of the chemical affinity of oxygen with Zr and Ti, to the point of forming numerous nanoscale LCO complexes but not yet second-phase precipitates [27]. Both approaches further increase the inhomogeneity and roughen the dislocation pathways, and have been shown to markedly increase the yield strength (by ~50% [27] up to doubling [8]). An alternative is to use prolonged processing (annealing) to develop short-range-clustering [37], which also raised the yield strength by 25% in FCC NiCoCr [13] and by 76% in a BCC HEA. Using a ternary model mimicking NiCoCr, Li *et al.* [12] systematically conveyed the idea that LCO inhomogeneity is an expected trend unless it is disallowed kinetically, and simulated what could happen to the flow strength if LCO is allowed to reach high levels. Note, however, that such simulations employing the Monte-Carlo scheme artificially reached an extreme scenario of highly concentrated "A-B-C" solution with considerable LCOs; in laboratory experiments it may be practically difficult due to kinetic (temperature/time) constraints to develop heterogeneity with the degree/extent that can significantly change the strength of a HEA. Very recent experimental verification shows that the hardness and strength increase due to annealing induced LCOs in a lab-made NiCoCr MEA is on the order of 25%, together with elevated strain hardening rate [13]. In any case, a HEA cocktail is the way to go to push the limit of solid solution hardening, under the premise that it is still a single-phase solution short of second phase precipitation.

As the **tenth** and final point, by tweaking the HEA make-up the dislocation slip mode can be controlled to improve **strain hardening and ductility**, which is often as important as the strength aspect discussed above. As mentioned earlier, composition inhomogeneity comes in handy when it comes to changing the slip mode from planar slip to wavy slip. An example of enhanced compositional undulation is the case in Fig. 1, which resulted in massive cross-slip (Fig. 3). Another example is the addition of 2% oxygen into a BCC HEA, deploying widespread nano-complexes [27]. We mentioned in the preceding paragraph that both of these two cases led to appreciable elevation of strength [8, 27]. But interestingly and importantly, the tuned chemical inhomogeneity changes the slip mode as well, which helps to improve strain hardening and ductility at room temperature. This is because, as discussed earlier, massive cross-slip was set off by the numerous heterogeneities that block and re-route dislocations. Even though the cross-slip opens pathways for dislocation annihilation (dynamic recovery), it can also increase the dislocation interactions and multiplications, and therefore dislocation storage and the strain hardening capability. Specifically, the double cross-slip [27] and the formation of new Frank–Read sources provide a vehicle to accumulate dislocations [8,27]. In this regard, the switch from planar slip to wavy slip is desirable; it spreads the dislocation activities and delocalizes the plastic flow, which is essential for achieving high tensile ductility [38]. Other types of heterogeneities at various levels can also be added, on top of the inhomogeneity we have discussed here in the single-phase solutions; this is discussed separately elsewhere [39].

Before concluding, we reiterate that HEAs as heavily concentrated solutions will always have composition undulations, even if just due to statistical fluctuations or inadequate homogenization during processing. On top of those, if one quenches in, from high temperature, a truly random and homogeneous solution, given sufficient ageing at intermediate temperatures there would be a tendency towards local chemical order and lattice-strain-induced modulation in these complex alloys, as the multiple principal elements interact in a complex way to lower enthalpy. Taken together, inhomogeneity becomes the norm, and is therefore inherently important: it influences defect energetics and configurations, as we discussed above (especially the first five points). This has consequences in the resistance to dislocation motion and accumulation (see for example the latter five points). Of course, all these ten points intertwine, and sometimes one feature may reflect/dictate multiple and overlapping effects, so we have not separated and designated each as either the cause or the consequence; one should look at the overall picture as a whole. In terms of strengthening, solid solution hardening now takes the form of what we call "cocktail hardening", which pushes up the boundary of strengthening potency: the dislocation in HEAs no longer runs into an individual solute every now and then like in conventional dilute solutions, but instead has to craw forward in a cocktail solution that contains heavily mixed ingredients varying in the make-up on (sub)nanometer scale everywhere.

In terms of previously-unexplored materials science, the ten points we have made above testify that there are indeed novel insights that can be garnered, and new questions that have been opened, regarding the emerging HEAs. In the meantime, our

summary of the “ten differences from traditional solutions" provide a specific set of comparisons to help gauge what may be the low-hanging fruits that the HEA field has to offer. As an outlook, the issues raised and understanding achieved so far may prove helpful, to envision new opportunities for alloy design that may lead to advances beyond conventional alloys.

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