A dual-phase alloy with ultrahigh strength-ductility synergy over a wide temperature range

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High-entropy alloys (HEAs), as an emerging class of materials, have pointed a pathway in developing alloys with interesting property combinations. Although they are not exempted from the strength-ductility trade-off, they present a standing chance in overcoming this challenge. Here, we report results for a precipitation-strengthening strategy, by tuning composition to design a CoNiV-based face-centered cubic/B2 duplex HEA. This alloy sustains ultrahigh gigapascal-level tensile yield strengths and excellent ductility from cryogenic to elevated temperatures. The highest specific yield strength (~150.2 MPa·cm^3/g) among reported ductile HEAs is obtained. The ability of the alloy presented here to sustain this excellent strength-ductility synergy over a wide temperature range is aided by multiple deformation mechanisms i.e., twins, stacking faults, dynamic strain aging, and dynamic recrystallization. Our results open the avenue for designing precipitation-strengthened lightweight HEAs with advanced strength-ductility combinations over a wide service temperature range.

INTRODUCTION

Metallurgists have dreamed to fabricate alloys with ultrahigh tensile strength and excellent ductility (>30% plastic strain) simultaneously at one temperature, even better over a wider temperature range (1). The mechanical response of alloys is remarkably dependent on the service temperature, and it is extremely difficult for a specific alloy to have excellent strength-ductility combinations over a wide temperature range. Advanced structural materials must have a high damage tolerance without premature fracture. The emergence of high-entropy alloys (HEAs) or medium-entropy alloys (MEAs) in recent years presents an unconventional concept and approach as compared with conventional alloy designing (2). These HEAs and MEAs are a new class of materials composed of multiprincipal elements with the possible potential for a wide range of property combinations (2, 3). The face-centered cubic (fcc) CrCoNi family of HEAs is characterized by their low yield strength at room temperature, while the body-centered cubic (bcc) refractory element family of HEAs exhibits insufficient ductility (4–6). Because early HEA research was primarily focused on single-phase microstructures, their transition from laboratory-designed materials to advanced structural materials became a major drawback due to the limited mechanical property. Although the fcc-structured HEAs become stronger when cold via the activation of additional deformation mechanisms such as twinning, phase transformations, and, quite recently, amorphization, obtaining ultrahigh yield strength at ambient temperature has been a challenge (7–10).

Vanadium has been reported to be an important element in M/HEAs, providing strong solid-solution hardening via severe lattice distortion and large misfit volumes in both fcc- and bcc-structured groups (11–13). The fcc-structured CoNiV exhibits the highest yield strength among fcc-based M/HEAs (12) in addition to its high corrosion and hydrogen embrittlement resistance (13). This renders V-contained M/HEAs a target group to be used for potential structural applications. A commonly exploited technique in obtaining M/HEAs with excellent strength-ductility synergy is via controlled Al addition, acting as a bcc phase stabilizer, to an inherently fcc-structured matrix to promote the formation of secondary-phase nanoparticles (i.e., in situ precipitation reinforced alloy) (14, 15). This presents an exciting promise in developing lightweight novel alloys (due to the low density of Al) with ultrastong mechanical properties. While this paper was under review, a recent paper was published, where Al addition to VCoNi resulted in a transition from an fcc phase to an fcc + bcc dual-phase structure (16).

In this study, we use a precipitation-strengthening strategy via simple Al minor alloying to the fcc-structured CoNiV MEA to obtain a dual-phase fcc/B2 structure. Despite the poor plasticity of the intermetallic B2 phase (17), its controlled precipitation from the fcc matrix allows for additional deformation mechanisms to be triggered to sustain high work hardening rates until failure. Hence, gigapascal yield strength levels combined with excellent ductility are obtained from cryogenic deformation temperatures to elevated temperatures by our alloy.

RESULTS

Microstructure

Here, we prepared our alloy by adding 5 atomic % (at %) Al to the equiatomic fcc-structured CoNiV MEA (12), herein, the Al5(CoNiV)95 alloy. The alloy was prepared by arc-melting and suction-casting high-purity metals in a vacuum furnace. The cast sheets were then subjected to a simple thermomechanical process involving homogenization, cold rolling, and annealing. Details are given in Materials and Methods. X-ray diffraction (XRD) results in fig. S1 show the presence of an fcc crystal structure and the ordered bcc (B2) structure, evident from the typical [100] superlattice reflection at 2θ = 30.8° (18). Figure 1 depicts the microstructure of the Al5(CoNiV)95 alloy. The electron backscatter diffraction (EBSD) phase map (Fig. 1A)
The tensile mechanical properties of the studied alloy from 77 K to 1073 K (liquid nitrogen temperature) resulted in even higher yield strength and ultimate tensile strength values of 1.4 and 1.9 GPa, respectively, with an increased fracture elongation of ~38%. A high yield strength of 1.1 GPa was still achieved at an intermediate temperature of 723 K, with a slight drop-off in ductility and an obvious serrated response. The ductile nature of the alloy deformed at 77, 298, and 723 K is confirmed by their dimple fracture morphology (fig. S2). The exceptional ductility exhibited at all three temperatures can be attributed to their excellent strain hardening rates (SHRs) at high yield strengths (see Fig. 2B). Most notably, the SHR at 77 K is higher than that at 298 K. The serrated response at 723 K is also evident by the SHR combined with a very attractive SHR similar to the 77 K, resulting in a high tensile strength at a high temperature. At 1073 K, the alloy showed an enhanced ductility (~74%) and ultimate tensile strength of 0.28 GPa. This tensile response at 1073 K is typical of ultrafine-grained materials that reach their ultimate tensile strengths at very low strains but can sustain very high elongations until failure without strain hardening (19). Crystalline materials usually deform at high temperature via dynamic recrystallization (DRX), which involves grain boundary sliding to accommodate stresses (i.e., superplasticity) (20, 21). Figure 2C illustrates a superior tensile strength and ductility combination of our newly developed alloy at room temperature compared to several established alloys such as transformation-induced plasticity (TRIP) steels, dual-phase steels, and even martensitic steels (14, 22, 23). Our alloy also exhibited the highest ambient-temperature specific yield strength, i.e., yield strength per unit density (~150.2 MPa·cm³/g) compared to other reported ductile M/HEAs (Fig. 2D) (10, 12, 24, 25). The significance of Al minor alloying was seen with the specific yield strength increasing from ~125 MPa·cm³/g in the CoNiV alloy to ~150.2 MPa·cm³/g in our Al₅(CoNiV)₉₅ alloy. This opens the pathway for designing low-density high-strength alloys with potentially improved energy efficiency, which are in demand today. Even at cryogenic temperature (fig. S3), the strength-ductility combination of this alloy is comparable to most cryogenic steels and several reported M/HEAs, thus making the AlCoNiV MEA an attractive candidate for engineering applications.

**Deformation mechanisms**

Before tensile deformation, XRD measurements (fig. S4) showed that the fcc/B2 dual-phase structure of the alloy was maintained at 77, 298, 723, and 1073 K, indicative of no temperature-induced phase transformation. To understand the tensile deformation mechanisms of the studied alloy at the different deformation temperatures, we combined synchrotron high-energy XRD, with EBSD and TEM measurements of the tensile-fractured samples. After 77- and 298-K tensile deformations, the original “fcc + B2” microstructure was sustained as seen by the synchrotron XRD (fig. S4). These preliminary high-energy XRD results echo the possibility of no stress/deformation-induced phase transformations at these deformation temperatures. Figure 3 shows the EBSD and TEM images of the tensile-fractured sample deformed at 77 and 298 K. EBSD phase maps (Fig. 3, A and E) after deformation at cryogenic and room temperatures show the fcc matrix with dispersed B2 precipitates. The corresponding kernel average misorientation (KAM) maps are also presented in Fig. 3 (B and F). The overall average KAM

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**Fig. 1.** Microstructural characterization of AlCoNiV MEA. (A) EBSD phase map. (B) EBSD image quality map (inset shows the average grain size distribution of the fcc phase). (C) Scanning transmission electron microscopy (STEM) image showing the details of the B2 nanoparticle (~200 nm in size) embedded in the fcc matrix. The corresponding SAED patterns of both phases and the EDS maps are also shown. (D) EDS line scan [extracted from the STEM image in (C) from position 1 to position 2] exhibiting variation in elemental concentrations of the B2 phase compared to the fcc matrix. The size of the B2 grains ranges from several hundred nanometers to ~1 µm, while the average grain size of the fcc matrix is about 1.9 µm (see Fig. 1B). The morphology and elemental composition of precipitates were further studied using transmission electron microscopy (TEM) and energy-dispersive x-ray spectroscopy (EDS). The crystal structure of the nanoprecipitates was confirmed to be of B2 type, evident from its [100] superlattice reflection in the selected area electron diffraction (SAED) patterns in Fig. 1C. EDS maps and line scans also reveal that the B2 nanoprecipitate is Al rich and Ni and V poor relative to the fcc matrix, while the Co content remains relatively unchanged.

**Cryogenic to high-temperature tensile properties**

The tensile mechanical properties of the studied alloy from 77 to 1073 K are represented in Fig. 2. Uniaxial tensile engineering stress-strain curves at various temperatures are shown in Fig. 2A. At room temperature, the alloy exhibited ultrahigh yield strength and ultimate tensile strength of 1.1 and 1.5 GPa, respectively, and a large tensile elongation of ~34%. Decreasing temperature to 77 K (liquid nitrogen temperature) resulted in even higher yield strength and ultimate tensile strength values of 1.4 and 1.9 GPa, respectively, while the fully recrystallized region is about 85%. The estimated volume fractions of the fcc and B2 phases are ~94% and ~6%, respectively, that the fcc/B2 dual-phase structure of the alloy was maintained at ~125 MPa·cm³/g in the CoNiV alloy to ~150.2 MPa·cm³/g in our Al₅(CoNiV)₉₅ alloy. This opens the pathway for designing low-density high-strength alloys with potentially improved energy efficiency, which are in demand today. Even at cryogenic temperature (fig. S3), the strength-ductility combination of this alloy is comparable to most cryogenic steels and several reported M/HEAs, thus making the AlCoNiV MEA an attractive candidate for engineering applications.
value at 77 and 298 K is 0.92 and 0.81, respectively, indicative of higher dislocation activities at 77 K. In Fig. 3C, twins were observed in the vicinity of the B2 precipitate after 77-K deformation. The observed nanotwins have an average twin spacing between 20 and 200 nm. The B2 phases are confirmed by their corresponding SAED patterns and EDS maps. It is expected that at larger strains, the primary twins would bundle up and grow in thickness with their thickness increasing from a few tens of nanometers to hundreds of nanometers (26), as shown by the low-angle annular dark-field scanning TEM image in Fig. 3D. These thicker twins are formed when partials are transmitted between multiple stacking faults (SFs), transforming them into twins (see fig. S5) (27). Meanwhile, the 298-K tensile-deformed sample also shows nanolamellar twins close to the B2 particle as shown in Fig. 3G. The twin spacing ranges from tens to hundreds of nanometers (~50 to 400 nm), and the twins are confirmed by their corresponding indexed SAED pattern (see inset of Fig. 3G). The dislocation pinning ability of the B2 phase is also shown in fig. S6. The high-resolution TEM (HRTEM) image in Fig. 3H demonstrates that the twinned fcc matrix is rotated around the common crystallographic [011] axis.

Figure 4 shows the microstructures of the tensile-fractured samples deformed at 723 and 1073 K. A transition from an fcc-B2 duplex structure to an fcc-B2-σ triplex structure was observed after 723-K tensile deformation. This new σ phase is V rich (see fig. S7) and also exhibits a high dislocation pinning effect, seen by the dislocation pileup around the σ nanoparticle in Fig. 4A. The σ phase was not captured by the synchrotron XRD measurement, suggesting its low volume fraction. Figure 4B shows numerous intersected SFs forming nanometer-spaced SF networks with an average spacing between 23 and 60 nm in the fcc matrix. These SFs are also captured in the HRTEM image (Fig. 4C), and the corresponding fast Fourier transform confirms that no nanotwins or hexagonal close-packed phase is formed in the 723-K deformed sample. The observation of high-density dislocations in the SF networks likely contributes to the superior strain hardening at high strength (>1 GPa) (i.e., SF-induced plasticity) at 723 K via two contributions: (i) dynamic Hall-Petch effect with the SF networks pinning mobile dislocations and (ii) formation of Lomer-Cottrell (L-C) locks due to mutual interactions between SFs (28). The microstructure after 1073-K deformation was composed of equiaxed fcc grains coupled with δ and B2 grains as shown in Fig. 4D. This newly formed δ phase was captured by the HE (High-energy)–XRD in fig. S4 and shows a similar compositional variation compared to the fcc matrix (Fig. 4E). It has been reported that for the CoNiV-based ternary systems, the BaPb3-type δ phase exists during heat treatment at temperatures between 973 and 1273 K (29).

**DISCUSSION**

In the precipitation strategy approach used in M/HEAs, a near-equiatomic matrix usually results in the formation of brittle phases (i.e., B2 and σ), while a nonequiatomic matrix leads to the formation of ductile L12 precipitates (30). Of course, both precipitation strategies require careful control to compensate for the mutual exclusivity between strength and ductility. In our design strategy, the addition of Al to the equiatomic matrix stimulated the formation of B2 nano precipitates. The formation of this intermetallic B2 phase is assisted by the mixing enthalpy difference between sub-binary compounds.
in the multicomponent alloy, most notably, the extremely negative enthalpy between Ni and Al (31). These B2 particles contribute an added precipitation-strengthening effect to the yield strength.

The dual-phase soft fcc–hard B2 structure of our studied AlCoNiV alloy causes an inhomogeneous deformation during straining. The soft fcc matrix deforms first, but the presence of hard B2 precipitates exerts a deformation constrain that is overcome via the pileup of geometrically necessary dislocation, consequently generating back stresses for the overall increase in yield strength (14). Because of the high shear stresses exerted by the hard B2 phase, the interface between the B2 phase and the fcc matrix serves as a stress concentration site to allow for easy crack initiation and propagation. The appearance of twins close to B2 precipitates here plays a crucial role in offsetting this problem, via the mitigation and accommodation of high fcc-B2 interfacial stresses to prolong strain hardening due to limited plasticity of B2 precipitates (32). The twin boundaries reduce the dislocation mean free path, thus sustaining strain hardening to delay the onset of fracture. This phenomenon is referred to as the “dynamic Hall-Petch” effect (33). The increase in yield strength at a lower temperature (77 K) means that critical twinning stress is easily reached because of the relative temperature insensitivity of twinning stress, and thus, nanotwinning occurs at relatively lower strains (34). This improves the SHR and delays failure at 77 K, even in the presence of the B2 phase that usually causes strong embrittlement problems at cryogenic temperature. A nanoscale toughening mechanism where twin boundaries are transformed into impenetrable dislocation walls upon interaction with necklace-like dislocations emitted from crack tips, resulting in an enhanced resistance to crack propagation via crack bridging and arresting, has been observed in Cu thin films (35) and CoCrFeMnNi HEA (36). The depleted amount of Al in the fcc matrix suggests a decrease in the stacking fault energy (SFE) of the fcc matrix, on the basis of the inverse relationship between Al content and SFE in Al-based M/HEAs (32, 37). The formation of the Al-rich B2 phase in studied AlCoNiV MEA triggers the transition from dislocation slip-induced plasticity in CoNiV MEA (12) to twinning-induced plasticity (TWIP) in studied AlCoNiV MEA at 298 K. There have been reports of decreased SFE with decreasing temperature, leading to a transition from TWIP to TRIP (38, 39). However, because we have not observed any phase transition at 77 K but rather enhanced twinning activities, this suggests that although the SFE is lowered at 77 K, the critical stress to trigger stress-induced martensitic transformation is still not reached. Therefore, the dynamic twin-dislocation and twin-twin interactions are vital to the exceptional strain hardening behavior at both 77 and 298 K, similar to other low SFE M/HEAs (26, 32, 40).

As already mentioned, our alloy shows an ultrahigh yield strength (~1.1 GPa) at an intermediate temperature of 723 K. The formation of the newly formed V-rich $\sigma$ nanoparticles provides an extra hardening mechanism to keep the yield strength at gigapascal levels. The SHR at this temperature is in the same order as the SHR at 77 K, which is abnormal, because SHR usually decreases with increasing temperature because of high-temperature dynamic recovery (19). On the basis of the observed flow serrations at 723 K, it is reasonable to assume dynamic strain aging (DSA) as a strengthening mechanism behind the enhanced SHR (41). In M/HEAs alloys, DSA typically proceeds via the arrest of dislocations by solute atoms (typically...
dislocations during the plastic deformation. These SF networks low enough to assist the dissociation of full dislocations into partial mediate temperature indicate that the SFE of the fcc matrix is still

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solute Al atoms), SFs, or twins (42). During plastic deformation, once dislocations are impinged by an obstacle, solute atoms diffuse to the vicinity of these pinned dislocations, which creates an energy barrier for the dislocations to overcome. To overcome this energy barrier, the frictional stress of the dislocations increases, after which the friction stress of the dislocations abruptly decreases once they overcome the barrier, hence the serrated deformation (43). The SF networks (Fig. 4B) acting as L-C locks actively pin mobile dislocations and contributes to DSA. The SF networks formed at this intermediate temperature indicate that the SFE of the fcc matrix is still low enough to assist the dissociation of full dislocations into partial dislocations during the plastic deformation. These SF networks are beneficial for strain hardening because they obstruct mobile
dislocations to delay the onset of necking and eventual fracture. This SF-mediated deformation mechanism has been observed in the Mg alloy (44), Al0.1CoCrFeNi (28), and Ni0.3Co0.9Fe1.3Cr1.5Al6Ti6 (45) HEAs at ambient temperature, which makes our finding fascinating.

At 1073 K, the alloy shows typical superplastic elongation where strain softening occurs and deformation is mostly accompanied by grain boundary sliding (19). The variation in grain sizes and the decreased dislocation density compared to the undeformed material and other deformed samples at different temperatures signify the occurrence of DRX to aid superplastic deformation. However, our alloy shows insufficient superplastic elongation (~74%) compared to its microduplex χ/σ CoNiV alloy with 330 to 450% (21). The lower superplastic elongations exhibited by our sample are attributed to a deformation-enhanced grain growth, which could be overcome by using high strain rates to enhance DRX (21).

We further demonstrate that our design strategy can be adapted to achieve superior strength-ductility combinations over a wide temperature range by slightly increasing the Al content to obtain another fcc/B2 duplex Al5(CoNiV)93 alloy. Here, the increased Al content was expected to increase the B2 phase volume fraction. From the tensile tests performed at the different studied temperatures, yield strengths greater than 1 GPa were again sustained from 77 to 723 K (fig. S9). The fracture elongation decreased from ~31% at ambient temperature to ~23% at cryogenic temperature, while an extremely high superplastic elongation (~280%) was seen at 1073 K, contrary to that of Al5(CoNiV)95. Again, 723-K tensile deformation exhibited a serrated response. As anticipated, the volume fraction of the B2 phase increased (~21%), while the recrystallized ultrafine fcc grains were randomly oriented (fig. S9) This increased density of B2 precipitates means a reduced spacing between the fcc matrix canals, which will further increase the critical stress for twin nucleation to promote strain hardening (46). Hence, the ductile-brittle transition is observed when the temperature is lowered from 298 to 77 K.

In conclusion, the gradual shift from phase stability–centered to optimized–property studies in M/HEA research has intensified over the past decade. This has opened the door to using existing conventional strengthening principles such as gradient structures, secondary nanoparticles, short-range order structures, and phase metastability in M/HEA design to develop potential cutting-edge alloys for structural applications. We use a simple precipitation–strengthening approach here to develop the Al5(CoNiV)93 MEA with sustained excellent strength-ductility synergy from cryogenic to intermediate temperatures and the superplastic behavior at high temperatures. We gain insights into the multiple toughening mechanisms at play at different deformation temperatures. We suggest that the chemical complexity offered by M/HEAs can be combined with existing conventional strengthening approaches to develop high-performance alloys to justify the relatively high costs, which will come with M/HEA industrial manufacture.

**MATERIALS AND METHODS**

**Alloy preparation**

Al5(CoNiV)93 (at %) alloy ingots were prepared by arc-melting high-purity (>99.9 at %) Al, Co, Ni, and V metals. To ensure chemical homogeneity, the ingots were flipped and remelted at least five times and then suction-cast into a Cu mold with dimensions 80 mm by 10 mm by 2.5 mm. The alloy sheets were vacuum-sealed in...
quartz tubes and homogenized at 1373 K for 8 hours followed by water quenching. The homogenized sheets were cold-rolled (65% reduction in thickness), followed by annealing at 1173 K for 1 hour and water quenching. Al₅(CoNiV)₉₅ alloy samples prepared by the same treatment were also studied. Al₇(CoNiV)₉₃ exhibits an fcc/B2 dual-phase microstructure confirmed by XRD and EBSD measurements shown in fig. S9. On the basis of the Archimedes principle, bulk density (ρ) measurements of the prepared alloys were performed using a Mettler Toledo XS105 microbalance with a 0.01-mg sensitivity. The densities were found to be ~7.72 and ~7.62 g/cm³ for Al₅(CoNiV)₉₅ and Al₇(CoNiV)₉₃, respectively.

**Microstructural characterization**

Phase identification of the as-prepared samples was performed by XRD using Cu-Kα radiation, while the phases of the tensile-deformed samples were identified via high-energy synchrotron radiation–based XRD at the Shanghai Synchrotron Radiation Facility. Two-dimensional (2D) diffraction patterns were collected at the 13-W beamline using a wavelength of about 0.1788 Å. The 2D diffraction patterns were then integrated using the FIT2D software (www.esrf.eu/computing/scientific/FIT2D/) to obtain 1D XRD patterns.

Thin TEM foils were obtained using a focused ion beam (FEI Strata 400S Dual-Beam) from the as-prepared samples and different regions of the fractured tensile specimens. A JEM Grand ARM300F transmission electron microscope operating at 300 kV was used to analyze the microstructures of the TEM foils. Microstructure and fracture morphologies of the samples were studied using a Zeiss Supra 55 scanning electron microscope equipped with a TexSEM Labs - Orientation Imaging Microscopy (TSL-OIM) EBSD system.

**Tensile testing**

Flat dog bone–shaped tensile specimens were cut from the annealed samples by electrical discharge machining. The tensile samples had a gauge length, a width, and a thickness of 8.5, 2.2, and 0.85 mm, respectively. Uniaxial tensile tests at room temperature (298 K), liquid-nitrogen temperature (77 K), and high temperatures (723 and 1073 K) were conducted using a universal testing machine (CMT5205 SANS) at a fixed strain rate of 10⁻³ s⁻¹. A thermocouple was used to observe the stability of the testing temperature, and an extensometer was attached to estimate the strain. Each tensile test was repeated three times to ensure data reproducibility.

**Yield strength contributions**

The room temperature yield strength (σₚ) of an alloy is estimated on the basis of the individual strengthening contributions, expressed as

\[ \sigma_y = \sigma_o + \sigma_{gb} + \sigma_d + \sigma_p + \sigma_{ods} \]  

(1)

where σₒ is the intrinsic lattice friction stress, σ_gb is the grain boundary–strengthening contribution from grain, σ_d is the dislocation-strengthening contribution, and σ_p is the precipitation-strengthening contribution. Here, the intrinsic lattice friction stress of CoNiV is adopted, so σₒ = 383 MPa (12). Because of the absence of any dispersed oxides, σ_{ods} = 0 MPa.

The grain boundary–strengthening contribution, σ_gb, is estimated on the basis of the Hall-Petch relation: \( \sigma_{gb} = k \cdot d^{-0.5} \). However, because of the contributions from fully recrystallized and nonrecrystallized grains, \( \sigma_{gb} \) is calculated by

\[ \sigma_{gb} = f_{RX} \cdot k \cdot d^{-0.5} \]  

(2)

where \( f_{RX} \) is the volume fraction of the recrystallized grains, \( k \) is the Hall-Petch coefficient, and \( d \) is the average grain size of the matrix grains. Here, \( f_{RX} = 85\% \), \( k = 864 \text{ MPa·μm}^{0.5} \) (12), and \( d = 1.87 \text{ μm} \). The grain boundary strengthening is equal to 537 MPa.

The strengthening contribution from dislocations is calculated according to the Taylor hardening law (47) by

\[ \sigma_d = f_{NRX} \cdot \alpha \cdot M \cdot G \cdot b \cdot \alpha \]  

(3)

where \( M \) is the Taylor factor of fcc materials (3.06), \( f_{NRX} \) is the volume fraction of nonrecrystallized grains (15%), \( G \) is the shear modulus (72 GPa), \( \alpha \) is a constant (0.2), \( b \) is the magnitude of the Burger’s vector, and \( \rho \) is the dislocation density. Here, \( b = (\sqrt{2}/2) \cdot a \), so using the calculated lattice constant \( a_{fcc} = 0.3606 \text{ nm} \) from XRD, \( b = 0.255 \text{ nm} \). The dislocation density \( \rho \) is expressed as \( \rho = 2\sqrt{3}\pi/\text{db} \) (48), where \( d \) is the microstrain estimated from the XRD according to the Williamson-Hall method (ε = 0.249%) (49), \( d \) is the average grain size, and \( b \) is the magnitude of the Burger’s vector. Hence, \( \rho = 1.8089 \times 10^{13} \text{ m}^{-2} \). Therefore, the dislocation-strengthening contribution, \( \sigma_d = 7.17 \text{ MPa} \).

On the basis of these, we can also estimate the precipitation-strengthening contribution as

\[ \sigma_p = \sigma_y - \sigma_{gb} - \sigma_d - \sigma_o \]  

(4)

Using the calculated σₚ from the tensile test (1175 MPa), the estimated \( \sigma_p = \sim247.8 \text{ MPa} \). This shows that the grain boundary strengthening (45.7%) offers the largest contribution to the yield strength of Al₅(CoNiV)₉₅, followed by the intrinsic lattice friction stress (32.6%), then the precipitation strengthening (21.1%), and lastly, the dislocation strengthening contribution with a ~0.61%.

Therefore, we see that by carefully introducing precipitates, an extra ~250 MPa is added to the yield strength while the ductility is still relatively high. Of course, increasing the precipitate amount in the matrix will inevitably further increase the yield strength but at the cost of ductility, because of the strength-ductility trade-off.

**SUPPLEMENTARY MATERIALS**

Supplementary material for this article is available at http://advances.sciencemag.org/cgi/content/full/7/34/eabi4404/DC1

**REFERENCES AND NOTES**

1. R. O. Ritchie, *The conflicts between strength and toughness*. *Nat. Mater*. 10, 817–822 (2011).
2. J. W. Yeh, S. K. Chen, S. J. Lin, T. S. Chin, T. T. Shun, C. H. Tsau, S. Y. Chang, Nanostructured high-entropy alloys with multiple principal elements: Novel alloy design concepts and outcomes. *Adv. Eng. Mater*. 6, 299–303 (2004).
3. B. Canter, I. T. H. Chang, P. Knight, A. J. B. Vincent, Microstructural development in equiatomic multicomponent alloys. *Mater. Sci. Eng. A* 375–377, 213–218 (2004).
4. E. P. George, D. Raabe, R. O. Ritchie, High-entropy alloys. *Nat. Rev. Mater*. 4, 515–534 (2019).
5. Y. Zhang, T. T. Zuo, Z. Tang, M. C. Gao, K. A. Dahmen, P. K. Liu, Z. P. Lu, Microstructures and properties of high-entropy alloys. *Prog. Mater. Sci*. 61, 1–93 (2014).
6. D. B. Miracle, O. N. Senkov, A critical review of high entropy alloys and related concepts. *Acta Mater*. 122, 488–511 (2017).
7. B. Gludovatz, A. Hohenwarter, D. Catoo, E. H. Chang, E. P. George, R. O. Ritchie, A fracture-resistant high-entropy alloy for cryogenic applications. *Science* 345, 1153–1158 (2014).
8. S. Zhao, Z. Li, C. Zhu, W. Yang, Z. Zhang, D. E. J. Armstrong, P. S. Grant, R. O. Ritchie, M. A. Meyers, Amorphization in extreme deformation of the CrMnFeCoNi high-entropy alloy. *Sci. Adv*. 7, eabb3108 (2021).
19. Exceptional damage-tolerance of fcc and bcc high-entropy alloys. Acta Mater. 188, 486–491 (2020).

24. D. P. J. Blackett, Y. S. Chen, X. X. Z. He, P. D. Raabe, and R. K. N. performed tensile tests. Q. S. Sun, J. Z. Jiang, and K. Raj, Ultrastrong high-entropy alloys for lightweight compositionally complex steels via dual-nanoprecipitation. Acta Mater. 194, 1743–1753 (1973).

35. S. W. Kim, X. Li, L. Gao, S. Kumar, In situ observations of crack arrest and bridging by nanoscale twins in copper thin films. Acta Mater. 60, 2959–2972 (2012).

Z. Zhang, M. M. Mao, J. Wang, B. Gloduzotto, Z. Zhang, S. X. Mao, E. P. George, Q. Yu, R. O. Ritchie, Nanoscale origins of the damage tolerance of the high-entropy alloy CoCrFeNiAl. Nat. Commun. 6, 10143 (2015).

D. Choudhuri, M. Komarasamy, V. Aheg, R. S. Mishra, Investigation of plastic deformation modes in Al25CoFeCrNi high entropy alloy. Mater. Chem. Phys. 217, 308–314 (2018).

D. G. Kim, Y. H. Jo, Y. Zhang, W. M. Choi, H. S. Kim, B.-J. Lee, S. S. Sohn, S. Lee, Ultrastrong duplex high-entropy alloy with 2 GPa cryogenic strength enabled by an accelerated martensitic transformation. Sci. Mater. 171, 67–72 (2019).

J. Liu, X. Guo, Q. Lin, Z. He, X. An, L. Li, P. K. Liaw, L. Xiao, L. Yu, J. Lin, L. Xie, J. Ren, Y. Zhang, Excellent ductility and serration feature of metastable CoCrFeNi high-entropy alloy at extremely low temperatures. Sci. China Mater. 62, 853–863 (2019).

Y. H. Jo, S. Jung, W. M. Choi, S. S. Sohn, H. S. Kim, B. J. Lee, N. J. Kim, S. Lee, Cryogenic strength improvement by utilizing room-temperature deformation thinning in a partially recrystallized VCrMnFeCoNi high-entropy alloy. Nat. Commun. 8, 15719 (2017).

S. Y. Chen, L. Wang, W. D. Li, Y. Tong, K. K. Tseng, J. W. Yeh, Y. Ren, W. Guo, J. D. Poplawsky, P. K. Liaw, Peerless barrier characteristic and anomalous strain hardening provoked by dynamic strain-aging strengthening in a body-centered-cubic high-entropy alloy. Mater. Res. Lett. 7, 475–481 (2019).

J. Brecht, S. Y. Chen, X. Xie, Y. Ren, J. W. Qiao, P. K. Liaw, S. J. Zinkle, Towards a greater understanding of serrated flows in an Al-containing high-entropy-based alloy. Int. J. Plast. 115, 71–92 (2019).

H. Y. Yasuda, K. Shigeno, T. Nagase, Dynamic strain aging of Al25CoFeCrNi high entropy alloy single crystals. Sci. Mater. 108, 80–83 (2015).

W. W. Jian, G. M. Cheng, W. Z. Xu, H. Yuan, M. H. Tsai, Q. D. Wang, C. C. Koch, Y. T. Zhu, S. N. Mathaudhu, Ultrastrong Mg alloy via nano-spaced stacking faults. Mater. Res. Lett. 1, 61–66 (2013).

T. Yang, Y. L. Zhao, J. H. Luan, B. Han, J. Wei, J. J. Liu, K. T. Liu, Nanoparticles-strengthened high-entropy alloys for cryogenic applications showing an exceptional strength-ductility synergy. Sci. Mater. 164, 30–35 (2019).

L. E. Murr, A. Ayala, C.-S. Niou, Microbands and shear-related microstructural phenomena associated with impact craters in 6061-T6 aluminum. Mater. Sci. Eng. A 216, 69–79 (1996).

Z. Wang, W. Lu, H. Zhao, C. H. Liebscher, J. He, D. Ponge, D. Raabe, Z. Li, Ultrastrong lightweight compositionally complex steels via dual-nanoprecipitation. Sci. Adv. 6, eaba9543 (2020).

A. W. Zhu, E. A. Starke Jr., Strengthening effect of unshearable particles of finite size: A computer experimental study. Acta Mater. 47, 3263–3269 (1999).

A. Khorasand Zad, W. H. Abd Majid, M. E. Abrishami, R. Yousefi, X-ray analysis of ZnO nanoparticles by Williamson–Hall and size–strain plot methods. Solid State Sci. 13, 251–256 (2011).

T. Yang, Y. L. Zhao, Y. Tong, Z. B. Jiao, J. Wei, X. Cai, X. D. Han, D. Chen, A. Hu, J. J. Kai, K. Liu, Y. Liu, C. T. Liu, Multicomponent intermetallic nanoparticles and superlative mechanical behaviors of complex alloys. Science 362, 933–937 (2018).

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A dual-phase alloy with ultrahigh strength-ductility synergy over a wide temperature range

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