



# Tensile yield strength of a single bulk $\text{Al}_{0.3}\text{CoCrFeNi}$ high entropy alloy can be tuned from 160 MPa to 1800 MPa

Bharat Gwalani<sup>a,b</sup>, Stephane Gorsse<sup>c,d</sup>, Deep Choudhuri<sup>a,b</sup>, Yufeng Zheng<sup>e</sup>,  
Rajiv S. Mishra<sup>a,b</sup>, Rajarshi Banerjee<sup>a,b,\*</sup>

<sup>a</sup> Advanced Materials and Manufacturing Processes Institute, University of North Texas Denton, TX-76207, USA

<sup>b</sup> Department of Materials Science and Engineering, University of North Texas Denton, TX-76207, USA

<sup>c</sup> CNRS, ICMCB, UMR 5026, 33600 Pessac, France

<sup>d</sup> Bordeaux INP, ENSCBP, 33600 Pessac, France

<sup>e</sup> Department of Materials Science and Engineering, The Ohio State University, Columbus, OH 4310, USA

## ARTICLE INFO

### Article history:

Received 21 September 2018

Received in revised form 13 October 2018

Accepted 17 October 2018

Available online xxxx

### Keywords:

High entropy alloys

Characterization

Mechanical properties

Complex concentrated alloys

Clusters

## ABSTRACT

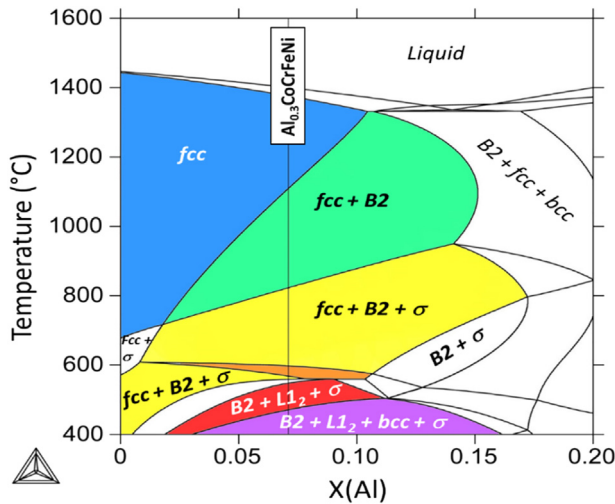
While there have been multiple recent reports in the literature of exceptional combinations of yield strength and ductility in high entropy alloys, there have been no reports discussing the extraordinary tunability of the mechanical properties in the same alloy in these systems. This paper shows that the tensile yield-strength of a single  $\text{Al}_{0.3}\text{CoCrFeNi}$  high entropy alloy (or complex-concentrated alloy), can be enhanced from 160 MPa to over 1800 MPa (1.8 GPa), a 1025% increase, via microstructural engineering enabled by thermo-mechanical processing of the bulk alloy. Such strength variations for the same composition are unprecedented in any other class of alloys.

© 2018 Acta Materialia Inc. Published by Elsevier Ltd. All rights reserved.

The investigation of high entropy alloys (HEAs) began with the pioneering work of Cantor [1] and Yeh [2] in 2004. They simultaneously showed that a fully austenitic microstructure, consisting of single face-centered cubic (fcc) phase, was obtained in equiatomic mixtures of multiple elements, having different crystal structures. These pioneering efforts created the groundwork for a new concept in alloy design by advocating that severe lattice distortion in concentrated disordered solid solutions can promote high strength and retain ductility, without the deleterious effect of embrittling intermetallic or other compound phases in structural alloys. Subsequently, a substantial effort has been directed towards finding new equiatomic combinations of elements to promote the formation of *fcc*, *bcc* or *hcp* single-phase-disordered solid solutions, presumably stabilized by the configurational entropy of mixing. However, despite the exceptional fracture toughness properties exhibited by the  $\text{CoCrFeNiMn}$  alloy [3], also referred to as Cantor alloy, the ability of the HEA concept to revolutionize alloy design, and lead to exceptional mechanical performance, seems limited since it involves only one principal strengthening mechanism, i.e. solid solution strengthening.

Recently, the HEA community returned to the basics of physical metallurgy, i.e. microstructural engineering by altering alloy composition (chemistry) and thermomechanical processing (simultaneous application of temperature and stresses) to enhance the balance of mechanical properties in HEAs [4–7]. The result was the adoption of two well-known microstructure templates, found in dual-phase (fcc austenite + body centered tetragonal or hexagonal martensite) TRIP steels [8], and age-hardenable Ni alloys (fcc austenite + ordered intermetallic  $\text{L}_{12}$  or  $\gamma'$  precipitates) [9], to these new alloys. These studies elucidated the applicability and efficacy of traditional metallurgical approaches for the development of the second generation of multi-phase HEAs, also referred to as complex concentrated alloys (CCAs) [4–6] or multi-principal element alloys (MPEAs), with often unequal proportion of principal elements and microstructures comprising multiple phases. However, the composition of these “HEAs”, e.g.  $\text{Fe}_{50}\text{Mn}_{30}\text{Co}_{10}\text{Cr}_{10}$  HEA [8] become comparable to high-Mn steels [10], and somewhat deviate from the notion of HEAs, because their compositions shift towards the edges of the composition space where conventional alloys already lie. We emphasize that the HEA design concept mandates searching for alloy compositions that are located in the interior regions of multi-component phase diagrams [6]. In the present study, we highlight a different approach that takes advantage of the compositional complexity of these concentrated alloys. This complexity leads to an exceptionally vast array of possible microstructures and phases that can be generated,

\* Corresponding author at: Advanced Materials and Manufacturing Processes Institute, University of North Texas, Denton, TX 76207, USA.  
E-mail address: [Rajarshi.Banerjee@unt.edu](mailto:Rajarshi.Banerjee@unt.edu) (R. Banerjee).

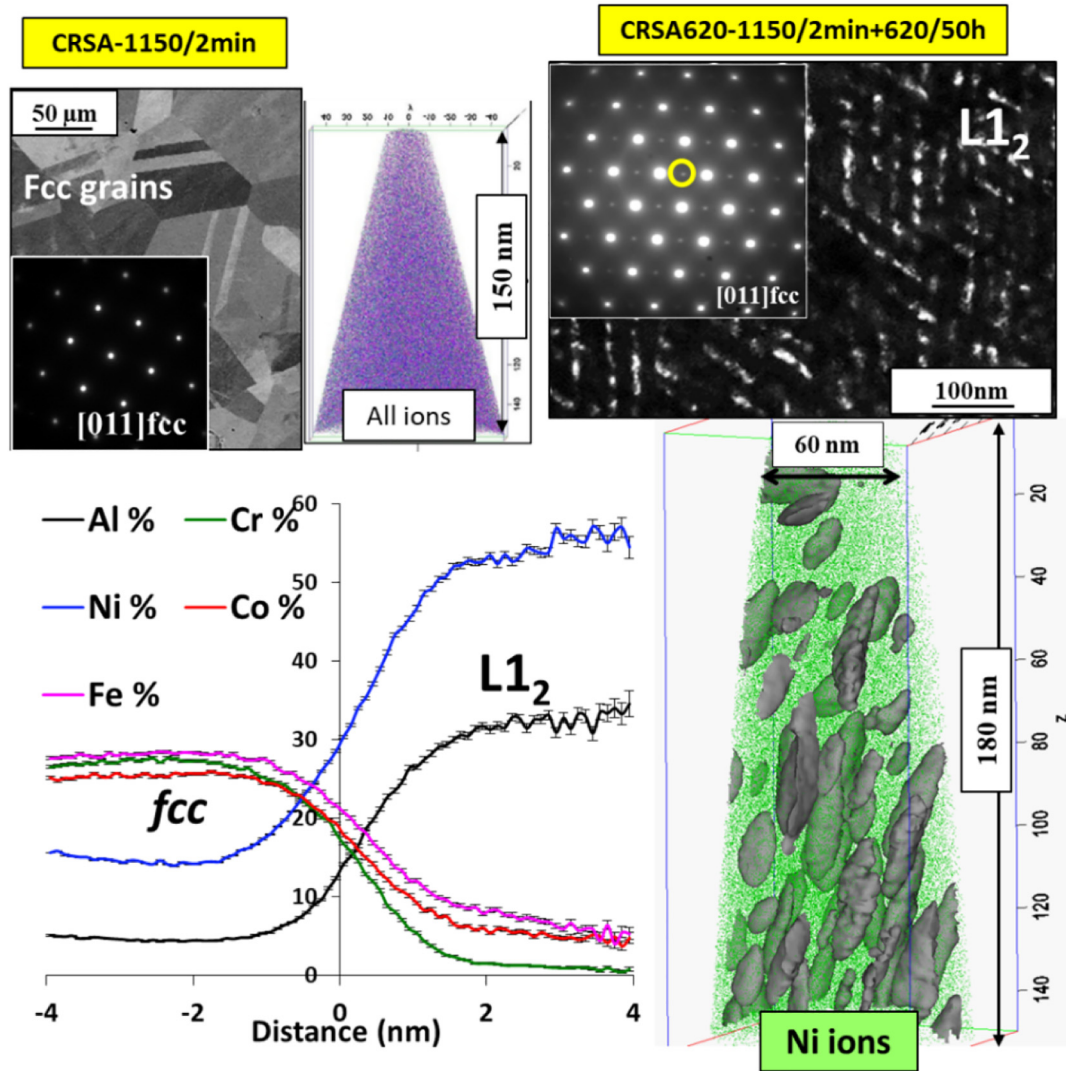


**Fig. 1.** Isoleth for  $\text{Al}_x\text{CoCrFeNi}$ . This section of the Al-Co-Cr-Fe-Ni quinary system was calculated using the TCHEA2 thermodynamic description from ThermoCalc. It shows possible phases as a function of Al content and temperature.

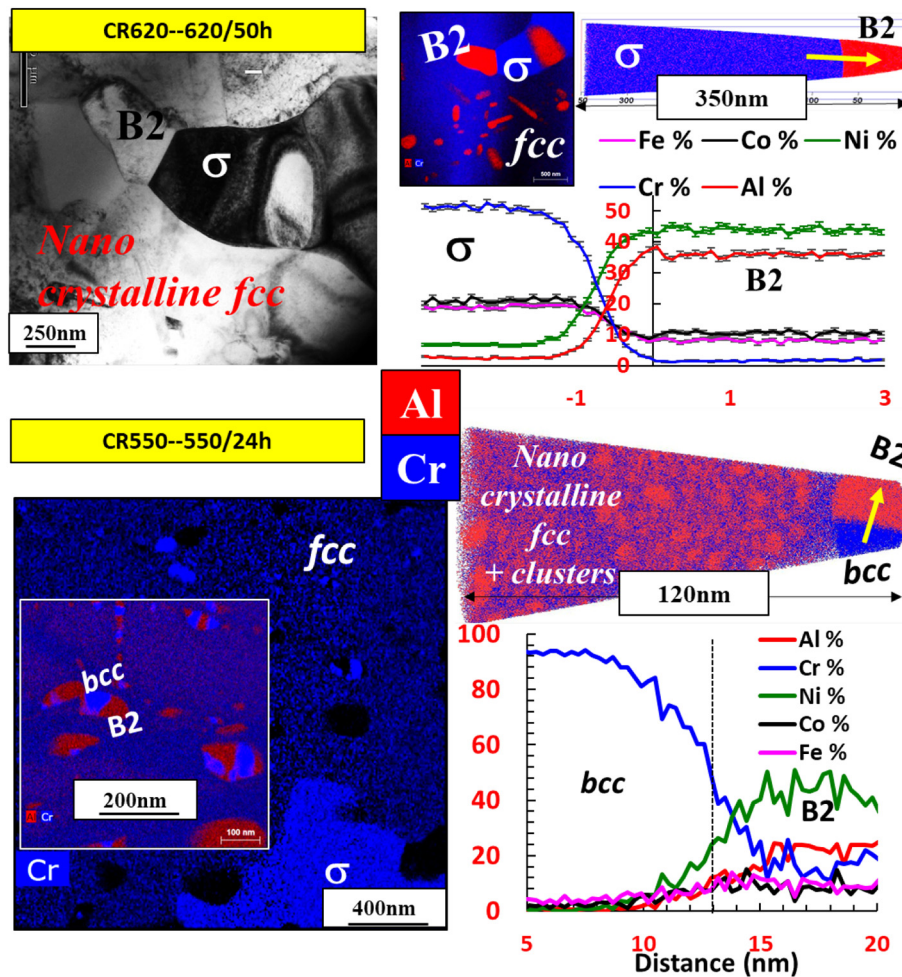
in a selected phase field of operation, for the same alloy composition. This distinguishing feature of HEAs, which has not been explored yet, opens up the avenue for materials with properties that can be tuned over a broad range of values without changing the composition.

To explore the concept of highly tunable HEA, we have selected the  $\text{Al}_{0.3}\text{CoCrFeNi}$  composition, as a candidate alloy, because it encompasses a remarkably large number of phase fields as a function of temperature. For example, with decreasing temperature, this HEA composition traverses from a single *fcc* phase field to *fcc* + B2 (ordered bcc), *fcc* + B2 + sigma ( $\sigma$ ), B2 +  $\text{L}_{12}$  +  $\sigma$ , and finally at lower temperatures, a B2 +  $\text{L}_{12}$  + bcc +  $\sigma$  phase field (Fig. 1). Such a unique richness of the complex multi-phase landscape offers the designer more options to tailor the phase combinations and the microstructural length scales, with one or more intermetallic/ordered phases, employing different thermo-mechanical treatments. Accordingly, such a vast array of multi-phase microstructures offers a unique opportunity to tune the properties over a staggeringly wide range in the same alloy.

The  $\text{Al}_{0.3}\text{CoCrFeNi}$  HEA was synthesized via arc melting using pure metals, solutionized at 1150 °C for 10 h and cold-rolled to 90% thickness reduction. Subsequently, sections of the cold-rolled alloy were subjected to four different thermal treatments: (a) recrystallization at 1150 °C (hereafter referred to as CRSA), and (b) annealing at 620 °C for 50 h to promote second phase precipitation (hereafter referred to



**Fig. 2.** Variety of the microstructures of the  $\text{Al}_{0.3}\text{CoCrFeNi}$  CCA for different processing conditions. SEM image showing the recrystallized grains from sample CRSA (A) along with the [011] *fcc* zone axis diffraction pattern, and atom probe tomography (APT) showing the 3D reconstructions of Al, Co, Cr, Fe, and Ni ions. The results show the presence of a single solid solution *fcc* phase.



**Fig. 3.** Variety of the microstructures of the  $\text{Al}_{0.3}\text{CoCrFeNi}$  CCA for four processing conditions. BFTEM and STEM-EDS maps of Al and Cr, and 3D reconstructions of atoms/ions Al and Cr, as marked, in a typical APT tip from samples CR620 (C) and CR550 (D). The compositional profile using one-dimensional cylinder across the various interfaces as marked in the figure shows the compositional partitioning of various elements.

as CRSA620–50), or direct annealing of the heavily deformed cold-rolled specimens at (c) 620 °C for 50 h (hereafter referred to as CR620–50) or (d) 550 °C for 24 h (hereafter referred to as CR550–24) and 50 h (hereafter CR550–50) to promote concurrent recrystallization and second phase precipitation.

Microstructural characterization was performed using scanning electron microscopy (SEM) (FEI Nova-NanoSEM 230™). The tensile tests were carried out by setting the initial strain rate at  $10^{-3} \text{ s}^{-1}$ , with respect to change in length. The tensile test specimens (gauge length ~ 3 mm, width ~ 1 mm and thickness ~ 0.5–1 mm) were machined using electric discharge machine (EDM). At least five independent tests were carried out to obtain tensile elongation and strength data. Conventional transmission electron microscopy (TEM) studies were carried out on a FEI Tecnai G2 TF20™ operating at 200 kV. Precipitate characterization was done using conventional and high angle annular dark field-scanning TEM (HAADF-STEM) modes. To measure the composition, energy dispersive spectroscopy (Super-X system) equipped on an FEI-TITAN G2 TEM microscope was used in the HAADF STEM mode operating at 300 kV and the results were analyzed with FEI's ES vision software version 6. TEM foils were prepared by an FEI Nova Nanolab 200 dual-beam focused ion beam (FIB) instrument using a Ga ion beam for milling. Standard lift-out techniques were used for Atom Probe Tomography (APT) sample preparation in the FIB before mounting the small sections of the samples on suitable holders for analysis. APT experiments were conducted on a CAMECA local electrode atom probe 3000× HR instrument.

A single fcc phase with uniform elemental distribution was obtained after recrystallization at 1150 °C (Fig. 2A, CRSA condition). Subsequent annealing at 620 °C for 50 h (CRSA620 condition), lead to a homogeneous distribution of ordered (Al,Ni)-rich  $\text{L}_{12}$  precipitates throughout the fcc matrix (Fig. 2B & C). When concurrent recrystallization and precipitation occur by direct annealing the samples at 620 °C after cold-rolling (CR620 condition), the alloy has a triplex, or three-phase microstructure comprising of fcc phase, (Al,Ni)-rich B2 precipitates, and Cr-rich  $\sigma$  precipitates (Fig. 3A). Finally, when the same alloy is annealed at 550 °C directly after cold-rolling (CR550–24 condition), the microstructure revealed two new additional features as compared to the CR620 condition: fine scale B2/bcc two-phase precipitates, and even finer, nanometer-scale (Al,Ni)-rich clusters uniformly distributed within the fcc matrix (Fig. 3B). The formation of conjoint B2/bcc on annealing at low temperature in a  $\text{Al}_x\text{CoCrFeNi}$  alloy system is also shown by Rao et al. [11]. The CR550–50 condition exhibited a similar microstructure as the CR550–24 sample. Interestingly, in all cases, the fcc matrix retained a high entropy like composition i.e.  $\text{Al}_{0.1}\text{CoCrFeNi}_{0.5}$  even after second (or third) phase precipitation (for example refer to Fig. 2C). Notwithstanding, these four types of microstructures, obtained for the same alloy composition, elucidate the rationale of this work; i.e. the widely varying combination of phases and microstructural length scales that can be achieved for the same HEA composition.

Dark field transmission electron microscope (DFTEM) image along with the corresponding [011] zone axis diffraction pattern, and atom probe tomography (APT) showing the 3D reconstructions of Al, Co, Cr,



Fe and Ni ions and 15 at.% Al isoconcentration surface outlining  $L1_2$  precipitates in a typical APT tip from sample CRSA620 (B). The compositional profile using a proximity histogram across the  $fcc$ - $L1_2$  interface shows the compositional partitioning of various elements.

Fig. 4A shows the room temperature engineering stress-strain curves for these five conditions. The corresponding true stress-strain curves for the same conditions are shown in Fig. S1A. The single  $fcc$  microstructure HEA (CRSA) exhibits an ultimate tensile strength (UTS) of ~400 MPa and a ductility (elongation to failure under tension) of

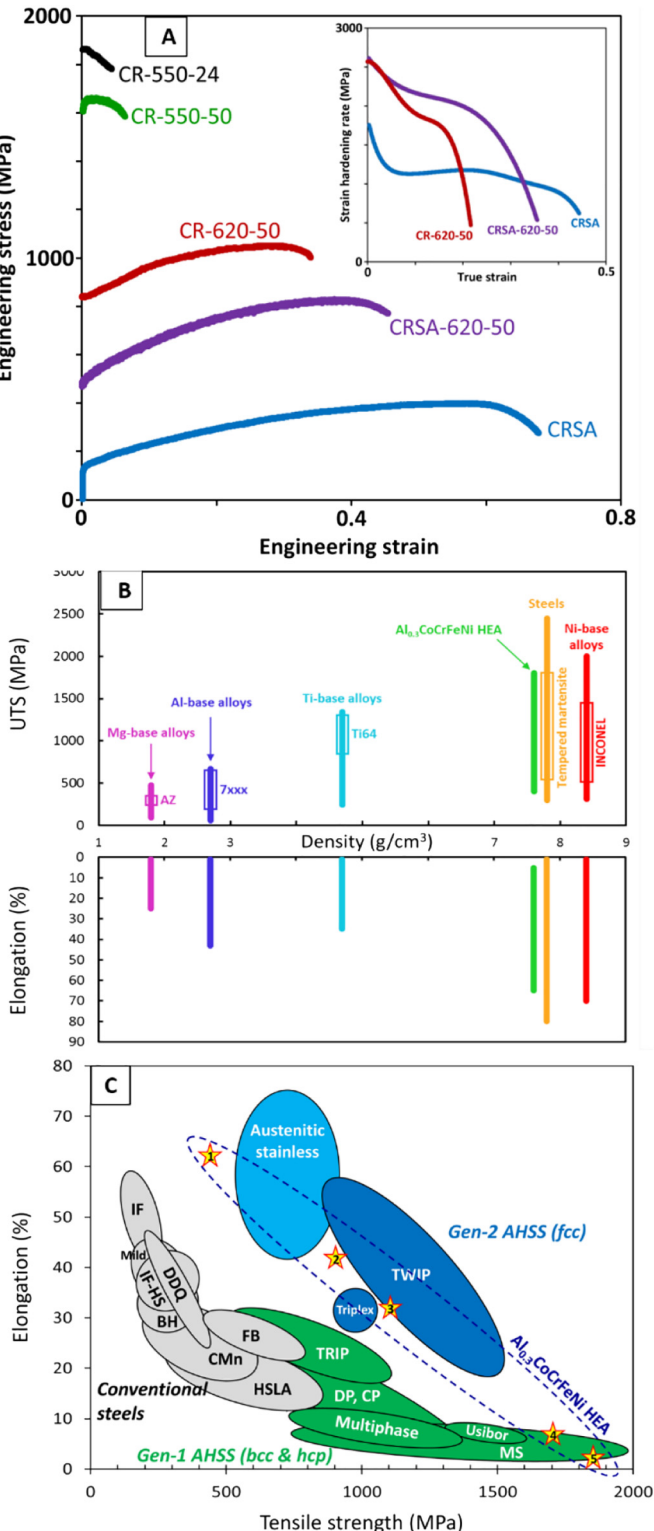
~65%. The introduction of precipitation hardening ( $L1_2$  precipitation) in this alloy (CRSA620–50 condition), resulted in a UTS increase by 40%, reaching 850 MPa, while maintaining an excellent ductility of ~45%. The triplex CCA (CR620–50 condition) exhibits a remarkable strength-ductility compromise of 1050 MPa and ~35%. The CR550–24 condition with the most complex multi-scale microstructure, exhibits an even higher UTS (1850 MPa), an exceptionally high value for an  $fcc$ -based “austenitic-like” alloy, but a lower tensile ductility (~5%). Similarly, the CR550–50 condition exhibits a very high UTS value of 1650 MPa and a ductility ~7%. The main difference between the CR550–24 and CR550–50 conditions is that while there is virtually no strain hardenability remaining in case of the former condition, the latter condition exhibits some limited strain hardenability.

Fig. 4A (inset) reveals a large range in the mechanical properties of  $Al_{0.3}CoCrFeNi$  as a function of processing conditions (or microstructures; see Figs. 2 and 3) and a notable difference in the strain hardening responses. In the single-phase HEA, the strain hardening rapidly drops and then remains nearly constant at a rate of 1000–1300 MPa before decreasing again. This behavior during plastic deformation is similar to that of low-stacking fault energy alloys such as  $CoCrFeMnNi$ -based HEA [12],  $Al_{0.1}CoCrFeNi$  [14] and TWIP steels [15], which harden at room temperature at a rate of 1000–1500 MPa, 740 MPa, and 800–1200 MPa, respectively. The strain hardening of the multi-phase CCAs, i.e. CRSA620 and CR620, is initially much higher than the single-phase HEA, i.e. between 1500 and 2000 MPa, (see inset Figs. 4 and S1B). These observations indicated that differences in strain hardening result from the formation of secondary-phases (e.g.  $L1_2$  or  $B2/\sigma$ ) and the corresponding change in the  $fcc$  matrix composition i.e.  $Al_{0.15}CoCrFeNi_{0.5}$  and  $Al_{0.1}CoCrFeNi_{0.5}$  of CRSA620 and CR620 conditions respectively. The consequent reduction in both Ni and Al, both of which typically have high SFEs (146 mJ/m<sup>2</sup> for Al (lattice constant of 0.405 nm), 120.3 mJ/m<sup>2</sup> for Ni (lattice constant of 0.352 nm)) [16], is expected to reduce the SFE in the  $fcc$  matrix of the HEA. Furthermore, such reduction in SFE is typically associated with enhanced twinning in steels [17]. A similar observation was also noted in our studies, where the  $fcc$  matrix of CR620 had coarser deformation twins as compared to the simple phase in CRSA (compare supplementary Figs. S2 and S3). It may be emphasized that twinning is often associated with high strain-hardenability and good ductility [17,18], and both properties are readily evident in the CR620 (containing  $fcc$ - $B2-\sigma$ ) and CRSA620 ( $fcc$ - $L1_2$ ) conditions. Therefore, it is likely that deformation twins, in the two conditions, will interact with mobile dislocations during quasi-static straining, and enhance strain-hardenability by increasing dislocation storage per the Kock-Mecking criterion [19].

A simplistic strength model can be employed to estimate the strengthening contribution of different microstructural features.

The yield strength can be approximated using the following equation:

$$\sigma_y = \sigma_{HP} + \Delta\sigma_{dis} + \Delta\sigma_{ppt} \text{ (MPa)} \quad (1)$$



**Fig. 4.** (A) Mechanical behavior at room temperature of the  $Al_{0.3}CoCrFeNi$  CCA. Engineering stress-strain and strain hardening vs true strain curves (inset). (B) Ranges of ultimate tensile strength and ductility achieved by  $Al_{0.3}CoCrFeNi$  in comparison with the main classes of structural alloys including present generation automotive and aerospace engineering alloys. (C) Materials property space for room temperature elongation vs tensile strength of steels and  $Al_{0.3}CoCrFeNi$  CCA. This strength-ductility diagram shows the range of properties achieved by  $Al_{0.3}CoCrFeNi$  CCA (represented by stars 1–4) in comparison with all steel grades which are classified in three groups in addition to stainless steels. Conventional steels include low-strength steels, (e.g. interstitial-free (IF)) and mild steels, and conventional high strength steels, e.g. bake hardenable, high-strength low alloy (HSLA) and ferritic-bainitic (FB) steels. The first generation of advanced high-strength steels (AHSS) consist of ferritic matrix containing a hard martensite phase, i.e. dual phase (DP), retained austenite, i.e. transformation induced plasticity (TRIP), or small amounts of austenite and pearlite in ferrite/bainite matrix, i.e. complex phase (CP), or martensitic matrix containing small amount of ferrite/bainite, i.e. martensitic (MS) steels. The second generation of AHSS is austenitic steels with twinning induced plasticity (TWIP) and triplex steels.

where  $\sigma_{HP}$  is the combined effect of strengthening contributions from lattice friction, solid solution, and grain size (Hall-Petch).  $\Delta\sigma_{dis}$  and  $\Delta\sigma_{ppt}$  are the additional strengthening contributions from dislocations and precipitates, respectively.

In the CR-620 condition, the composition of the *fcc* matrix is  $Al_{0.1}CoCrFeNi_{0.5}$  (from APT analysis). Kumar et al. [21] developed a Hall-Petch type equation for the yield strength as a function of grain size ( $\sigma_{HP} = 174 + \frac{371}{\sqrt{d}}$  MPa) for the single *fcc* solid solution phase  $Al_{0.1}CoCrFeNi$  HEA. The CR-620 condition in the present study exhibits an average grain size  $\sim 1.2 \mu m$  (refer Fig. S4), and consequently the Hall-Petch strengthening can be estimated to be  $\sim 512$  MPa (based on [21]). Hence, it can be assumed that  $850$  MPa (YS)– $512$  MPa (HP) =  $338$  MPa is the strengthening contribution due to the hard intermetallic B2/sigma precipitates and dislocations.

Likewise, in the CR-550-24 condition, (YS =  $1850$  MPa) average grain size is  $\sim 0.4 \mu m$  (refer to Fig. S4). The Hall-Petch strengthening contribution can be estimated to be  $\sim 760$  MPa. Based on simple image analysis (using the ImageJ software) the approximate fraction of B2 + sigma phase in CR-620 and CR-550 are 19% and 15% and therefore the strengthening contribution can be assumed similar and  $\sim 338$  MPa. Therefore, the additional strength enhancement in case of the CR-550-24 condition, can be attributed to the presence of nano-clusters or precipitates, and this enhancement can be estimated to be  $1850$  MPa (YS) –  $760$  MPa (HP) –  $338$  MPa (B2/Sigma/dislocations) =  $752$  MPa.

The strength enhancement due to nano-precipitates can be estimated by using a simple Orowan strengthening model, based on the shearing of these nano-precipitates by dislocations. It can be described by the equation below:

$$\sigma_{orowan} = \frac{G * b}{\lambda} \quad (2)$$

$$\lambda = \frac{4(1-f)r}{3f} \quad (3)$$

where  $\lambda$  is average planar center-to-center distance between nano-precipitates,  $G$  is the shear modulus of the matrix,  $b$  is the Burgers vector (assumed to be  $80$  GPa and  $0.255$  nm for  $CoCrFeNiMn$  respectively [22]),  $r$ , the radius of the nano-precipitates ( $2$  nm), and  $f$ , the volume fraction of precipitates ( $10\%$  by vol.), assessed from the APT results. Based on Eqs. (2) and (3), the Orowan strengthening can be estimated to be  $\sim 850$  MPa, which compares reasonably well with the experimentally observed strengthening value of  $752$  MPa. Therefore, the extraordinarily high value of yield strength  $\sim 1850$  MPa in case of the CR-550 condition is a result of the contributions from hierarchical features ranging from thousands of nanometers (grain boundaries) to hundreds of nanometers (B2/sigma) to few nanometers (clusters).

DongYue et al. showed the impact toughness of single phase  $Al_{0.3}CoCrFeNi$  was much superior to steels [20]. An estimation of the toughness for the different types of microstructures in the same HEA ( $Al_{0.3}CoCrFeNi$ ) has been carried out by calculating the area under the stress-strain curve (refer to Fig. S5). Quite remarkably, after second phase precipitation, the toughness of the CRSA620 and CR620 alloys increased, as compared to the CRSA condition.

The multiple combinations of hierarchical multi-phase microstructures, observed for a single HEA composition, provide the largest range of strength-ductility combinations observed in any engineering class of structural metallic materials in their bulk form (Fig. 4B), which is only attainable by accounting several grades of advanced high-strength-steels (AHSS) as shown by the strength-ductility diagram for steels (Fig. 4C). The vast range of mechanical properties achieved by the single  $Al_{0.3}CoCrFeNi$  HEA has been further put into perspective by plotting the strength-ductility values for the five conditions on a plot encompassing all presently known steels, as illustrated in Fig. 4C.

Remarkably, the range of mechanical properties exhibited by the single  $Al_{0.3}CoCrFeNi$  HEA encompasses three different groups of steels.

The first group belongs to ductile austenitic stainless steels with which the present HEA shares several common major elements (i.e. Fe, Cr, Ni) and the same single-phase austenitic, i.e. face-centered cubic (*fcc*), microstructure. The second group combines high strength and high ductility due to intense work hardening. This includes Twinning-Induced Plasticity (TWIP) and Triplex (e.g. Fe-Mn-C-Al) grades of steels for which strain hardening mechanisms are controlled by the stacking fault energy (SFE). These steels have a high manganese content that causes the stabilization of the austenite and adjusts the SFE at the appropriate value to activate deformation twinning, resulting in multi-stage strain hardening [8,10–12]. The third group, e.g. Martensitic steels (MS), have higher strength and lower ductility achieved by a martensitic matrix containing small amounts of ferrite and/or bainite [13].

The exceptionally large variety of mechanical behaviors and microstructures observed in the single  $Al_{0.3}CoCrFeNi$  HEA emphasizes one of the unique features arising from the concept of HEAs as complex concentrated alloys.

The tunability of the mechanical response exhibited by the  $Al_{0.3}CoCrFeNi$  HEA has the potential to solve two major problems. Today structural applications employ several materials to meet the requirements that often conflict. For example, in the automotive industry, this involves the welding and bonding of different grades of steels, which is challenging because they differ in chemical composition. The exceptionally high tunability at constant composition of  $Al_{0.3}CoCrFeNi$  HEA avoids this problem while meeting various functional requirements. The second concerns the development of architected materials. This concept has been depicted in the seminal work of O. Bouaziz et al. [23–25] as an innovative strategy for structural materials development to provide otherwise conflicting properties. The idea is to tailor the spatial distribution of the microstructure in a given topology and geometry in order to vary the mechanical properties in the overall component. Basu et al. emphasized on role of interfaces between crystallographically dissimilar phases on the size-dependent plasticity in multiphase HEAs [26–28]. Such a strategy has been underexploited because it relies on the fabrication of compositionally graded materials, which often require new processing methods. This obstacle could be overcome by the present HEA ( $Al_{0.3}CoCrFeNi$ ) which features a unique range of microstructures attainable using thermo-mechanical processing without changing the alloy composition.

Supplementary data to this article can be found online at <https://doi.org/10.1016/j.scriptamat.2018.10.023>.

## Acknowledgement

The work was performed under a cooperative agreement between the Army Research Laboratory and the University of North Texas (W911NF-16-2-0189). The authors also acknowledge the Materials Research Facility (MRF) at the University of North Texas for use of microscopy and atom probe tomography facilities, and the Center for Electron Microscopy and Analysis (CEMAS) at the Ohio State University, for microscopy facilities.

## References

- [1] B. Cantor, I.T.H. Chang, P. Knight, A.J.B. Vincent, *Mater. Sci. Eng.* 375 (2004) 213–218.
- [2] J.-W. Yeh, S.-K. Chen, S.-J. Lin, J.-Y. Gan, T.-S. Chin, T.-T. Shun, C.-H. Tsau, S.-Y. Chan, *Adv. Eng. Mater.* 6 (2004) 299–303.
- [3] B. Gludovatz, A. Hohenwarter, D. Catoor, E.H. Chang, E.P. George, R.O. Ritchie, *Science* 345 (2014) 1153–1158.
- [4] Gwalani, Bharat, Stephane Gorsse, Deep Choudhuri, Mark Styles, Yufeng Zheng, Rajiv S. Mishra, Rajarshi Banerjee, *Acta Mater.* 153 (2018) 169–185.
- [5] D. Miracle, O. Senkov, *Acta Mater.* 122 (2017) 448–511.
- [6] S. Gorsse, D.B. Miracle, O.N. Senkov, *Acta Mater.* 135 (2017) 177–187.
- [7] O. Bouaziz, S. Allain, C.P. Scott, P. Cugy, D. Barbier, *Curr. Opin. Solid State Mater. Sci.* 15 (2011) 141–168.
- [8] Z. Li, K.G. Pradeep, Y. Deng, D. Raabe, C.C. Tasan, *Nature* 534 (2016) 227–231.
- [9] J.Y. He, H. Wang, H.L. Huang, X.D. Xu, M.W. Chen, Y. Wu, X.J. Liu, T.G. Nieh, K. An, Z.P. Lu, *Acta Mater.* 102 (2016) 187.
- [10] Gutierrez-Urrutia, D. Raabe, *Acta Mater.* 60 (2012) 5791–5802.

- [11] J.C. Rao, H.Y. Diao, V. Ocelík, D. Vainchtein, C. Zhang, C. Kuo, Z. Tang, et al., *Acta Mater.* 131 (2017) 206–220.
- [12] Y. Deng, C.C. Tasan, K.G. Pradeep, H. Springer, A. Kostka, D. Raabe, *Acta Mater.* 94 (2015) 124–133.
- [13] R.L. Klueh, A.T. Nelson, *J. Nucl. Mater.* 371 (2007) 37–52.
- [14] X.D. Xu, P. Liu, Z. Tang, A. Hirata, S.X. Song, T.G. Nieh, P.K. Liaw, C.T. Liu, M.W. Chen, *Acta Mater.* 144 (2018) 107–115.
- [15] S. Curtze, V.-T. Kuokkala, *Acta Mater.* 58 (2010) 5129–5141.
- [16] Liming Xiong, Qian Deng, Garritt J. Tucker, David L. McDowell, Youping Chen, *Int. J. Plast.* 38 (2012) 86–101.
- [17] T.S. Byun, *Acta Mater.* 51 (2003) 3063–3071.
- [18] B.C. De Cooman, Y. Estrin, S.K. Kim, *Acta Mater.* 142 (2018) 283–362.
- [19] Yu Estrin, H. Mecking, *Acta Metall.* 32 (1984) 57–70.
- [20] Li, DongYue, Yong Zhang, *Intermetallics* 70 (2016) 24–28.
- [21] N. Kumar, M. Komarasamy, P. Nelaturu, Z. Tang, P.K. Liaw, R.S. Mishra, *JOM* 67 (2015) 1007–1013.
- [22] J.Y. He, W.H. Liu, H. Wang, Y. Wu, X.J. Liu, T.G. Nieh, Z.P. Lu, *Acta Mater.* 62 (2014) 105–113.
- [23] O. Bouaziz, Y. Bréchet, J.D. Embury, *Adv. Eng. Mater.* 10 (2008) 24–36.
- [24] B. Chéhab, H. Zurob, D. Embury, O. Bouaziz, Y. Brechet, *Adv. Eng. Mater.* 11 (2009) 992–999.
- [25] D. Embury, O. Bouaziz, *Annu. Rev. Mater. Res.* 40 (2010) 213–241.
- [26] Indranil Basu, Václav Ocelík, Jeff Th.M. De Hosson, *Acta Mater.* 150 (2018) 104–116.
- [27] Indranil Basu, Václav Ocelík, Jeff Th.M. De Hosson, *Acta Mater.* 157 (2018) 83–95.
- [28] Indranil Basu, Václav Ocelík, Jeff Th.M. De Hosson, *J. Mater. Res.* (2018) 1–22.