Thermo-mechanical response of single-phase face-centered-cubic Al$_x$CoCrFeNi high-entropy alloy microcrystals

Quan Jiao$^{a, *}$, Gi-Dong Sim$^{b, *}$, Mageshwari Komarasamy$^c$, Rajiv S. Mishra$^e$, Peter K. Liaw$^d$ and Jaafar A. El-Awady$^{a,*}$

$^a$Department of Mechanical Engineering, Johns Hopkins University, Baltimore, MD, USA; $^b$Department of Mechanical Engineering, Korea Advanced Institute of Science and Technology, Daejeon, Republic of Korea; $^c$Advanced Materials and Manufacturing Processes Institute, Department of Materials Science and Engineering, University of North Texas, Denton, TX, USA; $^d$Department of Materials Science and Engineering, The University of Tennessee, Knoxville, TN, USA

ABSTRACT

The response of [100]-oriented single-crystal face-centered-cubic Al$_{0.1}$CoCrFeNi and Al$_{0.3}$CoCrFeNi high-entropy alloy (HEA) microcrystals tested from 293 to 573 K by means of \textit{in situ} microcompression is reported. At all temperatures, plasticity is governed by dislocation slip with significant strain hardening and intermittent strain bursts observed. A model, which is in good agreement with experimental measurements, is also developed to predict the effects of Al concentration, temperature, and crystal size on the strength of HEAs. The model interestingly predicts a softening response with an increase in Al concentration when the crystal size is $\leq 0.1 \mu$m. Finally, this model can guide the development of advanced HEAs for small-scale applications.

IMPACT STATEMENT

\textit{In situ} scanning electron microscopy (SEM) experiments are performed to quantify the thermo-mechanical response of Al$_x$CoCrFeNi microcrystals. A physics-based model is also proposed to predict the strength of high-entropy alloys as a function of crystal size, temperature, and Al concentration.

High-entropy alloys (HEAs) are nominally single-phase random solid solutions formed from equal or nearly equal atomic concentrations of multiple constituents [1–7]. These innovative alloys have drawn substantial attention in recent years due to the vast available composition space for tuning their properties. Several HEA systems have been shown to demonstrate immense potential for a wide range of functional and structural applications. Examples include the metastable dual-phase FeMnCoCr HEAs that demonstrate extensive hardening and massive strengthening [8], NbMoTaWV HEAs that exhibit better high-temperature strength than currently available commercial superalloys [9], and AlCoCrCuFeNi HEAs that possess significantly improved fatigue resistance [10].

Despite extensive efforts that have been devoted to study the mechanical behavior of various HEA systems [11–17], the underlying deformation mechanism, in particular at high temperatures [18,19], is not yet fully addressed. The Fleischer model for dilute concentration alloys cannot be applied to HEA systems, and statistical-based Labusch-type models that may work for all concentrations are still in their infancy [20–22].

In Al$_x$CoCrFeNi HEA systems, the Al atoms have a much larger radius, compared to the other atoms comprising this alloy. Therefore, the increase in the Al content...
would lead to a significant increase in the lattice constant and distortion of this HEA [23]. On the other hand, increasing the Al concentration will also lead to solid-solution hardening via the formation of strong covalent bonds with neighboring atoms, and the generation of hard atomic clusters [24].

In this paper, the effect of Al concentration on the temperature-dependent mechanical response of [100]-oriented single-crystal face-centered-cubic (FCC) Al_{0.1}CoCrFeNi HEA (in molar ratio) microcrystals is investigated using in situ SEM micro-compression experiments. If the Al composition is such that $x > 0.4$, a secondary body-centered-cubic (BCC) phase will solidify [25,26], the chemical compositions are thus limited in this study to $x = 0.1$ and $x = 0.3$ to ensure that the crystal is single-phase FCC.

The Al_{0.1}CoCrFeNi HEA, referred to hereafter as the Al_{0.1} HEA, was procured by Sophisticated Alloys Inc. in the form of cast blocks. Electrical discharge machining was used to cut a block into $\sim 4$-mm thick sheet. From that sheet, $\sim 12.7$ mm diameter cylinders were machined using the Tormach PCNC 1100 mill. The Al_{0.3}CoCrFeNi HEA, referred to hereafter as the Al_{0.3} HEA, was prepared by arc-melting pure elements under a high-purity argon atmosphere on a water-cooled Cu hearth [27].

Both as-forged polycrystalline HEA alloys were mechanically polished using the silicon carbide grinding paper (400, 600, 800, and 1200 grit) and 0.02 μm colloidal silica formula. The chemical composition was confirmed using energy-dispersive spectrometry and the crystal orientation of the HEA was characterized using electron backscatter diffraction inside a Tescan Mira Field Emission-Scanning Electron Microscope (FE-SEM).

Pillar-like microcrystals having a diameter $D = 5.0$ μm were fabricated within a [100]-oriented single grain by the focused ion beam (FIB, FEI Strata DB235) annular milling. This size was chosen as an intermediate pillar size to avoid complications associated with smaller pillar sizes in terms of large taper and FIB-induced damage, while not being too large to avoid excessive fabrication times [28]. An aspect ratio of 2:1 (height:mid-plane diameter) was chosen to avoid buckling for higher aspect ratios and non-uniform stress along the length for lower aspect ratios [29].

Micro-compression tests of the single-crystal HEA microcrystals were performed using the InSEM HT (Nanomechanics Inc.) in situ indenter, equipped with a 15-μm diameter diamond flat punch tip in a Tescan Mira FE-SEM. All experiments were carried out at a nominal strain rate of $10^{-3}$ s^{-1} and at three temperatures: $T = 293$, 423, and 573 K. While the instrument is inherently load controlled, the displacement rate was controlled via feedback from the load signal in order to achieve a constant strain rate. If a displacement jump larger than 10 nm was recorded during the deformation (e.g. due to a large strain burst), the microcrystal is unloaded by 70% for the 293 and 423 K case, and by 10% for the 573 K case, then reloaded with the same procedure discussed above. Engineering stress and engineering strain were calculated by dividing the load and displacement by the initial mid-plane cross-sectional area and pillar height, respectively. The raw displacement data, measured at a data acquisition rate of 100 Hz, were corrected to account for the additional compliance due to deformation of the base along with the pillar during loading [30].

Representative stress–strain curves at different temperatures and the typical deformation morphologies at 293 and 573 K for the Al_{0.1} and Al_{0.3} HEAs are shown in Figures 1 and 2, respectively. The average and standard deviation of the CRSS at 2% and 5% strains from all tested samples are summarized in Table 1 and Figure 3, and accelerated in situ videos can be found in Supplementary Files. The apparent vibrations in the videos at higher temperatures are only shimmering effects. Irrespective of the test temperature, the in situ experiment revealed that as the load is applied, the microcrystals elastically deform until the onset of plasticity, at which point the first slip band and strain burst are observed. After plastic flow initiates, the microcrystals exhibit intermittent plasticity events that are evident by the successive strain bursts (i.e. load drops). Since the microcrystals are oriented in a multi-slip orientation, both samples exhibit strong work hardening. Furthermore, for both compositions, the flow stresses are observed to decrease modestly with increasing temperature. It is also observed that the Al_{0.3} HEA microcrystals are on average stronger than the Al_{0.1} HEA ones at all tested temperature due to the solid-solution strengthening effect of Al. However, since the difference in Al concentration is only less than 5% between the two HEAs (i.e. 2.4% versus 7%), this enhanced strengthening effect is small.

As expected for the deformation of [001]-oriented crystals, symmetric-slip deformation from all four [111] slip planes is observed. Twinning has also been commonly observed in coarse-grained (CG) polycrystalline Al_{0.1}CoCrFeNi HEAs after large deformation and extensive work hardening at a wide range of strain rates ($10^{-3}$ to $10^{3}$ s^{-1}) [31]. However, in the current study, TEM lift-outs from the center of the deformed microcrystals did not show any indication of deformation twinning. This lack of deformation twinning can be rationalized by the low strain levels reached in the current experiments (~10% strain), as compared to bulk crystals. In addition, it is commonly observed in other crystal systems that the twinning
Figure 1. (a) Representative stress–strain curves for the Al$_{0.1}$CoCrFeNi HEA tested at various temperatures. Representative SEM micrographs of the deformed microcrystals at (b) 293 K and (c) 573 K.

Figure 2. (a) Representative stress–strain curves for the Al$_{0.3}$CoCrFeNi HEA tested at various temperatures. Representative SEM micrographs of the deformed microcrystals at (b) 293 K and (c) 573 K.

Table 1. Summary of the experimentally measured critical resolve shear stress (CRSS) as well as predicted from Equation (1) for both tested compositions at different temperatures.

| HEA system | Temperature (K) | 2% Flow stress (MPa) | 5% Flow stress (MPa) | CRSS (MPa) | $\tau_{ss}$ (MPa) | $\tau_{thermal}$ (MPa) | CRSS from Equation (1) (MPa) |
|------------|----------------|----------------------|---------------------|------------|----------------|------------------------|-----------------------------|
| Al$_{0.1}$CoCrFeNi | 293 | 276 ± 7 | 352 ± 40 | 112.6 ± 2.9 | 48.9 | 69.6 | 118.5 |
| 423 | 260 ± 40 | 313 ± 37 | 106.1 ± 16.3 | 33.7 | 66.3 | 100.0 |
| 573 | 239 ± 16 | 356 | 97.5 ± 6.5 | 21 | 62.4 | 83.4 |
| Al$_{0.3}$CoCrFeNi | 293 | 307 ± 24 | 395 ± 19 | 125.3 ± 9.8 | 58 | 64.8 | 122.8 |
| 423 | 267 ± 40 | 307 ± 40 | 108.9 ± 16.3 | 43.9 | 61.7 | 105.5 |
| 573 | 244 ± 9 | 244 ± 9 | 99.6 ± 3.7 | 28.1 | 58.1 | 86.2 |

*CRSS is calculated from the 2% flow stress based on a Schmid factor of 0.408.

nucleation stress increases with decreasing sizes [32,33]. In addition, recent studies on single-crystal and polycrystalline materials reported that the power-law exponent for deformation-twinning-mediated plasticity is much larger than that for dislocation-mediated plasticity [32–36]. That is, deformation twinning shows stronger crystal-size effects. Therefore, even if deformation-twinning-mediated plasticity governs the mechanical response at the bulk scale, it is plausible that there is a crystal size below which dislocation-mediated
plasticity dominates over deformation twinning (i.e., when \( \text{CRSS}_{\text{slip}} < \text{CRSS}_{\text{twinning}} \)).

Under the premise that deformation-twinning-mediated plasticity is excluded from such micron-sized samples, in the following a dislocation-based model is developed to describe the effect of Al concentration \((C_{\text{Al}})\), temperature \((T)\), dislocation density \((\rho)\), strain rate \((\dot{\varepsilon})\), and crystal size \((D)\) on the CRSS of Al\(_x\)CoCrFeNi microcrystals. In this model, the CRSS, \(\tau_{\text{CRSS}}(T, \dot{\varepsilon}, D, \rho)\), is decomposed into two components, such that:

\[
\tau_{\text{CRSS}}(T, \dot{\varepsilon}, D, \rho) = \tau_{\text{thermal}}(D, \rho) + \tau_{\text{ss}}(T, \dot{\varepsilon}),
\]

where \(\tau_{\text{thermal}}(D, \rho)\) is the contribution from athermal mechanisms and \(\tau_{\text{ss}}(T, \dot{\varepsilon})\) is the contribution from thermally activated mechanisms during plastic deformation. The athermal contribution is directly associated with the long-range elastic interactions of dislocations, which is well described by the generalized size-dependent Taylor-strengthening law [28]:

\[
\tau_{\text{thermal}} = \mu \left( \frac{\beta \dot{\varepsilon}}{\rho^{1/2}} \right) + \alpha \rho^{1/2},
\]

where \(\beta = 1.76 \times 10^{-3}\), \(\alpha = 0.57\) are dimensionless constants, and \(\mu\) is the shear modulus as a function of temperature and Al concentration. The first term on the right-hand side of Equation (2) is the stress required to activate the weakest dislocation source, while the second term accounts for hardening arising from dislocation forest interactions. Experimental and ab initio simulation results suggest that by treating CoCr-FeNi as an effective medium, the material parameters can be approximated as a bilinear function of temperature, \(T\), and Al concentration, \(C_{\text{Al}}\). By fitting data in the literature, the shear modulus can be shown to be: \(\mu(T, C_{\text{Al}}) = (1 - 1.51C_{\text{Al}}) \times (-0.031T + 93.5)\), while Poisson’s ratio \(\nu\) and atomic volume \(V\) are relatively temperature insensitive and only depend on the Al concentration according to: \(\nu(C_{\text{Al}}) = 0.28 + 0.2C_{\text{Al}}\) and \(V(C_{\text{Al}}) = (1 - C_{\text{Al}}V_{\text{CoCrFeNi}}) + C_{\text{Al}}V_{\text{Al}}\) [22,37].

The thermal contribution, \(\tau_{\text{ss}}(T, \dot{\varepsilon})\), originates from the local energy barrier for dislocation slip that can be overcome by local thermal fluctuations. In the current HEA system, \(\tau_{\text{ss}}\) is predominantly dominated by the lattice friction originating from solid-solution strengthening. Here, a recently developed Labusch-type theory is adopted to estimate the solid-solution strengthening [38]. In this theory, \(\tau_{\text{ss}}\) can be computed at a finite temperature, \(T\), and a given strain rate of \(\dot{\varepsilon}\), by considering each elemental component, \(n\), as a solute atom embedded in an effective matrix of surrounding alloy, such that [22]:

\[
\tau_{\text{ss}}(T, \dot{\varepsilon}) = \begin{cases} 
\frac{T}{\Delta E_b} \left[ 1 - \left( \frac{kT}{\Delta E_b} \ln \frac{\dot{\varepsilon}_0}{\dot{\varepsilon}} \right) \right]^{2/3} & \text{for } \tau_{\text{ss}}/\tau_{\text{ss}}^0 > 0.5, \\
\frac{T}{0.57 \Delta E_b} \left[ \frac{1}{\frac{1}{\dot{\varepsilon}_0} + \frac{kT}{\Delta E_b} \ln \frac{\dot{\varepsilon}_0}{\dot{\varepsilon}} \right] & \text{for } \tau_{\text{ss}}/\tau_{\text{ss}}^0 < 0.5,
\end{cases}
\]

where \(k\) is the Boltzmann constant, \(\dot{\varepsilon}_0 = 10^4s^{-1}\) is a reference strain rate and the apparent zero-temperature flow
stress, $\tau_{0}$, is

$$\tau_{0}^0 = 0.051\alpha^{-1/3}K_{f}\tau_{f}\left[\sum_{n=1}^{N} c_n \Delta V_n^2\right]^{2/3},$$  \hspace{1cm} (4)

while the energy barrier, $\Delta E_b$, is

$$\Delta E_b = 0.274\alpha^{1/3}K_{\Delta E}\frac{f_{\Delta E}}{\Delta b}\left[\sum_{n=1}^{N} c_n \Delta V_n^2\right]^{1/3},$$  \hspace{1cm} (5)

and

$$K_{f} = \mu\left(\frac{1+v}{1-v}\right)^{4/3}b^{-4}, \quad K_{\Delta E} = \mu\left(\frac{1+v}{1-v}\right)^{2/3}b, \quad f_{\tau} = 0.35, \quad f_{\Delta E} = 5.7.$$  \hspace{1cm} (6)

Here, $N$ is the number of elements in the HEA, and the atomic volume can be calculated as $V = b^3/\sqrt{2}$. In this study, $f_{\tau}$ and $f_{\Delta E}$ are treated as two constants related to the dislocation core structures \[22,38\].

Using Equation (1), the variation of $\tau_{\text{CRSS}}$ as a function of temperature and Al concentration is shown in Figure 3(a,b) for $D = 5.0 \mu m$, $\rho = 3 \times 10^{13}/m^{-2}$ (which is a representative value for FIB-milled microcrystals), and $\dot{\varepsilon} = 10^{-3}s^{-1}$. The results are also summarized in Table 1 for the tested temperatures and Al concentrations for comparison with the experimental predictions. It is clear that the predicted $\tau_{\text{CRSS}}$ as a function of temperature and Al concentration is in excellent agreement with those measured experimentally. The slightly higher CRSS at 573 K compared to the model prediction can be attributed to possibly a slightly higher dislocation density in those experiments \[28\]. It is also observed that the effect of Al concentration decreases with increasing temperature, which can be rationalized by the fact that the energy barrier, $\Delta E_b$, is an increasing function of the overall atomic-size misfit from all elements (see Equation (4)). Thus, the increase in the Al concentrations raises the lattice distortion, which consequently leads to a higher energy barrier and stronger temperature dependency for dislocation slip. It should be noted that the small discontinuities in the predicted $\tau_{\text{CRSS}}$ at different temperatures and Al concentrations are a result of the piece-wise Arrhenius-type relation described by Equation (3), depending on the value of $\tau_{ss}/\tau_{0}^{ss}$. Nevertheless, this discontinuity is very small and has no implications for the current discussion.

It is also observed in Figure 3(b) that by increasing the Al concentration up to 8%, the effect of solid-solution
strengthening does not increase significantly. This can be attributed to two competing effects. On one hand, the relative large atomic misfit of Al atoms leads to solid-solution strengthening, while on the other hand, it leads to elastic softening. Considering that an Al concentration higher than 8% will lead to the formation of a BCC phase [25,26], the focus here is only on the Al concentration below this limit.

The currently proposed model demonstrates a similar strengthening effect of Al at all three testing temperatures and is in good agreement with the current experimental results. There is a slight discrepancy in the overall CRSS level for the highest temperature tested here (i.e. 573 K), although the strengthening effect of Al is well captured qualitatively. Nevertheless, a better approximation could be readily achieved by slightly adjusting \( f_c \) and \( f_{SE} \) in Equations (5) and (6), which would give a better dislocation core structure description at elevated temperatures [39].

The relative contribution of \( \tau_{\text{thermal}} \) and \( \tau_{\text{ss}} \) on the overall CRSS of the HEA microcrystal is summarized in Table 1 and as a stacked bar chart in Figure 3(c). It is clear that the significant drop in strength is predominantly associated with the decrease in \( \tau_{ss} \), which on the other hand is attributed to thermal softening. It is also clear that for the crystal size studied here, the contribution of \( \tau_{\text{thermal}} \) and \( \tau_{ss} \) to the total CRSS of the microcrystal is on the same order (tens of MPa) due to the sample size effect. This trend might not be the case for other crystal sizes, for example, for bulk single-crystal or CG polycrystalline sample, the contribution from \( \tau_{ss} \) will dominate the CRSS.

The model, represented by Equation (1), can give a general understanding of the relative effects of Al concentration, temperature, and crystal size on the CRSS of Al\(_x\)CoCrFeNi, as shown in Figure 4. It is clear that for a given concentration and temperature, a decrease in the crystal size leads to an increase in the CRSS in agreement with the ‘smaller is stronger’ phenomenon. It is also observed that for \( D \geq 20 \mu\text{m} \), the effect of size becomes irrelevant, and the crystal strength reaches a steady state that is equivalent to the bulk strength. Furthermore, the increase in temperature for a given crystal size and Al concentration will lead to a decrease in the crystal strength. On the other hand, it is interesting to note that due to the elastic softening effect of the added Al, when \( D \leq 0.1 \mu\text{m} \), the increase in the Al concentration leads to a decrease in the predicted CRSS (Figure 4(a)). However, for larger crystals, the increase in Al concentration will always lead to strengthening in agreement with the solid-solution hardening effects. It should be noted that although in the current study only experiments on micron-sized samples are used to verify the model, the model could be extended to bulk sample naturally. By assuming a dislocation density of \( 10^{12} \) and using a sample diameter of 2 mm, this model yields a prediction of 68.2 MPa CRSS which is very close to 75.2 MPa measurement [40].

In summary, in situ micro-mechanical testing technique was utilized to study the effect of Al composition and temperature on the mechanical properties and deformation mechanism for Al\(_x\)CoCrFeNi microcrystals. The experimental results show that, unlike bulk samples, plasticity is mediated by dislocation slip without any observation of deformation twinning. In addition, a physics-based model was developed to predict the effect of Al concentration, temperature, and crystal size on the strength of Al\(_x\)CoCrFeNi microcrystals. In this model, the crystal strength is proposed to be dominated by both the long-range athermal and short-range thermal contributions in a linearly additive manner. The model was shown to match well with the experimentally measured CRSS. The model also interestingly predicts that when the sample size decreases below 0.1 \( \mu\text{m} \), the increase in Al concentration leads to a softening effect. Finally, the currently proposed model is most likely to have a more general validity for a wider range of HEAs, and provides a guide for the development of advanced alloys for small-scale applications.

**Disclosure statement**

No potential conflict of interest was reported by the authors.

**Funding**

This work was supported by Air Force Office of Scientific Research [AFOSR, grant number FA9550-15-1-0070].

**ORCID**

Jaafar A. El-Awady [http://orcid.org/0000-0002-5715-2481](http://orcid.org/0000-0002-5715-2481)

**References**

[1] Cantor B, Chang ITH, Knight P, et al. Microstructural development in equiatomic multicomponent alloys. Mater Sci Eng A. 2004;375–377:213–218.
[2] Yeh J-W, Chen S-K, Lin S-J, et al. Nanostructured high-entropy alloys with multiple principal elements: novel alloy design concepts and outcomes. Adv Eng Mater. 2004;6:299–303.
[3] Tsai M-H, Yeh J-W. High-entropy alloys: A critical review. Mater Res Lett. 2014;2:107–123.
[4] Senkov ON, Wilks GB, Miracle DB, et al. Refractory high-entropy alloys. Intermetallics. 2010;18:1758–1765.
[5] Santodonato LJ, Zhang Y, Feygenson M, et al. Deviation from high-entropy configurations in the atomic distributions of a multi-principal-element alloy. Nat Commun. 2015;6:5964.
[6] Zhang Y, Zhou YJ, Lin JP, et al. Solid-solution phase formation rules for multi-component alloys. Adv Eng Mater. 2008;10:534–538.
[7] Zhang Y, Yang X, Liaw PK. Alloy design and properties optimization of high-entropy alloys. JOM. 2012;64:830–838.
[8] Li Z, Pradeep KG, Deng Y, et al. Metastable high-entropy dual-phase alloys overcome the strength–ductility trade-off. Nature. 2016;534:227–230.
[9] Senkov ON, Wilks GB, Scott JM, et al. Mechanical properties of Nb25Mo25Ta25W25 and V20Nb20Mo20Ta20W20 refractory high entropy alloys. Intermetallics. 2011;19:698–706.
[10] Hempfill MA, Yuan T, Wang GY, et al. Fatigue behavior of Al0.5CoCrCuFeNi high entropy alloys. Acta Mater. 2012;60:5723–5734.
[11] Shun T-T, Chang L-Y, Shiu M-H. Microstructures by mechanical alloying and spark plasma sintering. Intermetallics. 2012;26:44–51.
[12] Ji W, Wang W, Wang H, et al. Alloying behavior and novel properties of CoCrFeNiMn high-entropy alloy fabricated by mechanical alloying and spark plasma sintering. Intermetallics. 2015;32:24–27.
[13] Shun T-T, Chang L-Y, Shiu M-H. Microstructures and mechanical properties of FeCoNiCuAl alloys. Mater Sci Eng A. 2012;556:395–399.
[14] Gludovatz B, Hohenwarter A, Thurston KVS, et al. The influences of Al addition on the microstructure and mechanical properties of Nb25Mo25Ta25W25 and V20Nb20Mo20Ta20W20 high-entropy alloys. Intermetallics. 2011;19:698–706.
[15] Liu WH, Wu Y, He JY, et al. Grain growth and the Hall–Petch relation in commercial purity Al alloys. Scr Mater. 2013;68:526–529.
[16] Gludovatz B, Hohenwarter A, Thurston KVS, et al. Exceptional damage-tolerance of a medium-entropy alloy CrCoNi at cryogenic temperatures. Nat Commun. 2016;7:10602.
[17] Moon J, Hong SI, Bae JW, et al. On the strain rate-dependent deformation mechanism of CoCrFeMnNi high-entropy alloy at liquid nitrogen temperature. Mater Res Lett. 2017;5:472–477.
[18] Otto F, Dlouhý A, Somsen C, et al. The influences of temperature and microstructure on the tensile properties of a CoCrFeMnNi high-entropy alloy. Acta Mater. 2013;61:5743–5755.
[19] Wu Z, Bei H, Pharr GM, et al. Temperature dependence of the mechanical properties of equiatomic solid solution alloys with face-centered cubic crystal structures. Acta Mater. 2014;81:428–441.
[20] Labusch R. A statistical theory of solid solution hardening. Phys Status Solidi. 1970;1:659–669.
[21] Labusch R. Statistische theoreien der mischkristallhär tung. Acta Metall. 1972;20:917–927.
[22] Varvenne C, Curtin WA. Strengthening of high entropy alloys by dilute solute additions: CoCrFeNiAlx and CoCrFeNiMmAlx alloys. Scr Mater. 2017;138:92–95.
[23] Komarasamy M, Kumar N, Mishra RS, et al. Anomalies in the deformation mechanism and kinetics of coarse-grained high entropy alloy. Mater Sci Eng A. 2016;654:256–263.
[24] Tang Z, Gao MC, Diao H, et al. Aluminum alloying effects on lattice types, microstructures, and mechanical behavior of high-entropy alloys systems. JOM. 2013;65:1848–1858.
[25] Wang W-R, Wang W-L, Wang S-C, et al. Effects of Al addition on the microstructure and mechanical property of Al0.5CoCrFeNi high-entropy alloys. Intermetallics. 2012;26:44–51.
[26] Wang W-R, Wang W-L, Yeh J-W. Phases, microstructure and mechanical properties of Al0.5CoCrFeNi high-entropy alloys at elevated temperatures. J Alloys Compd. 2014;589:143–152.
[27] Rao JC, Diao HY, Ocelik V, et al. Secondary phases in Al0.5CoCrFeNi high-entropy alloys: An in-situ TEM heating study and thermodynamic appraisal. Acta Mater. 2017;131:206–220.
[28] El-Awady JA. Unravelling the physics of size-dependent dislocation-mediated plasticity. Nat Commun. 2015;6:5926.
[29] Zhang H, Schuster BE, Wei Q, et al. The design of accurate micro-compression experiments. Scr Mater. 2006;54:181–186.
[30] Sneddon IN. The relation between load and penetration in the axisymmetric Boussinesq problem for a punch of arbitrary profile. Int J Eng Sci. 1965;3:47–57.
[31] Kumar N, Ying Q, Nie X, et al. High strain-rate compressive deformation behavior of the A10.1CrFeCoNi high entropy alloy. Mater Des. 2015;86:598–602.
[32] Yu Q, Shan Z-W, Li J, et al. Strong crystal size effect on deformation twinning. Nature. 2010;463:335–338.
[33] Sim GD, Kim G, Lavenstein S, et al. Anomalous hardening in magnesium driven by a size-dependent transition in deformation modes. Acta Mater. 2018;144:11–20.
[34] Meyers MA, Vöhringer O, Lubarda VA. The onset of twinning in metals: A constitutive description. Acta Mater. 2001;49:4025–4039.
[35] Stanford N, Carlson U, Barnett MR. Deformation twinning and the Hall–Petch relation in commercial purity Ti. Metall Mater Trans A Phys Metall Mater Sci. 2008;39 A:934–944.
[36] Fan H, Aubry S, Arsenlis A, et al. Grain size effects on dislocation and twinning mediated plasticity in magnesium. Scr Mater. 2016;112:50–53.
[37] Tian F, Delczeg L, Chen N, et al. Structural stability of NiCoFeCrAlx high-entropy alloy from ab initio theory. Phys Rev B. 2013;88:S518.
[38] Varvenne C, Luque A, Curtin WA. Theory of strengthening in fcc high entropy alloys. Acta Mater. 2016;118:164–176.
[39] Varvenne C, Leysen GPM, Ghaziaaedi M, et al. Solute strengthening in random alloys. Acta Mater. 2017;124:660–683.
[40] Ma SG, Zhang SF, Qiao JW, et al. Superior high tensile elongation of a single-crystal CoCrFeNiAlx3 high-entropy alloy by Bridgman solidification. Intermetallics. 2014;54:104–109.