

Research Article

Tensile deformation behavior and mechanical properties of a bulk cast $\text{Al}_{0.9}\text{CoFeNi}_2$ eutectic high-entropy alloyHui Jiang^a, Dongxu Qiao^b, Wenna Jiao^b, Kaiming Han^{a,*}, Yiping Lu^{b,c,**}, Peter K. Liaw^d^a College of Mechanical and Electronic Engineering, Shandong University of Science and Technology, Qingdao 266590, China^b Key Laboratory of Solidification Control and Digital Preparation Technology (Liaoning Province), School of Materials Science and Engineering, Dalian University of Technology, Dalian 116024, China^c State Key Laboratory of Solidification Processing, Northwestern Polytechnical University, Xi'an 710072, China^d Department of Materials Science and Engineering, the University of Tennessee, Knoxville, TN37996, USA

ARTICLE INFO

Article history:

Received 6 April 2020

Received in revised form 8 May 2020

Accepted 12 May 2020

Available online 9 July 2020

Keywords:

Eutectic high-entropy alloy

Microstructure

Mechanical properties

Deformation behavior

ABSTRACT

In this study, a new $\text{Al}_{0.9}\text{CoFeNi}_2$ eutectic high entropy alloy (EHEA) was designed, and the microstructures as well as the deformation behavior were investigated. The bulk cast $\text{Al}_{0.9}\text{CoFeNi}_2$ EHEA exhibited an order face-centered cubic (FCC) (L_{12}) and an order body-centered cubic (B2) dual-phase lamellar eutectic microstructure. The volume fractions of FCC (L_{12}) and B2 phases are measured to be 60 % and 40 %, respectively. The combination of the soft and ductile FCC (L_{12}) phase together with the hard B2 phase resulted in superior strength of 1005 MPa and ductility as high as 6.2 % in tension at room temperature. The $\text{Al}_{0.9}\text{CoFeNi}_2$ EHEA exhibited obvious three-stage work hardening characteristics and high work-hardening ability. The evolving dislocation substructures during uniaxial tensile deformation found that planar slip dominates in both FCC (L_{12}) and B2 phases, and the FCC (L_{12}) phase is easier to deform than the B2 phase. The post-deformation transmission electron microscopy revealed that the sub-structural evolution of the FCC (L_{12}) phase is from planar dislocations to bending dislocations, high-density dislocations, dislocation network, and then to dislocation walls, and Taylor lattices, while the sub-structural evolution of the B2 phase is from a very small number of short dislocations to a number of planar dislocations. Moreover, obvious ductile fracture in the FCC (L_{12}) phase and a brittle-like fracture in the B2 phase were observed on the fracture surface of the $\text{Al}_{0.9}\text{CoFeNi}_2$ EHEA. The research results provide some insight into the microstructure-property relationship.

© 2021 Published by Elsevier Ltd on behalf of The editorial office of Journal of Materials Science & Technology.

1. Introduction

Recently, high entropy alloys (HEAs) [1,2] as a new class of multicomponent metallic materials have received extensive attention due to their attractive properties [3–8]. Although HEAs contain multiple principal elements (at least four elements), they frequently form a simple structure, such as face-centered-cubic (FCC), body-centered-cubic (BCC), and hexagonal-close-packed (HCP) solid-solution structures, or their ordered derivatives. Usually, single-phased FCC structured HEAs are ductile but not strong

enough. Single-phased BCC structured HEAs, on the other hand, can be very strong but at the price of brittleness. Alloys with poor ductility or strength are not suitable for industrial applications. Moreover, the simple structure HEAs usually have the poor castability and compositional segregation, which further reduces their mechanical properties. In order to improve the liquidity and castability and overcome the strength-ductility trade-off, eutectic HEAs (EHEAs) were proposed and produced by Lu et al. via the eutectic alloy concept [9].

After the concept of EHEA was proposed, many EHEAs components have been found [10–24], e.g., $\text{AlCoCrFeNi}_{2.1}$, $\text{CoCrFeNiNb}_{0.45}$, $\text{CoCrFeNiTa}_{0.4}$, $\text{CoCrFeNiZr}_{0.55}$, $\text{CoCrFeNiHf}_{0.4}$, $\text{CoCrFeNi}_2\text{Nb}_{0.74}$, $\text{CoCrFeNi}_2\text{Ta}_{0.65}$, $\text{CoCrFeNi}_2\text{Zr}_{0.6}$, $\text{CoCrFeNi}_2\text{Hf}_{0.55}$, AlCrFeNi_2 , $\text{Ni}_{30}\text{Co}_{30}\text{Cr}_{10}\text{Fe}_{10}\text{Al}_{18}\text{W}_2$, $\text{Al}_{0.7}\text{CoCrFeNi}$, $\text{AlCoCrFe}_2\text{Ni}_2$, $\text{Fe}_{20}\text{Co}_{20}\text{Ni}_{41}\text{Al}_{19}$, and $\text{Nb}_{25}\text{Sc}_{25}\text{Ti}_{25}\text{Zr}_{25}$. Some EHEAs displayed surprisingly high strengths, occasionally in combination with the high ductility and good thermal stability. For example, the $\text{AlCoCrFeNi}_{2.1}$ EHEA exhibits good mechanical properties at both

* Corresponding author.

** Corresponding author at: Key Laboratory of Solidification Control and Digital Preparation Technology (Liaoning Province), School of Materials Science and Engineering, Dalian University of Technology, Dalian 116024, China.

E-mail addresses: hankaiming@126.com (K. Han), luyiping@dlut.edu.cn (L. Yiping).

Table 1
Compositions of the as-cast $\text{Al}_{0.9}\text{CoFeNi}_2$ alloy by EDS (at.%).

Regions	Al	Co	Fe	Ni
FCC	15.29	23.23	23.38	38.1
BCC	27.73	16.53	15.37	40.37

room and elevated temperatures up to 700 °C [9,25]. The tensile fracture stress and elongation at room temperature were 944 MPa and 25.6 %, where tensile fracture stress and elongation at 700 °C were 538 MPa, 22.9 %, respectively. A bulk $\text{CoCrFeNiNb}_{0.45}$ EHEA showed the outstanding thermal stability and high temperature softening resistance up to 1100 °C [26]. Up to now, the composition design, microstructures, and mechanical properties of EHEAs have been attracting much attention. However, the deformation mechanisms of EHEAs were not fully studied.

In the present study, a new FCC ($L1_2$) + B2 two-phase EHEA with excellent mechanical properties was designed. The deformation behavior of the EHEA was investigated by a series of interrupted tests at varying strains during uniaxial tensile tests. Moreover, the sub-structural evolution of both the FCC ($L1_2$) and ordered B2 phases was investigated in this article.

2. Experiment procedures

A bulk $\text{Al}_{0.9}\text{CoFeNi}_2$ EHEA was designed and produced using high-purity raw materials [Al, Co, Fe, Ni $\geq$ 99.5 wt%] by a medium-frequency induction melting furnace under the Ar-atmosphere. Bulk $\text{Al}_{0.9}\text{CoFeNi}_2$ alloy ingots (about 2.5 kg) with a tapered cylinder with a height of 150 mm, top diameter of 55 mm, and bottom diameter of 48 mm were obtained. Some tensile tests were performed up to failure using an 810 Material Test System (MTS, Cary, NC, USA) at a strain rate of 10^{-3} s^{-1} , while others were interrupted at predetermined strains to characterize the sub-structural evolution of both the FCC and BCC phases at different strain levels. The tensile samples have a gauge dimension of $\Phi 8 \text{ mm} \times 40 \text{ mm}$. The crystalline structure was identified by the X-ray diffraction (XRD, Ettlingen, Germany) using a Bruker D8 diffractometer with the $\text{CuK}\alpha$ radiation scanning from 20° to 100° at a scanning speed of 4°/min. The microstructures of the samples were characterized by electron back-scattered diffraction (EBSD) and the Philips Tecnai G^2 transmission electron microscopy (TEM, Amsterdam, Netherlands). The distributions of chemical compositions of the specimens were examined by Zeiss supra55 scanning electron microscope (SEM, Oberkochen, Germany) equipped with an attached X-ray energy dispersive spectrometer (EDS).

3. Results

3.1. Microstructure

Fig. 1(a) shows the XRD pattern of the bulk $\text{Al}_{0.9}\text{CoFeNi}_2$ EHEA. An ordered FCC ($L1_2$) and ordered BCC (B2) dual-phase structure can be found, which is identical to the Cr-containing $\text{AlCoCrFeNi}_{2.1}$ EHEA, indicating that the Cr element in the alloy has no effect on the crystal structure. The microstructures of the $\text{Al}_{0.9}\text{CoFeNi}_2$ EHEA are confirmed by the EBSD phase mapping and TEM-SAED (select area electron-beam diffraction) patterns, and the results are given in Fig. 1(b–e), respectively. From Fig. 1(b), the $\text{Al}_{0.9}\text{CoFeNi}_2$ EHEA exhibits a full eutectic cell morphology. The eutectic cell consisted of a flat structure in the center surrounding by a regular FCC ($L1_2$) and ordered B2 two-phase lamellar structure with a width of 1–2 μm . Moreover, the ordered FCC phase was enriched with Fe and Co elements, while the order BCC phase was enriched with Al and Ni elements based on the SEM-EDS measurements (as shown in Table 1). The volume fractions of FCC ($L1_2$) and B2 phases were

about 60 % and 40 %, respectively. Fig. 1(c–e) shows the TEM micrograph and the SAED patterns of the as-cast alloy. The alloy was composed of an ordered FCC (denoted by A) and ordered BCC (B2) (denoted by B), which are consistent with the XRD results.

3.2. Mechanical properties

Engineering tensile stress-strain curves of the $\text{Al}_{0.9}\text{CoFeNi}_2$ EHEA are presented in Fig. 2(a). It can be found that the alloy had a high yield strength (σ_Y) of 559 MPa, high ultimate tensile strength (σ_{UTS}) of 1005 MPa, and large room-temperature tensile plasticity of 6.2 %. The alloy showed a superb work-hardening ability ($\sigma_{UTS} - \sigma_Y = 446 \text{ MPa}$), which is an effective delay of necking (and failure) and vital for reliability in engineering applications. Therefore, the superb work-hardening capacity accounts for the large ductility. Moreover, the true stress and strain hardening rate ($d\sigma/d\varepsilon$) against true strain curves are given in the Fig. 2(b). From the ($d\sigma/d\varepsilon$)- ε curve, it can be found that the ($d\sigma/d\varepsilon$) first dramatic drop in the range 0–0.009 (stage I), and then slightly decreases in the wide strain range 0.009–0.058 (stage II), finally, sharply falls until tensile failure over the strain of 0.058 (stage III). A similar three-stage work hardening described above is observed both in some HEAs [24,27] and steels [28].

In order to understand the excellent mechanical properties of the $\text{Al}_{0.9}\text{CoFeNi}_2$ EHEA, the fracture surface was observed by SEM (as shown in Fig. 2(c) and (d)). From Fig. 2(c), the fracture surface is mainly composed of the ductility tear edge and elongated groove microstructure, which is similar to the $\text{AlCoCrFeNi}_{2.1}$ EHEA [25]. By connecting the FCC ($L1_2$) and ordered BCC (B2) composite structure in the $\text{Al}_{0.9}\text{CoFeNi}_2$ EHEA, the formation mechanism of these trench-like microstructures can be concluded as follows. During the tensile deformation, the soft FCC phase is stretched, which leads to the FCC ($L1_2$) phase becoming thinner and the edge protruding. Meanwhile, the hard B2 phase with bare deformation was separated from the soft FCC ($L1_2$) phase at the bottom of the groove, which leads to the formation of this slender groove structure. It is suggested that the elongated groove microstructure making much energy was absorbed during ductile failure, which plays significant roles on the fracture toughness.

In order to further analyze the fracture model of the alloy, the high magnification SEM image of the fracture surfaces is shown in Fig. 2(d). There are some small holes on the fracture surface (marked with red arrows). Some of the holes were produced during the casting process of the alloy. The others were caused by the plasticity mismatch between the FCC ($L1_2$) and B2 phases in the alloy. In these critical interfaces, voids prefer to nucleate, accumulate and then grow into micro-cracks. Several micro-cracks found in the present EHEA should be classified as follows. One was the crack nucleated at the end of the B2 phase and then propagated (marked with red solid lines), in a certain direction. This trend is similar to the fracture mechanism of the B2 phase in the $\text{AlCoCrFeNi}_{2.1}$ EHEA [29]. Another was the crack nucleated at the boundaries of FCC ($L1_2$) and B2 phases and then propagated (marked with a red box) in the direction of 45°. The B2 phase was broken into several parts by the crack resulting in an early shear fracture of the B2 phase. With the increase of the strain, the deformation inconsistency of FCC ($L1_2$) and B2 phases increased, resulting in the two-phase separation at the phase interface. Thus, the cracks at the FCC ($L1_2$) and B2 phase boundaries will lead to an ultimate fracture. Moreover, the river pattern appears in the B2 phase (shown in the blue circle), which indicates that the fracture mechanism of the B2 phase is of a brittle type. In summary, the coupling growth of soft FCC and hard B2 phases contributed to the excellent mechanical properties of the $\text{Al}_{0.9}\text{CoFeNi}_2$ EHEA.

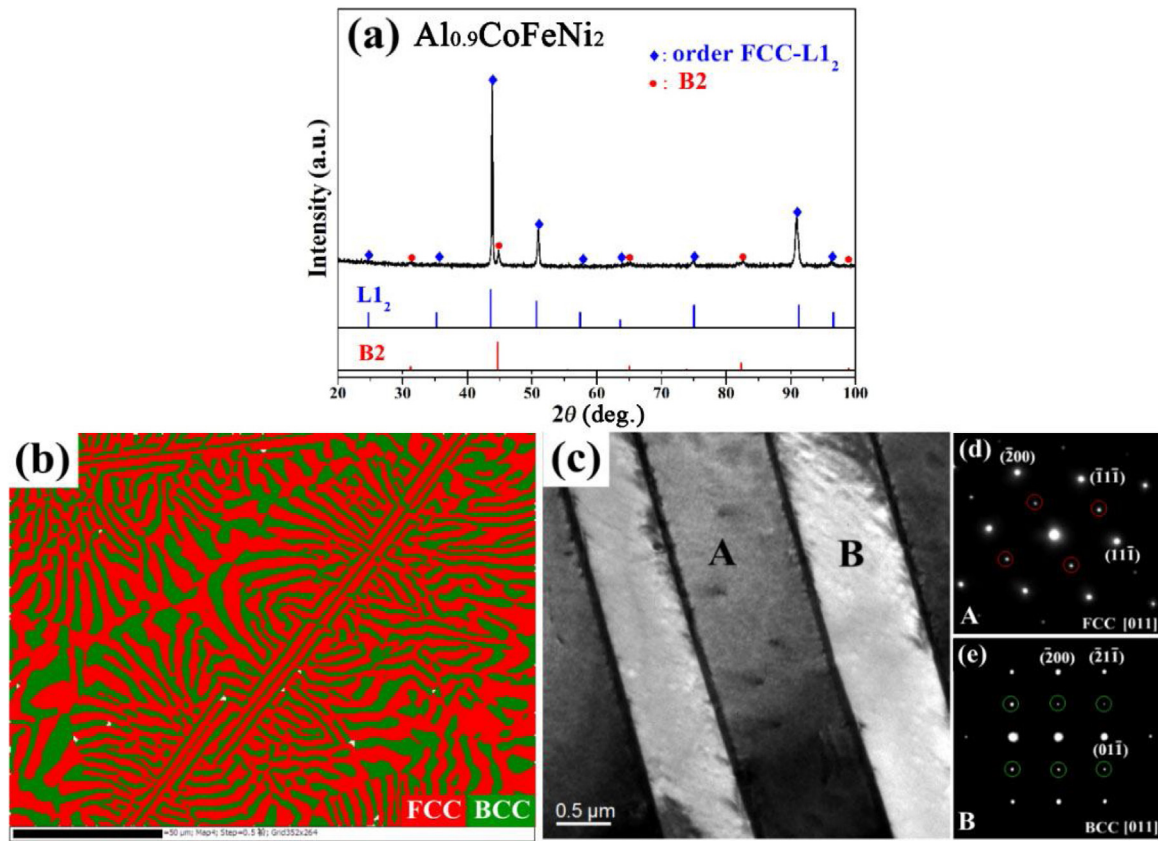


Fig. 1. (a) XRD pattern, (b) EBSD phase mapping and (c-e) TEM micrographs and SAED patterns.

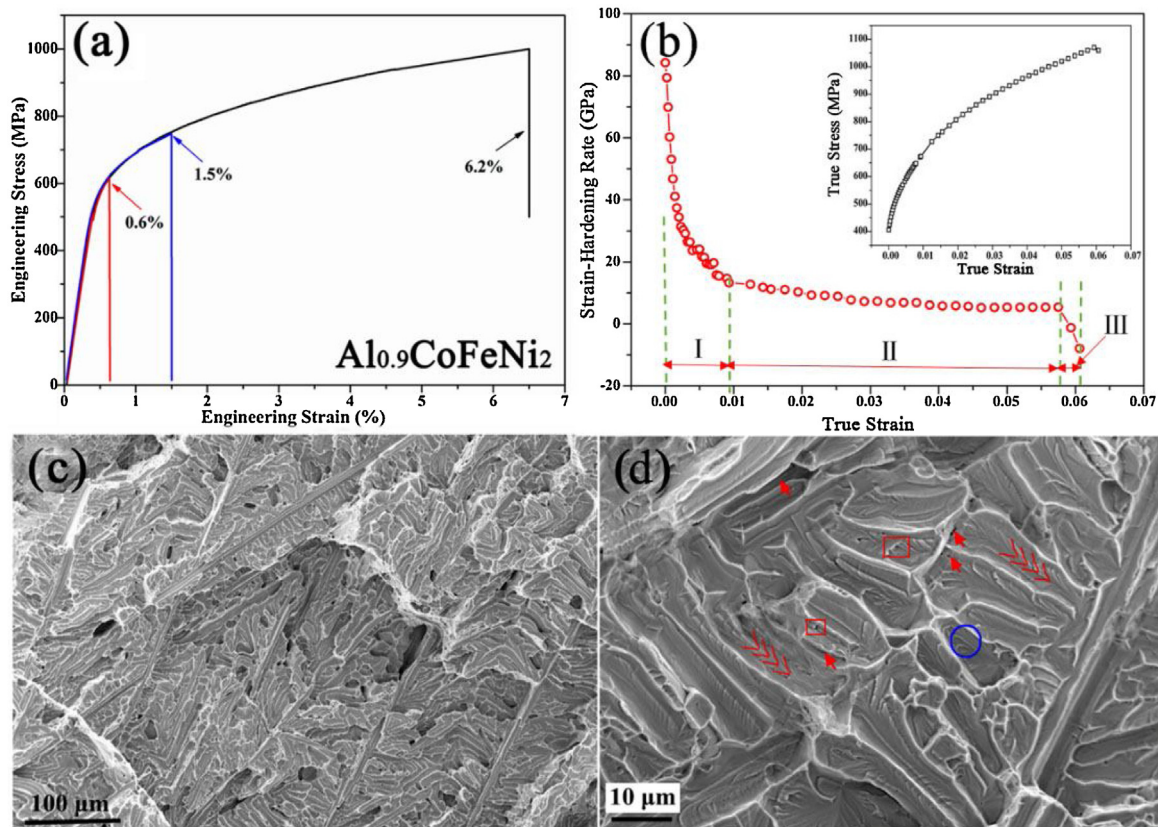


Fig. 2. (a) Tensile stress-strain curve of the $\text{Al}_{0.9}\text{CoFeNi}_2$ EHEA, (b) true stress and strain hardening rate ($d\sigma/d\varepsilon$) against true strain curves of the $\text{Al}_{0.9}\text{CoFeNi}_2$ EHEA and (c) low and (d) high magnification SEM images of the tensile fracture surface of the $\text{Al}_{0.9}\text{CoFeNi}_2$ EHEA at room temperature.

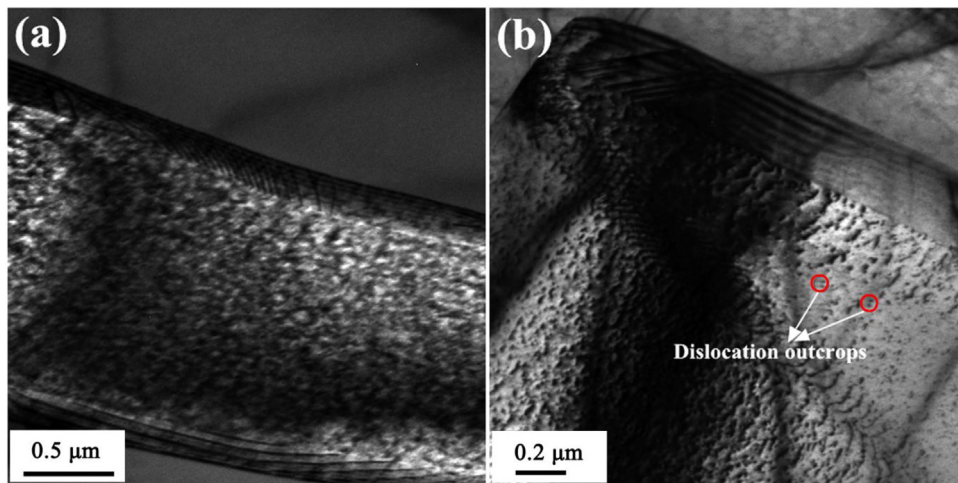


Fig. 3. TEM micrographs of the $\text{Al}_{0.9}\text{CoFeNi}_2$ alloy at a strain of 0.6 % during tensile deformation: (a) dislocation pileup and (b) dislocations configurations in the FCC (L_{12}) phase.

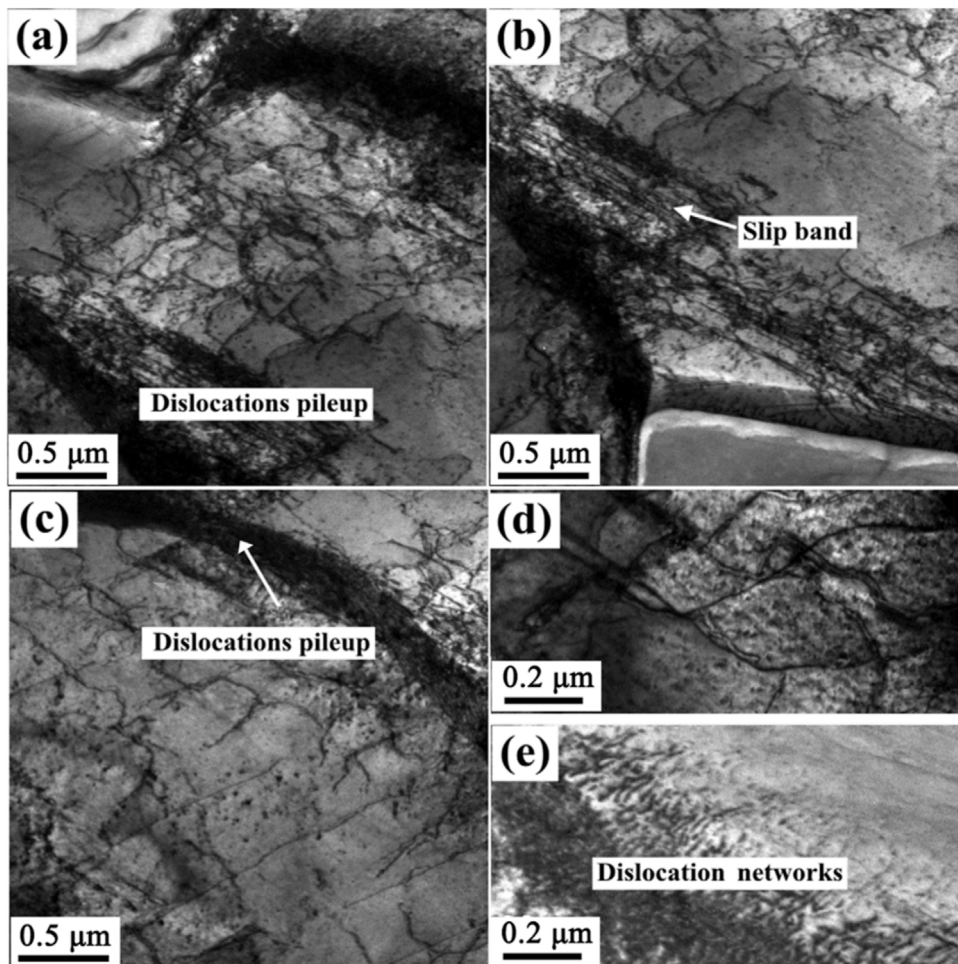


Fig. 4. TEM micrographs of the $\text{Al}_{0.9}\text{CoFeNi}_2$ alloy at a strain of 1.5 % during tensile deformation: (a) dislocations configurations in the FCC (L_{12}) phase; (b) slip band; (c) dislocation pileup; (d) intersecting dislocation; (e) dislocation networks.

3.3. The sub-structural evolution during tensile deformation

The dynamic evolution of deformation substructures at different strains with TEM (Figs. 3–5) were carefully investigated to decipher the deformation characteristics of FCC (L_{12}) and B2 phases in the $\text{Al}_{0.9}\text{CoFeNi}_2$ EHEA. At an initial deformation of 0.6 %, a number of

dislocations occur in the soft FCC (L_{12}) phase, while no obvious dislocations are observed in the hard/brittle B2 phase (see Fig. 3(a)). This feature indicated that dislocation movement is more active in the FCC (L_{12}) phase than that in the B2 phase. As presented in Fig. 3(b), the FCC (L_{12}) phase exhibited obvious planar deformation characteristics, such as the dislocation planar slip and dislocation

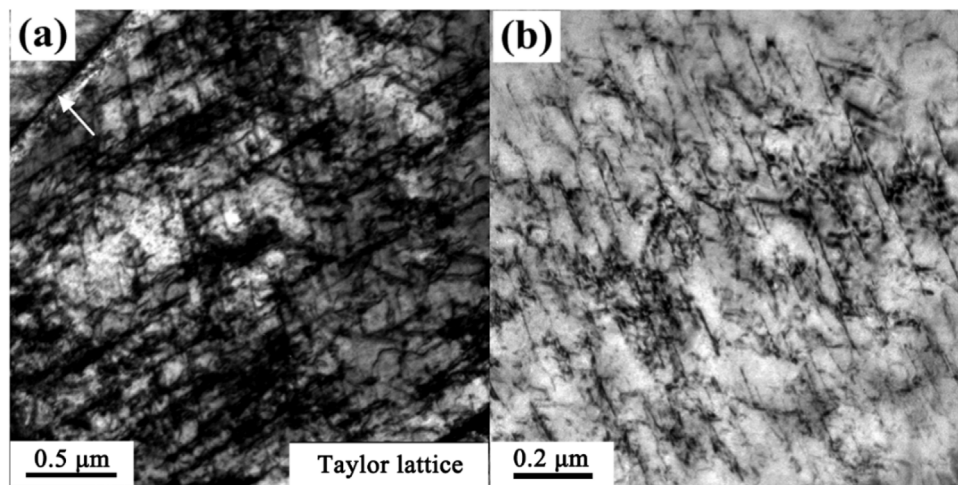


Fig. 5. TEM micrographs of the $\text{Al}_{0.9}\text{CoFeNi}_2$ alloy after tensile fracture: (a) dislocations configurations in the FCC (L_{12}) phase; (b) dislocations configurations in the B2 phase.

outcrop (marked with a red circle). Furthermore, stacking dislocations are also seen along the interface of FCC (L_{12})/B2 phases, which is due to the large internal stress caused by the lattice mismatch between FCC (L_{12}) and B2 phases. In order to release this internal stress, dislocations pile up at the interface. The block-up of dislocations at the FCC (L_{12})/B2 phase boundaries provides the EHEA with the high yield strength. It can be seen that in the initial deformation stage, dislocation slip first occurs in the soft FCC (L_{12}) phase, indicating that the dislocation planar slip in the FCC (L_{12}) phase contributed to the deformation of the early stage of the $\text{Al}_{0.9}\text{CoFeNi}_2$ EHEA.

As the deformation amount increases to $\sim 1.5\%$, still no obvious dislocation are observed in the hard/brittle B2 phase, while the dislocations density is significantly increased in the FCC (L_{12}) phase (shown in Fig. 4(a)). Meanwhile, some planar dislocation configurations, for instance, dislocation slip band (shown in Fig. 4(b)), dislocation build-up (presented in Fig. 4(c)), dislocation intersection (exhibited in Fig. 4(d)) dislocation networks (represented in Fig. 4(e)) are observed in the FCC (L_{12}) phase. These configurations appear only when two or more coplanar slip systems act in one grain and are usually found in some low-SFE FCC metals, indicating that the cross slip is quite difficult [28,30]. With the increase of the plastic strain, the dislocation in FCC phase is hardened and the dislocation gradually accumulates. Hence, it is necessary for the greater stress to promote the dislocation to continue to move. In order to adapt to the increase of the stress, the following two phenomena may occur in the FCC phase, one is to cause a more slip system to start, and the other is that the deformation of the alloy extends to other regions, which leads to the formation of slip bands. It is illustrated that the above complex dislocation morphology appears in the FCC (L_{12}) phase is beneficial to adapt to the increase of the stress.

With the further deformation to fracture for the $\text{Al}_{0.9}\text{CoFeNi}_2$ EHEA, a fast dislocation multiplication takes place in both FCC (L_{12}) and B2 phases. The density of dislocations is higher in the FCC (L_{12}) phase than in the B2 phase at this deformation level. As shown in Fig. 5(a), the FCC (L_{12}) phase exhibits a dense dislocation network arrangement formed by typical Taylor lattices, in which the parallel dislocation configuration is called the dislocation wall. The formation of high dense dislocation walls is mainly attributed to the increasing stress concentration, thus improving the tensile strength of the $\text{Al}_{0.9}\text{CoFeNi}_2$ EHEA. Moreover, the interaction of different slip systems in the dislocation wall also improves the strain hardening ability, delays the occurrence of necking, and increases the plasticity of the $\text{Al}_{0.9}\text{CoFeNi}_2$ EHEA. At the same time, grain boundaries

are known to constitute strong obstacles to the dislocation motion, resulting in the formation of dislocation pile-ups (denoted by the arrow in Fig. 5(a)). With the increase of deformation, the dislocation density at the two FCC (L_{12}) and B2 phases boundaries increases, resulting in the increased internal stress. When the internal stress reaches the yield strength of the B2 phase, the B2 phase will begin to experience plastic deformation. Thus, the straight and parallel dislocation morphology can be seen in the B2 phase at this deformation level, as shown in Fig. 5(b). Moreover, it can also be seen that the dislocation is transferred from the interface to the B2 phase, and the stress concentration of FCC (L_{12})/B2 boundaries is released, which makes the alloy experience larger strain. In general, when the deformation reaches fracture, dislocation slip occurred in both FCC (L_{12}) and B2 phases, which contributed to the strength and plasticity of the $\text{Al}_{0.9}\text{CoFeNi}_2$ EHEA.

With increasing the amount of deformation, the main sequence of the sub-structural evolution could be summarized as follows. In the FCC (L_{12}) phase, the dislocation changes from planar dislocations to bending dislocation, high-density dislocations, dislocation networks, then to dislocation walls and Taylor lattices. In the B2 phase, the evolution of dislocation is from a very small number of short dislocations to a number of planar dislocations.

For the $\text{Al}_{0.9}\text{CoFeNi}_2$ EHEA, the three-stage work hardening characteristics can be explained by the sub-structural evolution during tensile deformation. In stage I, there existed a deformation delay in the B2 phase, while the FCC (L_{12}) phase controlled the deformation of the alloy. And obvious planar deformation characteristics is observed in the FCC (L_{12}) phase. In stage II, the larger deformation occurred in the FCC (L_{12}) phase, resulting the significantly increase of dislocations density and formation of different dislocation configurations, such as dislocation slip band, dislocation build-up, dislocation intersection, dislocation networks. The formed above complex dislocation configurations in the FCC (L_{12}) phase required higher stress to move further. Moreover, the increased dislocations density indicated the FCC (L_{12}) phase underwent a work hardening process at this deformation stage, and then causing the increase in the work hardening rate. In a word, in the stage I and II, the continuously reduced work-hardening rate is due to dislocation-control of plastic deformation processes.

In stage III, straight and parallel dislocation dominates the deformation in the B2 phase, while Taylor lattice and dislocation wall appeared in the FCC (L_{12}) phase. The high density dislocation network is difficult to adapt to further strain. Thus, deformation inconsistency of FCC (L_{12}) and B2 phases increased, leading to the micro-cracks form along two-phase boundaries, and then propa-

gated resulting in the ultimate fracture. Thus the sharply falls of the work hardening rate could be attributed to the serious strain localization.

4. Conclusions

In the present study, the bulk cast $\text{Al}_{0.9}\text{CoFeNi}_2$ EHEA exhibiting high strength/ductility was designed and prepared, and the corresponding evolving dislocation substructures as well as the deformation behavior were investigated. The following conclusions can be drawn.

- (1) The as-cast bulk $\text{Al}_{0.9}\text{CoFeNi}_2$ HEA exhibits a characteristic of the full eutectic cell morphology, where a flat structure appears in the eutectic cell center surrounding by a regular lamellar structure. Furthermore, the two lamellar eutectic phases belong to FCC (L_{12}) and B2 structures, respectively. The width of the FCC (L_{12}) and B2 eutectic lamellae is 1–2 μm , and the volume fractions of FCC (L_{12}) and B2 phases are measured as 60 % and 40 %, respectively.
- (2) The as-cast $\text{Al}_{0.9}\text{CoFeNi}_2$ EHEA exhibits the superior strength of 1005 MPa and ductility as high as 6.2 % in tension at room temperature. The excellent mechanical properties originated from the coupling effect between the ductile FCC (L_{12}) and brittle B2 phases during tensile deformation.
- (3) The $\text{Al}_{0.9}\text{CoFeNi}_2$ EHEA exhibits obvious three-stage work hardening characteristics. In stage I and II, the continuously decrease in work-hardening rate is mainly attributed to the dislocation-control of plastic deformation in the FCC (L_{12}) phase. In stage III, the sharply falls of the work hardening rate could be attributed to the serious strain localization in the FCC (L_{12}) and B2 phases.
- (4) The analyses of deformation substructures at different strains revealed that planar slips dominates in both FCC (L_{12}) and B2 phases, and the FCC (L_{12}) phase is easier to deform. The sub-structural evolution of the FCC (L_{12}) phase is from planar dislocations to bending dislocations, high-density dislocations, dislocation networks, then to dislocation walls, and Taylor lattices. The sub-structural evolution of the B2 phase is from a very small number of short dislocations to a number of planar dislocations. The formation of these substructures in both FCC (L_{12}) and B2 phases contributed to the high strength and larger plasticity of the $\text{Al}_{0.9}\text{CoFeNi}_2$ eutectic high entropy alloy.

Acknowledgments

The present work was supported financially by the National Natural Science Foundation of China (Nos. 51901116, 51822402 and 51671044), the National Key Research and Development Program of China (Nos. 2019YFA0209901 and 2018YFA0702901), the Fund of the State Key Laboratory of Solidification Processing in NWPU (No. SKLSP201902), the Liao Ning Revitalization Talents

Program (No. XLYC1807047), the National MCF Energy R&D Program (No. 2018YFE0312400), the Fund of Science and Technology on Reactor Fuel and Materials Laboratory (No. STRFML-2020-04). P.K. Liaw thanks (1) the U.S. Army Research Office for the support of the present work through projects Nos. W911NF-13-1-0438 and W911NF-19-2-0049, and (2) the National Science Foundation for the support of the present work through projects Nos. DMR-1611180 and 1809640.

References

- [1] J.W. Yeh, S.K. Chen, S.J. Lin, J.Y. Gan, T.S. Chin, T.T. Shun, C.H. Tsau, S.Y. Chang, *Adv. Eng. Mater.* 6 (2004) 299–303.
- [2] B. Cantor, I.T.H. Chang, P. Knight, A.J.B. Vincent, *Mater. Sci. Eng. A* 375–377 (2004) 213–218.
- [3] Z. Li, S. Zhao, R.O. Ritchie, M.A. Meyers, *Prog. Mater. Sci.* 102 (2019) 296–345.
- [4] M.C. Gao, J.W. Yeh, P.K. Liaw, Y. Zhang, *High Entropy Alloys Fundamentals and Applications*, Springer International Publishing, Switzerland, 2016.
- [5] Y. Zhang, T.T. Zuo, Z. Tang, M.C. Gao, K.A. Dahmen, P.K. Liaw, Z.P. Lu, *Prog. Mater. Sci.* 61 (2014) 1–93.
- [6] Z. Li, K.G. Pradeep, Y. Deng, D. Raabe, C.C. Tasan, *Nature* 534 (2016) 227–230.
- [7] Q. Yu, Z.W. Shan, J. Li, X. Huang, L. Xiao, J. Sun, E. Ma, *Nature* 463 (2010) 335–338.
- [8] B. Gludovatz, A. Hohenwarter, D. Catoor, E.H. Chang, E.P. George, R.O. Ritchie, *Science* 345 (2014) 1153–1158.
- [9] Y.P. Lu, Y. Dong, S. Guo, L. Jiang, H.J. Kang, T.M. Wang, B. Wen, Z.J. Wang, J.C. Jie, Z.Q. Cao, H.H. Ruan, T.J. Li, *Sci. Rep.* 4 (2014) 6200.
- [10] S. Guo, C. Ng, C.T. Liu, *J. Alloys. Compd.* 557 (2013) 77–81.
- [11] H. Jiang, L. Jiang, D.X. Qiao, Y.P. Lu, T.M. Wang, Z.Q. Cao, T.J. Li, *J. Mater. Sci. Technol.* 33 (2017) 712–717.
- [12] H. Jiang, K.M. Han, X.X. Gao, Y.P. Lu, Z.Q. Cao, M.C. Gao, J.A. Hawk, T.J. Li, *Mater. Des.* 142 (2018) 101–105.
- [13] Y.P. Lu, H. Jiang, S. Guo, T.M. Wang, Z.Q. Cao, T.J. Li, *Intermetallics* 91 (2017) 124–128.
- [14] H. Jiang, H.Z. Zhang, T.D. Huang, Y.P. Lu, T.M. Wang, T.J. Li, *Mater. Des.* 109 (2016) 539–546.
- [15] Y.M. Tan, J.S. Li, J. Wang, H.C. Kou, *Intermetallics* 85 (2017) 74–79.
- [16] B. Gwalani, S. Gangireddy, Y. Zheng, V. Soni, R.S. Mishra, R. Banerjee, *Sci. Rep.* 9 (2019) 6371.
- [17] X. Chen, J.Q. Qi, Y.W. Sui, Y.Z. He, F.X. Wei, Q.K. Meng, Z. Sun, *Mater. Sci. Eng. A* 681 (2017) 25–31.
- [18] X. Jin, Y. Zhou, L. Zhang, X.Y. Du, B.S. Li, *Mater. Des.* 143 (2018) 49–55.
- [19] X. Jin, Y. Zhou, L. Zhang, X.Y. Du, B.S. Li, *Mater. Lett.* 216 (2018) 144–146.
- [20] Q.F. Wu, Z.J. Wang, T. Zheng, C. Da, Z.S. Yang, J.J. Li, J.J. Kai, J.C. Wang, *Mater. Lett.* 253 (2019) 268–271.
- [21] Ł. Rogal, J. Morgiel, Z. Świątek, F. Czerwiński, *Mater. Sci. Eng. A* 651 (2016) 590–597.
- [22] F. He, Z.J. Wang, P. Cheng, Q. Wang, J.J. Li, Y.Y. Dang, J.C. Wang, C.T. Liu, *J. Alloys. Compd.* 656 (2016) 284–289.
- [23] W.Y. Huo, H. Zhou, F. Fang, Z.H. Xie, J.Q. Jiang, *Mater. Des.* 134 (2017) 226–233.
- [24] P. Shi, W. Ren, T. Zheng, Z. Ren, X. Hou, J. Peng, P. Hu, Y. Gao, Y. Zhong, P.K. Liaw, *Nat. Commun.* 489 (2019) 1–5.
- [25] Y.P. Lu, X.Z. Gao, L. Jiang, Z.N. Chen, T.M. Wang, J.C. Jie, H.J. Kang, Y.B. Zhang, S. Guo, H.H. Ruan, Y.H. Zhao, Z.Q. Cao, T.J. Li, *Acta Mater.* 124 (2017) 143–150.
- [26] H. Jiang, D.X. Qiao, Y.P. Lu, Z. Ren, Z.Q. Cao, T.M. Wang, T.J. Li, *Scr. Mater.* 165 (2019) 145–149.
- [27] T. Yang, Y.L. Zhao, Y. Tong, Z.B. Jiao, J. Wei, J.X. Cai, X.D. Han, D. Chen, A. Hu, J.J. Kai, K. Lu, Y. Liu, C.T. Liu, *Science* 362 (2018) 933–937.
- [28] L. Zhang, R. Song, C. Zhao, F. Yang, *Mater. Sci. Eng. A* 640 (2015) 225–234.
- [29] X.Z. Gao, Y.P. Lu, B. Zhang, N.N. Liang, G.Z. Wu, G. Sha, J.Z. Liu, Y.H. Zhao, *Acta Mater.* 141 (2017) 59–66.
- [30] X. Feaugas, *Acta Mater.* 47 (1999) 3617–3632.