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Ultrahigh strength triggered by BCC and B2 eutectic-phase interfaces in a novel FeCrVNiAl high-entropy alloy

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ABSTRACT

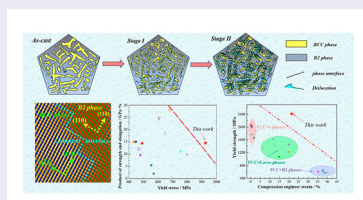
The strength–ductility tradeoff and composition homogeneity have been the obstacles to obtaining advanced structural alloys. To overcome the limitations, we develop a novel coherently eutectic high-entropy alloy, i.e. Fe₃₀Cr₁₅V₁₅Ni₂₀Al₂₀. By introducing a BCC phase with a low lattice misfit (0.13%) and constructing the BCC/B2 eutectic-phase interfaces, the alloy shows high mechanical stability. The low lattice mismatch contributes to the good stability of the phase interfaces and inhibits crack initiation. Thus, the as-cast Fe₃₀Cr₁₅V₁₅Ni₂₀Al₂₀ alloy shows an excellent tensile yield strength of 929 MPa and a high yield compressive strength of 2423 MPa. In addition, it shows an excellent compressive strain of 22.3%.

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KEYWORDS

Coherent eutectic high-entropy alloy; alloy design; coherent interface; mechanical properties



1. Introduction

Eutectic alloys belong to an important group of materials in engineering applications owing to their high rupture strength [1], excellent liquidity and castability [2]. Recently, massive research has been focused on developing a variety of eutectic systems [2,3]. However, only a few eutectic alloys are commercially significant, as more of them have shown weak deformability and have limited toughening mechanisms. High-entropy alloys (HEAs) represent an important break-through in alloy design, providing rich likelihoods for expanding the chemical composition of eutectic alloys. Lu et al. [4] first proposed the concept of eutectic high-entropy alloys (EHEAs), and achieved a good balance between strength and ductility.

EHEAs [4–7], as a branch of HEAs [8–10], concurrently have the advantages of HEAs and eutectic alloys. To date, numerous EHEA systems have been developed, and they are roughly classified into three types. The first type of EHEAs mainly consists of the face-centered-cubic (FCC) and B2 diphasic, such as AlCoCrFeNi_{2.1} [4],

Al_{19.3}Co₁₅Cr₁₅Ni_{50.7} [11], Al₁₉Co₂₀Fe₂₀Ni₄₁ [12] and Ni₄₉Fe₂₈Al₁₇V₆ [6]. This kind of EHEA generally has fine lamellar microstructure and excellent tensile behavior. The lamellar microstructure can be regarded as a kind of heterostructure [13], and the high density of lamellar interfaces is conducive to triggering large strain gradients during deformation. Thus, the work-hardening ability of EHEAs is effectively enhanced. Consequently, AlCoCrFeNi_{2.1} exhibits a yield strength of 610 MPa and an ultimate tensile strength of 1100 MPa, while retaining an excellent ductility of 17% at room temperature [14]. The excellent tensile strength of the above duplex EHEAs mainly stems from the strong back-stress effect.

The second type of EHEAs consists of the FCC and uncommon intermetallic compounds (IMCs), such as Laves phase [15], Ni₇Hf₂ [15], (Co, Ni)₂Ta [16], and (Co, Ni)₂Nb phases [17]. This type of EHEAs includes CoCrFeNiCuNb [17], CoFeNi_{1.5}VZr_x [18], CoCrFeNiZr_xHf_x [15], CoCrFeNi₂V_{0.89}B_{0.45}Si_{0.15} [19], etc. They show the high hardness (~ 5.42 GPa) and compressive yield

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strength (1925 MPa) [19], mainly resulting from the complex crystal structures of IMCs [5,19–21]. However, the crystal structures of IMCs are generally unstable and easy to transit to other structures under loading and heat treatment. Also, the phase interfaces between FCC and IMCs have high lattice misfit, which leads to less independent slip systems and makes the EHEAs brittle [22]. This is why it is difficult to test the tensile strength of EHEAs. It is known that similar crystal structures and low lattice misfits in eutectic phases can decrease the nucleation barrier, thus enhancing the stability of the eutectic phase interface [23]. The other type of EHEAs mainly consists of body-centered-cubic (BCC) and B2 diphasic, such as $\text{Al}_{0.75}\text{CrFeNi}$ [24], $\text{Al}_{1.25}\text{CrFeNi}$ [25] and $\text{Al}_2\text{CrCuFeNi}_2$ [26]. They usually consist of a B2 sunflower core and alternating BCC + B2 dendritic eutectic structures, in which the dendritic eutectic structure dominates mechanical properties. However, the presence of sunflower core structure (i.e. B2 phase) distinctly induces the brittleness of EHEAs. When the lattice mismatch between BCC and B2 phases is significantly decreased to 0.2%, the sunflower core structure almost completely disappears while the dual phases are distributed more evenly [27]. Consequently, the TiZrVNbAl HEA with a low lattice mismatch (0.17%) between BCC and B2 displays a high yield strength of 1072 MPa and good ductility of 16%. Evidently, the decreased lattice mismatch can significantly improve the mechanical property of HEAs with the BCC matrix.

In this study, a newly coherent eutectic high-entropy alloy (CEHEA, $\text{Fe}_{30}\text{Cr}_{15}\text{V}_{15}\text{Ni}_{20}\text{Al}_{20}$) with a low lattice misfit was developed. In general, B2 and BCC phases have similar crystal structures and a small difference in the lattice parameters, which is essential for constructing a coherent phase interface. Hence, the stable $\text{Fe}_{50}\text{Cr}_{25}\text{V}_{25}$ -BCC and NiAl -B2 phases are selected to form the coherent phase. Subsequently, the eutectic point between the two phases is determined by using the ThermoCalc software, and the results are shown in Figure S1d (Supplementary data). Thus, a full eutectic $\text{Fe}_{30}\text{Cr}_{15}\text{V}_{15}\text{Ni}_{20}\text{Al}_{20}$ CEHEA is designed (see the Supplementary data for more detail). Significantly, the alloy exhibits a superior balance between tensile strength and ductility. We conceive that the design strategy will explore a novel avenue for developing a new class of HEAs with advanced properties.

2. Materials and methods

The CEHEA with the chemical compositions of $\text{Fe}_{30}\text{Cr}_{15}\text{V}_{15}\text{Ni}_{20}\text{Al}_{20}$ was prepared by using arc-melting, and high-purity raw materials (> 99.9 wt.%) were used in an argon atmosphere. A Ti ingot was put in a separate

pot to reduce the oxygen content in the furnace. The thermal properties of the CEHEA were investigated by differential scanning calorimetry (DSC) with 96 Line equipment (SETARAM instrumentation, France) operated in an N_2 atmosphere from 50 to 1400 °C at a rate of $0.17\text{ }^\circ\text{C}\cdot\text{s}^{-1}$. Chemical compositions were detected by an electron probe microanalyzer (EPMA, JXA-8530F, JEOL). Microstructures were characterized, utilizing a JEOL JSM-7001 F scanning electron microscope (SEM) equipped with an energy-dispersive X-ray spectrometer (EDS, Oxford Ltd.) and the HKL electron-backscatter-diffraction (EBSD) detector with an EDS, where the samples were prepared by argon ion beam polishing. This is because, after argon ion polishing, EBSD can effectively identify B2 phases [28] (experimental information can be seen in the Supplementary data). Transmission electron microscopy (TEM) observations were conducted on a JEOL JEM-2100 F TEM in combination with an EDS, operated at 200 kV. Samples for TEM observations were mechanically polished to a thickness of $\sim 45\text{ }\mu\text{m}$ and then further thinned, using a Precision ion polisher (Gatan 691). Digital-micrograph software was applied to perform high-angle annular dark field scanning (HAADF), STEM-EDS, high-resolution TEM (HRTEM), Fast Fourier transform (FFT) and inverse Fast Fourier transform (IFFT). Strain++ software was utilized to perform geometric phase analysis (GPA).

To conduct tensile tests, dog-bone-shaped specimens with a dimension of $15 \times 5 \times 1.5\text{ mm}^3$ were cut from the middle of the as-prepared CEHEA by using an electrical discharge machine (the gauge size is 5 mm). Chong et al. [29] indicated that tensile properties obtained from the micro-tensile specimens were comparable to those acquired from the specimens with standard dimensions. The cylinder specimens with a dimension of $\Phi 5\text{ mm} \times 10\text{ mm}$ were cut for the uniaxial compression tests. The uniaxial tensile and compressive tests were performed on an AG-X plus pc-controlled mechanical testing system (Shimadzu Co. Ltd, Japan) with a constant strain rate of $1.0 \times 10^{-3}\text{ s}^{-1}$. Three parallel samples were tested and the average values were used to evaluate the mechanical performances. Nanoindentations were performed on a Hysitron Incorporated Tribo-indenter at room temperature (experimental information can be found in the Supplementary data).

3. Results

3.1. Microstructures and phase components of $\text{Fe}_{30}\text{Cr}_{15}\text{V}_{15}\text{Ni}_{20}\text{Al}_{20}$ CEHEA

The phase distribution map (Figure 1a) obtained from EBSD shows that the as-cast $\text{Fe}_{30}\text{Cr}_{15}\text{V}_{15}\text{Ni}_{20}\text{Al}_{20}$

CEHEA mainly has a lamellar microstructure, exhibiting the maze-like morphology. This is because the laths of BCC and B2 phases are alternatively arranged, with an average thickness of 1 μm . The area fractions of BCC and B2 phases are 59.5% and 40.5%, respectively, which agrees well with the prediction (Figure S1d). In general, the maze-like morphology is a characteristic of spinodal decomposition (SD). In the $\text{Fe}_{30}\text{Cr}_{15}\text{V}_{15}\text{Ni}_{20}\text{Al}_{20}$ alloy, an eutectic reaction first occurs and then SD is triggered. Hence, the specimen is still regarded as a eutectic HEA (see the Supplementary data in detail). Meanwhile, only one apparent endothermic peak and one exothermic peak are identified during either heating or cooling (Figure 1b). This means that the alloy only undergoes one phase transition, corresponding to the eutectic reaction between the BCC and B2 phases.

Besides, EPMA is performed to reveal the compositional distribution of the alloy in different phase regions. Elemental-mapping images showing the distributions of Cr, Fe, V, Al and Ni are shown in Figure 1(c–h). Combining the results of EBSD, the BCC lamellar is rich in Cr, Fe and V elements, while the B2 lamellar is enriched by Al and Ni elements. Moreover, as channels to long-range diffusion, Cr and V promote the diffusion of the main formation elements (Ni, Al) of the B2 phase [30]. Thus, these results convincingly reveal that $\text{Fe}_{30}\text{Cr}_{15}\text{V}_{15}\text{Ni}_{20}\text{Al}_{20}$ CEHEA consists of BCC (rich Cr-Fe-V) and B2 (rich Ni-Al) phases. The HAADF image (Figure 1i) demonstrates an alternating eutectic microstructure consisting of regions A and B, and the STEM-EDS analysis shows that Fe, Cr and V are enriched in region A, while Al and Ni are enriched in region B. The corresponding SAED patterns of region A taken along the (100) and (110) zone axes are shown in Figure 1(j and k), respectively. Meanwhile, the SAED patterns of region B taken along the (100) and (110) axes are shown in Figure 1(j1, k1) to confirm the B2 crystal structure. Compared to the SAED patterns taken along the (100) and (110) axes for region A (Figure 1j and k), superlattice-ordered diffraction spots can be found in region B (Figure 1j1, k1, indicated by red arrows). The position of these spots usually corresponds to the ordered B2 phase crystal structure. According to HADDF and SAED images, both regions are actually composed of FeCrV-BCC and NiAl-B2 phases.

Systematic HRTEM observations were conducted to explore the microstructures and phase components of the alloy (Figure 2). A HRTEM image of the interface region is shown in Figure 2(a), and a BCC/B2 coherent interface can be observed clearly, where the crystal lattice of the ordered B2 phase is very close to the disordered BCC phase. To see more clearly, the corresponding schematic illustration of the coherence interface between

BCC and B2 phases is given in Figure S3. In addition, Figure 2(b) shows the GPA image deduced from HRTEM (Figure 2a), which clearly demonstrates the uniform strain distributions in the BCC and B2 phases. In addition, the corresponding SAED images taken along the (110) and (100) zone axes indicate that the B2 phase is coherent with the BCC matrix (Figure 2c, d). The IFFT image (Figure 2e) deduced from the HRTEM (Figure 2a) displays that atomic arrangement at the phase interfaces is completely coherent on the (110) and ($\bar{1}\bar{1}0$) planes. Meanwhile, the lattice parameters of the BCC and B2 phases obtained from the IFFT image are identical, with a value of 0.28 nm (Figure 2e1, e2).

The lattice mismatch (δ) between the two phases is determined to be 0.13% (Table S1, supplementary materials). There is a coherent orientation relationship between the B2 phase and the matrix, i.e. $110_{\text{BCC}}//110_{\text{B2}}$ and $\langle 100 \rangle_{\text{BCC}} // \langle 100 \rangle_{\text{B2}}$, which is well consistent with that reported in [31,32]. The corresponding FFT patterns taken along the 001 zone axes for B2 and BCC phases are shown in Figure 2(e3 and e4). The strain distributions of the dual phases are shown in Figure 2(f–g). In order to further determine the strain distribution in the BCC and B2 phases, the detailed strain maps in different directions for the different phases of HRTEM are shown in Figure 2(a). Here, the interface direction is selected as the x direction while the vertical orientation to the interface is regarded as the y direction. The surface strain distributions from the ϵ_{xx} surface show that the two phases have relatively uniform strain. Moreover, strain of the both phases is also uniform, and no significant difference in the ϵ_{yy} and ϵ_{xy} surfaces. This implies that plastic deformation can occur simultaneously in the duplex-phase region, leading to uniform and continuous work hardening during plastic deformation.

3.2. Mechanical properties

Figure 3(a) shows the engineering stress–strain curve of the as-cast $\text{Fe}_{30}\text{Cr}_{15}\text{V}_{15}\text{Ni}_{20}\text{Al}_{20}$ CEHEA obtained from tensile testing, which exhibits a yield strength of 929 MPa and an ultimate tensile strength of 1142 MPa, respectively, accompanied by a total elongation of 6.8%. Hence, we construct a map of the product of ultimate tensile strength and total elongation vs. yield strength, as shown in Figure 3(b). The purpose is to measure the energy absorption capacity, which demonstrates that the present alloy has superior tensile performances, compared with the as-cast EHEAs [33–41]. For example, the highest product of strength and ductility was 27.7 GPa·% for the $\text{Ni}_{40}\text{Co}_{20}\text{Fe}_{10}\text{Cr}_{10}\text{Al}_{18}\text{W}_2$ [41], unfortunately, the yield strength was poor (680 MPa). In contrast, the highest yield strength of 806 MPa is found

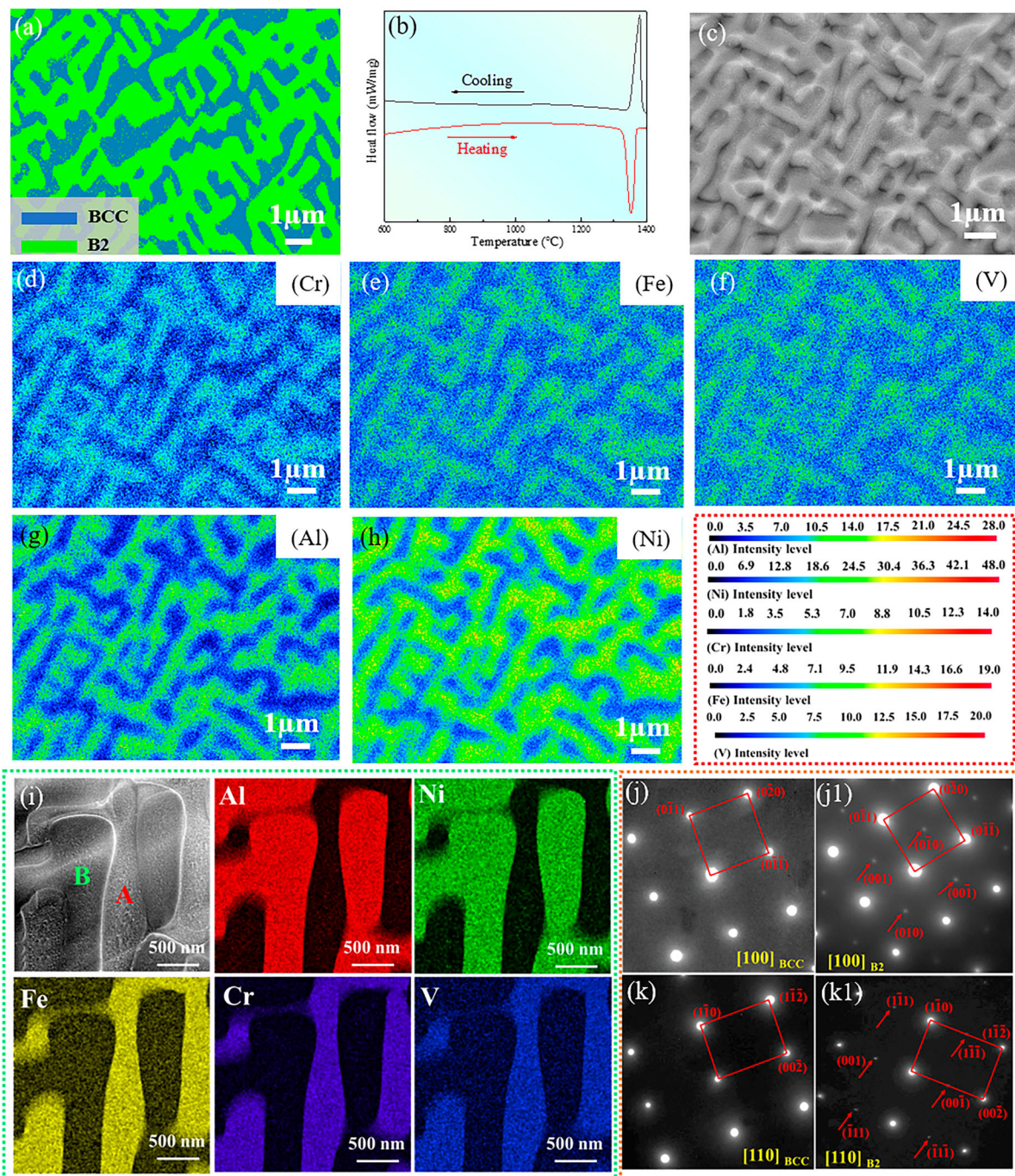


Figure 1. EBSD phase distribution map (a), the DSC curves (b), and EPMA maps (c-h) show elemental distributions of as-cast $\text{Fe}_{30}\text{Cr}_{15}\text{V}_{15}\text{Ni}_{20}\text{Al}_{20}$ CEHEA. (i) HAADF-STEM image and EDS maps for individual elements of Al, Ni, Fe, Cr and V. The SAED patterns taken along the (100) and (110) zone axes in region A (j-k); The SAED patterns taken along the (100) and (110) zone axes in region B (j1-k1).

in the $\text{Ti}_{1.6}\text{ZrNbAl}_{0.2}\text{V}_{0.2}$ HEA, nevertheless, its product of strength and ductility were as low as 10.8 GPa·% [39].

Concurrently, we conducted the compressive test to evaluate the mechanical properties of the as-cast $\text{Fe}_{30}\text{Cr}_{15}\text{V}_{15}\text{Ni}_{20}\text{Al}_{20}$ CEHEA, and the obtained stress-strain curve is displayed in Figure 3(c). Beyond expectation, the ultrahigh yield strength of 2423 MPa and compressive strength of 2871 MPa are achieved, while

good plasticity of 22.3% is maintained. By incorporating the reported data on EHEA, we can see that the present EHEA shows excellent comprehensive mechanical properties (Figure 3d). It is evident that EHEAs consisting of the FCC and σ phases exhibit poor compression strains, which are less than 5% [42–45]. Although the EHEAs consisting of the FCC + Laves phases [46–49] or the FCC + B2 phases [50] have high compression strain (>

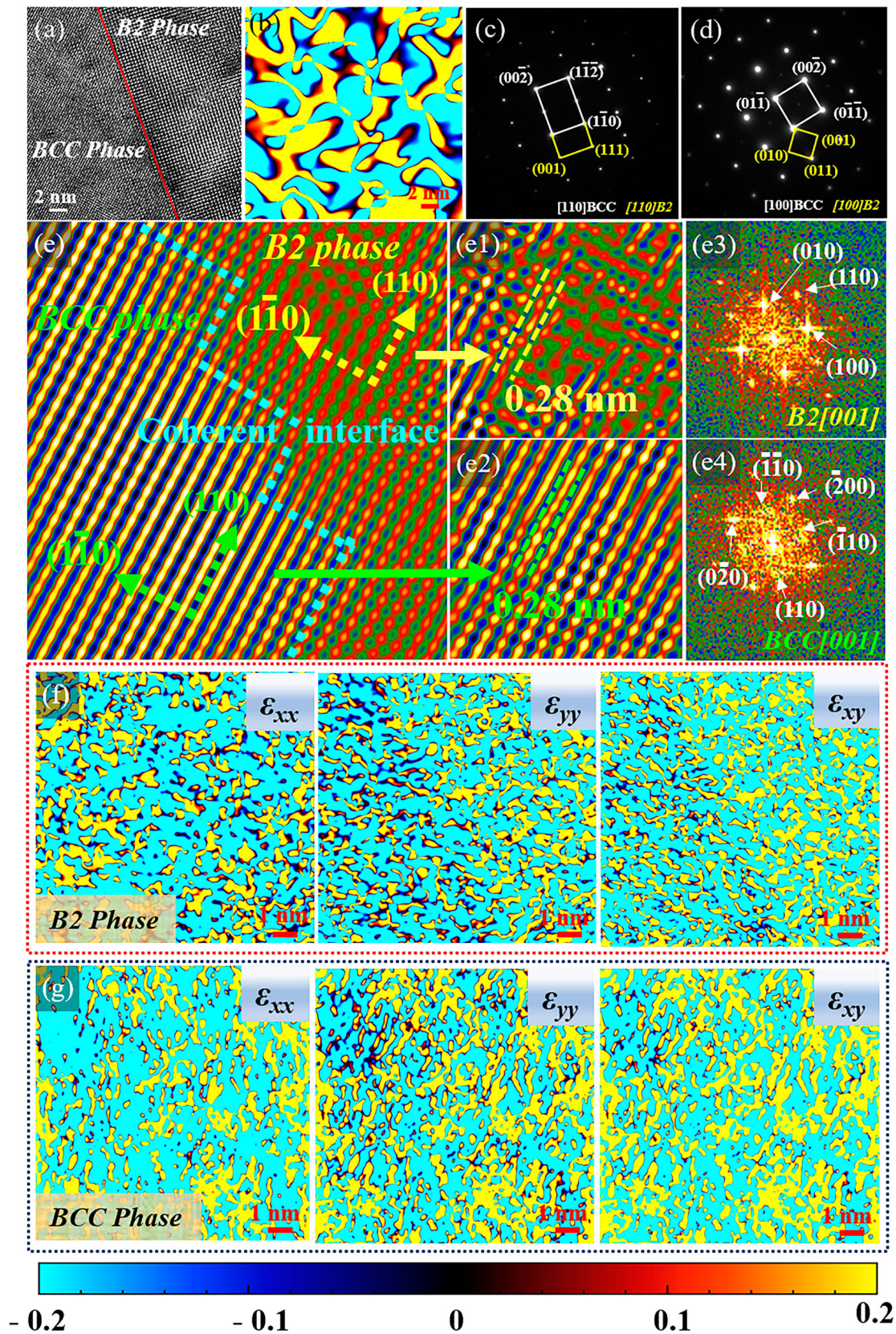


Figure 2. HRTEM image of the as-cast $\text{Fe}_{30}\text{Cr}_{15}\text{V}_{15}\text{Ni}_{20}\text{Al}_{20}$ CEHEA (a) and the GPA maps show strain distributions (b) deduced from the HRTEM image (Fig. 2a), with the SAED patterns taken along the (110) and (100) zonal axes at the phase interface region (c-d). An IFFT image (e), the magnified view of the B2 (e1) and the BCC phases (e2) as well as the corresponding FFT patterns (e3, e4) are shown. The GPA maps show strain distributions in different directions of the B2 and BCC phases, respectively (f-g).

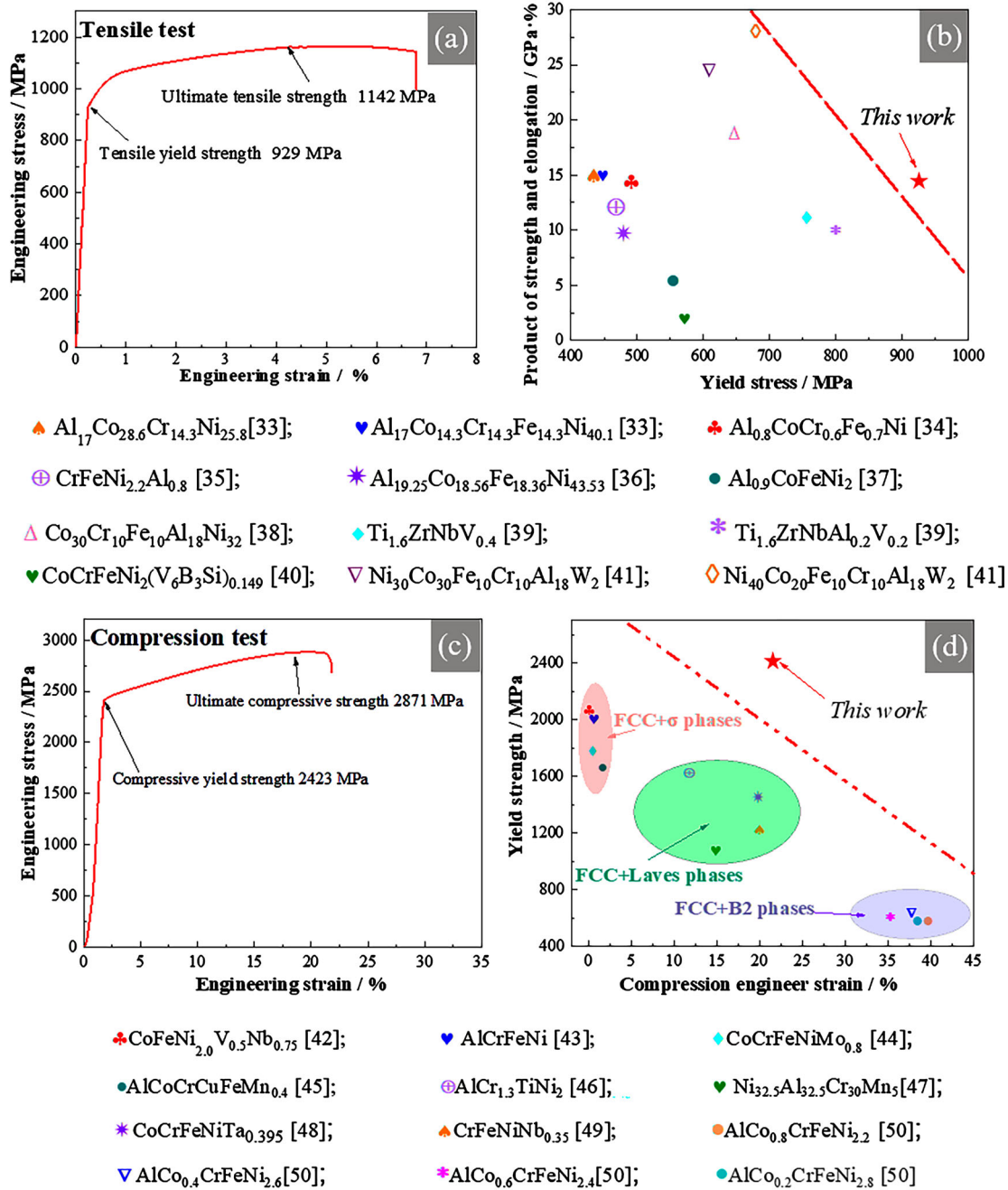


Figure 3. (a) Tensile stress–strain curve, (b) comparison of tensile yield strength versus the products of strength and ductility of the reported EHEAs [33–41], (c) compression engineering stress–strain curve and (d) comparison of compression properties of the as-cast $\text{Fe}_{30}\text{Cr}_{15}\text{V}_{15}\text{Ni}_{20}\text{Al}_{20}$ CEHEA with the reported EHEAs [42–50].

10%), the compressive yield strength is still lower than 1800 MPa. Namely, the $\text{Fe}_{30}\text{Cr}_{15}\text{V}_{15}\text{Ni}_{20}\text{Al}_{20}$ CEHEA demonstrates both encouragingly high tensile properties and superior compressive performances. Besides, there is a significant difference in the mechanical properties during tension and compression. When a metallic material is under tensile testing, the direction of the force on the material itself is the same as the tensile direction. However, when it is compressed, the force direction on the

material itself is opposite to the compressed direction. Thus, the compressive yield strength is generally higher than the tensile yield strength [51]. In addition, HEA with the [110] oriented grains has a good compressive performance, which is about 4.4 times higher than the tensile yield strength [52]. Here, $\text{Fe}_{30}\text{Cr}_{15}\text{V}_{15}\text{Ni}_{20}\text{Al}_{20}$ CEHEA shows an extensive [110] oriented microstructure (Figure 2e). Thus, the alloy has a better compressive yield strength compared to tensile yield strength.

Table 1. Young's modulus (E_r), shear modulus (G) and Vickers hardness (H) of BCC and B2 phases in the $\text{Fe}_{30}\text{Cr}_{15}\text{V}_{15}\text{Ni}_{20}\text{Al}_{20}$ HEA.

	E_r (GPa)	G (GPa)	H (GPa)
BCC	167.9	71.24	5.4
B2	167.3	71.22	5.4

4. Discussion

4.1. Strengthening effect of coherent interface

Nanoindentation tests are conducted to evaluate the hardening behaviors of the BCC and B2 phases, respectively. Figure 4(a, b) shows the load–displacement curves of the BCC and B2 phases, respectively. At the same maximum load, the BCC matrix and B2 phase have an identical depth of 140 nm. Young's modulus (E_r), shear modulus (G) and Vickers hardness (H) of the BCC and B2 phases are measured by nanoindentations (Table 1). Obviously, the BCC and B2 phases have similar values for E_r , G and H , approximately 167, 71.2 and 5.4 GPa, respectively. The microstructural morphology of the specimen after nanoindentation is shown in Figure S4 a. At least five points were tested for each phase to ensure data reliability. Figure S4 (b, c) clearly shows that the indentations of BCC and B2 are triangular. That is to say, the BCC and B2 phases have identical modulus and hardness in this study. So, due to the low-misfit coherent interfaces ($\delta = 0.13\%$), same strain distribution (Figure 2f–g) and identical hardness in the BCC and B2 phases, plastic deformation could simultaneously occur in the different phases in the initial stage of deformation, which had been proved in an Al–Cr–Fe–Ni–V HEA [30]. This means that no significant loading redistribution occurs between the BCC and B2 phases, enabling more homogenous deformation of the alloy and hence a good plasticity.

To validate this viewpoint, we carefully examined the deformation structure of the fractured sample after tensile testing. Dislocation behavior at 3% and 6% strains are presented in Figure 4(c, c1 and d, d1), respectively. At the early stage of the straining ($\varepsilon = 3\%$), dislocations tend to be activated in the locations away from the phase interfaces, and the regions close to the interfaces remain clean (Figure 4c). With further deformation to 6% strain, dislocation pile-ups and dense dislocation tangles can be clearly observed near the phase interfaces (Figure 4d). This is because that the dislocations move to the phase interfaces and latterly they are blocked, leading to the formation of dense dislocation tangles (DDTs) with increasing deformation (Figure 4d). This deformation process is conducive to enhancing the work-hardening ability, thus retaining a good ductility of the alloy. As shown in Figure 4(c), dislocations are mainly located at the phase interfaces, which facilitates dislocation slip

along in the (011) and (112) planes [53]. The Burgers vectors of these dislocations are $b = \pm a/2 [11\bar{1}]$ [54]. Under the higher loading strain, more dislocations are found at the phase interfaces and dislocation slip along the 110 plane is dominant (Figure 4c1–d1).

The initiation of dislocations can be schematically illustrated in Figure 4(e1, e2). In stage I, due to no significant load redistribution occurring between BCC and B2 phases, plastic deformation simultaneously occurs in the both phases, which makes the alloy deform more uniformly. In stage II, dislocations tend to slip along the (011) and (112) slip traces in the deformed sample (Figure 4c, d), indicating that the dislocations with Burgers vectors of $b = \pm a/2 [11\bar{1}]$ could be pinned at the coherence interfaces [53]. Previous studies have shown that dislocation slip can be inhibited by coherent twin boundaries [55]. When coherent twin boundaries show obvious stress concentration, inducing a dislocation dissociation reaction. This situation significantly blocks dislocation movement from different slip planes. Thus, the arbitrarily distributed coherent phase interfaces behave more or less like coherent twin boundaries that inhibit dislocation motion during plastic deformation [56]. Hence, deformation could occur in the different phases at the initial stage of deformation (stage I). The two eutectic phases deform through gliding on their respective slip systems. In stage II, when the applied stress reaches yield strength, dislocations move to the coherent interfaces. The coherent interface can significantly block the dislocation movements from different slip planes and lead to the formation of DDTs [57]. Subsequently, more dislocations move to the coherent interfaces with further straining, leading to an increasing amount of pinned dislocations and dislocation accumulation at the interfaces. As a result, a high yield strength can be achieved in the alloy, while good ductility is retained concurrently. A similar situation in the coherent interface can significantly block the dislocation movements from different slip planes and has also been found in a $\text{Al}_{17}\text{Ni}_{34}\text{V}_{32}\text{Ti}_{17}$ EHEA with a significant improvement in yield strength [57].

4.2. Strengthening mechanisms

In general, solid-solution strengthening ($\Delta\sigma_S$), grain-boundary strengthening ($\Delta\sigma_G$) and dislocation strengthening ($\Delta\sigma_D$) are the basic factors responsible for the strengthening of bulk materials. For the current alloy, yield strength can be formulated as the following microstructure-relate model (see the Supplementary data in detail):

$$\sigma_y = \sigma_0 + \Delta\sigma_S + \Delta\sigma_G + \Delta\sigma_D + \Delta\sigma_P. \quad (1)$$

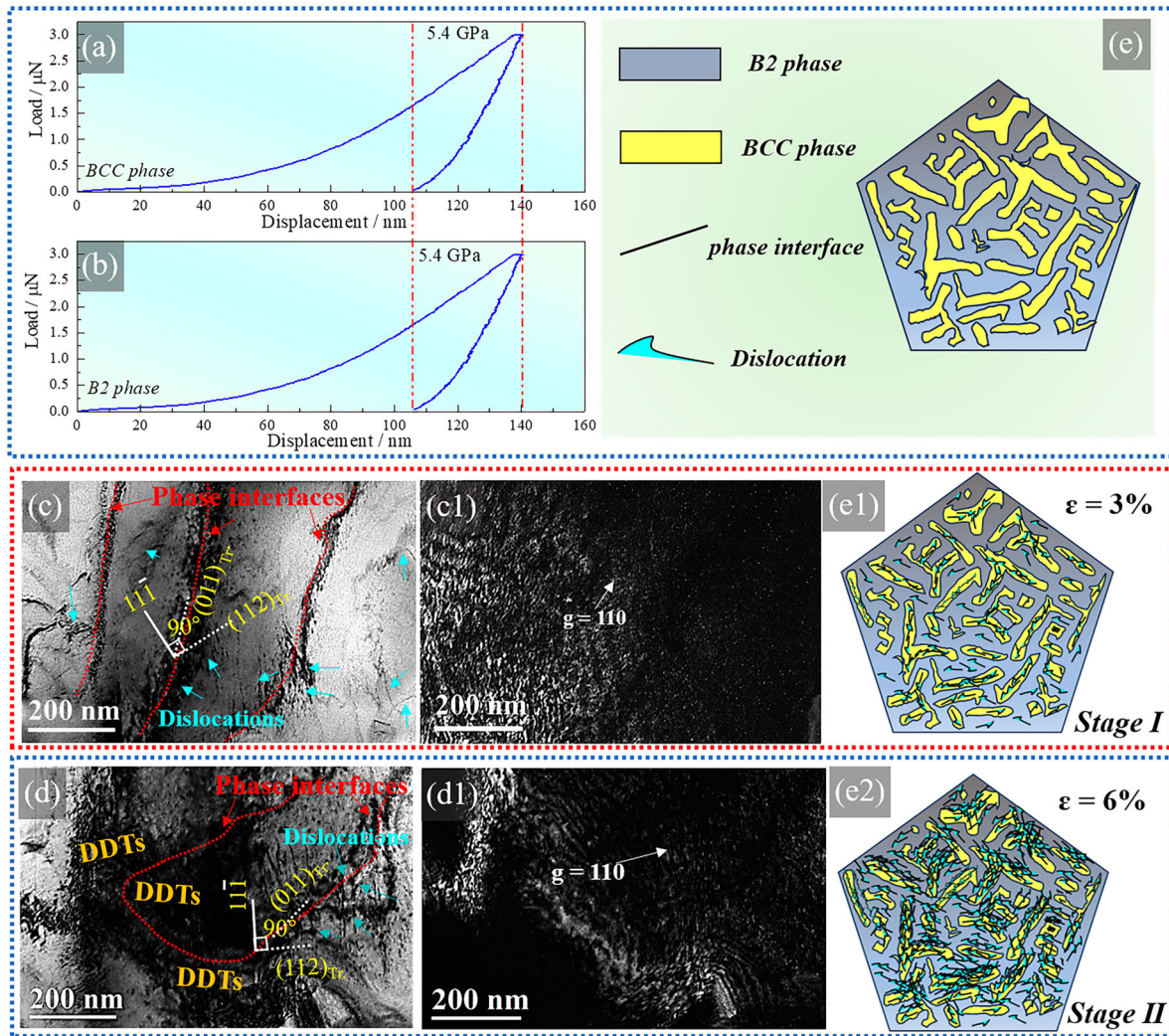


Figure 4. Typical nanoindentation load–displacement curves of the BCC matrix (a) and B2 phase (b) in the as-cast $\text{Fe}_{30}\text{Cr}_{15}\text{V}_{15}\text{Ni}_{20}\text{Al}_{20}$ CEHEA. TEM images exhibit microstructures and dislocation structures after tensile straining of 3% (c, c1) and 6% (d, d1), and the corresponding schematic illustrations show the dislocation evolution during deformation (e–e2).

The summation of coherency strengthening ($\Delta\sigma_{cs}$) and modulus mismatch strengthening ($\Delta\sigma_{ms}$) that contribute to precipitation strengthening ($\Delta\sigma_p$) of the $\text{Fe}_{30}\text{Cr}_{15}\text{V}_{15}\text{Ni}_{20}\text{Al}_{20}$ CEHEA is calculated to be 762 MPa, which accounts for 83% of the calculated yield strength. This confirms that the coherent BCC/B2 interfaces can effectively strengthen the current alloy. In addition, σ_s , σ_G , and σ_D are evaluated as 28, 11 and 19 MPa, respectively. The sum of the calculated strengths (915 MPa) is slightly lower than the measured value of 929 MPa, which may be caused by the following reasons: (1) there is no suitable strength calculation method for HEAs at present, hence only the strength calculation method for traditional metallic materials is applied for a rough estimation. (2) Several intrinsic material parameters (e.g. ΔG and α_ϵ) are referred to the related alloys. In summary, the $\Delta\sigma_{cs} + \Delta\sigma_{ms}$ induced by the coherent

interfaces significantly render excellent mechanical performances of the $\text{Fe}_{30}\text{Cr}_{15}\text{V}_{15}\text{Ni}_{20}\text{Al}_{20}$ CEHEA.

In addition, the mixing rule of the dual-phase EHEAs is a powerful way to estimate the distributions of strain in the different phases [58]. Hence, the strain in this study is estimated by considering the mechanics of duplex system EHEAs (see the Supplementary data for more details). The summation of ϵ_{EHEA} , ϵ_{BCC} , and ϵ_{B2} contributes to the true strain of the $\text{Fe}_{30}\text{Cr}_{15}\text{V}_{15}\text{Ni}_{20}\text{Al}_{20}$ CEHEA is calculated to be 0.058, 0.062 and 0.056, respectively (Table S3). Generally, the average true strain levels in EHEAs for the BCC and B2 phases are very close, indicating homogeneous strain distribution during deformation. That is to say, plastic deformation could occur in the two phases simultaneously. Based on the above discussion, it is suggested that the coherent BCC/B2 interfaces can effectively strengthen

the Fe₃₀Cr₁₅V₁₅Ni₂₀Al₂₀ CEHEA and the stress distribution in the two phases is uniform.

5. Conclusions

In this study, we proposed a valid and universal strategy for developing coherent and eutectic HEAs. Based on this, we successfully designed the Fe₃₀Cr₁₅V₁₅Ni₂₀Al₂₀ CEHEA composed of BCC and B2 phases. The main conclusions can be drawn as follows:

- (1) The as-cast CEHEA has ultrahigh tensile yield strength and ultimate tensile strength of 929 and 1142 MPa, respectively, while retaining a good ductility of 6.8%. In particular, it shows an unprecedented combination of yield strength of 2423 MPa and ultimate strength of 2871 MPa, under compression, accompanied with an excellent plasticity of 22.3%.
- (2) The ultrahigh strength mainly derives from the synergism between coherent interfaces and BCC and B2 duplex strengthening phases, whereas good ductility and unprecedented plasticity correlate to a homogeneous work-hardening regulated by the coherent interfaces.
- (3) The CEHEA shows a feature of two-stage deformation. In stage I, plastic deformation simultaneously occurs in the two-phase region because of the low-misfit coherent interfaces, same strain distribution and similar hardness values are found in the dual phases. When the applied stress reaches the yield strength, dislocations move toward the coherent interfaces. The coherent interfaces strongly block dislocations, thus promoting dislocation accumulation at the interfaces. This makes plastic deformation of the CEHEA more uniform and ensures good ductility for the alloy.

Disclosure statement

No potential conflict of interest was reported by the author(s).

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