



Deformation mechanisms and work-hardening behavior of transformation-induced plasticity high entropy alloys by *in-situ* neutron diffraction

S. Fu, H. Bei, Y. Chen, T.K. Liu, D. Yu & K. An

To cite this article: S. Fu, H. Bei, Y. Chen, T.K. Liu, D. Yu & K. An (2018) Deformation mechanisms and work-hardening behavior of transformation-induced plasticity high entropy alloys by *in-situ* neutron diffraction, Materials Research Letters, 6:11, 620-626, DOI: [10.1080/21663831.2018.1523239](https://doi.org/10.1080/21663831.2018.1523239)

To link to this article: <https://doi.org/10.1080/21663831.2018.1523239>



© 2018 The Author(s). Published by Informa UK Limited, trading as Taylor & Francis Group.



View supplementary material [↗](#)



Published online: 23 Sep 2018.



Submit your article to this journal [↗](#)



Article views: 5427



View related articles [↗](#)



View Crossmark data [↗](#)



Citing articles: 21 View citing articles [↗](#)



ORIGINAL REPORT



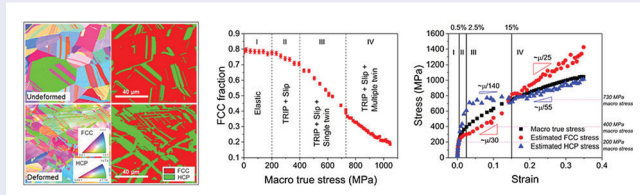
Deformation mechanisms and work-hardening behavior of transformation-induced plasticity high entropy alloys by *in-situ* neutron diffraction

S. Fu^{a,b}, H. Bei^c, Y. Chen^a, T.K. Liu^{a,c,d}, D. Yu^{a,d} and K. An^a

^aNeutron Scattering Division, Oak Ridge National Laboratory, Oak Ridge, TN, USA; ^bSchool of Chemical Engineering and Technology, Tianjin University, Tianjin, People's Republic of China; ^cMaterials Science and Technology Division, Oak Ridge National Laboratory, Oak Ridge, TN, USA; ^dDepartment of Materials Science and Engineering, The University of Tennessee, Knoxville, TN, USA

ABSTRACT

A full picture of tensile deformation mechanism evolution in the FCC-to-HCP transformation-induced plasticity high entropy alloy (TRIP-HEA) was revealed by real-time *in situ* neutron diffraction. Three transition points, i.e. the triggering of TRIP in the FCC phase and the activation of single and multiple twinning in the HCP phase, were identified to result in significant stress redistribution. Accordingly, four deformation stages with distinct phase-specific work-hardening behaviors were recognized. It was concluded that the easily-triggered persisting TRIP and the work-hardening potential of the HCP contribute together to the persisting bulk work-hardening.



IMPACT STATEMENT

The deformation mechanism evolution and phase-specific work-hardening behaviors of the FCC-to-HCP TRIP-HEA were unraveled, which provided a comprehensive understanding of the TRIP-assisted superior strength and ductility.

ARTICLE HISTORY

Received 8 April 2018

KEYWORDS

In situ neutron diffraction; high entropy alloy; dual phase; phase transformation; twinning

1. Introduction

Transformation-induced plasticity (TRIP) is an efficient strategy to obtain a combination of high strength and ductility through strong enhancement of the work-hardening capacity of metallic materials [1,2]. It is achieved by a mechanically induced transformation of a metastable phase at ambient temperature to a low-temperature phase. During such phase transformation, the TRIP material acts as a dynamic composite that incorporates multiple hardening mechanisms including interface hardening, enhanced dislocation hardening and second-phase hardening into additional work-hardening [3]. As the phase metastability at ambient temperature can be tuned by varying element composition, this strategy has potential to be applied in various metallic

material systems, such as austenite or TRIP-assisted multiphase steels, high-Mn steels and Ti alloys [4–7]. Recently, the TRIP effect was successfully introduced into high entropy alloys (HEAs) that outshined with supreme solid solution strengthening [8,9] to achieve a strength-ductility combination superior to most engineering metallic materials [1]. Improved work-hardening capacity underlying the ductility enhancement was experimentally evidenced for the various novel-developed TRIP-HEAs. For example, in non-equiatomic FeMnCoCr(Ni or C) alloys [1,10,11], the face-centered cubic (FCC) phase gradually transformed to the hexagonal close-packed (HCP) phase, which provided additional mechanisms for plastic deformation accompanied by a higher work-hardening rate than single-phase FeMnCoCrNi alloys.

CONTACT H. Bei hbei1@utk.edu Materials Science and Technology Division, Oak Ridge National Laboratory, Oak Ridge, TN USA; K. An kean@ornl.gov Neutron Scattering Division, Oak Ridge National Laboratory, Oak Ridge, TN, USA

Supplemental data for this article can be accessed here. <https://doi.org/10.1080/21663831.2018.1523239>

© 2018 The Author(s). Published by Informa UK Limited, trading as Taylor & Francis Group.

This is an Open Access article distributed under the terms of the Creative Commons Attribution License (<http://creativecommons.org/licenses/by/4.0/>), which permits unrestricted use, distribution, and reproduction in any medium, provided the original work is properly cited.

Similarly, in the body-centered cubic (BCC) Ti-rich $\text{Ti}_{35}\text{Zr}_{27.5}\text{Hf}_{27.5}\text{Nb}_5\text{Ta}_5$ [12] and $\text{Ta}_{0.4}\text{HfZrTi}$ [13] alloys, plasticity was accommodated by transitioning to the orthorhombic and HCP phases, respectively, exhibiting a double yielding behavior with enhanced work-hardening capacity.

One of the key issues to promote the application and advancement of TRIP materials is to understand the work-hardening mechanisms of the co-deforming multiphases under dynamic transformation. Using *in situ* neutron diffraction, researchers have unveiled the significant stress redistribution as the essential feature of TRIP [5], and manifested the importance of dynamic load partitioning in the bulk hardening behavior [5,13,14]. However, it is common and even promoted that TRIP occurs with concurrent stacking faults, slip and twinning in constituent phases to postpone the plasticity exhaustion. It is still challenging in fully understanding their work-hardening mechanisms due to the complicated cooperation and interactions among multiple deformation mechanisms. The newly reported FCC-to-HCP TRIP-HEA, for instance, presented high ultimate tensile strength above 850 MPa and more than 70% elongation to fracture [1]. The underlying multiple deformation mechanisms, including dislocation slip and twinning in the transformed HCP phase in addition to slip and TRIP in the parent FCC phase, were discretely recognized and visualized by methods like electron backscatter diffraction (EBSD), electron channeling contrast imaging (ECCI) and transmission electron microscopy (TEM) [1,11,15]. However, the whole picture of continuous mechanism evolutions and their interactions are still not revealed, and an unambiguously understanding of the high work-hardening capacity of the material is still left blank.

In this work, using the real-time *in situ* time-of-flight (TOF) neutron diffraction, we unraveled the full picture of deformation mechanism evolution, especially transformation and twinning, and the resulted phase-specific work-hardening behavior of TRIP-HEA $\text{Fe}_{50}\text{Mn}_{30}\text{Co}_{10}\text{Cr}_{10}$ under tension. The feasibility of *in situ* neutron diffraction on the deformation behaviors of HEAs has been proved by several studies [16–18], but few work on TRIP-HEAs is available [13]. Unlike the localized and post-test measurements of EBSD, ECCI or TEM, the real-time *in situ* TOF neutron diffraction enables a continuous acquisition of full diffraction patterns during deformation, which provides quantitative bulk average information of deformation over a large gauge volume at the lattice level, including the lattice spacing, lattice rearrangement, and reorientation. Accordingly, we were able to trace the activation and changes of deformation mechanisms such as TRIP and twinning, thus bridging

the gap of interrupted measurements by those conventional post-test techniques, and describing a full picture of deformation mechanisms [5,16,19]. Four deformation stages characterized by distinct work-hardening mechanisms were identified by the significant stress redistribution at critical transition points of deformation mechanisms, which is insightful and necessary to further efforts in engineering application and material optimization.

2. Experimental

The studied TRIP-HEA $\text{Fe}_{50}\text{Mn}_{30}\text{Co}_{10}\text{Cr}_{10}$ (at. %) was firstly reported by Li et al. [1]. In this study, the alloy with the same nominal compositions was first mixed with pure metals (> 99.9 wt. % pure) by vacuum arc melting and then drop-casted to form a $12.7 \times 12.7 \times 75 \text{ mm}^3$ bar. Details of alloy preparation can be found in Ref. [20]. The as-cast bar was homogenized at 1150 °C for 4 h in vacuum before room-temperature rolling to 75% thickness reduction. It was further annealed at 950 °C for 1 h followed by air cooling to achieve desired microstructures. The initial microstructures were examined by EBSD as shown in Figure 1(a), presenting equiaxed FCC grains with an average grain size of $\sim 40 \mu\text{m}$ and lath-shaped HCP grains with an average thickness of $\sim 4 \mu\text{m}$. The phase composition was determined by EBSD and neutron diffraction independently as shown in Figure 1. Both resulted 79% FCC and 21% HCP in volume fraction. *In situ* neutron diffraction under tension was conducted on VULCAN, at the Spallation Neutron Source, Oak Ridge National Laboratory [21,22]. The TOF diffraction patterns along the loading direction (LD) and transverse direction (TD) were continuously recorded by the -90° and $+90^\circ$ detector banks, respectively. The detailed set-up of the *in situ* diffraction measurement on VULCAN can be found elsewhere [22]. And details of tensile test can be found in the supplementary material. The phase transformation during tension was qualified by Rietveld refinement on the real-time neutron diffraction patterns as illustrated in Figure 1(b). The diffraction intensity and lattice spacing of separate diffraction peaks were obtained by single peak fit for grain reorientation and lattice strain analysis, respectively.

3. Results and discussion

The applied stress induces phase transformation and the kinetics can be observed by the evolution of phase composition upon loading. Figure 2 shows the tensile stress-strain behavior and the accompanied phase transformation of the TRIP-HEA. The material yielded at ~ 200 MPa (by the observable deviation onset of the elastic regime), and then showed continuous hardening at

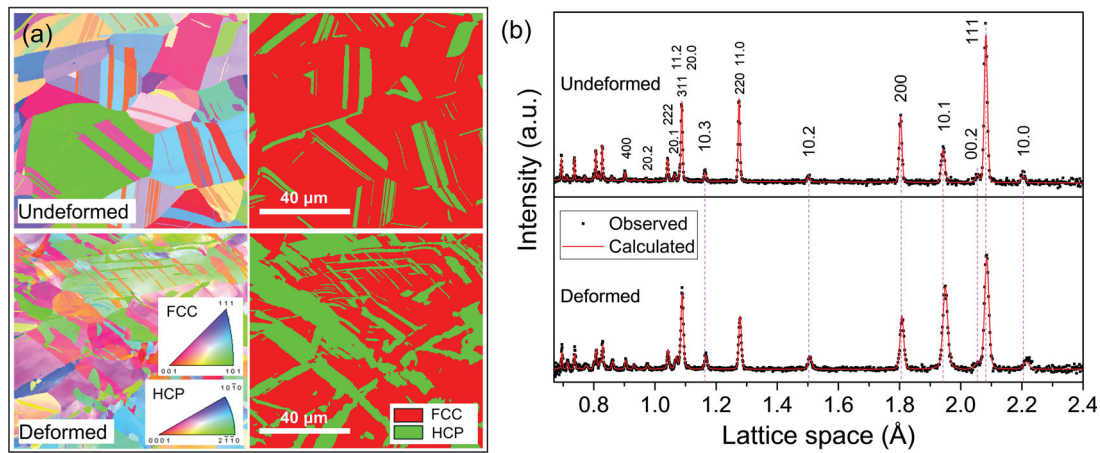


Figure 1. (a) EBSD inverse pole figures and corresponding phase maps and (b) neutron diffraction patterns in comparison with the Rietveld refinements before and after $\sim 6\%$ deformation.

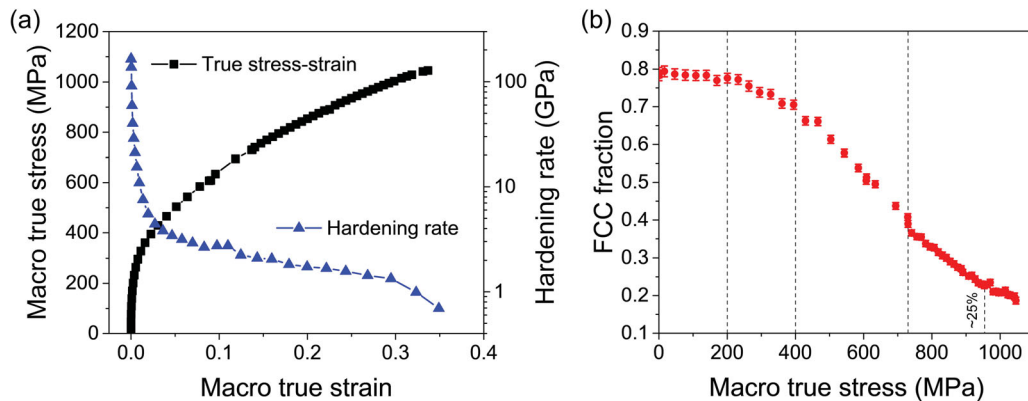


Figure 2. (a) The tensile stress-strain curve and evolution of work-hardening rate and (b) the dynamic FCC fraction under tension.

a gradually decreasing rate until it reached the ultimate strength of 1046 MPa and fractured at the elongation of 34%, which is comparable with that of the 45 μm -grain size sample in the literature [1]. During tension, the FCC-to-HCP transformation was triggered at ~ 200 MPa that was just at the yield point. It was indicated to account for the first plastic transition of the material. Upon loading, the FCC transformed at an increasing rate with macro stress and quickly reached a stable stage with a nearly constant rate at ~ 400 MPa. Considerable nucleation and growth of HCP with multiple variants occurred in the stable stage as illustrated by the EBSD image of $\sim 6\%$ deformation in Figure 1(a). As the macro stress increased to ~ 730 MPa with $\sim 15\%$ strain, the transformation started to develop at a lower rate than the previous stage. The small drop of the FCC fraction could be due to the change of control mode as described in the supplementary. These changes of transformation behavior at ~ 400 and ~ 730 MPa correspond to the transitions of primary deformation mechanisms as discussed. As presented, the FCC-to-HCP transformation had a close relation with the applied stress. This could be attributed to

the transition mechanism that nucleation and growth of HCP are accomplished by overlapping of stacking faults thus controlled by the movement of Shockley partial dislocations [23]. The applied stress provides the shear stress to move partial dislocations by overcoming the obstruction of lattice defects formed through the prior plastic deformation. Thus it directly stems the kinetics of phase transformation. The transformation was gradually suppressed when the plastic strain increased over $\sim 25\%$, which was related to the impeding effect of high dislocation density at large strains [23–25]. This resulted in the incomplete transformation with $\sim 17\%$ FCC remaining at fracture.

As traced by the evolution of phase transformation, the TRIP persisted in the FCC till fracture. Besides TRIP, dislocation slip was identified to be the main accompanying deformation mechanism in the FCC, which leads to the deformation texture development similar with the FCC single-phase HEA. A detailed discussion is in the supplementary material. It was thus suggested that the FCC deformed with the continuous evolution of deformation mechanisms. The changes in the transformation

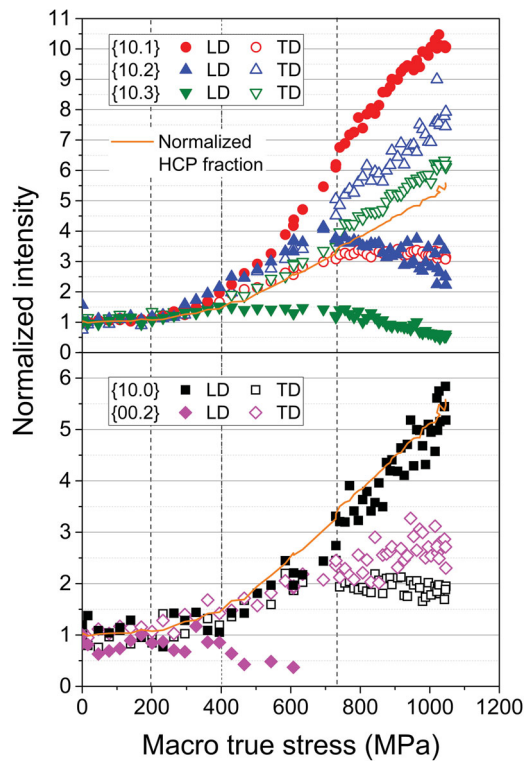


Figure 3. The normalized diffraction intensity evolution of the various-oriented HCP grains with macro stress in LD and TD. The evolution of normalized HCP phase fraction is given in orange solid line as the average evolution of grain intensity due to phase transformation. Critical transition load points were marked by dash lines. Note that the intensity was corrected by the gauge volume change due to deformation as described in the Supplementary.

behavior after being triggered were contributed by the plastic deformation of the produced HCP phase.

After the triggering of TRIP, deformation twinning in the HCP phase was involved in the material along with TRIP after hardening up to 400 MPa, which was evidenced by the splitting of grain diffraction intensities. As shown in Figure 3, the normalized diffraction intensity of {00.2} and {10.3} grains bifurcated in LD and TD at ~ 400 MPa. Upon loading, the grain intensities in TD increased with the decreasing intensities in LD, indicating a $\sim 90^\circ$ reorientation due to the {10.2} tensile twinning of HCP. Nevertheless, the other grains showed nearly equal intensity in LD and TD. It was suggested that the {00.2} and {10.3} grains along LD having their c-axes close to the loading axis were favored in tensile twinning due to relatively high Schmid factor [26]. The twinned {10.3} grains have a high chance to become {10.1} grains oriented *nearly* along LD, due to the $\sim 90^\circ$ angle between the lattice planes and the 86° twinning angle. This thus leads to an additional increase of {10.1} grain intensity, as seen above the average evolution. It should be noted that dislocation slip may also

contribute to the intensity variation. As reported for HCP Mg alloys under tension when the dislocation slip dominated the early plastic deformation, the {10.1} grain intensity decreased in LD [27,28]. Considering that the phase transformation is on-going in the studied TRIP-HEA, if dislocation slip dominates the early plasticity of HCP, one could expect a below-average increase in {10.1} intensity in LD. However, that is not the case regarding the data in Figure 3. Therefore, the tensile twinning rather than dislocation slip dominates the early plasticity of the HCP. Note that the evolution of {00.2} grain intensity along LD disappears above 600 MPa because the {00.2} grains oriented in the LD were all twinned to the directions perpendicular to the LD. As seen from Fig. S1, the {00.2} peak became extremely weak and barely recognized at 600 MPa and completely disappeared at higher stress.

The following stage of deformation twinning in the HCP began at ~ 730 MPa ($\sim 15\%$ strain) as featured by an abrupt split of the intensity of various grain families in LD and TD, indicating the start of dramatic reorientation in a broad range of grain orientations. This is related to the activation of compression twinning and multiple twin systems. The compression twinning could be identified by the reorientation of {10.1} grains in TD by the resolved compressive stress and the concurrent increase of {10.2} and {10.3} grains in this direction. The intensity change reflected a $\sim 50^\circ$ rotation along $\langle 12.0 \rangle$. This hardly activated {10.1} compression twin with boundary misorientation of 57° has been spotted at engineering strain higher than 26% for TRIP-HEA by EBSD [15]. Local stress heterogeneities at large deformation would further facilitate broad distribution in parent orientation for activating multiple twin systems [26]. Meanwhile, large-scale dislocation slip under large strain could also contribute to the change of intensity of various oriented HCP grains. The continuous significant decrease of {10.2} intensity in LD starting at ~ 730 MPa is considered the result of dislocation slip as reported in an HCP Mg alloy [27]. Therefore, multiple twinning and dislocation slip are the correlative deformation mechanisms in HCP under large strain.

With the power of *in situ* diffraction on obtaining lattice evolution, the internal load partitioning between the two constituent phases can be differentiated to further uncover the deformation dynamics of TRIP-HEA. Four deformation stages are determined by stress redistribution at the aforementioned critical mechanism-transition load points, i.e. 0–200, 200–400, 400–730, and 730 MPa till fracture. At those transition points, lattice strains showed corresponding changes in behaviors in Figure 4(a). The lattice strain is defined as the relative change of lattice spacing averaged for a certain family of

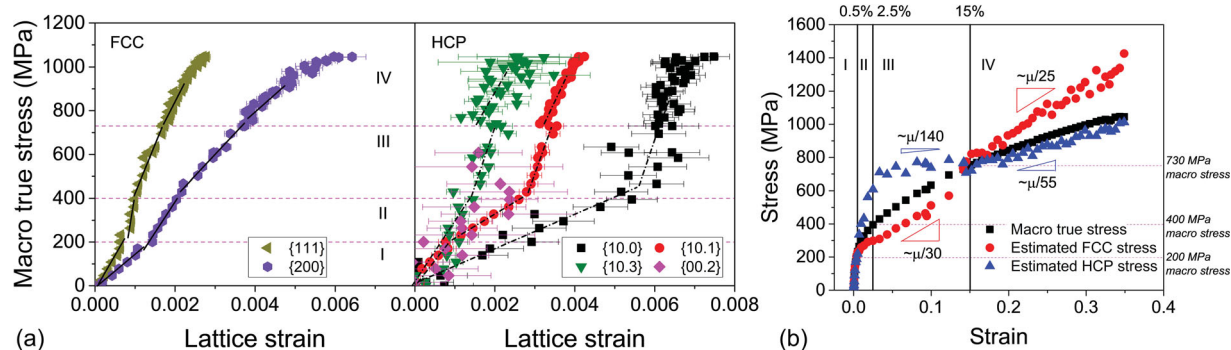


Figure 4. (a) Evolution of lattice strains of $\{111\}_{\text{FCC}}$, $\{200\}_{\text{FCC}}$, $\{00.2\}_{\text{HCP}}$, $\{10.0\}_{\text{HCP}}$, $\{10.1\}_{\text{HCP}}$ and $\{10.3\}_{\text{HCP}}$ grains with macro stress in LD and (b) estimated phase-specific stress–strain behaviors. The average hardening rate is calculated for both phases in stages III and IV. μ is the shear modulus of the FCC.

grains at loading with respect to the value at zero load [5,16,21,22]. Based on Hooke's law, it indicates the average stress of a certain grain family in the mesoscopic scale. By comparing the evolution of lattice strain with macro stress, the elastic or plastic anisotropy and resulted load partitioning between constituent phases can be easily traced. In stage I, both the FCC and HCP deformed elastically as their lattice strains increased linearly with the macro stress. In stage II, the lattice strains of $\{111\}$ and $\{200\}$ grains first showed an upward trend at ~ 200 MPa, indicating the yield point of the FCC. Similar to other TRIP materials, FCC-to-HCP phase transformation basically involves the movement and overlapping of stacking faults [29], which normally introduces stress relaxation and irreversible shear strain, corresponding to the yielding of parent phase, i.e. FCC in this study. As inferred from the phase fraction evolution in Figure 2(b), the trigger of phase transformation should account for the yielding of the FCC. Meanwhile, the lattice strain of $\{10.1\}$ grains abruptly went downward, suggesting that the produced HCP elastically took over an additional load that was relaxed by the plastic deformation of the FCC. The HCP grains first showed an upward trend at ~ 400 MPa in stage III, indicating the yield point of the HCP. It is activated by tensile twinning as inferred by the intensity analysis in Figure 3. The FCC grains simultaneously regained the relaxed stress as indicated by the downward trend of lattice strain evolution. Such additional hardening of the FCC was achieved by the increased rate of TRIP to the stable transformation stage (Figure 1b) besides dislocation slip. In the final stage IV, the lattice strain evolution of the FCC grains barely changed comparing with the previous stage, which is related to the limitation of the work-hardening capacity of the FCC as discussed in the following section. It is noted that the $\{10.1\}$ and $\{10.3\}$ HCP grains showed a small recovery of lattice strain at 730 MPa, which could be directly associated with the change of control mode as described

in the supplementary. Above 730 MPa, the HCP grains showed an additional increase of lattice strain as indicated by the slightly rightward slope line, indicating the work-hardening of the HCP as discussed below.

Accordingly, the work-hardening behavior of the TRIP-HEA can be analyzed by the estimated stress–strain behaviors of the constituent phases as shown in Figure 4(b). The stress–strain curves of individual phases were derived from the rule of mixture on stress and Taylor's assumption of uniform strain [30,31]. Calculation details are given in the supplemental material. The FCC was weakly hardened by the progressively initiated TRIP besides dislocation slip in stage II. After the HCP yielded in stage III, the hardening rate of the FCC rapidly increased to $\sim \mu/30$ through the increasing rate of TRIP and dislocation slip. The FCC further kept a high hardening rate of $\sim \mu/25$ that is close to the maximum of $\mu/20$ for cubic metals [32] until fracture. Compared to the fast decreasing hardening rate of the single-phase HEA [1], such high work-hardening rate of the FCC in TRIP-HEA could be attributed to the reduction of mean free path for dislocation slip and suppression of cross-slip for dynamic recovery [33]. Phase transformation introduces phase boundaries in the parent grain, thus reducing the mean free path for the slip. Inhibition of cross-slip decreases the probability of dislocation annihilation and dynamic recovery. Both the reduction of mean free path for slip and the inhibition of cross-slip lead to high storage capacity of dislocations, thus high work-hardening rate. Meanwhile, the work-hardening rate seems to saturate, implying limited work-hardening capacity of the FCC, which was probably resulted from the restrained slip by small mean free path for slip and restricted TRIP by high dislocation density and small volume fraction. While the FCC was strongly hardened since stage III, the HCP showed quite large plastic flow by single twinning and then increasingly hardened at a rate of $\sim \mu/55$ due to the multiple twinning and dislocation

slip until fracture. It is shown by the nearly overlapped stress–strain curves of the HCP and bulk material in stage IV that the work-hardening of the HCP dominated the large deformation behavior of the material. Collectively, the work-hardening of both the transformable FCC and the produced HCP persisted in a wide deformation range. Such specific work-hardening behavior is the result of similar modulus and close yield point of the two phases, in addition to persisting TRIP in FCC and deformation twinning in HCP and dislocation slip in both phases. Consequently, the TRIP-HEA presents monotonic hardening rate evolution (Figure 2a) without double yielding which is typical for the FCC-to-BCC/BCT and BCC-to-HCP TRIP materials [4,13,14].

4. Conclusion

In summary, the full picture of tensile deformation mechanism evolution and work-hardening behavior of the FCC-to-HCP TRIP-HEA Fe₅₀Mn₃₀Co₁₀Cr₁₀ is revealed by the real-time *in situ* neutron diffraction. The TRIP-HEA demonstrates multiple deformation mechanisms that differently function in four deformation stages, which could be taken advantages of in future engineering applications. The easily triggered persisting TRIP plus the work-hardening potential of the HCP through multiple twinning and dislocation slip contribute to the persisting work-hardening of the material, which suggests that the FCC-to-HCP transformation would be favored for material design. It should be noted, however, that the high hardening rate of the FCC with persisting TRIP could lead to high-stress localization that may assist cracking. Multiple deformation mechanisms besides TRIP and slip would thus be desired to introduce into the FCC for ductility improvement.

Acknowledgments

This work was supported by the US Department of Energy, Office of Science, Basic Energy Sciences, Materials Sciences and Engineering Division. Neutron scattering was conducted at the Spallation Neutron Source (SNS), which is a DOE Office of Science User Facility operated by ORNL. We thank Dr. Q. Xie and Dr. A.D. Stoica in ORNL for their help with the data reduction of texture measurement by neutron diffraction.

Disclosure statement

No potential conflict of interest was reported by the authors.

Funding

This work was supported by the US Department of Energy, Office of Science, Basic Energy Sciences, Materials Sciences and Engineering Division. Neutron scattering was conducted at the

Spallation Neutron Source (SNS), which is a DOE Office of Science User Facility operated by ORNL.

References

- [1] Li Z, Pradeep KG, Deng Y, et al. Metastable high-entropy dual-phase alloys overcome the strength-ductility trade-off. *Nature*. 2016;534(7606):227–230.
- [2] Pereloma E, Gazder A, Timokhina I. Retained austenite: transformation-induced plasticity. encyclopedia of iron, steel, and their alloys. New York: Taylor and Francis; 2016; p. 3088–3103.
- [3] De AK, Speer JG, Matlock DK, et al. Deformation-induced phase transformation and strain hardening in type 304 austenitic stainless steel. *Metall Mater Trans A*. 2006;37A:1875–1866.
- [4] Maréchal D, Sinclair CW, Dufour P, et al. *In situ* measurements of load partitioning in a metastable austenitic stainless steel: neutron and magnetomechanical measurements. *Metall Mater Trans A*. 2012;43(12):4601–4609.
- [5] Muránsky O, Šittner P, Zrník J, et al. In situ neutron diffraction investigation of the collaborative deformation–transformation mechanism in TRIP-assisted steels at room and elevated temperatures. *Acta Mater*. 2008; 56(14):3367–3379.
- [6] Kwon EP, Fujieda S, Shinoda K, et al. Texture evolution and fcc/hcp transformation in Fe–Mn–Si–Cr alloys by tensile deformation. *Mater Sci Eng A*. 2010;527(24–25):6524–6532.
- [7] Sun F, Zangh JY, Marteleur M, et al. Investigation of early stage deformation mechanisms in a metastable titanium alloy showing combined twinning-induced plasticity and transformation-induced plasticity effects. *Acta Mater*. 2013;61(17):6406–6417.
- [8] 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(5):299–303.
- [9] Liu TK, Wu Z, Stoica AD, et al. Twinning-mediated work hardening and texture evolution in CrCoFeMnNi high entropy alloys at cryogenic temperature. *Mater Des*. 2017;131:419–427.
- [10] Li Z, Tasan CC, Springer H, et al. Interstitial atoms enable joint twinning and transformation induced plasticity in strong and ductile high-entropy alloys. *Sci Rep*. 2017;7:40704.
- [11] Li Z, Körmann F, Grabowski B, et al. Ab initio assisted design of quinary dual-phase high-entropy alloys with transformation-induced plasticity. *Acta Mater*. 2017;136:262–270.
- [12] Lilensten L, Couzinié J-P, Bourgon J, et al. Design and tensile properties of a bcc Ti-rich high-entropy alloy with transformation-induced plasticity. *Mater Res Lett*. 2016;5(2):110–116.
- [13] Huang H, Wu Y, He J, et al. Phase-transformation ductilization of brittle high-entropy alloys via metastability engineering. *Adv Mater*. 2017;29(30):1701678.
- [14] Jacques PJ, Furnémont Q, Lani F, et al. Multiscale mechanics of TRIP-assisted multiphase steels: I. characterization and mechanical testing. *Acta Mater*. 2007;55(11):3681–3693.

- [15] Li Z, Tasan CC, Pradeep KG, et al. A TRIP-assisted dual-phase high-entropy alloy: grain size and phase fraction effects on deformation behavior. *Acta Mater.* **2017**;131:323–335.
- [16] Wu Y, Liu WH, Wang XL, et al. *In situ* neutron diffraction study of deformation behavior of a multi-component high-entropy alloy. *Appl Phys Lett.* **2014**;104(5):051910.
- [17] Woo W, Huang EW, Yeh J-W, et al. *In situ* neutron diffraction studies on high-temperature deformation behavior in a CoCrFeMnNi high entropy alloy. *Intermetallics.* **2015**;62:1–6.
- [18] Cai B, Liu B, Kabra S, et al. Deformation mechanisms of Mo alloyed FeCoCrNi high entropy alloy: In situ neutron diffraction. *Acta Mater.* **2017**;127:471–480.
- [19] Cakmak E, Choo H, An K, et al. Radial distribution of martensitic phase transformation in a metastable stainless steel under torsional deformation: a synchrotron X-ray diffraction study. *Mater Lett.* **2011**;65(19–20):3013–3015.
- [20] Wu Z, Bei H, Otto F, et al. Recovery, recrystallization, grain growth and phase stability of a family of FCC-structured multi-component equiatomic solid solution alloys. *Intermetallics.* **2014**;46:131–140.
- [21] Wang XL, Holden TM, Rennich GQ, et al. VULCAN—the engineering diffractometer at the SNS. *Physica B.* **2006**;385–386. (Part 1):673–675.
- [22] An K, Skorpenske HD, Stoica AD, et al. First in situ lattice strains measurements under load at VULCAN. *Metall Mater Trans A.* **2011**;42(1):95–99.
- [23] Mahajan S, Green ML, Brasen D. A model for the FCC HCP transformation, its applications, and experimental evidence. *Metall Trans A.* **1977**;8A:283–293.
- [24] Tamura I. Deformation-induced martensitic transformation and transformation-induced plasticity in steels. *Met Sci.* **1982**;16:245–253.
- [25] Liu Y, Yang H, Tan G, et al. Stress-induced FCC $\leftrightarrow$ HCP martensitic transformation in CoNi. *J Alloys Compd.* **2004**;368(1–2):157–163.
- [26] Bingert JF, Mason TA, Kaschner GC, et al. Deformation twinning in polycrystalline Zr: insights from electron backscattered diffraction characterization. *Metall Mater Trans A.* **2002**;33A:955–963.
- [27] Muránsky O, Carr DG, Barnett MR, et al. Investigation of deformation mechanisms involved in the plasticity of AZ31 Mg alloy: in situ neutron diffraction and EPSC modelling. *Mater Sci Eng A.* **2008**;496(1–2):14–24.
- [28] Wang H, Wu PD, Tomé CN, et al. Study of lattice strains in magnesium alloy AZ31 based on a large strain elastic-viscoplastic self-consistent polycrystal model. *Int J Solids Struct.* **2012**;49(15–16):2155–2167.
- [29] Olson GB, Cohen M. A general mechanism of martensitic nucleation: part I. General concepts and the FCC-HCP transformation. *Metall Trans A.* **1976**;7A:1897–1904.
- [30] Taylor GI. Plastic strain in metals. *J Inst Met.* **1938**;62:307–324.
- [31] Weng GJ. The overall elastoplastic stress–strain relations of dual-phase metals. *J Mech Phys Solids.* **1990**;38(3):419–441.
- [32] Kocks U, Mecking H. Physics and phenomenology of strain hardening: the FCC case. *Prog Mater Sci.* **2003**;48:171–273.
- [33] De Cooman BC, Estrin Y, Kim SK. Twinning-induced plasticity (TWIP) steels. *Acta Mater.* **2018**;142:283–362.