



Crystallographically degenerate B2 precipitation in a plastically deformed *fcc*-based complex concentrated alloy

Deep Choudhuri, Shivakant Shukla, Whitley B. Green, Bharat Gwalani, Victor Ageh, Rajarshi Banerjee & Rajiv S. Mishra

To cite this article: Deep Choudhuri, Shivakant Shukla, Whitley B. Green, Bharat Gwalani, Victor Ageh, Rajarshi Banerjee & Rajiv S. Mishra (2018) Crystallographically degenerate B2 precipitation in a plastically deformed *fcc*-based complex concentrated alloy, Materials Research Letters, 6:3, 171-177, DOI: [10.1080/21663831.2018.1426649](https://doi.org/10.1080/21663831.2018.1426649)

To link to this article: <https://doi.org/10.1080/21663831.2018.1426649>



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



[View supplementary material](#)



Published online: 22 Jan 2018.



[Submit your article to this journal](#)



Article views: 3670



[View related articles](#)



[View Crossmark data](#)



Citing articles: 25 [View citing articles](#)



ORIGINAL REPORT



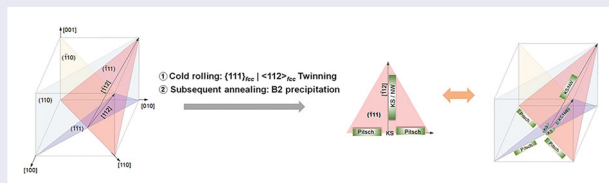
Crystallographically degenerate B2 precipitation in a plastically deformed fcc-based complex concentrated alloy

Deep Choudhuri^{a,b}, Shivakant Shukla^a, Whitley B. Green^a, Bharat Gwalani^{a,b}, Victor Ageh^{a,b}, Rajarshi Banerjee^{a,b} and Rajiv S. Mishra^{a,b}

^aDepartment of Materials Science and Engineering, University of North Texas, Denton, TX, USA; ^bAdvanced Materials and Manufacturing Processes Institute, University of North Texas Denton, TX, USA

ABSTRACT

Bcc-ordered B2 and fcc phases manifest three different orientation relationships (ORs) in the same microstructure: Kurdjumov–Sachs, Nishiyama–Wasserman and Pitsch. This unique microstructure was developed via conventional cold-rolling and subsequent annealing of an fcc-based Al_{0.3}CoCrFeNi complex concentrated alloy (CCA). The degeneracy in crystallographic ORs was caused by {111}⟨112⟩twins, on multiple {111}, from the prior cold-rolling step. Annealing produced B2 precipitates on all the major fcc slip-systems by heterogeneously nucleating B2 at twin-matrix interfaces and twin–twin intersections. Such a precipitation-hardenable microstructure is expected to increase the strength of fcc-based CCAs by effectively blocking 1/2⟨110⟩ and 1/6⟨112⟩ mobile dislocations.



IMPACT STATEMENT

Three different fcc-B2 orientation relationships (ORs) were observed for the first time in complex concentrated alloys. Such degenerate ORs in B2 precipitation can potentially block dislocation on multiple slip planes.

ARTICLE HISTORY

Received 26 October 2017

KEYWORDS

Complex concentrated alloys; precipitation strengthening; bcc-ordered b2; twinning; phase transformations

Main text

Complex concentrated alloys (CCAs, also known as high entropy alloys) with their ever-expanding compositional space [1] provide ample opportunities to engineer microstructures for structural applications. In recent years, studies have shown that phase formation in a large variety of CCAs is driven by enthalpy considerations; rather than entropy [1,2]. Under the circumstances, several alloy compositions are susceptible to secondary phase formation, thus allowing the design of precipitation-hardenable CCA compositions [3–6]. In most cases the strengthening precipitate nucleates homogeneously within the parent matrix, where nucleation involves a complex interplay among chemical, elastic (due to misfit strains) and interfacial energies [7,8].

Precipitation strengthening can also be achieved by pre-deforming, and, subsequent annealing, where dislocations introduced from plastic deformation act as heterogeneous nucleation sites for secondary phase formation [9,10]. This approach predisposes the nucleation of a select number of precipitate variants that are commensurate with the crystallography of deformation features, e.g. ⟨110⟩ and ⟨112⟩, and ⟨112̄⟩ directions in fcc and hcp, respectively [9–11]; meaning potentially fewer precipitate variants form compared to the undeformed condition [9,11]. However, low stacking fault energy fcc-based CCAs, like Al_x–Co–Cr–Fe–Ni ($x = 0–0.3$), are highly susceptible to {111}⟨112⟩ twinning on multiple {111} planes [12,13]. These twins along with dislocations can then act as triggers for precipitating secondary phases

CONTACT Rajiv S. Mishra rajiv.mishra@unt.edu

Supplemental data for this article can be accessed here. <https://doi.org/10.1080/21663831.2018.1426649>

that cover the majority of the *fcc* slip systems and that block mobile dislocations.

In the case of Al-containing CCAs, such precipitation tendencies are expected to be amplified, because Al facilitates the formation of a number of ordered secondary phases, e.g. *bcc*-ordered B2 and L2₁ and *fcc*-ordered L1₂ [1]. This notion was applied in this study to an *fcc*-based Al_{0.3}CoCrFeNi CCA which upon thermomechanical processing produced hard *bcc*-ordered B2 precipitates on multiple slip systems of the parent *fcc* matrix. Such a microstructure was caused by the manifestation of three different Bain transformation-related B2-*fcc* orientation relationships in the same microstructure. The processing conditions used in this study were guided by the current thermodynamic assessment of Al_xCoCrFeNi CCAs [14–16], and the known mechanical properties of Al_{0.3}CoCrFeNi CCA [17].

CCA of composition 6.69%Al–23.14%Cr–23.36%Cr–23.43%Fe–23.38%Ni (at.%) or Al_{0.3}CoCrFeNi was obtained from Sophisticated Alloys. The alloy was then homogenized within a single *fcc* phase-field at a temperature of 1150°C for 1 h (see supplementary Figure S1a), and subsequently cold-rolled (CR) to 50% and 85% thickness reductions (greater than the tensile ductility of the alloy [17]). The homogenized, and homogenized and cold-rolled specimens were then annealed at 800°C (B2 is stable, but L1₂ and sigma phases are unstable [14–16])

for the same length of time, i.e. 50 h. Microstructural characterization of the annealed and thermomechanically processed specimens was performed using scanning electron (FEI Nova NanoLab 200™) and transmission electron (FEI Tecnai F20-FEG™ operating at 200 kV, and outfitted with an EDAX model Tecnai 20 T/20ST energy-dispersive X-ray spectroscopy or energy dispersive spectroscopy (EDS) detector) microscopy. Transmission electron microscope (TEM) foils were prepared via focused ion beam (FIB) milling in an FEI Nova NanoLab 200™ by plucking site-specific FIB-lift-outs.

The effects of cold-rolling (0%, 50% and 85%) and subsequent annealing (at 800°C/50 h) were compared via backscattered electron (BSE) images (Figure 1 (a–c)). Note Figure 1(d) shows a higher magnification region than (b). Direct annealing (without cold-rolling: 0%CR-800°C/50 h (Figure 1(a)) produced only grain boundary B2 precipitates (inset) consistent with our past studies [15]. However, as presented in Figure 1(b,c), prior cold-rolling (50% and 85%) and subsequent annealing substantially altered the distribution of B2 precipitates, their morphology and the distribution of recrystallized *fcc*-grains. The microstructure in the 85%CR-800°C/50 h condition is comprised entirely of recrystallized grains (~5 µm dia.), while 50%CR-800°C/50 h features patchy recrystallized regions within the prior coarse *fcc* grains (supplementary Figure S1b). Regardless, the

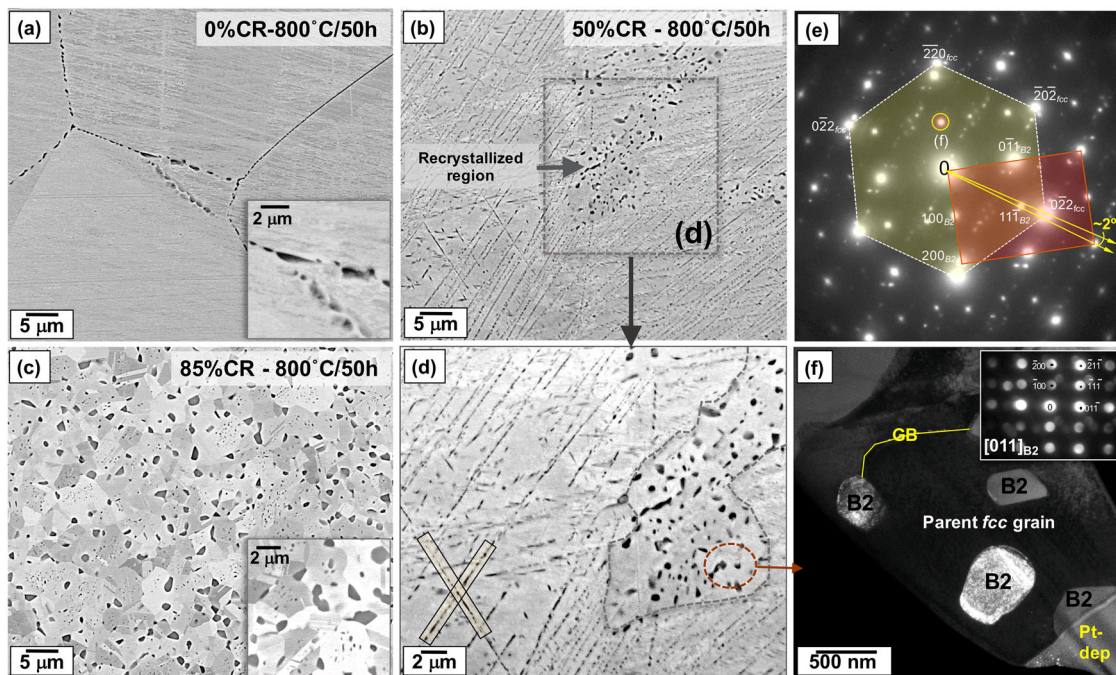


Figure 1. Backscattered secondary electron (BSE) micrographs showing the effect of prior cold-rolling (%CR) followed by subsequent annealing at 800°C/50h: (a) 0%, (b) 50% and (c) 85%. Insets in (a) and (c) show the high magnification view of grain boundary B2, and fine-grained *fcc* and *bcc*-ordered B2 phases, respectively. (d) Higher magnification BSE of 50%CR-800°C/50 h showing rod/plate-like and globular B2 precipitates in within the coarse *fcc* grains and the recrystallized regions, respectively. (e) SADP from *fcc*-B2, and (f) DFTEM of globular B2 particles and the [011] microdiffraction pattern from grain boundary B2.

BSE results demonstrate that prior plastic deformation influences the evolution of different microstructural entities in $\text{Al}_{0.3}\text{CoCrFeNi}$ CCA, and may impact the crystallography of B2 precipitation. This matter was probed next by comparing the 50%CR-800°C/50 h and 85%CR-800°C/50 h microstructures.

A high magnification BSE of 50%CR-800°C/50 h (Figure 1(d)) shows that the B2 precipitates broadly acquired two distinct morphologies. First, globular B2 precipitates (0.2–1 μm) clustered in small pockets (marked with the dotted square region in Figure 1(b) and an irregular dotted area in (d)) within the recrystallized regions; and second, rod/plate-like B2 precipitates (shaded 'X' in Figure 1(d)). Worth pointing out is that 30% CR and subsequently 800°C/50 h annealed microstructure contained rod/plate-like B2 precipitates (not shown). In stark contrast, the 85%CR-800°C/50 h microstructure (Figure 1(c) and inset) is comprised of larger B2 globules (0.5–2 μm) decorating the recrystallized *fcc* grain boundaries, and the intragranular regions (inset of Figure 1(c)). Furthermore, similarities in 50%CR-800°C/50 h and 85%CR-800°C/50 h indicated that globular B2 precipitation is linked with the recrystallization phenomenon, and that 85%CR-800°C/50 h is the extreme case of such microstructural manifestation. The globular B2 precipitates were further examined with a selected area diffraction pattern (SADP) (Figure 1(e)),

and were noted to have the $[111]_{fcc}$ and $[011]_{B2}$ zone axes nearly parallel ($\sim 2^\circ$ off); closest to the Kurdjumov–Sachs (KS) orientation relationship. A dark-field TEM or DFTEM (Figure 1(f)), recorded by selecting the superlattice $[001]_{B2}$ reflection from a $[011]_{fcc}$ SADP, illuminated the B2 globules at the *fcc* grain boundaries (GB) and within intragranular regions—consistent with BSE observations.

In stark contrast, the crystallography of the rod/plate B2 precipitates (shaded 'X' in Figure 1(d)) occupying the coarse un-recrystallized grains of 50%CR-800°C/50 h (Figure 1(b)) is more complicated, because the grains are present as multiple variants lying on different *fcc* planes. This *crystallographic complexity* is shown in Figure 2, containing a high-angle dark-field scanning TEM (HAADF-STEM), where the B2 precipitates can be categorized on the basis of morphology (ellipsoidal and plate/rod like, '1', '2/3'), and crystallographic orientation ('2' and '3'). In the BSE images rod/plate-like B2 precipitates appear as monolithic entities, however the higher magnification S/TEM indicated that each such entity is a cluster of multiple variants. While the details of the crystallographic orientations of the B2 precipitates marked '1', '2', and '3' in Figure 2(a) cannot be determined based on this HAADF-STEM image, the distinctly different two-dimensional projections of these three types of precipitates are clearly visible. Furthermore, STEM-EDS

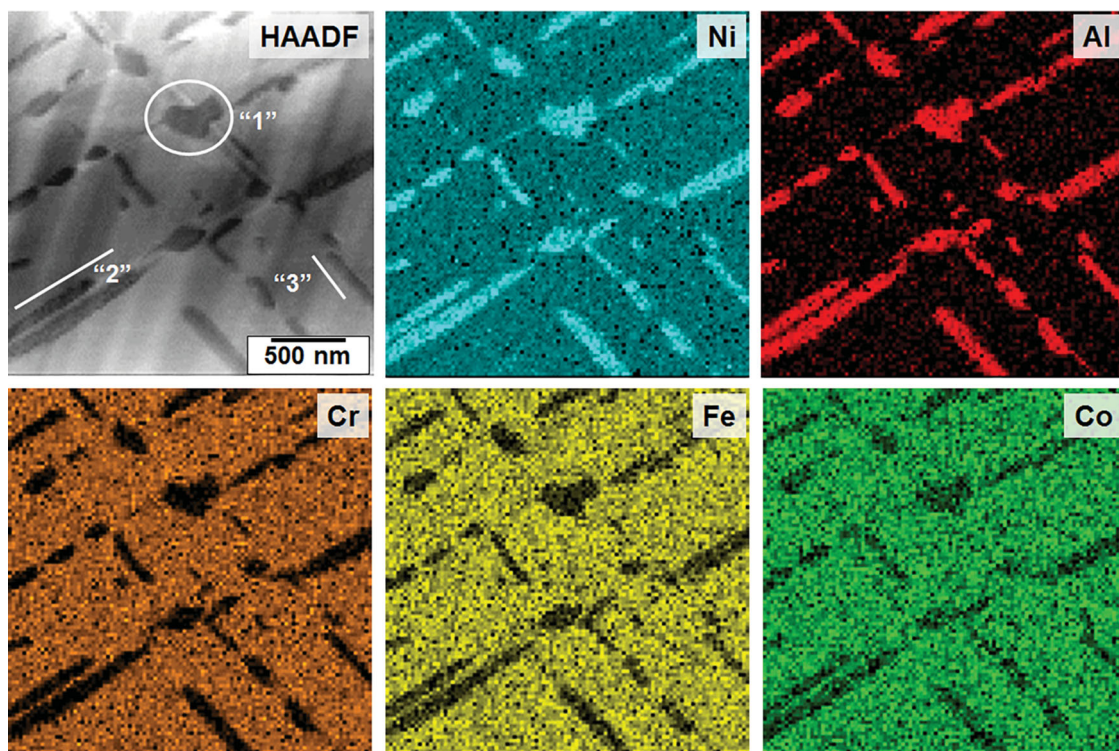


Figure 2. Elemental map of the rod/plate-like precipitates in the coarse grains of the 50%CR-800°C/50 h microstructure. The number marked '1', '2' and '3' marked in the HAADF image categorizes the B2 precipitates in terms of crystallographic alignment and morphology.

maps indicated that Ni and Al heavily partitioned into B2 precipitates (irrespective of their crystallographic orientations and morphology), while the remaining *fcc* matrix was richer in Fe and Cr. Multiple 1D concentration profiles revealed that the B2 composition was 55.5%Ni–19.0%Al–14.3%Co–9.2%Fe–2.0%Cr (in at.%), while the *fcc* matrix was 2.2%Al–28.0%Co–23.5%Cr–25.1%Fe–21.2%Ni (at.%) similar to a $\text{Al}_{0.1}\text{CoCrFeNi}$ CCA (supplementary Figure S2a). Similar compositional partitioning was noted in the globular B2 precipitates

(supplementary Figure S2b). Therefore, the influence of compositional partitioning on such B2 precipitation could not be assessed from the given heat treatments, and the remainder of the article will focus on the crystallography of these complex B2 precipitates.

The crystallographic alignment and morphology of B2 variants in 50%CR-800°C/50 h condition were further examined by recording SADPs along different *fcc* zone axes (Figure 3(ac and e)), and corresponding DFTEMs from the $[001]_{\text{B2}}$ superlattice reflections

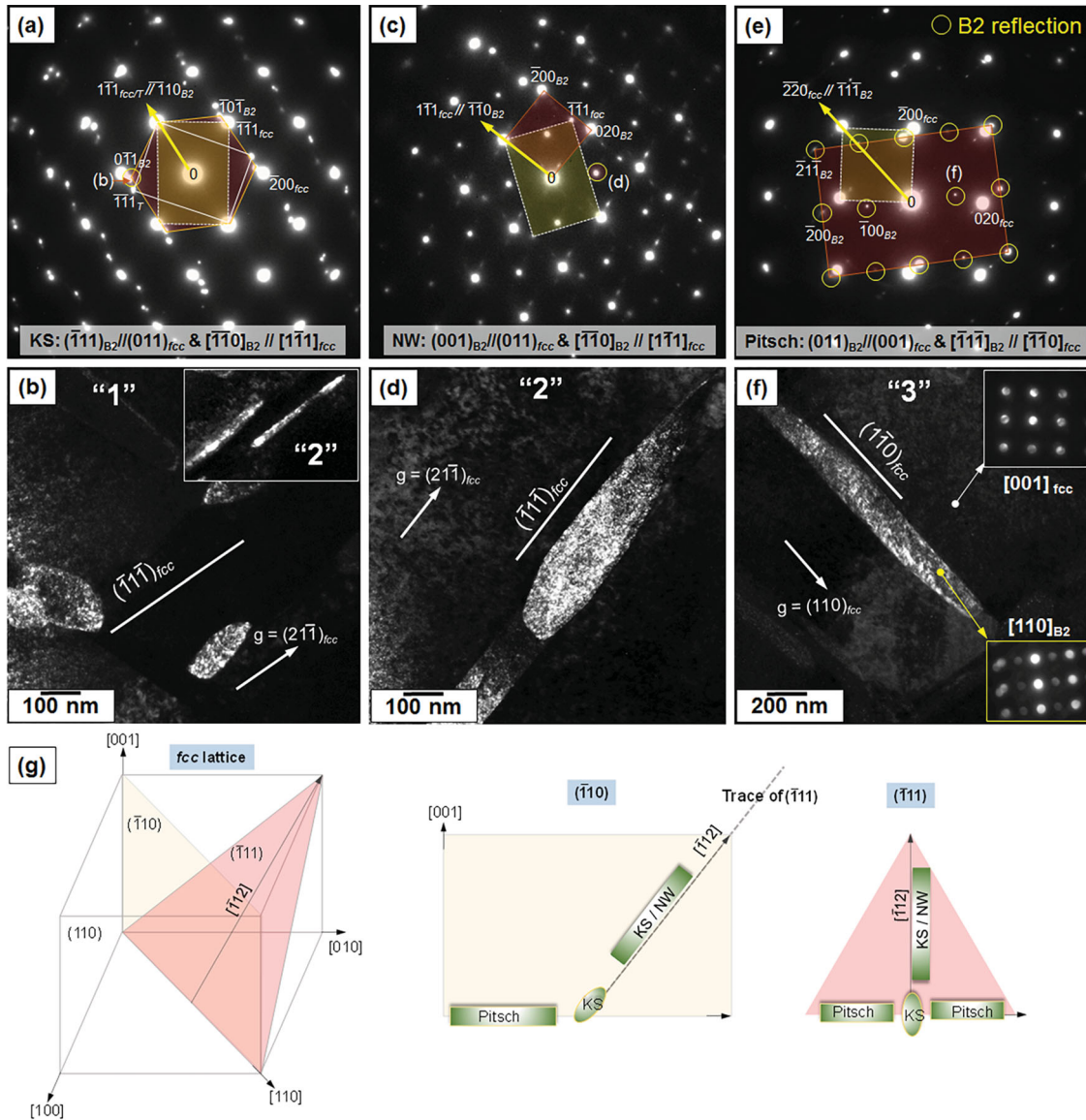


Figure 3. TEM results showing the degeneracy in fcc-B2 orientation relationships (ORs) inside the coarse grains of the 50%CR-800°C/50 h microstructure via SADP (a,c,e) and the corresponding DFTEMs (b,d,f). (a) SADP recorded along the $[011]_{\text{fcc}}$ zone axis and depicts a Kurdjumov–Sach (KS) OR between the (b) ellipsoidal and rod/plate-like (top-right inset) precipitates and the parent *fcc* matrix. (c) $[011]_{\text{fcc}}$ SADP indicating Nishiyama–Wasserman (NW) OR between the *fcc* matrix and (d) rod/plate-like B2, which had a different crystallographic alignment than NW-B2s. (e) $[001]_{\text{fcc}}$ SADP indicating Pitsch OR between the *fcc* matrix and (f) rod/plate-like B2. In all cases DFTEM were recorded by selecting the $[001]_{\text{B2}}$ super lattice reflections. All DFTEMs were recorded by selecting the $[001]_{\text{B2}}$ super lattice reflections. (g) Schematic depicting the approximate crystallographic locations of B2 precipitates based on their ORs with *fcc* lattice. In (a) twins also share a KS OR with B2 precipitates (marked with arrows in (a)).

(Figure 3(bdf)). The SAPD in Figure 3(a), which was recorded along $[011]_{fcc}$ was comprised of B2 (forming a $[111]_{B2}$ electron diffraction pattern), twin and the primary fcc reflections. As expected, these B2 precipitates shared a KS OR with both parent matrix and the twin, i.e. $(011)_{B2} // (111)_{fcc}$ and $[\bar{1}\bar{1}\bar{1}]_{B2} // [\bar{0}\bar{1}\bar{1}]_{fcc}$. The corresponding DFTEM in Figure 3(b) shows the crystallographic alignment of ellipsoidal ('1') and rod/plate-like ('2') B2 precipitates with the parent fcc matrix, which is oriented close to $[011]_{fcc}$. Interestingly, the rod/plate-like B2 precipitates also exhibited different ORs (other than KS) in the same microstructure. Figure 3(c,d) shows B2 rod/plate ('2') depicting Nishiyama–Wasserman or NW OR ($(011)_{B2} // (001)_{fcc}$ and $[\bar{1}\bar{1}\bar{0}]_{B2} // [\bar{1}\bar{1}\bar{1}]_{fcc}$), while the '3' variant (Figure 3(e,f)) shared Pitsch OR ($(001)_{B2} // (011)_{fcc}$ and $[\bar{1}\bar{1}\bar{1}]_{B2} // [\bar{1}\bar{1}\bar{0}]_{fcc}$) with the parent matrix. The SAPD in Figure 3(c) also contains reflections from twins and double diffraction. Discussion regarding the effect of twins on B2 formation will be discussed in a later section. In terms of crystallographic alignment within the parent matrix, the precipitates exhibiting KS and NW ORs appear to lie approximately on $\{111\}_{fcc}$ and are extended along $\langle 112 \rangle_{fcc}$ (Figure 3(b,d)). In comparison, the rod/plates displaying Pitsch OR were oriented along $\langle 110 \rangle_{fcc}$ and lay close to $\{110\}_{fcc}$ (Figure 3(f)). Based on TEM observations, Figure 3(g) schematically summarizes the 'approximate crystallographic locations' of the different B2 variants (i.e. '1', '2', and '3') on a pair of $\{110\}_{fcc}$ and $\{111\}_{fcc}$ planes. Crucially, we find that *significant crystallographic degeneracy in precipitate/matrix orientation relationships (KS, NW and Pitsch)— crystallographic alignment ($\langle 112 \rangle_{fcc}$ and $\langle 110 \rangle_{fcc}$) and habit planes (near $\{111\}_{fcc}$ and $\{110\}_{fcc}$)—can be achieved by the same precipitate phase in a CCA.*

To rationalize the crystallographic degeneracy in B2 formation, we carried out TEM examination of the deformation features in 50% CR-800°C/50 h processed (Figure 4(a,b)), and only 50% cold-rolled (Figure 4(c-e)) microstructures. The bright-field TEM (BFTEM) image in Figure 4(a) reveals the presence of twins on conjugate $\{111\}_{fcc}$ planes (shown with lines) and B2 precipitates. The DFTEM in Figure 4(b) and the inset $[011]_{fcc}$ SDAP confirm the presence of $\{111\}_{fcc} / \langle 112 \rangle_{fcc}$ twin in the annealed microstructure. These twins are remnants from the 50% cold-rolled microstructure in the DFTEMs presented in Figure 4(c,d) and the BFTEM in Figure 4(e). Note that two variants of twins in the 50% CR microstructure and, presumably, these variants, govern the crystallography of the B2 precipitates in the 50%CR-800°C/50H microstructure.

Here, B2 precipitates decorate the twin broad-faces (or twin-matrix interface) and are also present near

the twin–twin intersections (Figure 4(a)). Larger B2 rods/plates were also noted along the twin broad-faces in other regions (BFTEM inset in Figure 4(a) and DFTEM inset in Figure 3(b)). Further evidence of twin-mediated B2 precipitation can be gleaned from the SADP in Figure 3(a), where the B2 precipitates also share a KS OR with twins. The NW (Figure 3(c,d)) and Pitsch (Figure 3(e,f)) oriented B2 precipitates were well developed; consequently, in this case, a twin-precipitate relationship could not be established unambiguously. However, the NW and Pitsch crystallography provides some guidance. In case of NW, the $\{011\}_{B2}$ is parallel to the $\{111\}_{fcc}$ —twin habit plane; meaning twins may be associated with NW-B2 precipitation. However, such twin–Pitsch B2 relationship is absent, since that OR does not involve $\{111\}_{fcc}$ (Figure 3(c)). Thus, the TEM observations indicated that prior twinning features can be linked to B2 precipitation in the 50%CR-800°C/50 h microstructure.

Additional insights into B2 precipitation can be obtained by examining the 50%CR microstructure (Figure 4(c,e)). The DFTEM in Figure 4(c,d) shows two variants of twins on the conjugate (separated by $\sim 70/109^\circ$), which corresponds to the relative crystallographic alignments of the B2 precipitates and fcc matrix (Figure 2, 3, and 4(a)). Note that the two twin variants also differed in their respective widths. The wider variant (4(c)) appears to lie on the $(111)_{fcc}$, while the secondary thinner variant (Figure 4(d)) in certain locations 'curves' away from the conjugate $(111)_{fcc}$ plane (indicated with a straight line). The latter observation suggests that the secondary twins have been intersected by other deformation entities (e.g. dislocations, fine-scale tertiary twins) during the cold-rolling process [18,19]. A careful examination of the secondary twins (dotted rectangle in Figure 4(d)) also implies that they have emanated from a wider variant (#1). Such twin intersections denote extensive strain-hardening [18,19], which locally alters the matrix orientation. Evidence of such lattice misorientation can be seen in the 'smearing/streaking' of the primary fcc and twin reciprocal lattice reflections (inset Figure 4(e)). Under the circumstances, such regions with presumably high local strain energy potentially act as sites for B2 precipitation; e.g. ellipsoidal B2 at twin intersection (Figure 4(a)) or Pitsch-oriented B2 precipitates, and recrystallized fcc grains during annealing at 800°C (Figure 1(b)). Further studies are in progress to resolve this matter fully.

Finally, the detailed characterization of degenerate B2 containing 50%CR-800°C/50 h microstructure permitted us to postulate its deformation mechanism. The deformation behavior of the ductile fcc matrix will be comparable to $Al_{0.1}CoCrFeNi$ CCA (STEM-EDS,

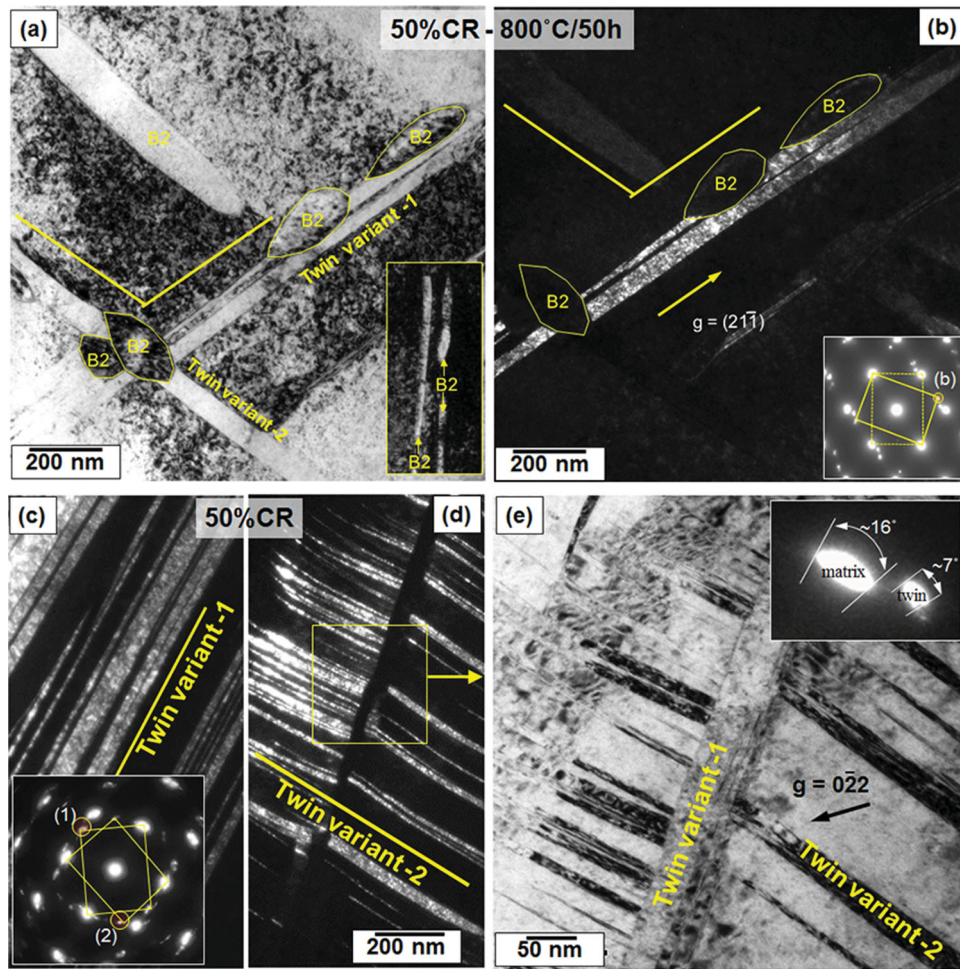


Figure 4. TEM comparing the deformation features in the (a) and (b) 50%CR-800°C/50 h (cold rolled and annealed), and (c)–(e) 50%CR (only cold-rolled) microstructures. (a) Bright-field TEM (BFTEM) image recorded with beam close to $[001]_{fcc}$ shows B2 precipitates next to two variants of twins, while (b) shows the corresponding DFTEM of one such twin. Note that B2 precipitates are decorating the twin broad-faces (twin matrix interface) and twin–twin intersections in (a). The bottom-right insets in (a) and (b) show B2 adjacent to a twin in another region and $[001]_{fcc}$ SADP containing the twin reflection. The straight lines in (a) and (b) are the conjugate $\{111\}$ planes. (c) and (d) show the DFTEMs of two different variants of deformation twins in the 50% cold-rolled microstructure. Inset in (c) shows the $[011]_{fcc}$ SADP with reflections from the two twin variants. (e) BFTEM recorded with $g = 0\bar{2}2$ shows a region of intersection between the two twin variants.

Figure 2 and supplementary Figure S2) or any other single-phase fcc CCAs, i.e. involving $\frac{1}{2} \langle 110 \rangle \{111\}$ planar slip, $\frac{1}{6} \langle 112 \rangle \{111\}$ partials, stacking fault formation and deformation twinning [20–22]. The B2 precipitates will provide additional strength by virtue of their placement on the $\{111\}_{fcc}$ planes and orientation along the fcc Burgers vectors, i.e. $\langle 110 \rangle_{fcc}$ and $\langle 112 \rangle_{fcc}$ (Figure 3(g)), which effectively blocks mobile dislocations. The recrystallized regions containing inter- and intragranular B2 precipitates (Figure 1(d)) are expected to further enhance the strength of $Al_{0.3}CoCrFeNi$. Preliminary tensile responses revealed that, compared to the single-phase fcc microstructure, B2 precipitation caused a fourfold increase in yield strength (~ 700 MPa), and high

ultimate tensile (~ 1000 MPa) strength while retaining reasonable ductility (see supplementary Figure S3).

Summarizing, a unique B2 precipitate-strengthened, fcc -based CCA microstructure was obtained by thermo-mechanical processing $Al_{0.3}CoCrFeNi$ alloy, where the precipitates were aligned along multiple crystallographic orientations that correspond to the slip systems in the fcc lattice. Such crystallographic/orientation degeneracy is manifested due to three different B2- fcc orientation relationships; i.e. Kurdjumov–Sachs, Nishiyama–Wasserman, and Pitsch in the same microstructure. Prior deformation twins were intimately linked with such crystallographic degeneracy, and grain boundaries also facilitated B2 precipitation. Together, these microstructural

entitles are expected to enhance strength while maintaining reasonable ductility of thermomechanically processed $\text{Al}_{0.3}\text{CoCrFeNi}$.

Acknowledgements

The work was performed under a cooperative agreement between the Army Research Laboratory and the University of North Texas (W911NF-16-2-0189). We also acknowledge the Materials Research Facility at UNT for microscopy facilities.

Disclosure statement

No potential conflict of interest was reported by the authors.

Funding

This work was supported by Army Research Laboratory: [Grant Number W911NF-16-2-0189].

References

- [1] Miracle DB, Senkov ON. A critical review of high entropy alloys and related concepts. *Acta Mater.* **2017**;122:448–511.
- [2] Choudhuri D G, Gorsse S, Mikler CV, et al. Change in the primary solidification phase from fcc to bcc-based B2 in high entropy or complex concentrated alloys. *Scripta Mater.* **2017**;127:186–190.
- [3] Zhang Y, Zuo TT, Tang Z, et al. Microstructures and properties of high-entropy alloys. *Prog Mater Science.* **2014**;61:1–93.
- [4] He JY, Wang H, Huang HL, et al. A precipitation-hardened high-entropy alloy with outstanding tensile properties. *Acta Mater.* **2016**;102:187–196.
- [5] Chuang MH, Tsai MH, Wang WR, et al. Microstructure and wear behavior of $\text{Al}_{x}\text{Co}_{1.5}\text{CrFeNi}_{1.5}\text{Tiy}$ high-entropy alloys. *Acta Mater.* **2011**;59:6308–6317.
- [6] Lu ZP, Wang H, Chen MW, et al. An assessment on the future development of high-entropy alloys: summary from a recent workshop. *Intermetallics.* **2015**;66:67–76.
- [7] Porter DA, Easterling KE, Sherif M. Phase transformations in metals and alloys. Boca Raton, FL: CRC Press; **2009**.
- [8] Khachatryan AG. Theory of structural transformations in solids. New York: Courier Corporation; **2013**.
- [9] Furuhashi T, Maki T. Variant selection in heterogeneous nucleation on defects in diffusional phase transformation and precipitation. *Mater Sci Eng A.* **2001**;312:145–154.
- [10] Choudhuri D, Dendge N, Nag S, et al. Role of applied uniaxial stress during creep testing on precipitation in Mg–Nd alloys. *Mater Sci Eng A.* **2014**;612:140–152.
- [11] Qiu D, Shi R, Zhang D, et al. Variant selection by dislocations during α precipitation in α/β titanium alloys. *Acta Mater.* **2015**;88:218–231.
- [12] Zaddach AJ, Niu C, Koch CC, et al. Mechanical properties and stacking fault energies of NiFeCrCoMn high-entropy alloy. *JOM.* **2013**;65:1780–1789.
- [13] Zhang Z, Sheng H, Wang Z, et al. Dislocation mechanisms and 3D twin architectures generate exceptional strength-ductility-toughness combination in CrCoNi medium-entropy alloy. *Nat Commun.* **2017**;8:14390–14397.
- [14] Zhang C, Zhang F, Chen S, et al. Computational thermodynamics aided high-entropy alloy design. *JOM.* **2012**;64:839–845.
- [15] Gwalani B, Soni V, Choudhuri D, et al. Stability of ordered L1₂ and B2 precipitates in face centered cubic based high entropy alloys- $\text{Al}_{0.3}\text{CoFeCrNi}$ and $\text{Al}_{0.3}\text{CuFeCrNi}_2$. *Scripta Mater.* **2016**;123:130–134.
- [16] Gwalani B, Choudhuri D, Soni V, et al. Cu assisted stabilization and nucleation of L1₂ precipitates in $\text{Al}_{0.3}\text{CuFeCrNi}_2$ fcc-based high entropy alloy. *Acta Mater.* **2017**;129:170–182.
- [17] Shun TT, Du YC. Microstructure and tensile behaviors of FCC $\text{Al}_{0.3}\text{CoCrFeNi}$ high entropy alloy. *J Alloys Compounds.* **2009**;479:157–160.
- [18] El-Danaf E, Kalidindi SR, Doherty RD. Influence of grain size and stacking-fault energy on deformation twinning in fcc metals. *Metall Mater Trans A.* **1999**;A 30:1223–1233.
- [19] Asgari S, El-Danaf E, Kalidindi SR, et al. Strain hardening regimes and microstructural evolution during large strain compression of low stacking fault energy fcc alloys that form deformation twins. *Metall Mater Trans A.* **1997**;A 28:1781–1795.
- [20] 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.
- [21] Laplanche G, Kostka A, Horst OM, et al. Microstructure evolution and critical stress for twinning in the CrMnFeCoNi high-entropy alloy. *Acta Mater.* **2016**;118:152–163.
- [22] Choudhuri D, Komarasamy M, Ageh V, Mishra RS. *manuscript under review*.