Microstructure and Texture Evolution in Thermomechanically Processed FCC Metals and Alloys: a Review

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Abstract: The stacking fault energy (SFE) of face-centered cubic (FCC) alloys is a critical parameter that controls microstructural and crystallographic texture evolution during deformation and annealing treatments. This review focuses on several FCC alloys, aluminum (Al), copper (Cu), austenitic stainless steels (ASSs), and high entropy alloys (HEAs), all of which exhibit varying SFEs. These alloys are often subjected to thermo-mechanical processing (TMP) to enhance their mechanical properties. TMP leads to the evolution of deformation-induced products, such as shear bands (SBs), strain-induced martensite (SIM), and mechanical/deformation twins (DTs) during plastic deformation, while also influencing crystallographic texture. High-medium SFE materials, such as Al and Cu, typically exhibit the evolution of Copper-type texture during room temperature rolling (RTR), while low SFE materials, such as ASSs and HEAs, display Brass-type texture at high reduction ratios. Moreover, the presence of second-phase particles/precipitates can also impact the microstructure and texture evolution in Al and Cu alloys. Particle-stimulated nucleation (PSN) during the annealing treatment has been reported for Al, Cu, ASSs, and HEAs, which causes texture weakening. Another interesting observation in severely deformed Cu alloys is the room-temperature softening phenomenon, which is discussed in the reviewed work. Additionally, plastic deformation and heat treatment of ASSs result in phase transformation, which was not observed in Al, Cu, or HEAs. Furthermore, the dependence of special boundaries in HEAs on plastic deformation temperature, strain rate, and annealing temperature is also discussed. Thus, this review comprehensively reports on the impact of TMP on microstructural and crystallographic texture evolution during plastic deformation and the annealing treatment of Al, Cu, ASSs, and HEAs FCC materials, using results obtained from electron microscopy.

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1. INTRODUCTION

FCC alloys have been extensively used in various industries, including nuclear plants, aerospace, information technology, the automobile sector, and food industry, due to their lightweight, favorable strength-to-weight ratio, good recyclability, low price, and high corrosion resistance in different environments [1]. Aluminum (Al) and Copper (Cu)
belong to the category of FCC materials and have attracted significant attention across diverse industries, such as the electronics, utensils, energy transportation, and electromechanical fields. Al alloys are typically manufactured using either direct-chill (DC) casting or twin-roll casting (TRC)/continuous casting. DC casting involves several steps, including casting, scalping, homogenization, hot rolling (HR), room-temperature rolling (RTR), and annealing. In contrast, TRC entails continuous casting, RTR, and annealing, which notably reduces production costs [2].

This review discusses differences in the microstructure and texture evolution in DC and TRC cast Al alloys, in subsequent sections. Due to the significant differences in stacking fault energy (SFE) between Al and Cu alloys, the evolution of deformation-induced products, such as deformation twins (DTs), strain localizations, and shear bands (SBs), varies considerably in these two FCC materials. Al alloys have high SFE, while Cu alloys are considered medium-low SFE materials. Pure Cu has an SFE of 78 mJ/m², while the addition of Al, in amounts up to 2% and 4.5%, decreases the SFE of Cu alloys to minimum values of 37 and 7 mJ/m², respectively [3]. Materials with low SFE have large stacking fault (SF) regions that hinder dislocation motion via slip during plastic deformation. However, the twinning mechanism is very active in low SFE materials and facilitates further deformation. Plastic deformation by twinning is reported to enhance the work-hardening rate while simultaneously increasing strength and ductility [3]. Since Cu alloys are predominantly used in applications requiring high electrical conductivity, achieving a trade-off between high conductivity and high strength is vital. Thermo-mechanical processing (TMP) of Cu-Fe-P, Cu-Cr, Cu-Mg, Cu-Sn, and Cu-Fe alloys strengthens the alloys through solid solution strengthening, work-hardening, precipitation hardening, and also enhances the conductivity of the final Cu products [4].

Further, this review discusses the evolution of microstructure in austenitic stainless steels (ASSs) and high entropy alloys (HEAs). These alloys fall under the category of FCC materials, with low SFE values of 18 mJ/m² and 3.5 mJ/m², respectively [5]. ASSs are principally divided into two series: the 300 series and the 200 series. In the 200 series, nickel is replaced by nitrogen and/or manganese to stabilize the austenitic structure [6]. Due to the rising cost of nickel, industries have pivoted towards low-nickel or nickel-free ASSs (200 series). ASSs are widely employed in the fabrication of high-performance pressure vessels and in medical/surgical tools due to their superior corrosion properties and high formability [7]. However, their low yield strength limits their applications; this limitation can be mitigated through TMP, strain hardening, and solid solution strengthening [8]. Additionally, since rolling of ASSs is an inevitable stage during the fabrication process, the evolution of strain-induced martensite (α’), deformation bands, microbands, and deformation twins (DTs) during rolling has been reported.

HEAs are alloys composed of multiple principal elements, with each element comprising between 5% and 35% of the total composition [9]. Initially, their high configurational entropy was believed to be the sole factor affecting phase stability. However, the role of mixing enthalpy was later found to be equally critical for phase stabilization [10]. The equiatomic composition Co_{20}Cr_{20}Fe_{20}Mn_{20}Ni_{20}, popularly known as the Cantor alloy, exists in a single FCC phase [11]. The SFE, estimated by density functional theory (DFT) calculations for the Cantor alloy, is found to be approximately 20-25 mJ/m², placing it in the low SFE regime [12]. Consequently, HEAs, which are classed as low SFE materials, characteristically undergo twinning during deformation (a twinning-induced plasticity effect [13]) and subsequent annealing heat treatment. Deformation twinning (DT) incrementally introduces more twin interfaces, which act as barriers (akin to grain boundaries (GBs)) to dislocation motion, and reduce the mean free path of dislocations,
culminating in the dynamic Hall-Petch effect [14]. The twinning-induced plasticity effect in HEAs, combined with the dynamic Hall-Petch effect, results in higher ductility and strain hardening rates, giving them excellent mechanical properties [15–17].

FCC metals/alloys are widely used in almost every sector, as highlighted above and depicted in Fig. 1. The requirement for good formability along with high strength, especially in the creation of automotive frames, has been met by employing Al alloys and ASSs (Fig. 1(a)). With the development of electric vehicles, the usage of lightweight materials such as Al and Cu alloys has increased (Fig. 1(b)). ASSs, known for their high corrosion resistance and strength, are used to manufacture containers for the chemical, food, and paper industries (Fig. 1(c)). HEAs have been developed to maintain mechanical strength at the high temperatures found in airplane engines, thereby enhancing engine efficiency (Fig. 1(d)). The fabrication of these alloys primarily consists of casting, rolling at high and room temperatures, followed by intermediate annealing or solution annealing, etc., to create the final product for various applications. Predominantly, these processes are used to produce metal sheets, and during the fabrication process, these alloys often undergo the evolution of various microstructures and crystallographic textures.

The microstructure and texture of all metals/alloys have a strong relationship with the material properties. For example, the balance between the deformation texture components (Copper (112)<111>, S (123)<634> and Brass (110)<112>) and annealing textures (011)<122> and Cube [001]<100> strongly influences the drawability of Al alloys. A higher fraction of deformation texture forms ears 45° to the rolling direction (RD), while a higher fraction of annealing texture forms ears 0°/90° to the RD. Besides manufacturing processes, processing methods such as severe plastic deformation (SPD), dissimilar rolling, and accumulative roll bonding can also be employed to tailor the microstructure and texture [18–20]. The subsequent sections discuss the differences in deformed and recrystallized microstructures, along with particle simulated nucleation (PSN) and texture evolution, for various FCC alloys. In this review, the authors provide insight into the generation of various microstructures and textures that evolve during the deformation and annealing treatment of FCC materials (such as Al and Cu alloys, ASSs, and HEAs). TMP, a process widely used to enhance the properties of alloys, results in the formation of a fine-grained structure, thus enhancing mechanical and chemical properties. Therefore, the authors discuss the impact of TMP on deformation and texture evolution in several FCC metals and alloys.

2. ALUMINUM ALLOYS

The deformation and recrystallization behavior of Al alloys is strongly influenced by the presence of precipitates or intermetallic particles. Achieving optimal deformability in Al alloys necessitates controlling the microstructure, especially the size of second-phase particles. The second-phase particles, which include α-Al(Mn,Fe)Si, β-Al(Mn,Fe)Si, θ-phase, Al6Mn, Mg2Si, Al2Cu, S-phase (Al2CuMg), η-phase (Mg(Al,Cu,Zn)2), and Al-Cu-Fe, influence the mechanical and chemical properties of Al alloys [21–25]. Non-deformable second-phase particles (size>1 mm) create strain incompatibility at the matrix-particle interface, leading to significant problems during formability operations. Processing routes define the shape and morphology of the second-phase particles. DC cast Al alloys contain intermetallics in the form of a Chinese script [26], whereas TRC exhibits center line segregation (CLS) [27,28]. For TRC AA8011, it has been noted that homogenization at ~580 ºC for 12 h can reduce the size of intermetallics in CLS, though complete elimination has not been reported [29].

Fig. 2 illustrates the differences in the microstructure and second-phase particles in DC cast and TRC Al alloys. Equiaxed grain morphology was observed for DC cast AA3000 alloys (Figs. 2(a)). For instance, the average grain size (GSavg) after DC cast was 153±47 µm for AA3003. Intermetallics (Chinese script) formed along the grain boundaries (GBs), as shown in Figs. 2(b). During grain formation, Fe/Si/Mn were ejected due to their low solubility in Al (Fe solubility in Al is 0.05 wt.% at 650 ºC, Si is 1.65 wt.% at 577 ºC, and Mn is 1.8 wt.% at eutectic temperature [21,23]), forming a Chinese script of Al-Fe-Mn-Si along the GBs [30]. Because of the continuous morphology of the intermetallics, it is difficult to measure their exact size. On the other hand, coarse elongated grains
Solidification occurred from either side of the roller towards the center of the cast strip. Therefore, Fe and Si, ejected at the center and due to their low solubility in Al, formed CLS directly in the center along the production direction, as shown in Fig. 2(d) [27]. CLS, also continuous in nature, consists of numerous branches of intermetallics along the transverse direction. Consequently, changes in the casting methods alter the types of microstructure and the formation of intermetallics in the Al alloys. Of course, a change in texture is expected for both the DC and TRC routes since the formation of grains varied under both conditions.

Fig. 3(a) shows the $\varphi_2 = 0^\circ$, $45^\circ$, and $65^\circ$ constant sections of orientation distribution functions (ODFs) for the FCC metals/alloys. Deformation texture in Al alloys is composed of $\beta$-fiber, which is a continuous tube of texture components that connects the Copper component to the Brass component through the S, while recrystallization texture consists of Cube and Goss components. However, the intensity of the textures depends on various factors, such as % deformation, strain rate, temperature/time of annealing [27], and intermetallic

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**Fig. 2.** Influence of different casting routes (DC cast/TRC) on microstructure evolution: (a, b) DC cast AA3003 [26], (c, d) TRC AA8011 [27]. Note: micrographs illustrating EBSD ND-IPFs (a, c) and corresponding SEM microstructures (b, d).

**Fig. 3.** Texture evolution: $\varphi_2 = 0^\circ$, $45^\circ$, $65^\circ$ constant sections of orientation distribution functions (ODFs) illustrating the (a) standard component of texture that present typically in Al alloys, (b) after DC-cast AA3003 [26] and (c) after TRC AA8011 [27].
After Deformation

Fig. 4. Influence of different types of deformation routes on microstructure development in Al alloys: (a, b) 95% hot rolled (HR) at 500 °C [26], (c, d) 72% room temperature rolled (RTR) [26], and (e, f) 72% RTR [27] after TRC.

2.1. Deformed microstructure and texture evolution

Deformation in Al alloys increases the stored energy (SE) due to the increment of dislocation density [31]. Differences in deformation routes lead to variations in the evolved microstructure (see Figs. 4(a-f)). After hot rolling (95% HR) at 500 °C and 72% RTR of DC-cast AA3003 [26], the grains transform into a lamellar/banded form (elongated in the rolling direction (RD)), as depicted in Figs. 4(a,c) [26]. Additionally, there are unindexed regions (black regions) due to high dislocations and intermetallic particles. The Chinese script also gets fragmented and scattered throughout the Al-matrix along the RD after the RTR, as demonstrated in Figs. 4(b,d). For instance, the intermetallic size varies from ~ 0.2 µm to ~ 10 µm after HR and RTR [26].

Further, it has been reported that the presence of coarse intermetallics of Al-Mn-Si (after casting) causes grain fragmentation, exhibiting intragranular misorientation after severe deformation. The coarse intermetallics also get fragmented along the RD during 98.84% RTR and refined to the size of ~ 0.2 to ~ 4 µm [32]. Conversely, after TRC, there is no need for additional processes such as HR, unlike with DC-cast alloy, since TRC produces a strip with a thickness of approximately 7-10 mm that does not require heavy deformation. However, RTR is used to deform the material up to the required thickness, for example, 2 mm in the case of utensil applications. During RTR, the elongated grains and center line segregation of TRC AA8011 were further deformed, leading to the development of a banded structure of grains and fragmented intermetallic particles, as shown in Figs. 4(e, f) [27]. Additionally, refinement of the elongated grains and a decrease in the CLS (the branches of the CLS come together) were observed [27].

In addition to RTR, equal channel angular pressing (ECAP) has also been used to tailor the microstructure. Pokova et al. [18] tailored the microstructure of TRC AA3003 via ECAP and, after four instances of ECAP, elongated grains transformed into ultra-fine grains with a GSavg of 0.5 µm. The transmission electron microscope (TEM) images also confirmed that the ECAP process evolved subgrains of 500 nm [18]. Therefore, it can be concluded that the HR/RTR/ECAP deformation route leads to the development of a banded/lamellar type of grain morphology and the fragmentation of coarse intermetallic particles along the RD. Additionally, plastic deformation increases the stored energy (SE) by enhancing dislocation density, which will aid in the recrystallization of new grains.
during subsequent annealing. With changes in microstructure, the crystallographic texture also undergoes significant alterations, dependent on the chosen deformation route. Figs. 5(a-c) display the constant sections ($\phi_2 = 0^\circ$, $45^\circ$, and $65^\circ$) of ODFs for two processing routes: DC cast (Figs. 5(a,b)) [26] and TRC (Fig. 5(c)) [27], across different Al alloys (AA3003, AA5XXX, AA8011) and post-processing methods (HR, RTR). The volume fraction of texture components is presented in Table 1. After deformation (HR/RTR), the rolling texture components, such as Brass, Copper, Goss, and S, are strengthened, while the Cube component decreases, as depicted in Figs. 5(a,b) for the DC cast [26,32].

Similar strengthening of deformation texture components (Brass, Copper, S) was observed after 72% RTR of TRC AA8011, as displayed in Fig. 5(c). This means the rolling texture components were strengthened with a weakening of the Cube component during HR/RTR deformation routes, which was independent of the casting route. However, the magnitude of each component varied [26,27,29,32] based on the strain rate, strain, and the HR temperature, as reported in Table 1.

The balance between rolling texture and recrystallization texture (Cube) is crucial for optimal deformability/drawability. Thus, the Cube component must be present after deformation to evolve again through strain-induced boundary migration during subsequent annealing [33]. For instance, up to ~1.7% and ~1.3% of Cube was retained after RTR of the DC-cast (Fig. 5(b)) and TRC (Fig. 5(c)) samples, respectively, as provided in Table 1.

2.2. Heat-treated microstructure and texture evolution

After deformation, the material becomes significantly work-hardened due to an increase in dislocations, making it difficult to form any product from the deformed sheets. Therefore, an annealing heat treatment is required to develop strain-free grains. Fig. 6 illustrates the microstructure and texture development during the annealing treatment of DC and TRC processed Al alloys. In the DC cast, after annealing at 450 °C for 16 h, the deformed banded structure was transformed into equiaxed grains of ~24 μm (Fig. 6(a)) [26]. The driving force for recrystallization was the stored energy of deformation. After annealing, the amounts of Fe, Si, and Mn in the intermetallics also increased, forming stable α-

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Table 1. Volume fraction of different texture components for different processing conditions in Al alloys.

<table>
<thead>
<tr>
<th>Sr. No</th>
<th>Sample Condition/Volume Fraction (%)</th>
<th>Brass</th>
<th>Copper</th>
<th>Goss</th>
<th>S</th>
<th>Cube</th>
</tr>
</thead>
<tbody>
<tr>
<td>1.</td>
<td>DC-AA3003 [26]</td>
<td>5.5</td>
<td>5.1</td>
<td>2.5</td>
<td>8.3</td>
<td>3.5</td>
</tr>
<tr>
<td>2.</td>
<td>DC-AA3003+95% HR [26]</td>
<td>12.8</td>
<td>12.1</td>
<td>5.9</td>
<td>19.0</td>
<td>1.9</td>
</tr>
<tr>
<td>3.</td>
<td>DC-AA3003+95% HR+72% RTR [26]</td>
<td>15.0</td>
<td>15.9</td>
<td>6.8</td>
<td>28.3</td>
<td>1.7</td>
</tr>
<tr>
<td>4.</td>
<td>DC-AA3003+95% HR+72% RTR+Annealed (450°C – 16 h) [26]</td>
<td>12.3</td>
<td>9.7</td>
<td>3.5</td>
<td>19.2</td>
<td>1.5</td>
</tr>
<tr>
<td>5.</td>
<td>TRC AA8011 [27]</td>
<td>5.7</td>
<td>12.2</td>
<td>3.4</td>
<td>17.4</td>
<td>3.7</td>
</tr>
<tr>
<td>6.</td>
<td>TRC AA8011+72% RTR [27]</td>
<td>12.5</td>
<td>16.1</td>
<td>8.8</td>
<td>24.3</td>
<td>1.3</td>
</tr>
<tr>
<td>7.</td>
<td>TRC AA8011+72% RTR+Annealed (375°C – 4 h) [27]</td>
<td>6.1</td>
<td>5.3</td>
<td>3.7</td>
<td>9.0</td>
<td>4.8</td>
</tr>
<tr>
<td>8.</td>
<td>TRC AA8011+50% RTR+Annealed (580°C – 4 h) [29]</td>
<td>10.1</td>
<td>2.7</td>
<td>9.3</td>
<td>5.0</td>
<td>7.7</td>
</tr>
</tbody>
</table>
Al_{12}(Fe, Mn)_{3}Si with a size range of ~6-8 μm [26], as shown in Fig. 6(b).

Higher deformation (ECAP) can result in a higher SE in the material, producing more fine grains during annealing. Sidor et al. [32] deformed AA5XXX up to ~98.84% (thickness reduction). In TRC, refinement of grains up to 14 μm was observed after the annealing treatment at 550ºC for 12 s (Fig. 6(c)). Due to the higher SE, recrystallization occurred at several sites rather than growth of the grains (i.e., nucleation competed with the growth of the grains) during annealing, resulting in the formation of finer grains [34].

In comparison to the DC cast microstructure, the annealing treatment of TRC AA8011 (72% RTR) at 375 ºC for 50 min displayed the evolution of strain-free, fine grains with a size of approximately 22±10 μm as shown in Fig. 6(d). However, previous studies reported that CLS remained unaffected at 375 ºC and that the intermetallic size was in the range of approximately 0.63 - 3.63 μm (away from the CLS) [27].

One study by Kumar et al. [29] focused on the dissolution of the second phase particles during the homogenization of TRC AA8011. They concluded that the particle size of centerline segregation was reduced to approximately 2 μm at 580 ºC, but there was no complete removal of centerline segregation.

ECAP is an effective method to refine grain sizes. 4-ECAP processed AA3003, annealed at 400 ºC for 8 hrs, showed a fine grain size of approximately 0.5±0.2 μm, which is considerably lower compared to other processed Al alloys [18]. Also, TEM confirmed that the ECAP resulted in the evolution of subgrains of 500 nm, which recovered during annealing at 400 ºC.

Moreover, TEM confirmed the presence of Al(Mn,Fe)Si intermetallics in TRC AA3003 that evolved during pre-annealing at 450 ºC [18]. After reviewing the microstructural development during annealing, it can be concluded that the evolution of GS during annealing depends on the amount of deformation and annealing temperature. A higher amount of deformation enhances the SE and results in the evolution of fine grains. Moreover, high-temperature annealing reduces the recrystallization texture, mainly...
consisting of Goss and Cube orientations, evolves at the expense of deformation texture.

Figs. 7(a-c) show constant sections of ODFs at $\varphi_2 = 0^\circ$, 45° and 65° after annealing at different temperatures and processing conditions. These conditions include 450 °C for 16 h (after DC-AA3003+HR+RTR) [26], 375 °C for 240 min (after TRC+72% RTR) [27] and 580 °C for 8 h (after TRC+50% RTR) [29]. Despite the different processing conditions, in most cases, the Cube orientation was enhanced during annealing, as shown in Figs. 7(a-c) and Table 1. However, the Cube orientation was found to deviate along the ND by 15-20° due to the instability of the non-octahedral slip system ({$1\bar{1}0$}$\angle011$) during RTR. As a result, the ND-Rotated Cube orientation evolved adjacent to S orientation from the deformed Cube bands [23,35]. Goss-oriented grains also evolved during recrystallization from the Brass orientation [36]. The % volume fraction of Cube orientation was found to differ for different processing conditions, which can be attributed to different strain accumulation associated with the processing conditions, GS, and intermetallic particle size. The control of texture is essential to achieve good deformability/drawability.

3. COPPER ALLOYS

Electrolytic tough pitch (ETP) and oxygen-free high conductivity are two common grades of pure Cu [37,38]. ETP Cu is electrolytically refined to a purity of 99.96%-99.99%, with the intentional addition of oxygen (100-600 ppm), and is known for its high electrical conductivity (96%-101% IACS) [39,40]. The intentional addition of oxygen forms metal oxides (Cu$_2$O), which react with other impurities to clean the Cu matrix and enhance its electrical conductivity. ETP Cu possesses excellent electrical conductivity and good formability [41] but has low tensile strength [42]. Conversely, the addition of solute elements in Cu enhances mechanical strength and electrical conductivity. Copper ferroalloys (Cu-Fe) have garnered significant interest due to their excellent magnetic and electromagnetic shielding properties [43]. Metastable d'-(Cu,Ni)$_2$Si precipitates were formed in the early stages of aging in the Cu-Ni-Si alloy, dictating the overall properties of the alloys [44]. Rolling deformation is a common technique for creating Cu alloy sheets, which serve as raw material for the further processing of designated products. Both RTR and cryogenic-temperature rolling (CTR) have attracted interest as emerging SPD methods for obtaining high work-hardening and ultra-fine grain microstructures [45,46]. RTR and CTR of Cu and Cu alloys can result in various microstructures, such as grain elongation, formation of DTs, SBs formation, and kink microstructure [47–49]. RTR-processed Cu-Zn alloy samples (75% RR) have shown an increase in dislocation density, mechanical twinning, and micro-strain [50]. In Cu alloys, the deformation sequence involves the formation of equiaxed cells of dislocations, microbands, clustering of microbands, and either SBs or strain localization formation [51].

3.1. Deformed microstructure and texture evolution

The deformation microstructures reported for Cu alloys are quite different from those in Al alloys, as, in this case, deformation only starts after forming the cast billet structure. In this section, the authors attempt to discuss the temperature effect on microstructure evolution. Deformation at cryogenic temperature (CT) restricts dynamic recovery and grain growth during deformation, maintaining higher dislocation density compared to RT-deformed (RTR) samples [52]. Differences were observed when comparing the evolved microstructure in RTR and CTR pure Cu [53]. Variations in GS, kernel-average misorientation and work-hardening were observed for RTR and CTR processed samples [54,55]. As shown in Fig. 8, kernel-average misorientation values were higher for CTR (3.33°) than RTR (2.91°) up to a 40% reduction ratio (RR), indicating a higher work-hardening rate for the CTR samples. Further deformation leads to a decrease in kernel-average misorientation values, which is known as work softening [53]. Work softening infers self-annealing in the deformed Cu samples when exposed to RT atmosphere [56]. Variations in work-hardening rate create differences in mechanical properties, and CTR-processed samples exhibit higher mechanical strength than RTR-processed samples. Significant differences were observed in the micro-hardness and electrical conductivity values in RTR and CTR-processed Cu-Mg alloys [57]. A higher rate of increase in micro-hardness values was observed for CTR samples.
compared to RTR samples. A 90% CTR sample showed hardness values of \(\sim 240\) Hv, whereas a 90% RTR sample showed \(\sim 180\) Hv.

For Cu alloys, CTR is more effective for improving mechanical strength than RTR. Based on the microstructural results obtained in the literature [57], several strengthening mechanisms have been suggested to be involved in Cu alloys i.e., grain boundary strengthening (\(\sigma_{\text{GB}}\)), twin boundary strengthening (\(\sigma_{\text{TB}}\)) and dislocation strengthening (\(\sigma_{\text{d}}\)). RTR-processed Cu-Mg samples showed higher electrical conductivity than CTR samples. The electrical conductivity (irrespective of deformation temperature) decreased with an increase in RR's because of increased GBs, twin boundaries, and dislocation density [57]. Compared with samples processed at RTR, CTR samples exhibit a larger number of grain/twin boundaries and high dislocation density, which were responsible for low electrical conductivity [57]. Solute addition in Cu alloys was able to increase mechanical strength upon RTR; however, electrical conductivity decreased due to the lattice distortion caused by those solute atoms [58].

ECAP is another well-known deformation route to apply SPD on Cu. Cobos et al. [42] reported the evolution of a large fraction of low angle grain boundaries (70%, sub-grain structure with GBs of 3-15\(^\circ\)) after the first ECAP pass for pure Cu. Ultrafine grain microstructures with \(GS_{\text{avg}}\) of \(~0.46\) \(\mu m\) and \(~0.49\) \(\mu m\) were obtained after 8 and 16 passes, respectively [42]. The SFE also affects the deformation mechanism, as low SFE Cu alloys (RTR-processed) show a deformation sequence involving the formation of stacking faults, DTs, and SBs [48]. One of the features of ETP Cu is the presence of \(\text{Cu}_2\text{O}\) particles. Fig. 9 shows the presence of \(\text{Cu}_2\text{O}\) particles in both the as-received and RTR-processed ETP Cu. From Fig. 9(d,e), it can be observed that even after an 80% RR, the \(\text{Cu}_2\text{O}\) particles do not fragment, indicating their hard nature [53]. Most \(\text{Cu}_2\text{O}\) particles had a size greater than \(1\) \(\mu m\), which could assist in PSN during heat treatment [53]. Voids were observed in the vicinity of these particles due to the deformation incompatibility between the Cu matrix and strong \(\text{Cu}_2\text{O}\) particles [59].

Texture evolution in pure Cu and Cu alloys can be characterized based on the SFE, where medium to high SFE materials (pure Cu, Al-alloy) exhibit a Copper-type texture and low SFE materials (Cu alloys, ASSs, HEAs) show a
Brass-type texture [60,61]. In the large-strain rolling of initially ultrafine-grain Cu (GS$_{avg}$ = 0.32 μm) obtained by eight passes in ECAP, a Brass-type texture, rather than a Copper-type texture, was observed. The formation of DTs was the main cause of the evolution of the Brass-type texture or transition from Copper-type to Brass-type texture [62,63]. The volume fractions of the major texture components were: 11.2% Copper, 39% S, and 34% Brass, which is a typical proportion for a Brass-type texture [63]. Visco-plastic self-consistent (VPSC) simulation results revealed that the development of a Brass-type texture in pure Cu at very large strains likely resulted from the activation of $\{11-1\}$<112> slip, in addition to the usual $\{1-11\}$<110> slip. The <112> slip was only significant in the ultrafine-grain regime, and only a small part (about 10%) of the partial dislocations formed DTs [63]. The evolution of Copper, Brass, and S components has been reported in 80% of RTR Cu-3Ag-0.5Zr alloys [64]. Similarly, RTR Cu-Cr-Zr alloys showed Goss/Brass ($\{110\}$<115>), Brass, and S components in the 60% RR sample, the intensity of which increased with an 80% RR [65].

3.2. Heat-treated microstructure and texture evolution

High SE and microstructural defects in deformed Cu alloy samples accelerate the recrystallization kinetics even at lower annealing temperatures [66,67]. Fan et al. [67] reported that an annealing temperature of 300 °C for spin-deformed Cu-Sn alloy was regarded as the critical point of static recrystallization (SRX). At this temperature, static recovery occurred, whereas SRX was observed in the annealing treatment from 400-600 °C. SE is gradually reduced at the GBs, leading to the formation of small grains near GBs through the process of SRX [67]. The microstructure evolution during the recrystallization and grain growth of CTR Cu was investigated at a temperature of 450 °C [68]. Primary recrystallization caused the evolution of very fine grains ranging from 1.59-21.65 μm with many annealing twins. However, long time (100 hours) annealing at 450 °C caused grain growth with bimodal grain size distribution, in which smaller grains ranged from 3.35-13.20 μm, with larger grains from 14.96-35.79 μm [68]. Along with high temperature annealing phenomena, self-annealing phenomena have also been reported for RTR and CTR of pure Cu/Cu alloys [53,69,70]. Self-annealing refers to the occurrence of recrystallization at RT [52]. Lapeire et al. [45] observed the evolution of recrystallized grains in CTR-processed ETP Cu when the samples were removed from -17 °C for metallographic preparation. A similar type of observation was reported by Konkova et al. [71,72]. They also observed a time-dependent softening phenomena in CTR-processed pure Cu. High defect density, such as vacancies and other lattice
defects, can arise during deformation at CT, which decreases the thermal stability of pure metals and shows GB migration at RT [73]. Recrystallization phenomena are divided into two types: (1) discontinuous SRX (DSRX) and (2) continuous SRX (CSRX). Differences in the recrystallization phenomena can be understood based on the area of nucleation sites and recrystallization texture. Self-annealing was observed in 80% RTR Cu alloy in which GS_{avg} increased from 57 nm to 280 nm after 1 month of RT exposure [74]. During self-annealing, nanosized/submicron grains were recrystallized in the unique matrix of a single Brass-oriented deformed grain (after 14 months) [74].

Self-annealing phenomena can be better understood using microstructural explanations. Differences between the deformed and partially recrystallized (self-annealed) grains in a CTR80 sample occurred based on the grain orientation spread (GOS) criteria [45,71]. The GOS criterion (2º) was used to differentiate the deformed and self-annealed grains. Grains with GOS>2º are deformed grains and grains with GOS>2º refer to self-annealed grains. The ODF of the deformed microstructure showed a plane strain texture (Fig. 10(a)). Plane strain texture refers to the presence of Copper, Brass, and S components which evolved during RTR deformation in Cu alloys. In the magnified region of the IPF, only deformed grains are shown, while the black partitioned regions indicate the locations of self-annealed (SRX) grains, which are marked with an arrow and discussed in Fig. 10(b). A high orientation gradient of 12º was also observed inside the deformed grains (along line 1 in the magnified region in Fig. 10(a)), which could cause the further formation of new grains inside the deformed grains during the process of self-annealing. Interesting results were observed in the case of the self-annealed microstructure. Fig. 10(b) shows the IPF map of SRX grains, predominantly elongated with an GS_{avg} of less than 1 μm. A GOS of > 2º was used to differentiate the SRX grains from the deformed microstructure. The ODF map in Fig. 10(b) illustrates the evolution of plane-strain texture in the SRX grains. The magnified region of the IPF map shows that some of the SRX grains were elongated, while some were equiaxed in shape. Gerber et al. [75] studied the microstructure and texture evolution in heat-treated pure Cu after 70% and 90% reduction in thickness (RTR). Annealing twins were observed, which affected the growth of the Cube nuclei after 70% rolling and heat treatment (300 ºC). In contrast, the growth of Cube grains was so fast after 90% strain that nothing else was detected in the electron backscatter diffraction (EBSD) measurement of the recrystallized sample (150 ºC) [75].

Recrystallization texture components, such as Cube and Goss, have been reported for Cu and Cu-alloys [45,76]. As previously discussed, variations in the SE or RRs can also cause differences in texture evolution. In deformed conditions (70% and 90% RR), a β-fiber texture with strong S and Brass components and a weaker Copper component was observed. However, after 70% rolling and heat treatment (300 ºC), both strong Cube and retained rolling texture components were observed in the ODF map, whereas only the Cube component was visible in the ODF map after 90% rolling and heat treatment (150 ºC) [75]. The effect of SE was also observed; the low RR (70% RR) sample was completely recrystallized in the temperature range of 200-300 ºC, whereas the severely deformed (90% RR) sample showed an occurrence of recrystallization at 120-150 ºC [75]. Of course, severe deformation induces a large SE inside the deformed material, which enhances the recrystallization kinetics.
Chen et al. [77] reported the evolution of deformation and recrystallization texture in 99% RTR Cu-Ni alloys. The RTR Cu-Ni alloy showed the evolution of a very strong S component, with medium Copper, Brass, and weak Goss components. However, the SRX phenomenon at 700-800 °C led to the evolution of a strong Cube texture, whose intensity increased with an increase in the annealing temperature. Copper, Brass and Goss orientations disappeared during the annealing treatment [77].

The evolution of PSN microstructure has also been reported for pure Cu and Cu-alloys [53,78]. Fig. 11 shows SEM micrographs of the Cu₂O precipitate particles (Fig. 11(b)) and their orientation map (Fig. 11(c)). Self-annealed grains were observed not only at the deformed GBs but also around the Cu₂O particles [79]. This exceptional behavior in ETP Cu is known as the PSN [53]. As discussed in the Al alloys section, PSN has been reported only for particles larger than 1 μm [53,78]. A nanostructured Cu matrix was achieved due to the occurrence of discontinuous dynamic recrystallization (DDRX) and PSN mechanisms after 96% RR asymmetric CTR [78]. The authors would like to inform readers that Figs. 10 and 11 are the results of experimental work (ETP Cu) performed for the present review, with detailed discussions about the sample preparation and results given elsewhere [53]. Fig. 11(a) show that a large number of Cu₂O particles observed in the CTR80 sample were very rigid, and the shapes of these particles were not deformed even after 80% RR. During the deformation of ETP Cu containing second-phase Cu₂O particles, dislocations bow around the Cu₂O particles. If the strength of the particle is less than the force exerted by dislocations, the particle will deform; otherwise, the dislocation acquires a semicircular configuration [80]. The dislocations then encircle the Cu₂O particles, leaving an Orowan loop. The generation of extra dislocations, known as geometrically necessary dislocations (GNDs), in the form of the Orowan loop at Cu₂O particles, enhances the SE around the Cu₂O particles [81].

Figs. 11(b,c) show an SEM micrograph and its corresponding EBSD scan, respectively. Self-annealed grains can be seen in Fig. 11(d); many grains nucleated around the Cu₂O particles (marked with a rectangular box). A magnified region near the Cu₂O particle in Fig. 11(c) displays the nucleation of small grains around the Cu₂O particles, as seen in Fig. 11(e). PSN (grains 1-10) was confirmed with the help of the GOS ≤ 2° criteria, as shown in Fig. 11(f). The orientation relationship between the PSN grains nucleated around the Cu₂O particles was also analyzed. As seen in the (100)-pole figure, grains 1-5 show different orientations to each other, which indicates the DSRX mechanism (Fig.
11(g)) [82]. Similarly, grains 6-10 show dissimilar orientations, indicating that the nucleations around the Cu$_2$O particles were always DSRX in nature (Fig. 11(h)). Generally, PSN causes recrystallization in a discontinuous manner, as also reported for the Al alloys [82].

4. AUSTENITIC STAINLESS STEELS (ASSs)

ASSs are some of the most widely used FCC materials due to their excellent properties, including non-magnetic behavior, excellent corrosion resistance, and good formability and weldability [83,84]. ASS of both the grade 200 and 300 series are metastable at RT, and during deformation, transform into strain induced martensite (SIM) [6,85]. The 200 series ASSs have a lower SFE compared to the 300 series ASSs, and therefore, differences in the deformation behavior can be observed between these two grades of alloys. During deformation, two types of SIM can be formed in ASS: ε-martensite (HCP) and α’-martensite (BCC) [86]. It has been reported that ε-martensite usually forms at lower strains, reaches its maximum volumetric fraction, and then decreases with further increases in the RR [87,88]. For instance, ε-martensite was formed at low strains and reached its maximum volumetric fraction at ε = 0.11. Further deformation decreases its fraction, whereas the α’-martensite fraction increased with an increase in the RRs for 201 ASS [86]. As can be understood, the deformed microstructure of ASSs consists of new phases, i.e., α’ and ε martensite along with the deformation induced products. Formations of SIM during deformation can occur via one of the following routes:

(a) γ → ε, (b) γ → α’, and (c) γ → ε → α’, which occurs via the displacement of atomic planes [89,90].

In the case of heat treatment/annealing treatment, grain refinement occurs, and SIM transforms back to a fine reverted austenitic structure (α’ → γ). SIM exhibits a higher strength level compared to γ, enhancing the strengthening of the steel. Thus, the mechanical properties, especially the strength of the steel, are mainly controlled by the formation of α’, contributing to pronounced strain-hardening. As discussed for the previous two materials (Al and Cu alloys), where an improvement in mechanical properties was attributed to the TMP, was also the case for ASSs. During the TMP process, the amount of SIM formed during deformation, and refined γ grains formed after heat-treatment, controlled the overall properties of the ASSs. Therefore, considerable attention must be paid to estimating the evolution of SIM and refined γ grains during TMP.

4.1. Deformed microstructure and texture evolution

The formation of ε or α’ martensite principally hinges on factors like chemical composition, GS, SFE, and degree of deformation [91]. It has been documented that a decrease in SFE hinders dislocation slip, which is a primary deformation mechanism, thus creating a window for the occurrence of mechanical twinning or martensite formation upon straining. 201 ASSs exhibit a notably low SFE value (approximately 10 mJ/m$^2$), and it has also been reported that an SFE greater than 20 mJ/m$^2$ inhibits ε formation, while a lower SFE leads to the γ → ε → α’ transformation [86]. Similarly, during the deformation of 316L ASSs, SBs, SIM (ε and α’), DTs, and deformation bands were formed [90,92]. The typical microstructure evolution of cold-rolled 316LN ASSs is
presented in Fig. 12 via optical micrographs (subsets a-d) and TEM micrographs (subsets e-g) [85]. The initial microstructure consists of equiaxed γ-grains (Fig. 12(a)). As can be seen, upon a small RR of 10%, SBs (Fig. 12(b)) and DTs (Fig. 12(c)) were formed, and dislocations were accumulated in them. With an increase in the RR to 30%, SIM was formed, and the martensite boundary inhibited dislocation movement, causing dislocations to cluster around the martensite boundary. Upon a further increase in RR to 50%, the untransformed γ structure was elongated to the RD (Fig. 12(c)), and martensite laths were also formed (Fig. 12(f)). At 90% RR, a large block of γ and a fine martensite phase mixed with the γ structure (white circle in Fig. 12(d)), and dislocation cell-type martensite was observed (Fig. 12(g)) [85].

Zhang et al. [93] investigated the microstructure evolution of cold-rolled and tensile-deformed Cr-Mn-Ni-N metastable ASS samples at temperatures of 0 °C, -15 °C, and -30 °C. The formation of α'-martensite was observed during tensile testing, which was sensitive to temperature but insensitive to RRs. Lowering the deformation temperature effectively promoted the formation of ε-martensite, which becomes a transitional phase of the subsequent transformation to α'-martensite, leading directly to the enhancement in strength [93].

For a more detailed view, Fig. 13 displays a schematic diagram of the various microstructural features that evolved with increasing strains [94]. At low strain, dislocation slip predominates, and twinning is uncommon because the critical resolved shear stress for slip is lower than that for twinning [94]. This microstructure evolved as a result of dislocation glide/slip and typically forms in alloys with low SFEs. The dislocation cell-block structure is created in high SFE materials like Al, where dislocations exhibit a high three-dimensional mobility and can easily cross-slip. Low strain also reveals domain boundaries and microbands (Fig. 13(a)). Microbands and domain boundaries are geometrically necessary barriers created to account for changes in orientations between adjacent domains. Compared to low strain, medium strain results in higher dislocation density, finer dislocation border spacing, and a larger twin volume fraction (Fig. 13(b)). Consequently, these boundaries interact more frequently, and the twin-matrix (T-M) lamellae become an important component of the microstructure. The twinning planes of the T-M lamellae rotate closely to the rolling plane under high strain. With locally concentrated slip, deformation becomes heterogeneous, resulting in SBs that are roughly 30° inclined to the RD (Fig. 13(c)). The majority of the plastic strain is carried by these SBs, which also aid in martensite nucleation. It should be emphasized that the scale, alignment, and local strain of SBs are markedly different from those of microbands (Fig. 13(c)).

As discussed in the previous sections, a Brass-type texture was observed for low SFE materials. The SFEs of 304L ASS, high Mg twinning-induced plasticity steel, and 316L ASS were 18 mJ/m², 40 mJ/m², and 64 mJ/m², respectively. Chowdhury et al. [95] reported that increasing the RRs for 316L ASS increased the volume fraction of DTs and SBs. The material initially showed a Copper-type texture, but with the increase in RRs, the Copper component diminished, and the Brass and Goss components increased. The mechanism of texture evolution from Copper component to Brass component in medium SFE materials has been reported by Hirsch et al. [96].
The following routes were observed during the texture transition: Copper → Copper twin (\{552\}<115>) → Goss → Brass components. Chowdhury et al. [95] reported that the formation of DTs was correlated with the texture transition from Copper → Brass-type texture transition. A change in the rolling direction or strain path can lead to changes in the microstructure and texture evolution in ASSs. The effect of strain path on the texture evolution of the γ and α' phases was reported in [97]. Texture evolution for the γ and α phases during unidirectional rolling (RTR) and the multistep cross rolling of 316L ASS alloy samples was discussed in [97]. Unidirectional rolling refers to samples being placed in the same direction at every rolling pass, whereas multistep cross rolling indicates that the samples were rotated by 90° at every rolling pass. Unidirectional rolling resulted in Brass, Goss, and γ-fiber texture, while multistep cross rolling mainly formed Brass texture for deformed austenite after 90% RR. The Copper component was not observed in the unidirectional and multistep cross-rolled samples. After 90% RR, \{112\}<110> and Rotated-Cube \{001\}<110> were the main texture components observed for the transformed martensite in 90% RR samples of unidirectional and multistep cross rolled samples [97]. On the other hand, during 60% RTR of 201 ASS, the austenite phase developed Goss, Brass, and S texture components, which did not change significantly upon further straining, and α' developed Rotated Cube, α- and γ-fibers [86]. Furthermore, the formation of DTs resulted in the development of the Brass component [86].

4.2. Heat-treated microstructure and texture evolution

As discussed in Section 4, plastic deformation induces SIM in metastable ASS, and reverting back to the γ phase (α' → γ) through heat treatment serves as a viable option for grain refinement [98]. The α' → γ reversion transformation upon heat-treatment has been explored by various authors [99,100], with the starting and ending temperatures of this reversal being dependent on the composition [99]. The reverted γ, forming through the TMP, inherits high densities of dislocations from the prior α'. Consequently, during reversion annealing, the dislocations reorganize, establishing finely micro-structured cells and subgrains in the reverted γ phase. Fig. 14 displays the image quality (IQ) maps of SA and RTR (15%, 30%, and 50% RR) samples after thermal aging at 900 °C for 6 h of 202 ASS alloy. Note: EBSD measurements were performed on the RD-TD plane.

![Fig. 14. EBSD micrographs (image quality maps) of solution annealed and RTR (15%, 30%, and 50%) samples after thermal ageing at 900 °C for 6 h of 202 ASS alloy. Note: EBSD measurements were performed on the RD-TD plane.](image-url)
with a very low volume fraction of SIM, resulting from the dissolution of α' or, put differently, the reversal of α' into γ [101]. In Fig. 14(d), at 50% RTR, grains fully recovered into reverted and refined γ, reducing the GS_{avg} from 100 m to less than 10 mm. It is noteworthy that an increase in deformation escalates the internal SE of the ASSs, thereby enhancing recrystallization kinetics and potentially facilitating a more rapid reversion. Diffusional α' → γ reversion is understood to primarily be induced by the nucleation and growth of fine grains at martensite lath boundaries, as observed in 16Cr-10Ni and 18Cr-9Ni metastable ASS samples [105].

In another study, Sun et al. [106] reported that continuous heating of 80% RTR 304 ASS at 700 °C, using heating rates of 2 °C/s, 20 °C/s, and 100 °C/s, did not significantly alter the texture components. For continuous heating to 700 °C at 2 °C/s, the γ phase texture was most pronounced in the Brass orientation, followed by the Goss and Copper components. When the heating rates were increased, the most prominent texture shifted to the Goss component, followed by Brass and Copper components. The authors concluded that the heating rate exerted no significant influence on the annealed γ phase texture [106]. Variations in texture components also became apparent in relation to the fluctuations in the GS_{avg} of ASS samples [107]. In one study, an as-received 201 ASS alloy sample underwent tensile deformation (ε=0.34) and subsequent annealing from 100 °C to 800 °C [108]. In the as-received state, the γ phase displayed a random texture. Upon tensile deformation (ε=0.34), the Goss and weak Copper texture components emerged in the γ phase, while in the SIM (ε=0.34), α- and γ-fiber, alongside the strong Rotated Cube texture component, were formed. Upon annealing at up to 590°C after deformation, the Goss component in the γ phase weakened, while new components, namely Brass, were formed; concurrently, the texture of the SIM weakened until it disappeared. Notably, the ODF of the reversed γ phase exhibited only the Brass and S components. Thus, it can be deduced that the new γ phase nucleated with different orientations. After the complete reversal of the γ phase, it exhibited a randomized texture [108]. The evolution of texture during annealing at temperatures ranging from 600 to 1000 °C in a 95% RTR 304L ASS was studied, with major components centered around the Goss and Copper components, as well as the BR component {236}<385>.

Entirely new orientations after the recrystallization of the γ phase were observed, correlating with deformed texture components through twin relationships. However, a decrease in texture intensity was observed concurrent with an increase in annealing temperature [109].

The evolution of texture can be linked to deformation texture through twin relationships. TMP resulted in refined and reverted γ grains, significantly enhancing mechanical properties in terms of both strength and ductility. Somani et al. [110] demonstrated that the strength of the RTR 301LN steel (and also 301 steel) was exceptionally high, reaching levels of 1600 to 1800 MPa at high RR, albeit with very low elongation. However, upon undergoing the reversion process, the strength diminished, while ductility sharply increased beyond an annealing temperature of 600 °C. Depending on the annealing temperature (700 to 900 °C), the yield strength ranged from approximately 600 to 1000 MPa, while the elongation varied from 27 to 52% [110].

5. HIGH ENTROPY ALLOYS (HEAs)

HEAs are characterized by the alloying of more than four elements, typically in equiatomic or near-equiaxial compositions. Despite their complex chemical compositions, HEAs crystallize in simple crystal structures (FCC, BCC, and HCP) and form non-ordered solid solutions. High mixing entropy is beneficial for increasing the stability of the solid solution against the formation of intermetallic compounds [111,112]. Just as in any other conventional alloy, the mechanical and chemical properties of HEAs are predominantly determined by the microstructure that evolves during TMP. The trade-off between strength and ductility in HEAs can be addressed by meticulously designing TMPs to modify the microstructural features to deliver the desired properties [113–116]. Consequently, a significant field of research has been dedicated to understanding microstructure and texture evolution in HEAs during various TMP [117]. A fine-grained microstructure can be achieved in HEAs by applying medium strain (50-70% RR) and annealing at relatively low homologous temperatures (~500-600 °C) or through a short-time annealing treatment [118]. This fine-grained microstructure suppresses DTs, resulting in higher yield strength but relatively low elongations [119]. SPD leads
to ultrafine-grained HEAs, which, when exposed to temperature ranges of 500-700°C, result in the formation of intermetallic compounds [120,121]. The degree of deformation prior to annealing (in the temperature range of 700-1000°C) impacts the recrystallization kinetics and its fraction. Higher strain imposed before annealing provides more nucleation sites and a larger recrystallized fraction at a given annealing temperature [122]. Moreover, a fully recrystallized microstructure showed a weak texture, which was attributed to annealing twinning [111,123–126], which does not depend on the initial GS [127]. Additionally, microstructures with varied proportions of special grain boundaries, such as coincident site lattice boundaries, can be obtained through different TMP treatments, which is often referred to as grain boundary engineering (GBE) [116,128,129]. The aforementioned descriptions about the microstructural evolution in HEAs during deformation and heat-treatment are discussed in the subsequent sections.

5.1. Deformed microstructure and texture evolution

Various microstructural features, including dislocation slip (⟨111⟩⟨110⟩), DTs (⟨111⟩⟨112⟩), and micro-SBs, appear as deformation progresses. The deformation mechanisms can be controlled by managing the GS_{avg} [118] and SFE [130]. Grain refinement suppresses the formation of DTs, as illustrated by the complete absence of DTs in an ultrafine-grained CoCrFeMnNi alloy (GS_{avg}=503 nm) due to the very high critical resolved shear stress required for twinning [119]. Shahmir et al. [131] produced HEAs with GS ranging from 0.05 µm to 70 µm and reported a critical GS of 2 µm below which DTs were not formed. Otto et al. [15] discussed the effect of deformation temperature and grain size on the tensile properties of the CoCrFeMnNi alloys. The equiatomic CoCrFeMnNi microstructure with coarse grains (GS_{avg}=157 µm) resulted in a lower yield strength and higher elongation compared to the fine-grained (GS_{avg}=4.4 µm) microstructure at various temperatures ranging from -196°C to 600°C. DTs were observed only at -196°C after a strain of 20%, whereas dislocation cell structures were developed at RT and high temperatures during tensile loading, regardless of the GS [15]. In a similar type of study, tensile and compression experiments on a single-crystal CoCrFeMnNi alloy also supported the fact that DTs only appear at -196°C and not at RT deformation [132,133]. In contrast, formation of DTs in
the same alloy with GS\textsubscript{avg} = 17 µm were reported prior to necking during a tensile test at RT [134]. DTs were found to be responsible for the increased strain-hardening of the coarse-grained (GS\textsubscript{avg} = 590.2±9.3 µm) non-equiaxial Fe\textsubscript{41}Mn\textsubscript{25}Ni\textsubscript{24}Co\textsubscript{8}Cr\textsubscript{2} alloy, whereas those features were not observed during deformation of the fine-grained microstructure sample (GS\textsubscript{avg} = 8.1±0.52 µm) [135]. Furthermore, Co-rich HEAs (Co\textsubscript{35}Cr\textsubscript{20}Mn\textsubscript{15}Ni\textsubscript{15}Fe\textsubscript{15}) exhibited higher DTs as compared to the Cant or alloy on account of its lower SFE (~11 mJ/m\textsuperscript{2}), resulting in higher strain hardening capabilities of Co-rich HEAs compared to that of the Cantor alloy [136]. In contrast, a highly coarse-grained (GS in the range of 500-1000 µm) non-equiaxial HEA (Co11.3Cr20.4Fe22.6Mn21.8Ni23.9) also demonstrated good hardness (3 GPa) along with a good strain-hardening rate (~2600 MPa/ε) [137].

Moreover, DTs appeared in the RTR CoCrFeMnNi alloy (initial GS\textsubscript{avg} = 81±39 µm) after a 20% RR (RR20), and the twin density increased with subsequent RRs of 40% (RR40) and 60% (RR60) [138], as shown by electron channeling contrast imaging micrographs in Fig. 15. When comparing the crystallographic texture with the DTs evolution, the Copper component was more favorable for the formation of DTs than other texture components. Severe deformation (RR60) can also cause the formation of micro-SBs and the breakdown of the dislocation cell structure via DTs, a common phenomenon in low-SFE materials at higher deformation strains [124,138,139].
The texture evolution of a CoCrFeMnNi alloy (with GS_{avg} of 35 µm) during cryogenic tensile loading was characterized by a predominant \{111\}<112> texture component, accompanied by a minor Rotated-Cube (\{001\}<110>) texture component. Additionally, the \{115\}<552> texture component evolved due to twinning of the \{111\}<112> texture component [140]. As shown in Fig. 16, the typical texture transition from a Copper-type texture to a Brass-type texture during the RTR of HEAs is consistent with that of other FCC materials with low SFE [111,138,141]. A similar transition in crystallographic texture was discussed for the rolling deformation of ASSs in previous sections [95]. The Brass-type texture is characterized by strong α-fiber components, such as Brass (\{110\}<112>) and Goss (\{110\}<001>) texture components [138]. The texture that evolved after 90% RR through cryo-rolling (CTR) of CoCrFeMnNi was not significantly different than the RTR texture, as both exhibited a typical Brass-type texture [124]. Sathiaraj et al. [127] reported that the initial GS had no significant effect on the texture evolved after 90-95% RTR. When subjected to shear loading during high-pressure torsion experiments, CoCrFeMnNi alloys (with GS_{avg} of 500 µm) exhibited a texture similar to that of other FCC materials, with major texture components such as \{111\}<112> and Copper orientations [125].

5.2. Heat-treated microstructure and texture evolution

The annealed microstructure of HEAs often exhibits numerous annealing twins, which are introduced due to multiple generations of annealing twinning. In HEAs a low-temperature annealing treatment causes the occurrence of static recovery, along with the formation of intermetallics/precipitations. Zheng et al. [142] conducted annealing experiments at various temperature conditions on cold-rolled CoCrFeMnNi with different reductions and reported the recovery (<600 °C), recrystallization (600-830 °C) and grain growth (>830 °C) regimes. Another similar research reported the recovery temperature to be around 650 °C and that grain growth occurred at a temperature above 800°C through grain boundary migration in the Cantor alloy [143]. Furthermore, Chen et al. [144] reported a two-stage recrystallization process in the Cantor alloy at 700 °C wherein the first stage (annealing time<5 min) was characterized by a transition from a cold-rolled microstructure to a mixture of eutectic-like microstructure and a new, non-recrystallized microstructure with a changed texture. The second stage was marked by complete recrystallization [144].

The generation of multiple twinning has also been observed in the Cantor alloy after 80% RTR followed by annealing at 700 °C for 1 h [123,145]. A quasi in-situ annealing experiment performed on the Cantor alloy after RTR revealed that recrystallized grains preferentially nucleated at SBs and grew by subgrain coarsening mechanisms [146]. Similarly, a reduction in area of 80% through rotary swaging produced SBs, that acted as nucleation sites during subsequent annealing treatment [127]. The fraction of twin boundaries was found to depend on the final annealed GS during the grain growth regime (annealing temperature: 800 and 1000 °C for various times) for 90% RTR CoCrFeMnNi [147]. In another study, the evolution of annealed microstructures and crystallographic texture were discussed when the Cantor alloy was subjected to 80% RR followed by annealing treatment for various time periods, such as 700 °C for 6 min (Fig. 17 (a,f)), 700 °C for 7.5 min (Fig. 17 (b,g)), 700 °C for 10 min (Fig. 17 (c,h)), 700 °C for 15 min (Fig. 17 (d,i)), and 700 °C for 1 h (Fig. 17 (e,j)) [145]. The annealing treatment led to the evolution of the Copper component and devolution of α-fiber texture for SRX grains. In the partially annealed samples, non-recrystallized/retained deformed grains belonged to Brass, Brass/Goss, and Goss orientations [145]. The heterogeneous microstructure underwent hierarchical nano-twinning and showed higher evolution rates of GNDs. Moon et al. [116] demonstrated an interesting processing route in which the DTs fraction was enhanced by pre-stretching the Cantor alloy at CT (-196 °C) and then exposing it to a low annealing temperature of 500°C, which only recovered dislocations while DTs were retained.

The effect of TMP on the microstructural and mechanical properties of carbon added Cantor alloy was studied by Stepanov et al. [115].The microstructure showed improved strain hardening due to the addition of carbon (0.2 wt.%) to the Cantor alloy [115]. The twin boundary fraction in the annealed microstructure was lower than that of undoped Cantor alloy [115]. The increase in SFE due to carbon addition was found to be responsible for this behavior. On
the other hand, the addition of Si (0.2 molar ratio) slowed down the recrystallization process by hindering GBs migration [148]. Consequently, an increase in Si content led to a more heterogeneous microstructure with a mix of recrystallized and non-recrystallized grains. Annealing treatment after plastic deformation promotes the formation of intermetallic phases in HEAs. For example, prolonged heating of a nano-crystalline Cantor alloy at 450°C resulted in the decomposition of a single FCC phase into NiMn, FeCo, and Cr-rich phases, which increased the hardness [120]. In another study, a 95% RTR Cantor alloy followed by annealing at various temperatures showed the formation of the σ-phase in the microstructure, which further increased hardness compared to the deformed microstructure [121]. Texture evolution in the Cantor alloy during heat-treatment was similar to the annealing behavior of other low SFE materials. Crystallographic texture evolution in the 50% RR Cantor alloy followed by annealing treatment at various
temperatures is mentioned in [111]. It was observed that annealing treatment up to 600 °C, resembles the features of the deformed texture. However, a further increase in the annealing temperature up to 700 °C, decreased the texture intensity [111]. The multiple generations of annealing twins introduced many new orientations in the microstructure and caused texture weakening during the recrystallization of HEAs, which can be seen for 800 °C and 900 °C [123]. The net effect was significant texture weakening rather than complete randomization. The effect of strain paths, specially unidirectional rolling and multistep cross rolling, on the annealing texture of CoCrFeMnNi alloy was studied by Reddy et al. [149]. The dependence of annealing texture evolution on CoCrFeMnNi alloy processed through different strain paths showed that the [236]<385> texture component did not evolve during the annealing of cross-rolled samples [149]. Differences in crystallographic texture after annealing were attributed to variations in the annealed GSavg. The evolution of the microstructure (GSavg) during annealing treatment was dependent upon substructure destabilization and misorientation build-up, which changed with the varying strain path. Additionally, a higher fraction of nucleation sites (SBs, GBs, DTs) leads to easier nucleation of recrystallization. In such cases, nucleation will dominate over the growth process, leading to a finer recrystallized GS. The deformation structure and misorientation build-up were affected by cross-rolling, which diminished the density of potential nucleation sites and adversely affected the nucleation of recrystallization [149]. Annealing of 50% and 80% RTR Al0.25CoCrFeNi was marked by the evolution of the Rotated-Cube component [122]. The intensity of the Rotated-Cube decreased as the annealing temperature increased from 700 to 1000 °C [122]. In the case of the 90% RTR MnFeCoNiCu alloy, the Brass and Goss components were found to be quite stable when the annealing time was increased from 1 to 16 hrs at 900 °C [150]. This evolved texture was attributed to the sluggish diffusion effect and the absence of oriented nucleation and growth during recrystallization. Furthermore, various GBE treatments, often a combination of deformation and thermal treatments, can also be employed to increase the fraction of CSL boundaries in the microstructure for enhanced mechanical and corrosion properties [116,128,129].

6. DISCUSSIONS

6.1. Particle stimulated nucleation (PSN)

Work-hardening and softening are the two main phenomena that occur during deformation and annealing, respectively, for FCC metals/alloys. Work-hardening refers to the increase in the fraction of dislocation density during RTR deformation, enhancing the strength of the materials [151], whereas PSN, recovery, and recrystallization processes lead to softening during annealing treatment. During deformation, the first aspect to change is the shape of the grains. Equiaxed grains transform into a lamellar/banded structure, enhancing the GBs area, which is accompanied by an increase in dislocation density. Hence, the SE of deformation is the summation of newly formed interfaces and the accumulation of dislocations [34]. The enhancement of dislocation density can be measured using EBSD in terms of GND density. GNDs are the dislocations required to accommodate plastic deformation and maintain compatibilities between the GBs. GND can be influenced by both deformation temperature and deformation rate/strain rate [152]. Zheng et al. [152] studied microstructural evolution in hot-deformed (tensile testing) AA6082 alloy samples. It was observed that decreasing the deformation temperature and increasing the strain rate resulted in increased GND density, leading to higher SE after deformation [152].

This SE is potential for recrystallization during annealing. In Al alloys, intermetallic particles (> 1 μm) play a vital role in defining the microstructure and crystallographic texture through PSN during recrystallization [153]. Fig. 18 illustrates the PSN mechanism under the influence of coarse intermetallic particles. After casting, grains are in an equiaxed shape with coarse second phases, as shown in Fig. 18. Further plastic deformation causes the breakdown of intermetallics distributed along the RD and results in higher SE. The hard particles induce particle deformation zones around them [152]. These particle deformation zones consist of higher dislocation density (also referred to as GNDs) and a large misorientation gradient, indicated in Fig. 18 by black dotted lines. Moreover, the size of particle deformation zones is directly influenced by particle size [154]. The size of particle deformation zones is almost equal to particle size. Larger particle deformation zones will have a higher
misorientation gradient (near the particle-matrix interface) that will lead to more randomly oriented grains (PSN) during annealing [153]. This accumulated strain around the particles will be the supporting force for recrystallization during subsequent annealing. Initially, the dislocations arrange themselves as subgrains during annealing and then nucleate the randomly oriented grains adjacent to the interface of the particles and matrix, as shown in Fig. 18 [153]. The higher the accumulated strains around the particles, the more random grains will form [154,156]. Similar PSN behavior was also reported for ETP Cu, which occurred due to self-annealing as well as high-temperature heat treatment around Cu$_2$O particles [79,157]. In the case of 7Mo super-ASS, s precipitates promoted dynamic recrystallization through the PSN mechanism [158]. Other literature have also reported the occurrence of the PSN mechanism during the heat-treatment of ASSs [159,160].

6.2. Self-annealing and Softening phenomena

The self-annealing phenomenon depends not only on material properties such as purity, melting point, and composition but also on external factors such as strain, strain rate, SFE, and deformation temperature [161]. Under high strain and strain rate and low SFE and deformation temperature, self-annealing can occur at RT [161]. This phenomenon, observed as SRX and grain growth in Cu, Pb-Sn, Zn-Al, and Al-Cu alloys, occurs shortly after SPD processing [161,162]. In Al alloys, self-annealing can be correlated with the natural aging process, which involves the nucleation of very fine precipitates. During natural aging, the size and number density of the Guinier-Preston (GP) zones in Al alloys change with time. In contrast to softening at RT in Cu alloys, natural aging enhances hardening in Al alloys [163]. Softening behavior in Cu alloys (medium to low SFE materials) is divided into SRV and SRX. As it is well known, a large number of dislocations are generated during plastic deformation, mainly accumulating around the GBs and particle interfaces. The difference in the mechanisms of SRV and SRX can be observed in terms of dislocations arrangement. Of course, SRX occurs at a higher temperature than the SRV process. In the SRV phenomenon, the following processes of dislocation arrangement can be observed: (a) dislocation tangles, (b) cell formation, (c) annihilation of dislocations within cells, (d) subgrain formation, and (e) subgrain growth. Similar types of features regarding dislocation arrangement during the heat treatment of Cu-Cr and Cu-Cr-Mg alloys are explained in [164]. The microstructure evolution in Cu-Cr alloy samples after aging treatment at 480 °C showed a cellular substructure after 15 min of heat treatment, while the annihilation of dislocations and the formation of annealing twins were observed after 4 hrs of heat treatment [164]. When comparing the softening phenomena in Cu-Cr-Mg alloy samples with Cu-Cr, a difference in the dislocation arrangement was observed. A
large dislocation density could be observed in the short-time-aged sample. Dislocation tangles were observed in the 15 min heat-treated sample, whereas a cellular substructure was present even after 4 hrs of heat treatment [164]. The good softening resistance performance of Cu-Cr-Mg alloy was due to the pinning effect of dislocations by fine precipitates and Mg atoms [164]. The addition of Ca, Sr and Y was reported [165,166]. The pinning effect of dislocations by fine precipitates and Mg atoms [164]. The addition of Ca, Sr and Y was reported [165,166].

6.3. Phase transformation

Phase transformations in Al, Cu and HEAs are unheard of during deformation [26,27,45,138,167]. Rather, annealing or aging treatment at respective temperature ranges causes the formation of other phase precipitates and/or intermetallics in Al, Cu and HEAs [120,168–173]. In contrast to the aforementioned FCC alloys, ASSs show phase transformation during deformation and annealing treatment [89,90]. The TMP of ASS can be performed in two stages: (i) rolling deformation and (ii) subsequent annealing. As discussed in Section 4, in the first stage, the transformation can occur via one of these routes: (a) $\gamma \rightarrow \epsilon$, (b) $\gamma \rightarrow \alpha'$, and (c) $\gamma \rightarrow \epsilon \rightarrow \alpha'$, which increases the yield strength and microhardness of the ASS [85,174]. The study showed that the $\epsilon$-martensite formed at lower RR; however, by increasing the RR, the $\epsilon$-martensite was observed to be coexisting with $\alpha'$-martensite, and $\alpha'$-martensite could transform from the surrounding $\epsilon$-martensite [174]. Also, it was reported that RTR promotes the formation of $\alpha'$ for various grades of ASSs [89,92]. Furthermore, as discussed above, the transformation cycling (deformation) results in an increase in dislocation densities; it has been reported that high dislocation densities significantly decreases $\gamma$-phase stability [175], and by decreasing the stability of the $\gamma$-phase, the kinetics of the martensitic transformation could be enhanced [175,176].

Mohammadzehi et al. [176] investigated the effect of RTR and $\gamma$-phase stability on the mechanical properties of 316L ASS. As is known, 301 and 201L ASSs have lower $\gamma$-phase stability compared to 316L ASS. The lower the $\gamma$-phase stability, the higher the chances of martensitic formation, even at low RRs. $\gamma$-phase stability depends upon the alloying content in ASSs; lower alloying contents in 301 and 201L ASSs compared to 304 and 316 ASSs effectively mask the effects of other variables such as varied GS and small changes in the rolling temperature. In the second stage, a reversion of the $\alpha' \rightarrow \gamma$ phase takes place, resulting in improved ductility, grain refinement, and enhanced corrosion properties. Proper execution of TMP leads to a fine-grained $\gamma$ structure, with $GS_{avg}$ ranging from nano to submicron, imparting excellent room-temperature strength and ductility to the ASSs [177,178]. Plastically deformed metastable ASSs can undergo two types of reverse transformation during TMP: an athermal ‘shear-type’ transformation and an isothermal (thermally activated 'diffusional') reverse transformation [179]. Both transformation types lead to $\gamma$-phase formation, but the resulting microstructures are completely different. The thermally activated isothermal transformation is time-dependent and leads to equiaxed, finely grained $\gamma$-phase with low dislocation density [179]. In contrast, the athermal transformation is temperature-dependent only and leads to a lath-shaped $\gamma$-phase with high dislocation density [180]. The type of reverse transformation experienced by an ASSs depends on its chemical composition and the heating rate applied during the heat treatment [181,182]. Higher heating rates and temperatures enable athermal transformation, whereas low $\gamma$ stability enables isothermal transformation [183].

The reversion mechanism of $\alpha' \rightarrow \gamma$ during thermal aging at 700 °C at different heating rates in the experimental steel is given in [106]. The study concluded that the diffusional reversion process dominated the transformation of $\alpha' \rightarrow \gamma$ at low heating rates (2°C/s), resulting in the development of equiaxed, defect-free, nano/ultrafine-grained austenite grains. While at a rapid heating rate (100°C/s in this work), the regulated process was a martensitic shear-type reverse transformation of $\alpha' \rightarrow \gamma$ that produces a lath-type reversed $\gamma$ structure with a high density dislocation [106]. Further, Mao et al. [175] reported that the two step rolling and annealing process was beneficial in achieving ultra-fine grained ASSs which exhibited a great combination of strength and ductility. The study reported that in a metastable Fe-24Ni-0.3C alloy sample, the first step rolling and annealing process (50% RTR + 600 °C annealing for 30 s) resulted in a sample consisting of a partially recrystallized $\gamma$-phase with $GS_{avg} \sim 1.5 \mu m$, and upon the second step rolling and annealing process (90% RTR + 600 °C annealing for 30 s), the sample consisted of a fully recrystallized $\gamma$-phase with $GS_{avg} \sim 0.5 \mu m$
Fine-grained ASS possesses better mechanical properties than coarse-grained ASS [107].

6.4. Grain boundary engineering

As discussed in Section 5, the concept of GBE is to control the fraction of special boundaries in the microstructure, such as CSL boundaries in low SFE FCC materials. Twin boundaries (<111>60°) in FCC materials belong to Σ3 CSL boundaries. The fact that FCC alloys are compliant with GBE imparts distinct mechanical and corrosion behaviors to the material due to the low GB energy of these special boundaries [116,128,129,184,185]. Chen et al. [128] reported that a CSL boundary fraction of over 50% was achieved during a one-step recrystallization of CoCrFeMnNi. It has been reported that the fraction of annealing twin boundaries decreases with increasing amounts of alloying elements. This is because the alloying effect and severe lattice distortion can decrease both the average GB energy and twin boundary energy [128]. However, Thota et al. [129] argued that a proper GBE treatment should disrupt the network of random HAGBs and, therefore, strain-annealing (low strain followed by annealing) should be preferred as a GBE treatment over the one-step recrystallization employed by Chen et al. [128].

The effect of annealing temperature and time on CSL boundary fraction for non-deformed (as-received) and RTR Cantor alloy samples was discussed in [129]. The authors reported an exceptional CSL boundary fraction (>70%) in HEAs obtained through 5% RR (RTR) followed by 1 h annealing at 950 °C when compared to other conditions, such as, as-received conditions, 10% RR and 15% RR (RTR) followed by 1 h annealing at 950 °C. Moon et al. [116] adopted a strategy to enhance the twin boundary fraction in HEAs by deformation at CT, followed by low-temperature annealing that was just sufficient for the recovery of the microstructure without affecting the twin boundaries. Moreover, Kaushik et al. [123] also reported a CSL fraction of more than 50% in Cantor alloy, which was RTR to 80% RR, followed by annealing at 700 °C for 1 hr. The microstructure consisted of several twin clusters with profuse Σ3 and Σ9 boundaries. Higher RTR deformation and annealing temperatures increased the twin boundary fraction in the HEAs [186]. For the case of hot deformation of ASSs, it was concluded that the generation of strain-free grains by dynamic recrystallization and the occurrence of CSL boundaries (Σ3 and Σ9) led to a random texture [187]. In another study on RTR ASSs [94], DTs occurred on planes with the highest twinning Schmid factors and showed a strong orientation dependence. In the case of RTR 316L ASS, DTs occurred preferentially in grains with near Copper orientation rather than the Brass orientation [94].

7. SUMMARY AND RECOMMENDATIONS

In the present review, TMP and its impact on microstructure and texture evolution have been thoroughly studied for Al and Cu alloys, ASSs, and HEAs. Various features of the deformed microstructure in FCC materials, such as grain-fragmentation, precipitates/particles distribution, SBs/DTs formation, and martensitic transformation, are discussed for the aforementioned alloys. Similarly, texture evolution during deformation and annealing treatment and the effect of second-phase particles, strain rate and deformation temperature, annealing twin boundaries (CSL boundaries), and phase transformation on evolved crystallographic texture have been discussed for various grades of FCC metals/alloys. A concise summary has been drawn for each alloy case as follows:

1. DC-cast and TRC Al alloys exhibit second-phase particles in the form of Chinese script and CLS, respectively. RT deformation fragments coarse intermetallic particles of Chinese script/CLS into smaller particles. Low-temperature (<450 °C) annealing evolves a greater fraction of second-phase particles, whereas at high-temperature annealing (> 500 °C), the size of intermetallics is reduced. PSN have been reported only for the coarse intermetallic particles (size > 1 μm) during annealing treatment. However, detrimental effects of PSN have been observed, as it reduces the likelihood of forming preferred orientations (especially the Cube {100}<001>) after heat treatment.

2. Grain fragmentation and the formation of SBs/ strain localizations and DTs have been reported during the plastic deformation of Cu alloys. Severe deformation enhances the SE of Cu alloys, which causes self-annealing or RT recrystallization phenomena. Softening occurs when the specimen is exposed to the RT atmosphere. The formation of
SRX grains has been observed at the deformed GBs and triple junction boundaries, which is due to the DSRX mechanism, whereas the formation of grains inside the deformed/parent grains is due to the CSRX mechanism. PSN has also been reported in ETP Cu, which causes the nucleation of grains around the hard Cu$_2$O particles via the DSRX mechanism. Pure Cu exhibits a Copper-type texture, whereas low SFE Cu alloys exhibit the Brass-type texture after RTR deformation.

3. In ASSs, the nucleation characteristics of SIM (phase transformation) during TMP were observed to occur via (a) $\gamma \rightarrow \varepsilon$, (b) $\gamma \rightarrow \alpha'$, or (c) $\gamma \rightarrow \varepsilon \rightarrow \alpha'$. During subsequent heat-treatment, recovery occurs, through which SIM reverts back to refined $\gamma$ grains. However, the process of recrystallization can only occur in the reverted $\gamma$ regions when $\alpha' \rightarrow \gamma$ transformation is complete. Upon deformation, the texture developed in 304L is a Copper-type texture, while 316L develops only the Brass-type texture. The texture transition during severe deformation (from Copper-type to Brass-type) in medium SFE materials follows the route: Copper component $\rightarrow$ Copper twin ($\{552\}<115>$) $\rightarrow$ GossBrass.

4. In single FCC phase HEAs, the evolution of DTs depends on GS, deformation temperature, and the state of stress. Larger GS, low deformation temperature, and deformation by rolling greatly promote twinnability during deformation behavior. In contrast, smaller GS, temperatures higher than RT, and low strain tensile/compression loading impede deformation twinning. The deformation texture exhibits a transition from the Copper-type texture to the Brass-type texture with an increase in RTR reduction. The SBs developed during severe deformation act as preferred nucleation sites for recrystallized grains. Complete recrystallization results in the evolution of a weak texture due to the introduction of several new orientations as a result of multiple annealing twinning in HEAs.

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