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Revealing the relationship between microstructures, textures, and mechanical behaviors of cold-rolled Al$_{0.1}$CoCrFeNi high-entropy alloys

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Abstract

Here we reveal and discuss the relationship among microstructures, textures, and mechanical behaviors of Al$_{0.1}$CoCrFeNi high-entropy alloys (HEAs) after cold rolling and annealing. The initially coarsen grains display profuse lamellar-structured slip bands upon cold rolling to 50% reduction with mostly extending vertically to the rolling direction. Meanwhile, cold rolling facilitates the evolutions of Goss ($\{011\} <100>$) and Brass ($\{110\} <112>$) component textures in the low stacking fault energy (SFE) HEAs accompanying with the formation of deformation twins. Interestingly, the rolling strengthen HEAs exhibit novel anisotropies of yield strength and strain hardening associating more with the direction of the slip lines and twins rather than the rolling-induced preferred orientations, attributing to the easier dislocations glide in

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between the slip lines and twins than across them. The microstructural characters including dislocation density, slip lines, and twins are quantitatively evaluated, which demonstrates that the dynamic grain refinement contributes much more for the overall strength, compared to the increased dislocation density. A lower strength (370 MPa for yield and 733 MPa for failure) and exceptional ductility (~ 55%) are achieved in the annealed samples with random grain orientations.

**Keywords**: High-entropy alloys; Cold rolling; Microstructure; Texture; Mechanical properties; Deformation mechanisms.
1. Introduction

High-entropy alloys (HEAs) are designed with four or more equiatomic or near-equiatomic elements, which breaks through the traditional alloy-design concepts [1-4]. Surprisingly, these alloys always have a simple and disordered single solid-solution structure, including face-centered-cubic (fcc: e.g., CoCrFeMnNi [2]), body-centered-cubic (bcc: e.g., TaNbHfZrTi [5]), and hexagonal close-packed structures (hcp: e.g., GdHoDyTbY [6]) rather than brittle intermetallic compounds. The representative single-phase fcc HEAs, Al$_x$CoCrFeNi ($x = 0 \sim 0.3$) and CoCrFeMnNi alloy systems, which can be produced by traditional casting, possess outstanding mechanical properties, such as the high cryogenic tensile strength and exceptional fracture toughness [7, 8]. However, the fcc-structured HEAs are relatively weak in strength, only around 200 MPa in the as-cast and homogenized states at ambient temperature [9, 10], which is far below the requirements of practical structural applications.

Strengthening mechanisms of the fcc-based HEAs have been proposed with respect to the solid-solution hardening, grain-boundary hardening, dislocation hardening, and precipitation-hardening mechanisms as in other conventional alloys. Note that nearly all the strengthening mechanisms are predominantly originated from the typical severe plastic-deformation processing (SPD) and thermo-mechanical processing (TMP) treatments. The basically understanding of SPD and TMP of HEAs enables numerous production, manufacturing, design, repair, and recycling pathways. For this reason, many studies have been dedicated to investigating the microstructure
and texture evolutions and the corresponding mechanical properties of HEAs during processing [11-13]. Gratifyingly, both CoCrFeMnNi and Al$_{0.1}$CoCrFeNi HEAs possess a low or medium stacking fault energy (SFE) ranging from 18.3 to 30 mJ/m$^2$ [14, 15], similar to low SFE metals, such as the brass and high-manganese steels [13, 16]. Usually, deformation in HEAs with the low-to-moderate SFE proceeds primarily by the dislocation glide. Need to add that, at high tensile strains [8], during heavy cryo or room temperature rolling [17], and under high speed tensile or compression conditions [15, 18], an assistant mechanism named deformation twinning can be activated, leading to twinning-induced plasticity (TWIP). Additionally, the activation of different deformation mechanisms, i.e., dislocation slip, deformation twin, and shear band, is proved to strongly affect the development of the crystallographic texture during cold rolling.

So far, in the preliminary studies of fcc-structured HEAs, heavily fragmented microstructure and Brass dominated texture have been found in the quaternary CoCrFeNi and quinary CoCrFeMnNi and Al$_{0.25}$CoCrFeNi HEAs after cold rolling with a reduction of 90%, attributing to the formation of twins or shear bands [13, 19, 20]. What’s more, strong Goss-Brass deformation texture was observed in 90% cold-rolled fcc-based MnFeCoNiCu HEAs due to short-range ordering (SRO) assisted planar slip without twinning [21]. By contrast, the ternary CoFeNi HEAs developed the lamellar deformation microstructure and pure metal type texture after heavy cold rolling [19]. After annealing, the CoFeNi, CoCrFeNi, CoCrFeMnNi, and Al$_{0.25}$CoCrFeNi HEAs showed fully recrystallized microstructures with numerous
annealing twins at lower annealing temperatures, while the MnFeCoNiCu HEAs presented higher recrystallization temperature, strengthen cube component, and rare annealing twins [13, 19-21]. The preliminary works have focused on either the microstructure and texture evolutions by cold/cryo rolling or the mechanical properties without quantitatively correlating the microstructure and mechanical properties after cold rolling and annealing. Especially, a great challenge for applications is to unearth the response of textures on the mechanical anisotropies of cold-rolled HEAs, and the contributions of microstructural characters, such as slip lines, dislocations density, deformation twins, and shear bands, to the overall mechanical properties are unknown until now.

Consequently, the aim of this work is to reveal the relationship among microstructures, textures, and mechanical behaviors in Al\textsubscript{0.1}CoCrFeNi HEAs upon cold rolling and annealing. The different deformation mechanisms - especially mechanical twinning - are quantitatively visualized. The influences of deformation mechanisms are studied by the crystallographic texture measurement, orientation intensity, and volume fraction plots. The microstructure features at different cold rolling strains help to explain the anisotropic mechanical properties in present HEAs during rolling processes. For the cold-rolled HEAs, plastic instability can be predicted by the Hart’s rule, and the local strain is detected by digital-image-correction (DIC) method. Subsequently, the grain size can be controlled by rolling and recrystallization for an improvement of mechanical property.

2. Experimental
2.1 Alloys processing

Raw Al, Co, Cr, Fe, and Ni metals with a purity exceeding 99.9 weight percent (wt.%) were used to produce Al_{0.1}CoCrFeNi HEAs. The HEAs were fabricated by suction casting into a copper mold with a dimension of 100 \times 20 \times 3 \text{ mm}^3 under an argon atmosphere. In order to promote the chemical homogeneity, the ingots were flipped and remelted at least four times. The as-cast ingots were first homogenization-treated at 1100 °C for 5 h and then cold rolled (CR) to the thickness of \sim 1.5 \text{ mm (} \sim 50\% \text{ reduction}) and 0.9 \text{ mm (} \sim 70\% \text{ reduction}) along the longitudinal direction. The 70\% reduction cold-rolled samples were isochronally annealed at temperatures in the range between 800 °C and 1000 °C for 1 h to tune the mechanical behavior. All the heat treatments were conducted in an air furnace, followed by water quenching to room temperature.

2.2 Characterization

2.2.1 Microstructure

The microstructures of the cold-rolled HEAs were observed by Leica CTR 6000 optical microscope (OM). The cube-shaped samples with a dimension of 10 \times 5 \text{ mm} were cut from the different rolling reduction of Al_{0.1}CoCrFeNi HEAs. Hence, its three orthogonal planes were perpendicular to the rolling direction (RD), transverse direction (TD), and normal direction (ND) labeled as S1, S2, and S3, respectively. The three-dimensional (3D) topography of the deformed alloys was presented by OM images, including S1, S2, and S3 planes. Due to the limited resolution of OM, further observation on the rolling-plane (S3) sections were carried out by transmission
electron microscope (TEM) using a JEM2100F operating at 200 kV. TEM specimens (~ 50 μm in thick, and 3 mm in diameter) were prepared on the rolling-plane section using a double jet Tenupol-5 electrolytic polisher with a voltage of 25 V at room temperature and further thinned by ion-milling to a thickness of the electron transparency.

2.2.2 Bulk & micro texture measurements

Phase identification and bulk texture measurements were performed by X-ray diffraction (PANalytical, Model: X’Pert PRO), using the Co K$_{α1}$ radiation (1.788965 Å) operating at 35 kV and 30 mA between 45° to 105° at a step size of 0.03°. The bulk texture measurements were carried out on the carefully mechanical polished rolling-plane (S3) section due to a limited thickness of the deformed sheets. Three incomplete pole figures, {111}, {200}, and {220} were measured in the rolling surface layer of the HEAs under the back-reflection mode. JTEX-Software was used to calculate the complete orientation distribution function (ODF) and pole figures [22]. In addition, the microstructure and microtexture of the cold-rolled and annealed alloys were characterized by the electron backscatter diffraction (EBSD) system, using a high-resolution field emission JSM-7100F SEM. During the EBSD acquisition, the working voltage was operated at 20 kV and a sample tilt of 70°. For the cold-rolled samples, the step size was varied between 0.4 μm and 1.8 μm. The step size of 0.22 μm ~ 1.45 μm was used for the annealed samples.

2.2.3 Mechanical property

Rectangular dog-bone-shaped tensile specimens, with a gauge dimension of 10
mm × 2 mm, were artificially machined by electrical discharge along the RD and TD. The samples were strained on an Instron 5969 materials testing machine at room temperature and at a strain rate of $1 \times 10^{-3} \text{s}^{-1}$ along the RD and TD. More than five specimens for each alloy were tensile tested, and one representative engineering stress-strain curve was plotted for each alloy. Nanoindentation tests were performed on S1 and S2 planes using an Agilent Nano Indenter (G200) with a maximum indentation depth of 2000 nm at a strain rate of 0.2 s$^{-1}$ to measure the elastic modulus on S1 and S2 planes. The C11, C44, Poisson's ratio, bulk modulus, Young’s modulus, and shear modulus on S1, S2, and S3 planes were measured by the Ultrasonic Material Characterization System (UMS-100) [23]. To observe the evolution of the local strain that occurred in the cold-rolled samples during deformation, a DIC method was applied.

3. Results

3.1. Microstructure evolution during cold rolling

Homogenizing treatment eliminates the dendritic segregation of the as-cast structure and obtains single fcc phase coarse-grained Al$_{0.1}$CoCrFeNi HEAs with average grain size of 128 μm. The slip lines were straightforwardly observed in the three sections of the CR50% sample in Fig. 1a. The microstructure on the rolling plane (S3) shows remarkable lamellar slip bands extended mainly along TD with average slip band spacing of ~ 4.58 μm. Note that the slip is such that a straight slip line will be parallel to the trace of the primary slip plane, extended over the whole grains, and terminated at grain boundaries because of the released stress concentration.
All the grains developed multiple slips. Occasionally, slip along both secondary and tertiary (111) slip planes becomes active in some grains for the ‘favorable’ orientation deformation, as indicated as cross slip lines, which divides these grains into several portions. Thus, it can be known that the activated crystal slip is the dominant deformation mechanism during present thickness reduction, which causes the preferred orientations.

With increasing the cold rolling reduction to 70% (Fig. 1b), the density of slip lines is increased significantly with finer slip bands spacing of ~ 3.25 μm. The cross slip lines become remarkably wide, and the cross slip bands gradually evolve into shear bands [24]. The grains are further elongated along RD on S3 and S2 planes. Longitudinal microstructural deformation features are frequently observed. While the boundaries of the elongated subgrains appear to be wavy, dark lines along the rolling direction (a river-like pattern) on the etched surfaces of the transverse-sectioned samples are shown, indicated as the strip structure in Fig. 1b.

The EBSD crystal-orientation maps for the different CR HEAs are presented in Fig. 2. The band-contrast map of the CR50% sample is characterized by slip bands in Fig. 2a, and the crystal orientation map shows the lamellae-structured deformation twins with volume percent of 8 vol.%, as indicated in Fig. 2b. The increase of shear stress resulted from cold rolling favors the initiation of deformation twins accompanied by forming twin-matrix (T-M) lamellae. A magnified region of microstructure in the CR50% sample is lifted out from the rectangular box in Fig. 2a, i.e., micro slip lines, cross slip lines, and deformation twins are together shown in Fig.
2f. The orientation characteristics of the deformation twins and matrix are depicted through the detailed analysis of misorientation changes along the dotted line marked and demonstrated by the selected circle area of the (111) pole figure in Fig. 2f. The local orientations were measured from the EBSD orientation map, as shown in Fig. 2g, to demonstrate a large increase of the misorientation angle between the matrix and twin lamellas. The (111) pole figure indicates the orientation between the matrix and deformation twins (Fig. 2h).

Interfacial shear slipping is developed with increasing the CR reduction to 70%. A large number of micro-scale shear bands are observed, and the T-M lamellae structure becomes continuously eliminated in Figs. 2c and d, indicating that the locally-inhomogeneous plastic deformation occurs during heavy cold rolling [25]. Thin shear bands (marked by the arrow) from band-contrast maps, aligned at 40 - 70° to the rolling direction, are observed in the matrix, which is larger than the results of Sathiaraj et al. [26] and Bhattacharjee et al. [27]. The orientation gradients within each individual grain manifested as color gradients on the crystal orientation map, indicating the development of local misorientations arisen from the dislocation accumulation, is attributed to the greater density of geometrically-necessary dislocations (GNDs).

Deciphering for a more detailed deformation mechanism of the CR HEAs, TEM examinations have been conducted and the bright-field images are shown in Fig. 3. At the CR50% reduction, the high-density planar slip of dislocations on {111} type fcc planes is a dominant feature, as shown in Fig. 3a. Deformation twins and planar
dislocation paths (DPs) are clearly observed in the micrograph (Fig. 3b). The formation of dislocation paths with high-density dislocations has been proposed to be slip lines at the crystal surface in many fcc metals and alloys [28]. The deformation mainly localizes on the slip lines and dislocation paths at two distinct scales. Once a slip line is formed it fills up with dislocations due to the restricted ability to cross-slip. Thus, the dislocation paths can impede the dislocation motion by trapping and blocking mobile segments in those paths/walls. With increasing the rolling extent to 70%, dislocations cells (DCs) are accumulated at the grain boundaries (GBs) and can be found to be tangled with such twins, revealing that slipping and twinning, as well as interactions between them, are active during cold rolling (Figs. 3b and c). Shear bands (SBs) are found between the dislocations paths and twins near GBs, which further accumulates the plastic deformation and suppresses the expansion of dislocation paths. T-M lamellaes within the grains are formed with thicker bundles and containing a high dislocation density in the between of matrix, as shown in Fig. 3d. Comparing with an earlier report [29], it is evident that the deformation twins are relatively finer with twin thickness of 40 - 60 nm, spacing of 0.41 μm with each other. With straining, importantly, the critical resolved shear stress acting on a single dislocation might not be sufficiently high to overcome the threshold stress of dislocation paths and deformation twins, resulting in the blocked dislocations, as confirmed by Fig. 3e and f.

Gholinia et al. [25] consider that when the alloy is subjected to conventional simultaneous cold rolling, i.e., the alloy produces a flat compressive strain, the
microstructure is usually elongated, and the grain boundaries are extruded to form a strip-structure, as shown in Fig. 1b. With increasing the CR reductions, the frequency of low-angle grain boundary (LAGB) increases gradually, as evidenced in Fig. 4. Plenty of LAGBs are responsible for dislocation slips in the interior of grains. Especially, the frequency of $\Sigma 3$ twin boundaries is increased after CR70%. Meanwhile, there is a gradual conversion of LAGB to high-angle grain boundary (HAGB) with an increase for geometrically as well as statistically stored dislocations as dynamic recovery sets in [30].

3.2 Texture evolution during cold rolling

The texture evolutions in the CR50% and CR70% HEAs in terms of (111) pole figures are exhibited in Fig. 5. In the case of CR50% samples, the bulk texture could be characterized by the strong Brass (Bs {110} <112>) and Goss (G {011} <100>) locations in Fig. 5a and b, while the micro-texture is basically consistent with the bulk texture, as displayed in Fig. 5c and d.

The evolution of macro-texture is further illustrated by the constant $\varphi_2 (\varphi_2 = 0^\circ, 45^\circ, \text{and} 65^\circ)$ sections of the ODF in Fig. 6. The important texture components are summarized in Table. 1. It can be inferred that CR50% results in the formation of the $\alpha$-fiber texture (<110>//RD), which is characterized by a strong Goss texture component, followed by the Brass texture component. The appearance of Copper (Cu {112}<111>) and S ({123}<634>) components are noticed in the $\varphi_2 = 45^\circ$ and 65$^\circ$ after CR50%. With increasing the CR reduction to 70%, the intensity of Goss/Brass (G/B {110}<110>) component increases, indicating the Goss orientation continuously
shifted toward Brass orientation. Besides, the main texture is further converted from Cu into Goss and Brass textures aided by widespread deformation twinning, as shown in Fig. 3d. Meanwhile, the Goss component reveals an improved scatter toward the Copper Twin (CuT {552}<115>) position at the 70% rolling reduction stage [31]. The texture components of the present HEAs are complex and depend on multiple deformation mechanisms, namely dislocation slipping, deformation twinning, and shear banding. Additionally, a weak γ-fiber (<111>//ND) begins to develop. In summary, the texture measurements turn out the characteristic α-fiber in low SFE fcc HEAs along with weaker γ fiber. The texture evolution in the present HEAs is similar to that in Fe-28Mn-0.28C TWIP steels [31].

3.3 Mechanical properties of the cold-rolled HEAs

Typical tensile stress-strain curves for the homogenized and cold-rolled samples, tested in uniaxial tensile loading along the RD and TD, are presented in Fig. 7a. The as-homogenized sample shows the highest attainable elongation (EL) of 71.4% with a yield strength (YS) of 124 MPa and ultimate tensile strength (UTS) of 405 MPa. As seen from the diagram, a distinct increase of the yield/tensile strength in CR50% and CR70% conditions occurs, and the elongation is sharply decreased from 71.4% to 6~12%. Notably, the tensile properties of cold-rolled HEAs presented in Table 2 are dependent on the loading directions, where the YS values for RD are higher than that for TD. In contrast, the UTS along TD after CR50% and CR70% reductions are 1010 MPa and 1186 MPa, respectively, which are 39 and 69 MPa higher than that along RD, respectively. There is no obvious difference in the elongation rate at both RD and TD,
as depicted in Fig. 7a. Expectedly, both samples exhibit weak work-hardening capability.

Fig. 7b shows force-displacement curves of nanoindentation tests along RD and TD of CR50% HEAs. The measured elastic modulus for the RD (197 ± 4 GPa) and TD HEAs (183 ± 4 GPa) are very close to each other, both of which are somewhat higher than FeCoCrNiMn fcc-structured HEAs (179 ± 4 GPa) [32]. These results signify that the present HEAs exhibit anisotropic behavior and suffer from anisotropy in mechanical properties. The anisotropy of mechanical properties in RD/TD is related to the texture component and corresponding volume fraction caused by the anisotropy in the atomic arrangement. Similar to the present HEAs, the cold-rolled Al-Mg alloy, AA 5005, shows obvious Goss and Brass texture components, the yield strength is very close, but the tensile strength is higher along TD than along RD [33]. Usually, the fcc metals with Brass texture own higher tensile strength along TD than along RD.

As shown in Fig. 7c, the cold-rolled HEA displays high strength, but its plasticity is limited. It lacks sufficient work-hardening ability. The tendency of plastic instability in the early stage of plastic deformation, which is called localized necking usually results in fracture, as shown in Fig. 7d. The beginning of localized flows is predicted fairly well by the well-known Considéré criterion [34] (see Supplementary materials).

3.4 Microstructure and texture evolutions during annealing

Fig. 8 displays the microstructure evolution of the CR70% samples after
annealing subjected to different temperatures. The HEAs show a completely recrystallized microstructure after annealing at 800°C. It is almost impossible to have a full recrystallization when annealing present heavily-deformed HEAs (not shown here) and Nb-dopped TWIP steels below 600°C for 1 h, partial recrystallization happened at 650 and 700°C for 1 h, and the complete recrystallization microstructure is also formed at 800°C in CoCrFeMnNi HEAs [13, 35]. Annealing at 900°C and 1000°C for 1 h, as expected, results in discernable grain growth. All the fully-recrystallized grains are separated by HAGBs and still remain single fcc phase. Abundant annealing twins could be easily observed in the annealed samples, which gives already evidence of the low value of the SFE (< 40 mJ/m²) [36]. The annealing twin, clearly exhibiting very strong peak at misorientation angle of 60° corresponding to the 60°<111> twinning relationship, could be confirmed from the misorientation distribution plots (Fig. 8d). The average grain size, excluding annealing twins (i.e. twins are not considered as separate grains), are 9.2, 18.2, and 47.1 μm after annealing at 800, 900, and 1000°C, respectively. The evolution of annealing textures could be understood via analyzing the micro texture components and corresponding volume fractions in the Fig. 8e. The dominant micro-annealing textures are inherited from the deformation textures after annealing, showing relatively high volume fraction of Goss and Copper textures. The increase of Cube texture indicates the randomization of crystallography. After increasing the annealing temperature up to 1000°C, the Goss and Copper textures decrease due to the randomization. However, the Brass shows the opposite trend of growth may be due to the preferential growth of Brass texture during
annealing.

4. Discussion

4.1 Relationship between the microstructure and texture evolutions during cold rolling

The evolution of the microstructure and texture for HEAs is tightly associated with the SFE, which determines the dislocation movement. Conventionally, Copper-type textures are formed in alloys with medium and high SFE, such as Al and Cu alloys, during cold rolling. As for the relatively-low SFE alloys, the Goss component has a higher intensity, compared to the Brass component after medium and high degrees of rolling [37]. It is well known that the twin formation begins to occur with decreasing SFE down to a range between 12 and 35 mJ/m$^2$ [38] or 14 and 50 mJ/m$^2$ [39]. Zaddach et al. [14] have studied the SFE of CoCrFeNi and CoCrFeNiMn HEAs (about 20 – 30 mJ/m$^2$), and with the minor addition of Al in CoCrFeNi, the SFE is expected to decrease further [15]. Earlier studies have reported the formation of the Brass texture component being the main texture in CoCrFeMnNi [12] and CoCrFeNi [19] HEAs after CR90%. Similar to the current CR50% results, in experiments by Tazuddint et al [21], the characteristic deformation texture components like Goss, Cu, Bs, and S start to appear after CR70% in MnFeCoNiCu HEAs. The Cu, S, and Goss components are increased as a result of dislocation slip being more active at the early stages of deformation. The presence of Brass texture is caused by partial deformation twinning, which is well coincided with the observations by Haase et al. [13].
The intensity of G/B components is increased, and the Cu texture is transformed to Goss texture gradually, illustrating the deformation twins after CR70% reduction. During the transformation of texture, an intermediate texture component formed, known as the CuT component. The Goss and CuT texture components are reported as an indication for deformation twinning in low SFE alloys [40]. Therefore, in the subjected HEAs with a 70%-rolling reduction, the grains deformed by mechanical twinning as substantiated by profuse T-M lamellae in Fig. 3d. Moreover, the agglomeration and collapse of the T-M lamellae and pronounced planarity of dislocation slip, developing the shear bands that are created as a necessity to accommodate more plastic deformation occurred infrequently (Figs. 2c and d). The appearance of shear bands on the microscale in this reduction condition leads to grain fragmentation. Accordingly, the appeared shear bands result in a weak γ fiber (F {111}<112> and E {111}<110> components) [41-43]. The development of a weak γ-fiber with increased E and F texture components is caused by not only related to the T-M lamellae but also the activation of shear deformation as shown in Fig. 2c, and as reported by Weidner et al [43]. A shear-band sliding (SBS)-dominated deformation mechanism has been previously reported in Al_{0.25}CoCrFeNi HEAs after CR90% [20]. The evolution of such texture components is similar to earlier studies in low SFE austenitic TWIP steels, but the microstructure did not present pronounced deformation twinning at low degrees of cold rolling.

Generally, the influences of microstructure on the response of texture components during rolling of the present low SFE HEAs can be summarized as
follows:

(i) The increase of Cu, S, and Goss components is due to more active dislocation slips at the early stages of deformation. The presence of Brass texture is caused by partial deformation twinning.

(ii) The increase in Brass, Goss, and CuT components can be correlated with the onset of mechanical twinning, the development of a weak γ-fiber with increased E and F components is resulted from the activation of shear deformation. Regarding the microstructure, the present HEAs have revealed that dislocations and slip lines prevailed at low deformation degrees and more deformation twinned and bands sheared at high degrees.

4.2 Relationships among microstructures, textures, and mechanical properties.

Profuse slip lines and shear bands on the microscale lead to grain fragmentation, and high-density dislocations and abundant mechanical twins result in nanoscale substructures, which enhances the yield and tensile strengths. The cold-rolled samples show anisotropies of mechanical properties in RD and TD mainly caused by the crystallographic texture [44]. The volume fraction of the main texture components during cold rolling is presented in Fig. 9. Texture-induced anisotropy has been widely studied and simulated with the crystal plasticity model [33]. The anisotropy of cold-rolled alloys can be calculated by the “in-plane anisotropy (AIP)” [45, 46] and estimated using the parameter of “anisotropy index (δ)” [47]:

\[
\text{AIP\%} = \left( \frac{S_{\text{max}} - S_{\text{min}}}{S_{\text{max}}} \right) \times 100\%
\]

(1)
\[
\delta = \frac{|\%El(R) - \%El(T)|}{\%El(R) + \%El(T)} \times 100\% 
\]  

(2)

where \( S_{\text{max}} \) and \( S_{\text{min}} \) are the maximum and minimum value of yield strength (\( \sigma_{0.2} \)) along various directions, respectively; while \% El (R) and \% El (T) are the percentage elongation of the tensile specimens along RD and TD, respectively. Texture components and the AIP of yield strength have been semi-quantitatively evaluated in 2524 T3 alloy sheets [46]. The calculated AIP and \( \delta \) of CR50% samples are 11.28% and 5%, respectively, which demonstrates the obvious anisotropy of the cold-rolled HEAs. The yield strength along RD is higher than that along TD. However, the tensile strength along RD is lower than that along TD. The results indicate that tensile deformation along RD is difficult than along TD in the initial deformation. Moreover, \( \delta \) shows that the elongation of TD is larger than RD, indicating the greater strain hardening ability at TD, which results in higher tensile strength. Additionally, upon applying the same indentation depth on the S1 and S2 planes, the S1 plane responds with a higher force, also demonstrating the relatively higher strain hardening ability. With increasing the rolling reduction to 70%, the AIP reduced to 3.2% due to the further enhancement by the dislocation density for the flow stress.

The initial grain orientation has been proved to have a significant effect on slipping, twinning, crystallographic texture, and anisotropic mechanical properties. According to Gutierrez-Urrutia et al. [48] and Sato et al. [49], the influence of the grain orientation detected at the initial deformation stage is governed entirely by the relative value of the Schmid factor (\( m \)). As known, the primary deformation mode in fcc alloys is the prismatic \{111\}<110> slipping and \{111\}<112> twinning. Because of
the high-density dislocations and numerous T-M lamellas microstructure, as presented in Fig. 3, it will be reasonable to assume that the deformation of cold-rolled Al_{0.1}CoCrFeNi HEAs occurs by slipping and twinning. Generally, the operative slip system under uniaxial tensile deformation is preferentially determined by the relative magnitudes between the critical resolved shear stress (CRSS) and the Schmid factor for each slip system. The texture components, volume fractions, loading directions, and corresponding Schmid factors are listed in Table 3.

The cold-rolled HEAs will be postulated in the following discussion to consist of governing texture component crystals with the orientations corresponding to the ODF figures. The maximal values of the Schmid factor for Goss and CuT components are almost same whether tension direction is along RD or TD. The Brass texture shows a lower m along TD than RD indicating larger critical stress for deformation at TD, i.e., the YS is higher at TD than that at RD [16]. Obviously, the above results indicate that for further exploration of the cold-rolled HEAs tensile behavior, a simplified view that considers only crystallography - neglecting microstructure morphology - is not sufficient [50]. Besides, the appearance of a set of parallel subgrain boundaries (slip lines/twin boundaries/shear bands) affects the selection of slip systems. The rolling plane observed in Fig. 1a shows that these grains are subdivided into the parallel-microsized slip band with most slip lines paralleling to TD, as can also be seen in Supplementary materials. The parallel slip lines divide the grains into micro-sized columnar grains, similar to the nanoscale-twins columnar-grained copper [51].
Fig. 10 presents the fracture morphology at the initial necking zones, the center of necking zones, and the crack-tips when tension along RD and TD, respectively (the whole deformed regions are shown in Supplementary materials). Evidently, there are two types of slip traces from the inspection. The first type is shallow, straight, and perpendicular to the initial slip lines in the CR50% HEAs, where the cold-rolling induced slip lines gradually evolve into cracks with an angle of 30° to the slip traces marked by blue arrows. This trend leads to some sub-crack surfaces perpendicular to the slip traces and the formation of cleavage fracture surfaces. By contrary, the secondary slip trace is long, straight, and located in the HEAs parallel to the loading direction. The sub-cracked planes are formed with an angle of 34° to the slip traces near the crack tips. It is probable that the slip traces resulted from dislocations emitted along the initial slip lines, and the sub-crack is formed by a cross slip trace near the crack tips.

From the above analyses, the underlying deformation mechanisms responsible for the special microstructure in present HEA are schematically illustrated in Fig. 11. Thus, the deformation response for this microstructure is divided into the in-plane slip and cross-plane slip. Dislocation glide in between the slip lines is easier than across the slip lines. The previously formed parallel slip lines, twins, and shear bands drastically reduce the effective glide distance of dislocations along RD. Because the slip band interfaces act as slip planes of dislocations and provide ample room for dislocation storage such as the coherent twin interfaces in fcc metals. It should be mentioned that the extended dislocations are firstly nucleated from the intersections
between GBs and twinning boundaries (TBs), and then slip across TBs/GBs strugglingly [51]. The critical resolved shear stress acting on a single dislocation is low to overcome the threshold stress of dislocation paths and deformation twins, resulting in the blocked dislocations, as seen from Fig. 3e and f. Thus, the easier dislocation motion of the in-slip plane slip system promotes slightly larger homogenize plastic deformation along TD, which results in a higher strain-hardening rate at TD. Conventionally, the maximum $m$ is generally shared by several slip or twin systems on the consideration of strong texture. This, combined with the columnar-like grain, the grain shape may active individual slip systems to strengthen differently, depending on the alignment of the slip plane and shear direction with respect to the grain principal diameters. Therefore, the cold-rolled HEAs exhibit novel anisotropic yield strength, ductility, and work hardening capacity.

The Poisson’s ratio, shear modulus, and Young’s modulus of CR50% HEAs are given in Table 4. Ultrasonic longitudinal and transverse wave velocities and longitudinal backscattering coefficients in S1, S2, and S3 planes were measured by a broadband 5-MHz focused transducer, and Young’s modulus was calculated on the various planes are 163, 172.5, and 176.5 GPa, respectively. Obviously, Young’s modulus of present rolled HEAs on the S1 plane is lower than that on the S2 plane, similar to the cold-rolled martensitic Ti-V-Sn alloys [52]. Considering the random texture for homogenized alloys and the preferred Brass and Goss rolling textures (Fig. 6), the anisotropy in Young’s modulus will be due to the rolling texture formation. But, the ultrasonic backscattering shows complicated anisotropic behavior along the
propagation direction as a result of grain shape and macroscopic texture, which essentially results in the anisotropy of Young’s modulus in the rolled HEAs [53].

4.3 Strengthening mechanisms of the dynamic slip band refinement and dislocations.

Engineering stress-strain curves for tensile testing along RD of the rolled and annealed Al$_{0.1}$CoCrFeNi HEAs are exhibited in Fig. 12a. True stress-strain and work-hardening rate curves of the HEAs are shown in Fig. 12b. The increases in the strengths (both YS and UTS) of present HEAs with reduction in thickness are believed primarily due to the cold rolling-induced increase in density of dislocations, the formation of refined slip bands and deformation twins (a dynamical Hall-Petch effect), and the reduced casting-induced pores/voids [54, 55]. The different mechanisms contributing to the overall strength of the present alloys during annealing are presented in Fig. 12c. Therefore it can be typically expressed, using the modified Hall-Petch equation [54, 56]:

\[
\sigma_{0.2} = \sigma_0 + \sigma_\rho + \sigma_{H-P}
\]  

(3)

where \(\sigma_{0.2}\) denotes the offset yield strength, \(\sigma_0\) is the frictional stress, \(\sigma_\rho\) contributes to the substructure strengthening, and \(\sigma_{H-P}\) is the Hall-Petch strengthening. The dislocations strengthening to all strength is typically expressed as the form:

\[
\sigma_\rho = M\alpha Gb\sqrt{\rho}
\]  

(4)

where \(M\) denotes the Taylor factor (3.06 for fcc materials), \(\alpha = 0.2\) stands a constant for fcc metals, \(b\) is the Burgers vector, \(G\) is the shear modulus, and \(\rho\) is the dislocation density [8]. The dislocation density of \(\rho\) was calculated from the average values of the
crystallite size, $d$, and micro strain of $\varepsilon$ ($\varepsilon_{\mathrm{CR}50\%} \sim 0.191$, $\varepsilon_{\mathrm{CR}70\%} \sim 0.237$) according to previous studies [20, 57]. The Hall-Petch contribution can be given as:

$$\sigma_{H-P} = K(d_S^{-1/2} + d_T^{-1/2})$$  \hfill (5)

where $K$ reflects the Hall-Petch coefficient, $d_S$ and $d_T$ represent the grain size of slip band and twin, respectively. In this work the following parameters were used: $G = 67$ GPa determined by UMS-100 on the S1 plane, $b = 2\sqrt{2}\times a = 0.253$ nm [20, 58]. The value of $\sigma_0 = 80$ MPa and $K = 825$ MPa $\mu$m$^{-1/2}$ were experimentally calculated as the yield stress of the HEAs in the coarse-grained (homogenized and annealed) state as shown in Fig. 12d, almost equal to the values obtained for Al$_{0.3}$CoCrFeNi and VCoNi HEAs [59, 60]. Especially notable is, the Hall-Petch coefficient of $K$ is remarkably large. This fiercely high dependence of strength upon grain size change, unlike most of the alloys, implies that grain refinement plays an important role in further improving the YS. The values of refined grain sizes, $d_S$ and $d_T$, are obtained from Fig. 1 and Fig. 3 as the spacing between slip lines and twins, respectively. In comparison with the experimental datas, the values of the YS calculated using Eqs. (3) - (5) are relatively lower. The refined slip lines near the initial GBs cannot be collected well in the CR50% HEAs, and the shear bands and dislocation cells are not considered in predicting the YS of CR70% HEAs, which decrease the theoritical values. The increasing tendency of YS appears to be slow, because the structural defects gradually become saturated with the improvement of rolling reductions.

From an application standpoint, although cold working significantly promotes the yield and ultimate tensile strength of HEAs to 1068 MPa and 1117 MPa,
respectively. It seems to be more attractive for the consideration of a reasonable balance between strength and ductility (Fig. 12a). By annealing the 70% cold-rolled HEAs, there is a gradual decrease in the YS and UTS, and a continuation increase in EL%. To begin with, upon annealing at 800 $\degree$C, static absolute recrystallized processes formed equiaxed grains with extensive annealing twins (Fig. 8a), which indicates a high ability to accumulate dislocations. Hence, the annealed HEAs at 800 $\degree$C are capable of storing further dislocations, leading to strongly increasing work-hardening rates. A relatively good balance among YS, UTS, and fracture EL is achieved with 370 MPa, 733 MPa, and 55 %, respectively. Partially recrystallized microstructure are formed in CoCrFeMnNi and current HEAs when annealing below 800 $\degree$C, which achieves the relatively higher yield and tensile strengths than the present work [13]. With increasing the annealing temperatures, the YS and UYS are decreased further whereas the total EL is increased accordingly.

The principal contribution to the high work hardening ability of the partially or fully recrystallized alloy restrain doubtlessly from the activation of multiple slip systems, the strong planarity of dislocation slip and slip-band formation, the appearance of deformation twins, and the related low ability for dislocation cross-slip, as usually occurred in fcc HEAs [8, 13, 61]. According to Laplanche and Haase, the work-hardening rate curves (Fig. 12b) can be divided into two stages, at the initial strain, A, it is the remarkably high overall strain hardening rate stage, where dynamic recovery processes facilitate a tremendously, continuously decreasing work-hardening rate. At a higher strain B, the reduced capability to store further dislocations lead to a
continuously decreasing slope in the work-hardening rate-true stress/strain curve [8, 42]. Although the 1000立方-annealed HEAs have a slower descent platform at stage B, it must be pointed out that, the strength is lower than the others due to the coarsened grains. Deformation twinning in recrystallized grains of CoCrFeNi, CoCrFeMnNi, and Al_{0.3}CoCrFeNi HEAs was only observed sparsely or no during room temperature tensile test [13, 62, 63]. The peculiar work-hardening rate of the partially recrystallized alloys usually was divided into five stages caused by numerous deformation twins [13, 64]. The present alloys present a monotonous decay of the strain hardening rate as a function of strain.

Summarizing the relationship between the rolling strain and the YS, as calculated by Eq (3), as well as the contributions of the grain size strengthening and the dislocation strengthening to the overall YS, it can be, therefore, concluded that the contribution of dynamic grain refinement in this studied HEAs is more important for the overall strengthening, compared to the increased dislocation density.

5. Conclusion

A low-SFE Al_{0.1}CoCrFeNi HEA has been subjected to cold rolling and annealing to form different microstructural characters with respect to dislocation, deformation twin, texture, and grain size states. Comprehensive microstructure and texture analyses have been performed after defined deformation degrees to reveal the deformation behaviors of these alloys. The conclusions drawn from the results can be summarized as follows:

1. The physical deformation of the HEAs is mainly controlled by planar gliding
of dislocation and partial deformation twinning upon CR50%, forming slip lines and twin bundles. When the HEAs are cold rolled to 70% reduction, the slip lines evolve into shear bands and the twin bundles are further activated into T-M lamellas, respectively.

2. During the rolling process, the development of texture components of present low SFE HEAs can be summarized as follows: (i) the increase of Cu, S, and Goss components is due to the more active dislocation slip at early stage of deformation. The presence of Brass texture is caused by the partial deformation twins; and (ii) the increase in Brass, Goss, and CuT components after the onset of mechanical twinning. The development of a weak $\gamma$ fiber with the increased F and E components is depend on the activation of shear deformation.

3. The dominant Brass texture has a lower Schmid factor at TD than at RD, indicating larger critical stress for deformation at TD, i.e., higher YS at TD than that at RD. However, the dislocation glide in between the slip lines/twins is easier than across them, which results in the relatively lower YS along TD than RD. The easier dislocation motion promotes the slightly larger homogeneous plastic deformation along TD, which obtains a higher strain hardening at TD, i.e., greater UTS. Moreover, the cold-rolled HEAs demonstrate the anisotropic Poisson’s ratio, Young’ modulus, and bulk modulus due to the special microstructures and textures.

4. The YS and UTS of the investigated HEAs are improved to 1068 MPa and
1117 MPa after CR70%, respectively. The improvement of YS are primarily originated from the cold rolling-induced increase in the density of dislocation and the dynamic formation of slip bands, shear bands, and deformation twins. The quantitative analysis favors that the dynamic grain refinement provides noticeably higher contribution in comparison with the dislocation hardening to the overall YS.

5. After annealing the cold-rolled HEAs at 800 °C, fully recrystallized grains are obtained with random orientation containing numerous annealing twins, a relatively good balance among YS, UTS, and fracture EL is achieved with 370 MPa, 733 MPa, and 55 %, respectively. The Al$_{0.1}$CoCrFeNi HEAs exhibit a monotonous decrease of the strain hardening rate as a function of strain especially in the finer-recrystallized grains.

**Acknowledgement**

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**Reference:**


[40] S. Vercammen, B. Blanpain, B.C. De Cooman, P. Wollants. Cold rolling behaviour of an austenitic


Figure captions
Fig. 1. Optical morphologies of Al$_{0.1}$CoCrFeNi HEAs after different CR reductions: (a) CR 50%; (b) CR 70%. The black dashed lines depicting the main direction of the slip lines. (GB = grain boundary).
Fig. 2. EBSD band contrast maps and crystal orientation maps on the rolling section with different CR reductions: (a), (b) CR50% and (c), (d) CR70%. The substantial
formations of slip lines and deformation twins were observed after CR50%, shown in the dotted box of (a) in (e) and (f). (g) Point-to-point misorientation profiles corresponding to the 50% cold-rolled sample. (h) (111) pole figure obtained from the yellow line area in (f).

Fig. 3. TEM bright field (BF) and dark field (DF) micrographs of the sample subjected to the 50% and 70% cold deformations: (a), (b) CR50% and (c), (d), (e), (f) CR70%. (DPs = Dislocation paths, DCs = Dislocation cells, GB = Grain boundary, SBs = Shear bands, Ds = Dislocations)
Fig. 4. The distribution of grain-boundary misorientation angles for different cold-rolled HEAs: (a) CR50% and (b) CR70%.

Fig. 5. The evolution of (111) polefigures of cold-rolled Al$_{0.1}$CoCrFeNi HEAs on the rolling planes: bulk textures determined by XRD (a) CR50% and (b) CR70%; the microtextures determined by EBSD (c) CR50% and (b) CR70%.
Fig. 6. Schematic illustration of the ideal texture components and fibers formed in the current HEAs, ODF sections (determined by XRD) at 0°, 45°, and 65°: (a) CR50% and (b) CR70%.

Fig. 7. (a) Engineering stress-strain curves of the tension along RD and TD. (b)
Force-displacement curves of CR 50% HEAs loaded by nanoindents on S1 and S2, respectively. (c) Strain-hardening rate of 70% CR HEAs with the magnified part (inserted in a) tensile along RD; and (d) Local strain levels with respect to true strain ($\varepsilon_T$).

Fig. 8. EBSD orientation micrographs of the CR70% HEAs after annealing at various temperatures: (a) 800 $^\circ$C, (b) 900 $^\circ$C, and (c) 1000 $^\circ$C. (d) The grain boundary misorientation for annealed HEAs as a function of annealing temperature. (e) Texture components and corresponding volume fractions of HEAs annealed at 800 $^\circ$C, 900 $^\circ$C, and 1000 $^\circ$C.
Fig. 9. Volume fractions of the main texture components developed during cold rolling.

Fig. 10. SEM images at the initial necking zones, the center of the necking zones, and crack tips for the CR50% reduction HEAs after fracture: (a), (b), and (c) for loading along RD; (d), (e), and (f) for loading along TD; the initial slip lines (caused by cold rolling, yellow arrows), slip traces (red arrows), and sub-cracks (blue arrows) are indicated.
Fig. 11. Schematic representation of deformation mechanisms in the cold-rolled HEAs as a combination of parallel slip lines and twin bundles, indicating the dislocation trace and crack-plane evolution during the tensile test.
Fig. 12. (a) Engineering stress-strain curves and (b) True stress-strain curves and work-hardening rate of Al_{0.1}CoCrFeNi HEAs; (c) Contribution of different strengthening mechanisms to the overall strength of rolled HEAs; and (d) Hall-Petch plot of the HEAs after recrystallization.
### Table captions

Table. 1 Texture component, Euler angles, and Miller indices in the cold-rolled alloys.

<table>
<thead>
<tr>
<th>Component</th>
<th>Symbol</th>
<th>Euler angles $(\phi_1, \Phi, \phi_2)$</th>
<th>Miller indices</th>
<th>Fiber</th>
</tr>
</thead>
<tbody>
<tr>
<td>Cube (C)</td>
<td>■</td>
<td>(0, 0, 0)</td>
<td>{001} &lt;100&gt;</td>
<td>/</td>
</tr>
<tr>
<td>Copper (Cu)</td>
<td>●</td>
<td>(90, 35, 45)</td>
<td>{112} &lt;111&gt;</td>
<td>$\gamma, \tau$</td>
</tr>
<tr>
<td>Brass (Bs)</td>
<td>♠</td>
<td>(35, 45, 0)</td>
<td>{110} &lt;112&gt;</td>
<td>$\alpha, \beta$</td>
</tr>
<tr>
<td>Goss (G)</td>
<td>◆</td>
<td>(0, 45, 0)</td>
<td>{110} &lt;001&gt;</td>
<td>$\alpha, \tau$</td>
</tr>
<tr>
<td>Goss/Brass (G/B)</td>
<td>★</td>
<td>(90, 45, 0)</td>
<td>{110} &lt;110&gt;</td>
<td>$\alpha$</td>
</tr>
<tr>
<td>Copper Twin (CuT)</td>
<td>▲</td>
<td>(74, 90, 45)</td>
<td>{110} &lt;115&gt;</td>
<td>$\tau$</td>
</tr>
<tr>
<td>E</td>
<td>△</td>
<td>(0/60, 55, 45)</td>
<td>{111} &lt;110&gt;</td>
<td>$\gamma$</td>
</tr>
<tr>
<td>F</td>
<td>◊</td>
<td>(30/90, 55, 45)</td>
<td>{111} &lt;112&gt;</td>
<td>$\gamma$</td>
</tr>
<tr>
<td>S</td>
<td>▼</td>
<td>(59, 37, 63)</td>
<td>{123} &lt;634&gt;</td>
<td>$\beta$</td>
</tr>
</tbody>
</table>

- $\alpha$-fiber $<110> \parallel$ ND
- $\beta$-fiber $<110>$ tilted 60° from ND toward RD
- $\gamma$-fiber $<111> \parallel$ ND
- $\tau$-fiber $<110> \parallel$ TD
Table 2. Yield strength (YS), ultimate tensile strength (UTS), and elongation to fracture (EL) for Al₀.₁CoCrFeNi HEAs with the different cold rolling reduction when the tensile direction is along RD and TD, respectively.

<table>
<thead>
<tr>
<th>Rolling reduction</th>
<th>Tensile direction</th>
<th>YS (MPa)</th>
<th>UTS (MPa)</th>
<th>EL(%)</th>
</tr>
</thead>
<tbody>
<tr>
<td>50 %</td>
<td>RD</td>
<td>939±8</td>
<td>975±5</td>
<td>12.3±2</td>
</tr>
<tr>
<td></td>
<td>TD</td>
<td>833±4</td>
<td>1010±4</td>
<td>13.6±0.5</td>
</tr>
<tr>
<td>70 %</td>
<td>RD</td>
<td>1068±4</td>
<td>1117±2</td>
<td>6.5±1</td>
</tr>
<tr>
<td></td>
<td>TD</td>
<td>1033±9</td>
<td>1186±10</td>
<td>8.2±0.2</td>
</tr>
</tbody>
</table>

Table 3. Main texture components, corresponding volume fraction, loading direction (RD/TD) and Schmid factors, the maximal value of the Schmid factor calculated for the\{110\}<1 1 1> slip systems and\{111\}<112> twinning systems.

<table>
<thead>
<tr>
<th>Texture component</th>
<th>Volume fraction (%)</th>
<th>Loading direction</th>
<th>Schmid factor (m)</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td>CR50%</td>
<td>CR70%</td>
<td>Slip</td>
</tr>
<tr>
<td>Cube</td>
<td>6.9</td>
<td>0.8</td>
<td>&lt;100&gt;/&gt;/RD</td>
</tr>
<tr>
<td></td>
<td></td>
<td></td>
<td>&lt;010&gt;/&gt;/TD</td>
</tr>
<tr>
<td>Goss</td>
<td>12.5</td>
<td>17.3</td>
<td>&lt;001&gt;/&gt;/RD</td>
</tr>
<tr>
<td></td>
<td></td>
<td></td>
<td>&lt;110&gt;/&gt;/TD</td>
</tr>
<tr>
<td>Brass</td>
<td>8.8</td>
<td>12.2</td>
<td>&lt;112&gt;/&gt;/RD</td>
</tr>
<tr>
<td></td>
<td></td>
<td></td>
<td>&lt;111&gt;/&gt;/TD</td>
</tr>
<tr>
<td>Cooper</td>
<td>6.6</td>
<td>5.2</td>
<td>&lt;111&gt;/&gt;/RD</td>
</tr>
<tr>
<td></td>
<td></td>
<td></td>
<td>&lt;110&gt;/&gt;/TD</td>
</tr>
<tr>
<td>CuT</td>
<td>6.7</td>
<td>13.3</td>
<td>&lt;115&gt;/&gt;/RD</td>
</tr>
<tr>
<td></td>
<td></td>
<td></td>
<td>&lt;110&gt;/&gt;/TD</td>
</tr>
<tr>
<td>Random</td>
<td>59.1</td>
<td>51</td>
<td>/</td>
</tr>
</tbody>
</table>
Table. 4 The C11, C44, Poisson's ratios, bulk modulus, Young’s modulus, and shear modulus of CR50% samples at RD, ND, and TD, respectively.

<table>
<thead>
<tr>
<th>Direction</th>
<th>C11</th>
<th>C44</th>
<th>Poisson ratio</th>
<th>Bulk modulus B (GPa)</th>
<th>Young's modulus E (GPa)</th>
<th>Shear modulus G (GPa)</th>
</tr>
</thead>
<tbody>
<tr>
<td>RD</td>
<td>263.64</td>
<td>60.30</td>
<td>0.35</td>
<td>183.24</td>
<td>163.03</td>
<td>60.30</td>
</tr>
<tr>
<td>ND</td>
<td>270.20</td>
<td>64.17</td>
<td>0.34</td>
<td>184.64</td>
<td>172.53</td>
<td>64.17</td>
</tr>
<tr>
<td>TD</td>
<td>237.43</td>
<td>67.93</td>
<td>0.30</td>
<td>146.86</td>
<td>176.56</td>
<td>67.93</td>
</tr>
</tbody>
</table>
(a) Engineering stress (MPa) vs. Engineering strain (%)
(b) True stress $\sigma$ (MPa) vs. Work-hardening rate $\frac{d\sigma}{d\varepsilon}$
(c) Yield strength (MPa) vs. Rolling reduction (%)
(d) Engineering stress (MPa) vs. $\frac{d^{-1/2}}{m(\mu m^{-1/2})}$

\[ \sigma = 80 + 825d^{-1/2} \]
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