Effect of Cobalt-Content on Mechanical Properties of Non-Equiatomic Co–Cr–Ni Medium Entropy Alloys

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Non-equiatomic high entropy alloys (HEAs) and medium entropy alloys (MEA)s are expected to have the potential to exhibit good mechanical properties due to abundant composition designs compared to equiatomic alloys. It has been reported that an equiatomic CoCrNi MEA shows better strength-ductility balance than CoCrFeMnNi HEA, and there is a possibility that the mechanical properties can be further improved by changing chemical composition. Among the constituent elements, cobalt (Co) has the effect of decreasing stacking fault energy (SFE). In this study, we clarified the effect of Co-content on mechanical properties of non-equiatomic Co-Cr-Ni MEAs with different amounts of Co through investigating deformation behaviors and deformation microstructures. Co0.5CrNi100-x (x = 20 (Co20), 40 (Co40), 60 (Co60) at%) MEAs were processed to very high plastic strains by high-pressure torsion (HPT) and subsequently annealed under proper conditions to obtain FCC single-phase and uniform fine grain sizes. Mechanical properties of the specimens with fully rerecrystallized microstructures were characterized by tensile tests at room temperature. Their deformed microstructures at different tensile strain levels were observed by electron microscopy. The result of the tensile tests showed that the work-hardening rate was enhanced with increasing the Co-content although early fracture before reaching plastic instability condition occurred in Co60. Planar slip of dislocations and deformation twinning were observed in Co20 (SFE = 30 mJ/m2), while, in addition to them, deformation-induced martensitic transformation to HCP ε-martensite was observed in Co40 having lower SFE (SFE = 10 mJ/m2), leading to higher work-hardening rate. By increasing Co-content (decreasing SFE) further, phase fraction of ε-martensite greatly increased in Co60 (SFE = 0 mJ/m2) compared with Co40, and early fracture occurred due to stress concentration at intersects between martensite and grain boundaries. The present results suggested that the mechanical properties of the present materials could be effectively designed by controlling the SFE.

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1. Introduction

Conventionally, alloys have been designed by adding small amounts of alloying elements to a base metal such as iron, aluminum, etc., of which chemical compositions locate in the edge parts of phase diagrams. On the other hand, a concept of high entropy alloys (HEAs) that enables a wide range of their chemical compositions deviating from an equiatomic concentration. HEAs are often referred to alloys composed of five or more principal elements with a (near-) equimolar composition, locating in the center part of phase diagrams. In this study, we clarified the effect of Co-content on mechanical properties of non-equiatomic Co–Cr–Ni MEAs with different amounts of Co through investigating deformation behaviors and deformation microstructures. MEAs were processed to very high plastic strains by high-pressure torsion (HPT) and subsequently annealed under proper conditions to obtain FCC single-phase and uniform fine grain sizes. Mechanical properties of the specimens with fully rerecrystallized microstructures were characterized by tensile tests at room temperature. Their deformed microstructures at different tensile strain levels were observed by electron microscopy. The result of the tensile tests showed that the work-hardening rate was enhanced with increasing the Co-content although early fracture before reaching plastic instability condition occurred in Co60. Planar slip of dislocations and deformation twinning were observed in Co20 (SFE = 30 mJ/m2), while, in addition to them, deformation-induced martensitic transformation to HCP ε-martensite was observed in Co40 having lower SFE (SFE = 10 mJ/m2), leading to higher work-hardening rate. By increasing Co-content (decreasing SFE) further, phase fraction of ε-martensite greatly increased in Co60 (SFE = 0 mJ/m2) compared with Co40, and early fracture occurred due to stress concentration at intersects between martensite and grain boundaries. The present results suggested that the mechanical properties of the present materials could be effectively designed by controlling the SFE. 

In fact, Li et al.7–9) reported that the strength-ductility balance of CoCrFeMnNi HEA and CoCrFeMn MEA were improved by changing the chemical composition. This was because the transformation-induced plasticity (TRIP) behavior and/or the twinning-induced plasticity (TWIP) behavior that form characteristic FCC–HCP dual-phase (DP) microstructures were controlled by chemical compositions. A similar attempt has been made by other research groups. For instance, Wei et al.10) developed a series of non-equiatomic Co-rich Co–Cr–Fe–Ni MEAs exhibiting superior tensile properties due to TWIP and TRIP. According to their results, Co has an effect to lower stacking fault energy (SFE) of alloys, leading to a high probability of twinning and ε-martensitic transformation during deformation.

In the present study, we studied the effect of Co-content on mechanical properties of Co–Cr–Ni ternary MEAs. Non-equiatomic Co–Cr–Ni MEAs with fully-rerecrystallized microstructures were fabricated and tensile-deformed at room temperature to obtain stress-strain curves. The relationship between deformation microstructures and mechanical properties was investigated by microscopy techniques. Finally, the effect of SFE on the deformation mechanisms operated in the MEAs were discussed.
2. Experimental Procedure

2.1 Materials and process

2.1.1 Fabrication of alloys

Ingots of $\text{Co}_x\text{(CrNi)}_{100-x}$ ($x = 20, 40, 60 \text{ at} \%$) were produced by arc-melting of pure Co, Cr, and Ni with purities higher than 99.9 mass\% in a water-cooled mold under inert Ar gas atmosphere. The three kinds of alloys are hereafter called as Co20, Co40, and Co60, respectively, in this paper. The ingots were flipped and re-melted at least four times to improve elemental homogeneity.

2.1.2 Thermo-mechanical processing

The as-cast ingots were cold-rolled by 90\% reduction in thickness and subsequently homogenized at 1100°C for 24 h in a vacuum atmosphere ($\sim 10^{-3}$ Pa) to remove segregation of elements. Afterward, disc-shaped specimens with a diameter of 10 mm and a thickness of 0.80 mm for subsequent heavy deformation by the high-pressure torsion (HPT) process$^{11}$ were cut from the homogenized materials. The HPT was performed under a compressive pressure of 6.0 GPa at a rotation speed of 0.20 rpm at room temperature to introduce lattice defects of high density homogeneously. A total shear strain of 196 in maximum was applied to the materials after 5 rotations of HPT. Since the thickness of our tensile specimens was 0.5 mm (as shown in the next section), the mean grain size of our materials should be less than 25 μm to obtain reliable stress-strain behavior avoiding the size effect on mechanical properties.$^{12,13}$ So that, subsequently, the HPT processed specimens were annealed at 850°C for 2.1–7.2 ks to obtain fully-recrystallized microstructure with a mean grain size of 8 μm (excluding twin boundaries).

2.2 Tensile tests

Tensile tests were performed at room temperature by using a universal tensile test machine (Shimadzu, AG-100 kn Xplus) with an initial strain rate of $8.3 \times 10^{-4}$ s$^{-1}$ to characterize mechanical properties of the materials. Small-sized tensile specimens with a gauge length of 2.0 mm and a gauge cross-section of 1.0 mm $\times$ 0.50 mm were cut from the annealed HPT specimens with a disc-shaped so that the center of the gauge part coincided with the region at a radial distance of 3.0 mm from the center of the discs. Surfaces of all the tensile specimens were painted with white and black inks to produce a random dot pattern (speckle pattern) for the digital image correlation (DIC) analysis.$^{13}$ The DIC technique was employed to precisely measure the tensile strain in the gauge part of the tensile specimens during deformation by tracking the speckle patterns via a CCD camera. The collected snapshots of the speckle patterns were analyzed by using Vic-2D software.$^{13}$ We have confirmed in the previous studies (Ref. 5) for example) that the stress-strain curves obtained by using the small-sized tensile specimens and DIC technique were equivalent to those obtained from large standard-size specimens of the same materials tested with a strain gauge. To investigate deformation microstructures of the materials, tensile tests were interrupted at different strain levels, and microstructures were observed using microscopy techniques described in the following section 2.3.

2.3 Microstructure observation

2.3.1 Scanning electron microscopy

Microstructure observation was performed in a field emission scanning electron microscope (FE-SEM) (JEOL, JSM-7100F and 7800F) equipped with backscattered electron (BSE) and electron backscattered diffraction (EBSD) detectors. Specimens for the observations were mechanically polished by using 1000–4000 grid sized fine sandpapers first and then electrically polished in a solution of 90 vol\% of ethanol and 10 vol\% of perchloric acid at 30 V for 15 s at room temperature to achieve flat mirror surfaces. The gauge sections of the tensile specimens normal to the disc surface were observed. Electron channeling contrast imaging (ECCI) technique$^{14,15}$ was employed to characterize deformation microstructures including images of dislocations, stacking faults, etc. The working distance was 15 mm for BSE and EBSD, and 3.0 mm for ECCI. The data of EBSD were collected by the software produced by TSL solution (TSL-OIM data collection, analysis ver. 5.31) and analyzed by the OIM analysis software (Tex SEM Laboratories, ver. 7.0). The step size of the EBSD measurement was set to be 50–60 nm, and one large area including more than 150 grains was examined for each condition to obtain statistically reliable data. Grain sizes excluding $\Sigma$3 twin boundaries of the specimens were determined by the line intercept method on SEM-BSE images. In EBSD phase maps, FCC and HCP phases are colored by red and green, respectively, and high-angle grain boundaries (HAGB) and $\Sigma$3 twin boundaries (TB) are shown by black and light-blue lines, respectively.

2.3.2 Transmission electron microscopy

A transmission electron microscope (TEM) (JEOL, JEM-2010) operated at 200 kV was used to observe deformation microstructures. To prepare thin-foils for observations, deformed tensile specimens were mechanically polished down to the thickness of 0.1 mm first. Afterward, rectangular-shaped sheets having a gauge part dimension of 1.0 mm (width) $\times$ 3.0 mm (length) $\times$ 0.1 mm (thickness) were cut from the specimens. Finally, the rectangular-shaped sheets were further thinned by twin-jet electropolishing system (Struers, Tenupol-5) with a solution of 70 vol\% of methanol, 20 vol\% of glycerine, and 10 vol\% perchloric acid at an applied voltage of 10 V and temperature of 243 K.

3. Results and Discussions

3.1 Initial microstructure

Figure 1 shows SEM-BSE microstructures of (a) Co20, (b) Co40, and (c) Co60 alloys after HPT and annealing at 850°C for 7.2 ks, 3.6 ks, and 2.1 ks, respectively. Mean grain sizes were $d = 7.36 \mu$m, 8.92 μm and 8.34 μm (excluding twin boundaries) in Co20, Co40, and Co60, respectively, which were very close to the designated grain size ($d = 8 \mu$m). A large number of annealing twins were observed in all the materials, and particularly the density of annealing twin in Co60 was apparently the largest among the MEAs. Mean grain sizes including twin boundaries were $d_m = 2.48 \mu$m, 2.50 μm and 1.14 μm in Co20, Co40, and Co60, respectively. As discussed later, it was confirmed by the EBSD analysis that all the alloys had FCC single-phase structures at this initial stage.
3.2 Mechanical properties

Figure 2(a) shows nominal stress–nominal strain curves of the Co20, Co40, and Co60 alloys. Yield strength (σY), ultimate tensile strength (σUTS), uniform elongation (εu) and total elongation (εtot) obtained from the stress-strain curves were summarized in Table 1. It was found that the yield strength of Co20 (373 MPa) and Co40 (364 MPa) were similar to each other, while that of Co60 (441 MPa) was somewhat higher than other two alloys.

In polycrystalline materials with a single-phase structure, grain size (including twin boundaries) dependence of yield strength follows the well-known Hall-Petch relationship,\(^{16,17}\)

\[ \sigma_Y = \sigma_0 + k d^{-1/2} \]  

where \( k \) is a constant so-called Hall-Petch slope. Since the density of annealing twins in Co60 was higher than others as mentioned above, the higher yield strength of Co60 could be mainly attributed to the narrower spacing (smaller \( d_{tw} \) as mentioned in 3.1) of annealing twin boundaries.

Figure 2(b) shows true stress–true strain curves together with work-hardening rate–true strain curves of Co20 (red line), Co40 (black line), and Co60 (blue line) tensile tested at room temperature.

\[ \frac{d\sigma}{d\varepsilon} \leq \sigma \]  

where \( \sigma \) is true stress and \( \varepsilon \) is a true strain. According to the eq. (2), the crossing point of true stress–true strain curves and work-hardening rate–true strain curves agrees with the plastic instability point of materials. It can be clearly seen from Fig. 2(b) that Co60 showed early fracture before satisfying the plastic instability condition, for which the details will be discussed in section 3.3.3. On the other hand, Co20 and Co40 reached to the plastic instability points, which agreed with the existence of post-uniform elongation in the nominal stress-strain curves shown in Fig. 2(a). It was also found from Fig. 2(b) that the work-hardening rate of Co40 was higher than that of Co20 in the entire range of tensile deformation before macroscopic necking occurred, and Co20 showed larger post-uniform elongation than Co40. On the other hand, the slope of the work-hardening rate–true strain curve of Co60 quickly increased at a true strain level of \(~0.05\) although the absolute value of work-hardening rate of Co60 monotonously decreased and was smaller than that of Co40.
3.3 Deformation microstructure and deformation mechanism

To investigate the microstructural origin of the deformation behaviors discussed in the previous section, we observed deformation microstructures of the tensile-deformed specimens.

3.3.1 Deformation microstructure of Co20

Figure 3 shows deformation microstructures of Co20 tensile-deformed to different tensile strains (ε) observed by the ECCI technique, and Fig. 4 shows EBSD phase maps of the tensile-deformed Co20. It was confirmed from Fig. 4 that Co20 showed FCC single-phase structures in the initial state and also during deformation, except that the fractured specimen (ε = 68.5%; Fig. 4(e)) showed a very small amount of deformation-induced ε-martensite with HCP crystal structure (fε = 0.4%). Just after yielding (ε = 0.9%) (Fig. 3(a)), dislocations piled-up at grain boundaries were observed. At a tensile strain of ε = 16.1% (Fig. 3(b1) and (b2)), planar arrays of dislocations confined in specific slip planes could be seen. At a tensile strain of ε = 21.2% (Fig. 3(c1)), it was found that the dislocation density increased, and multiple slip systems were operated. Furthermore, regions with plate-shaped defects were confirmed in some grains, as shown in Fig. 3(c2). Based on the boundary character analyzed by EBSD, the plate-shaped regions were considered to be deformation twins, as pointed by a yellow arrow in Fig. 4(d). It can be concluded that plastic deformation in Co20 was brought by dislocation slips and deformation twinning.

Planar morphology of dislocation microstructure can be more effective to increase the work-hardening rate than wavy morphology. This is because Lomer-Cottrel immobile dislocations can form at intersects of slip planes with a planar slip of dislocations and become obstacles for dislocation motion.\(^1\) Thus, mean free path of dislocations can decrease, and higher work-hardening rate can be achieved. In addition, when SFE (of \{111\} planes) in FCC materials decreases, it has been generally believed that deformation twins come to be formed easily,\(^2\) and the formation of deformation twins would increase the work-hardening rate of materials. This is the so-called dynamic Hall-Petch effect in TWIP, which can be formulated as

<table>
<thead>
<tr>
<th>Alloys</th>
<th>Yield stress (σ_Y) (MPa)</th>
<th>Tensile strength (σ_{UTS}) (MPa)</th>
<th>Uniform elongation (ε_u)</th>
<th>Total elongation (ε_{tot})</th>
</tr>
</thead>
<tbody>
<tr>
<td>Co20</td>
<td>373</td>
<td>680</td>
<td>0.51</td>
<td>0.69</td>
</tr>
<tr>
<td>Co40</td>
<td>364</td>
<td>750</td>
<td>0.49</td>
<td>0.53</td>
</tr>
<tr>
<td>Co60</td>
<td>441</td>
<td>666</td>
<td>0.16</td>
<td>0.16</td>
</tr>
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Table 1 Yield strength, tensile strength, uniform elongation, and total elongation of three kinds of the present alloys obtained by tensile tests at room temperature.

Fig. 3 SEM-ECCI images showing deformation microstructures of Co20 at different tensile strains (ε): (a) ε = 0.9%, (b1 and 2) ε = 16.1%, and (c1 and 2) ε = 21.2%. Loading direction in the tensile tests is indicated by an arrow.
At a strain level of 0.8%, as well as a planar slip of dislocations shown in Fig. 5(e), we confirm that the amount of deformation twins in Co40 was apparently larger than that in Co20, suggesting that the SFE of Co40 is lower than that of Co20. Figure 6 shows EBSD phase maps of the deformed Co60. At the initial state, the alloy had an FCC single-phase structure, while the existence of ε-martensite with island-like morphology was evident after a tensile strain of e = 38.2% (Fig. 6(d)). The fraction of ε-martensite increased up to fε = 11.1% after fracture (e = 53.4%; Fig. 6(e)). It was concluded from the results that plastic deformation of Co40 was brought by dislocation slips, deformation twinning, and deformation-induced (ε-) martensitic transformation.

Similar to Co20, dislocation hardening and deformation twinning (dynamic Hall-Petch effect) should contribute to the enhanced work-hardening in Co40. In addition, the widely-extended stacking faults could also act as obstacles for dislocation glides leading to hardening. Most importantly, ε-martensite (HCP phase) could have an impact on the work-hardening behavior in Co40 because FCC-HCP interface would also act as obstacles for dislocation motion, and possible slip systems in HCP structure are limited, which would result in the TRIP effect.

3.3.3 Deformation microstructure of C60

Figure 7 shows deformation microstructures of Co60 observed by the ECCI technique. Similar to Co20, dislocations piled up at grain boundaries were seen after yielding (e = 0.8%; Fig. 5(a)). After deformation of e = 11.9%, widely-extended stacking faults could be clearly observed (Fig. 5(b2)), as well as a planar slip of dislocations (Fig. 5(b1)). At a strain level of e = 38.2%, a planar slip of dislocations in multiple slip systems was observed (Fig. 5(c1)). Furthermore, plate-shaped lattice defects with a spacing of several tens nm were found. The plate-shaped defects were too thin to resolve in the EBSD analysis. In order to clarify the details of the plate-shaped lattice defects, we also made observations by TEM. Figure 5(d) shows a bright-field TEM image of a deformed microstructure clearly including plate-shaped defects. Based on the selective area diffraction (SAD) pattern with a zone axis (ZA) of [0 1 1] shown in Fig. 5(e), we confirmed that the plate-shaped defects were mainly deformation twins. It should be noted that the number of deformation twins in Co40 was apparently larger than that in Co20, suggesting that the SFE of Co40 is lower than that of Co20. Figure 6 shows EBSD phase maps of the deformed Co40. In the initial state, the alloy

\[
\sigma_f = \sigma_0 + K_{TB} \cdot \frac{1}{\lambda_{TB}}
\]

where \(\sigma_f\) is the flow stress, \(K_{TB}\) is a constant, and \(\lambda_{TB}\) is the mean spacing of twin boundaries. Since deformation twinning was observed in Co20, we consider that in addition to dislocation hardening (in the early stage of deformation), deformation twinning contributed to the strain-hardening (in the later stage of deformation), leading to the high strength and large ductility of this alloy. It has been also reported that deformation twinning in FCC HEAs and MEAs was effective to retard fracture, which might be related to the larger post-uniform elongation in Co20.

3.3.2 Deformation microstructure of C40

Figure 5(a–c2) show deformation microstructures of Co40 observed by the ECCI technique. Similar to Co20, dislocations piled up at grain boundaries were seen after yielding (e = 0.8%; Fig. 5(a)). After deformation of e = 11.9%, widely-extended stacking faults could be clearly observed (Fig. 5(b2)), as well as a planar slip of dislocations (Fig. 5(b1)). At a strain level of e = 38.2%, a planar slip of dislocations in multiple slip systems was observed (Fig. 5(c1)). Furthermore, plate-shaped lattice defects with a spacing of several tens nm were found. The plate-shaped defects were too thin to resolve in the EBSD analysis. In order to clarify the details of the plate-shaped lattice defects, we also made observations by TEM. Figure 5(d) shows a bright-field TEM image of a deformed microstructure clearly including plate-shaped defects. Based on the selective area diffraction (SAD) pattern with a zone axis (ZA) of [0 1 1] shown in Fig. 5(e), we confirmed that the plate-shaped defects were mainly deformation twins. It should be noted that the number of deformation twins in Co40 was apparently larger than that in Co20, suggesting that the SFE of Co40 is lower than that of Co20. Figure 6 shows EBSD phase maps of the deformed Co40. In the initial state, the alloy
Fig. 5 (a–c2) SEM-ECCI images showing deformation microstructures of Co40 at different tensile strains (ε): (a) ε = 0.8%, (b1 and 2) ε = 11.9%, and (c1 and 2) ε = 38.2%. (d) Bright-field (BF) TEM micrograph of Co40 tensile deformed to a strain of ε = 38.2% and (e) corresponding selected area diffraction (SAD) pattern along zone-axis (ZA) of [0 1 1]. Diffraction spot corresponding to matrix and twin are indexed as (h k l) and (h k l'), respectively. Loading direction in the tensile tests is indicated by an arrow.

Fig. 6 EBSD phase maps of deformed Co40 at different strains (ε): (a) ε = 0.0%, (b) ε = 0.8%, (c) ε = 11.9%, (d) ε = 38.2%, and (e) ε = 53.4%. FCC and HCP phases are colored by red and green, respectively. χ is also indicated in the figures. HAGB and Σ3 TB are shown by black and light-blue lines, respectively. Loading direction in the tensile tests is indicated by an arrow.
strain level was corresponding to the onset of the work-hardening rate plateau (after $e = 5\%$) in Co60 shown in Fig. 2(b). Since the plate-shaped martensite blocks dislocation slips at FCC/HCP interfaces and effectively reduces the mean free path of dislocation slips, work-hardening could be accelerated. After fracture, the fraction of $\varepsilon$-martensite was quickly increased to $f_\varepsilon = 43.2\%$ (Fig. 8(d)). Above all, the operating deformation mechanisms in Co60 seemed to be similar to those in Co40, but the fraction of $\varepsilon$-martensite was significantly larger than that in Co20 and Co40.

Figure 9 shows a fracture surface of Co60. It looked like an intergranular fracture surface, but many steps existed on the facets to form so-called step-like-ridge pattern. Such a fracture surface morphology has been reported in some high-Mn austenitic steel exhibiting deformation-induced $\varepsilon$-martensitic transformation. The formation mechanism of this morphology is believed to be related to stress concentration at intersections between grain boundaries and $\varepsilon$-martensite plates. Since Co60 exhibited deformation-induced $\varepsilon$-martensitic transformation significantly, it would be reasonable to consider that the same fracture mechanism was operated in Co60.

### 3.4 Effect of Co on the deformation mechanisms

Based on the present results, we found that with increasing Co-content in Co$_x$(CrNi)$_{\{100-x\}}$ MEAs, the number of deformation twins increased (Co20 $\rightarrow$ Co40). By increasing Co-content further, the deformation mechanism was changed from deformation twinning to deformation-induced martensitic transformation, and the phase fraction of $\varepsilon$-martensite increased (Co40 $\rightarrow$ Co60). Figure 10 shows the relationship between Co-content and SFE on $\{111\}$ plane of FCC structure in Co$_x$(CrNi)$_{\{100-x\}}$ MEAs, extracted from Refs. 25, 26. It was found that the SFE decreased with increasing the Co-content. Generally, it is said in FCC metals that, as SFE decreases, deformation mechanism can change from dislocation slip (High and medium SFE) to deformation twinning (Low SFE) and deformation-induced martensitic transformation (Very low SFE).

This is very consistent
4. Conclusion

In the current study, we fabricated Co₆₀(CrNi)₁₀₀₋ₓ (x = 20 (Co20), 40 (Co40), 60 (Co60)) MEAs, and investigated their mechanical behaviors by carefully conducting tensile tests at room temperature and observations of deformation microstructures by SEM-ECCI, EBSD, and TEM. The main finding of the present study are as follows:

1. In Co20, plastic deformation was brought by dislocation slips and deformation twinning. Planar morphology of dislocation microstructure could be effective for forming Lomer-Cottrell immobile dislocation, which could be obstacles for dislocation motion. Due to dynamic Hall-Petch effect (TWIP effect) as well as dislocation hardening, Co20 showed high strength (σ_{UTS} = 680 MPa) and large ductility (ε_{tot} = 69%).

2. In Co40, plastic deformation proceeded by dislocation slip and deformation twinning, similar to Co20. Widely-extended stacking faults could act as obstacles for dislocation motion, resulting in hardening. Moreover, deformation-induced ε-martensite with island-like morphology with an area fraction of 11.1% was observed after a tensile strain of ε = 53.4%. TRIP effect, as well as TWIP effect, seemed responsible for the highest tensile strength (σ_{UTS} = 750 MPa) and the highest work-hardening rate among the alloys studied in the current work.

3. In Co60, plastic deformation was brought by dislocation slips, deformation twinning, and deformation-induced ε-martensitic transformation. The phase fraction of ε-martensite was 43.2% after a tensile strain of 16.1%, which is significantly larger than Co20 and Co40. Morphology of ε-martensite was plate-shaped, which was different from that in Co40. Large phase fraction of ε-martensite was responsible for higher work-hardening rate, resulting in a plateau of work-hardening rate after tensile strain of about 5%. Also, it might cause stress concentration at the intersection of grain boundaries and martensite leading to early fracture (ε_{tot} = 16%) with step-like-ridge fracture surface morphology.

4. With increasing Co-content in Coₓ(CrNi)₁₀₀₋ₓ MEAs, the SFE decreased, and deformation mechanism changed from dislocation slip (High and medium SFE) to deformation twinning (Low SFE) and deformation-induced martensitic transformation (Very low SFE). This suggests that mechanical properties (deformation and fracture behavior) of the present alloys can be effectively controlled by changing the SFE.

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with the present results and previous studies of equiatomic CoCrNi MEA.²⁶,²⁹)

Figure 11 shows the relationship between nominal strain and phase fraction of HCP phase (ε-martensite) in tensile deformation of Co20, Co40, and Co60.
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