

Paragraph Ran in the Queries

Paper Title: Achieving a remarkable strength–ductility combination in a novel casting AlCoCrNi high entropy alloy

Content :

Results and discussion

Fig. 1a shows a synchrotron high-energy X-ray diffraction (SHE-XRD) pattern, revealing that the as-cast AlCoCrNi alloy has an L₁₂ (ordered face-centered cubic) and B₂ (ordered body-centered cubic) dual-phase microstructure. The electron backscatter diffraction (EBSD) grain orientation map and phase map of as-cast AlCoCrNi alloy are exhibited in Fig. 1b and c, respectively. Fig. 1b shows that different B₂ phase growth directions were observed in the adjacent coarse L₁₂ grains with different orientations. The results of the phase map reveal that the corresponding content of L₁₂ and B₂ phases in AlCoCrNi alloy is about 92 vol% and 8 vol%, as seen in Fig. 1c. According to the high-angle annular dark-field scanning transmission electron microscope (HAADF-STEM) image and selected area electron diffraction patterns (SADPs) in Figs. S1(a–b), the dark and light phases are L₁₂ and B₂ phases, respectively. This diffraction is also confirmed by SHE-XRD results. EDS maps and SADPs (Figs. S1(c) and (b)) reveal that the L₁₂ phases enriched in Co and Cr while the B₂ phases enriched in Al and Ni but depleted in Co and Cr. The chemical compositions of as-cast AlCoCrNi alloy are shown in Table 1. To better understand the microstructural features of as-cast AlCoCrNi alloy, we further conducted high-magnification TEM observations to investigate L₁₂ and B₂ phases. Fig. S1(d) and Fig. S1(e) indicate that substantial nano-particles inside the B₂ phases while no precipitates or other phases in the L₁₂ phases. The average size of the nano-particles is ~10.26 nm (Fig. S1(f)) according to analyzing the magnified HAADF-STEM image. EDS maps reveal that the nano-particle is only enriched in Cr (Fig. S1(g)). Additionally, the presence of Cr-riched nanoparticles within the B₂ phases was also detected in the AlCoCrFeNi_{2.1} [22], AlCoCrCuFeNi [23] and Al_{0.7}CoCrFeNi [24] HEAs. These Cr-riched nanoparticles are mainly caused by the spinodal decomposition of component modulation [22].

displays the tensile property of the as-cast AlCoCrNi alloy tested at room temperature. Our as-cast AlCoCrNi alloy shows a better strength–ductility balance (tensile strength of ~1025 MPa and a large uniform elongation of ~30 %) than typical as-cast AlCoCrNi_{2.1} eutectic high entropy alloy (EHEA) [25] and as-cast Fe₂₀Co₂₀Ni₄₁Al₁₉ EHEA [13] (Fig. 2a). Especially, the uniform elongation of our HEA is nearly twice that of two typical EHEAs (~16 %) under the similar tensile strength. The inset of Fig. 2a gives the strain-hardening rate of our as-cast AlCoCrNi HEA. The work-hardening rates of two

typical as-cast EHEAs decrease gradually with the increase of true strain until they reach zero. Moreover, no obvious work hardening platform was detected during the tensile deformation of two typical as-cast EHEAs (Fig. 2a, inset). Compared with two typical as-cast EHEAs, our HEA shows a more significant and persistent strain-hardening ability. In particular, Our HEA exhibits an abnormal strain hardening platform at a true strain of about 2.5 %–15 % (Fig. 2a, inset). For more clarity, as shown in Fig. 2b, a direct comparison with many as-cast HEAs include arc-melting and directly cast HEAs [12,18,22,[26], [27], [28], [29], [30], [31], [32], [33], [34], [35], [36]] is plotted in ultimate tensile strength versus uniform elongation. Apparently, our HEA presents a remarkable strength-ductility combination, when compared with other high-performance as-cast HEAs.

To unravel the mechanisms responsible for the marked improvement in the strain-hardening of our as-cast AlCoCrNi HEA, we studied its deformation substructures using interrupted SHE-XRD patterns, EBSD phase maps and kernel average misorientation (KAM) maps at different tensile strains. The XRD patterns at different interrupted strains are shown in Fig. 3a. With increasing true strain, the diffraction peaks of B2 phases shift to weaken. When the true strain approached ~15 %, a new diffraction peak around 3.3° (indicated by the red arrow) beside the B2 peak appears, indicating the occurrence of an obvious phase transition during the deformation. With further true strain increase, the new peaks intensity increases appreciably, and the initial B2 peaks entirely disappear at ~25 % true strain, suggesting the transition completion of the B2 phase. The new phase peaks can be well indexed into the BCT phase that was detected in Al_{0.6}CoCrFeNi HEA during compression at room temperature [37]. Therefore, we observed an obvious B2→BCT phase transformation in our as-cast AlCoCrNi HEA during the deformation. The EBSD results show that only L1₂ and B2 phases were observed in the as-cast AlCoCrNi HEA, and low KAM values in the 0 % strain state, were observed (Fig. 3b1). As the strain increased to 2.5 %, phase transformation was activated in several B2 phases (Fig. 3b2). As the strain increased to 15 %, only a very small amount of B2 phases was observed according to Fig. 3b3, indicating that most B2 phases have transformed into BCT structures at this strain stage. As the strain increased to 25 %, we detected that only L1₂ and BCT phases were in the fractured microstructure of our as-cast AlCoCrNi HEA, and also most BCT phases were stretched along the loading direction (Fig. 3b4). In addition, the average KAM value, as an obvious characteristic of the local misorientation, increases with increasing tensile strains. Generally, the average KAM value is positively correlated with the density of geometrically necessary dislocations (GNDs) in the deformed HEAs [38]. It appears that the local misorientation forms preferentially close to phase boundaries and becomes more intense as the strain increases, indicating the GNDs piled up and heterogeneously distributed in our deformed HEAs.

To further understand the deformation mechanisms and reveal the mechanical responses of our as-cast AlCoCrNi alloy, TEM observations were conducted to investigate the deformed substructures under different strains. The substructures at true strains about 2.5 % and 15 % were examined by TEM observations, corresponding to the monotonically increased strain hardening behavior. Fig. 4a1 shows a representative bright-field TEM image of the tensile tested L₁₂ phases. The dislocation density (within the L₁₂ phases) noticeably increases relative to the as-cast state. However, no visible defects can be detected in the B2 phases at the early plastic stage (Fig. 4a2). It is reasonable to attribute this deformation behavior to the B2 phase has a significantly higher hardness than the L₁₂ phase. This scenario agrees with that a larger critical stress is needed to operate the plastic deformation of the hard B2 phase than that of the L₁₂ phase [39]. More importantly, phase transformation (B2→BCT) in the B2 phase was detected at this strain stage, which is consistent with the EBSD results above. The square-arranged spots before deformation change to parallelogram-arranged spots following deformation according to the selected area electron diffraction (SAED) patterns of the B2 phases with the [001] zone axis (Fig. 4a3), revealing the different interplanar spacings between the (200) and (020) planes. Calculation results display the crystal structure decreases along the (200) direction while expands along the (002) direction following deformation. Additionally, twin martensite was observed in the B2 phase. Its dark-field image and SAED pattern are shown in Fig. 4a3. It is generally acknowledged that such phase transformation (B2→BCT) during the deformation may be caused by the nucleation and spread of 110 twinning planes [40].