



Mechanical and corrosion properties of lightweight $(\text{Ti}_{60}\text{Cr}_{30}\text{Nb}_{10})_{100-x}\text{Al}_x$ medium-entropy alloys

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Abstract: Microstructure, mechanical properties and corrosion behavior of lightweight $(\text{Ti}_{60}\text{Cr}_{30}\text{Nb}_{10})_{100-x}\text{Al}_x$ ($x=0, 5, 7.5$ and 10 , at.%) medium-entropy alloys (MEAs) were investigated in the as-cast and homogenized states. As the Al content increases from 0 to 10 at.%, all the as-cast MEAs consist of identical phase composition, i.e., a body-centered cubic (BCC) phase and a small amount of TiCr_2 Laves phase. After the homogenization, all the homogenized MEAs show a single BCC phase. The $(\text{Ti}_{60}\text{Cr}_{30}\text{Nb}_{10})_{95}\text{Al}_5$ MEA shows the most excellent mechanical properties both in as-cast and homogenized states among the studied alloys, displaying yield strengths of ~ 1048 MPa and ~ 1017 MPa along with the compression strain exceeding 50%. Additionally, compared with the TC4 and the as-cast $(\text{Ti}_{60}\text{Cr}_{30}\text{Nb}_{10})_{95}\text{Al}_5$ alloys, the homogenized $(\text{Ti}_{60}\text{Cr}_{30}\text{Nb}_{10})_{95}\text{Al}_5$ alloy shows superior corrosion resistance due to its uniform microstructure and chemical composition.

Key words: medium-entropy alloy; lightweight alloy; microstructure; mechanical properties; corrosion behavior

1 Introduction

The performance of conventional alloys is usually determined by their one or two major constituent elements, and thereby multi properties can be hardly achieved in these alloys. Recently, a new design concept of multi-principal element alloys (MPEAs), also called high-entropy alloys (HEAs) and medium-entropy alloys (MEAs), has been proposed by YEH et al [1] and CANTOR et al [2], which is expected to break through the multi-property bottleneck of the existing alloys. Such newly-developed alloys were initially defined as the alloys containing at least five principal elements with the configurational entropy greater

than or equal to $1.61R$ (R is molar gas constant). Owing to the high-entropy effect, these alloys always show simple solid-solution phases, such as face-centered cubic (FCC) phases, body-centered cubic (BCC) phases and/or close-packed hexagonal (HCP) phases [3–5]. These unique phase structure and composition give rise to superior properties of HEAs/MEAs, including high strength [6–8], good ductility [6–8], excellent corrosion resistance [9,10] and high fracture toughness [11].

Owing to the promising performance of these HEAs/MEAs, numerous researches have been performed on the high strength and ductility along with good corrosion resistance for the demands of structural and functional materials. The so-called refractory high-entropy alloys (RHEAs) consisting

of high-melting-point elements have attracted considerable attention in recent years. The pioneering RHEAs, MoNbTaW and MoNbTaVW, proposed by SENKOV et al [12] exhibit exceptionally high-temperature strength and hardness, which were considered as the promising next-generation of high-temperature materials. And the corrosion behavior of these two alloys was also investigated by BACHANI et al [13], suggesting that their corrosion resistance is comparable to that of SS304 in 0.5 mol/L H_2SO_4 solution. In addition, some RHEAs have also been proven to have excellent corrosion resistance. JAYARAJ et al [14] suggested that TaNbHfZrTi RHEA had high corrosion resistance in 11.5 mol/L HNO_3 solution. LI et al [15] pointed out that $\text{TiZr}_{0.5}\text{NbCr}_{0.5}\text{V}_x\text{Mo}_y$ RHEAs displayed excellent corrosion resistance no matter in NaCl solution or in H_2SO_4 solution. However, it is noted that the existing RHEAs suffer from their high density and disappointing ductility, and therefore their engineering applications have been greatly impeded. Hence, it is of great significance to develop RHEAs with low density, high strength and high plasticity, as well as outstanding corrosion resistance.

To address the problem of high density for the present RHEAs, considerable attentions have been paid to design light-weight HEAs/MEAs based on Al, Cr, Nb, Ti, V and Zr elements [16–19]. These light-weight HEAs/MEAs can exhibit excellent high-temperature strength. YURCHENKO et al [16] produced a series of equiatomic Al–Cr–Nb–Ti–V–Zr moderate-temperature HEAs (MT-HEAs) with high yield strength and low density: such as AlNbTiV (~1280 MPa and ~5.49 g/cm³), AlNbTiVZr (~1675 MPa and ~5.76 g/cm³), and AlCrNbTiV (~1570 MPa and ~5.82 g/cm³). SENKOV et al [17] reported that the low-density NbTiVZr (~6.52 g/cm³), CrNbTiZr (~6.67 g/cm³) and CrNbTiVZr (~6.57 g/cm³) exhibited compressive yield strength over 800 MPa at 873 K. In spite of high strength and low density, these equiatomic HEAs/MEAs still have limited room-temperature ductility. Recently, non-equiatomic HEAs/MEAs have been designed to overcome room-temperature strength–ductility trade-off. LIAO et al [20] produced novel lightweight $\text{Ti}_x(\text{AlCrNb})_{100-x}$ MEAs with high fracture strength (>1000 MPa) and high failure strain (>30%). QIAO et al [21] investigated the alloying effect of Nb element on mechanical

properties of low-density $\text{Ti}_2\text{ZrHf}_{0.5}\text{VNB}_x$ alloys, revealing that with the increasing of Nb content, the yield strength decreased from 1160 to 980 MPa without fracture. Unfortunately, the corrosion behavior of these non-equiatomic light-weight HEAs/MEAs remains poorly understood.

In this work, the low-density Al element was introduced into the Ti–Cr–Nb alloy system in the aim of reducing the density and increasing the effect of solid-solution strengthening. Therefore, a series of Ti-rich lightweight $(\text{Ti}_{60}\text{Cr}_{30}\text{Nb}_{10})_{100-x}\text{Al}_x$ ($x=0, 5, 7.5$ and 10 , at.%) MEAs were designed and prepared. Microstructure, mechanical properties and electrochemical performance of the as-cast and homogenized specimens were systematically investigated. The phase formation mechanisms, strengthening mechanisms and corrosion behavior of the studied MEAs were also discussed.

2 Experimental

Ingots of the as-cast $(\text{Ti}_{60}\text{Cr}_{30}\text{Nb}_{10})_{100-x}\text{Al}_x$ alloys ($x=0, 5, 7.5$ and 10 , at.%, thereafter named as #AC-Al_0, #AC-Al_5, #AC-Al_7.5 and #AC-Al_10, respectively) were prepared by vacuum arc-melting using the high-purity element particles (purity ≥ 99.95 wt.%) in a water-cooled copper cavity under high-purity argon atmosphere protection. Each ingot was remelted at least eight times to ensure chemical homogeneity, and then specimens with a geometry of 60 mm (length) $\times$ 25 mm (width) $\times$ 12 mm (thickness) were obtained. The compositions of the alloys measured by EDS/SEM are listed in Table 1, which are close to their nominal compositions. Specimens with a geometry of 40 mm $\times$ 15 mm $\times$ 12 mm were cut from the ingots using an electric discharge machine, and then homogenized in a quartz tube filled with high-purity argon atmosphere at 1200 °C for 12 h (homogenized specimens are thereafter named as #HT-Al_0, #HT-Al_5, #HT-Al_7.5 and #HT-Al_10, respectively).

The phase constitution and microstructure of the as-cast and homogenized alloys were studied using X-ray diffraction (XRD), scanning electron microscope (SEM) and transmission electron microscope (TEM). Specimens for XRD analysis were ground using SiC papers up to 2000 grit and then characterized by Panalytical X'Pert by using Cu K_α radiation at 2θ of 20°–90° and a speed of

Table 1 Chemical compositions and densities of as-cast $(\text{Ti}_{60}\text{Cr}_{30}\text{Nb}_{10})_{100-x}\text{Al}_x$ MEAs

Alloy	Nominal composition/at. %	Measured composition/at. %	Calculated density/ ($\text{g}\cdot\text{cm}^{-3}$)	Measured density/ ($\text{g}\cdot\text{cm}^{-3}$)
#AC-Al_0	$\text{Ti}_{60}\text{Cr}_{30}\text{Nb}_{10}$	$\text{Ti}_{61.3}\text{Cr}_{27.2}\text{Nb}_{11.5}$	5.72	5.66
#AC-Al_5	$\text{Ti}_{57}\text{Cr}_{28.5}\text{Nb}_{9.5}\text{Al}_5$	$\text{Ti}_{57}\text{Cr}_{25.2}\text{Nb}_{11.3}\text{Al}_{6.5}$	5.57	5.46
#AC-Al_7.5	$\text{Ti}_{55.5}\text{Cr}_{27.75}\text{Nb}_{9.25}\text{Al}_{7.5}$	$\text{Ti}_{55.5}\text{Cr}_{24.77}\text{Nb}_{10.54}\text{Al}_{9.19}$	5.49	5.41
#AC-Al_10	$\text{Ti}_{54}\text{Cr}_{27}\text{Nb}_9\text{Al}_{10}$	$\text{Ti}_{53.5}\text{Cr}_{23.1}\text{Nb}_{12.2}\text{Al}_{11.2}$	5.42	5.34

5 (°)/min. SEM investigations were carried out using a NOVA NANOSEM 430 SEM equipped with an energy dispersive spectrometry (EDS) detector and a backscattered electron (BSE) detector. Prior to testing, specimens were prepared by standard metallographic polishing and then etched with a mixed solution of $\text{HF}:\text{HNO}_3:\text{H}_2\text{O}=1:3:11$ in volume ratio. The specimens were conducted using an FEI Talos F200x TEM equipped with a selected area electron diffraction (SAED).

The densities of the as-cast alloys were measured using Archimedes' principle, and the measured values are listed in Table 1. Vickers' hardness was measured using a digital micro hardness tester (HVS-1000) with a load of 2.94 N and a dwell time of 15 s. Each specimen was tested for 10 times in order to ensure the accuracy of the data. Compression tests were examined using Instron 5500 testing system with an initial strain rate of $1\times 10^{-3}\text{ s}^{-1}$ at ambient temperature. Three cylindrical specimens ($d4\text{ mm}\times 6\text{ mm}$) were used to perform compression tests for each alloy to ensure the accuracy and reproducibility.

Potentiodynamic polarization (PDP) testing and electrochemical impedance spectroscopy (EIS) were conducted using Princeton Applied Research K-0235. The tests were carried out by means of a three-electrode flat-cell, coupled with a saturated calomel electrode (SCE) as the reference electrode, a platinum mesh as the auxiliary electrode, and the specimens as the working electrode. The test solution was 3.5 wt.% NaCl solution, and the measurements were carried out at room temperature. The specimens with a geometry of $d15\text{ mm}\times 3\text{ mm}$ were cut from #AC-Al_5, #HT-Al_5 and TC4 alloys for electrochemical tests. Each specimen was mechanically ground to 3000 grit number by using SiC paper, and then polished with 0.5 μm diamond suspension. Before the electrochemical testing, the specimens were placed in the test solution for 20 min to achieve approximately stable open circuit

potential (OCP). Potentiodynamic polarization tests were performed using a scanning rate of 1 mV/s from an initial potential of -1.2 V (vs SCE) to a final potential of 0.6 V (vs SCE). EIS tests were performed in a frequency range of 10^{-2} – 10^5 Hz with an applied AC amplitude of 10 mV. After the electrochemical testing, the surface corrosion morphologies were characterized by SEM.

The equilibrium step diagrams of #AC-Al_0, #AC-Al_5, #AC-Al_7.5 and #AC-Al_10 alloys were calculated using the Thermo-Calc version 2021b [22] with the TCHEA4 High Entropy Alloy database. The non-equilibrium solidification paths were predicted based on the Scheil–Gulliver model [23,24], which assumes infinitely fast diffusion in the liquid phase and no diffusion in the solid phases. Meanwhile, according to experimental observations, the C14_Laves phase was suspended during thermodynamic calculations.

3 Results

3.1 Phase compositions and microstructures

Figure 1(a) shows the XRD patterns of the as-cast $(\text{Ti}_{60}\text{Cr}_{30}\text{Nb}_{10})_{100-x}\text{Al}_x$ MEAs. Diffraction peaks corresponding to a single BCC phase without any other diffraction peaks can be observed, suggesting that the as-cast $(\text{Ti}_{60}\text{Cr}_{30}\text{Nb}_{10})_{100-x}\text{Al}_x$ MEAs may exhibit a single BCC solid-solution phase. Previous studies have shown that α and Laves phases would be formed in the HEAs/MEAs with the high Cr content [25,26]. Evidently, as the addition of Al content increases, the lattice constants increase from $\sim 3.179\text{ \AA}$ (#AC-Al_0 alloy) to $\sim 3.182\text{ \AA}$ (#AC-Al_10 alloy). Figures 1(b–e) display BSE images of the as-cast $(\text{Ti}_{60}\text{Cr}_{30}\text{Nb}_{10})_{100-x}\text{Al}_x$ MEAs. Dendritic segregations can be clearly observed in the #AC-Al_0 alloys, as shown in Fig. 1(b). With the addition of Al content, the #AC-Al_5, #AC-Al_7.5 and #AC-Al_10 alloys show dendritic segregations similar to the #AC-Al_0

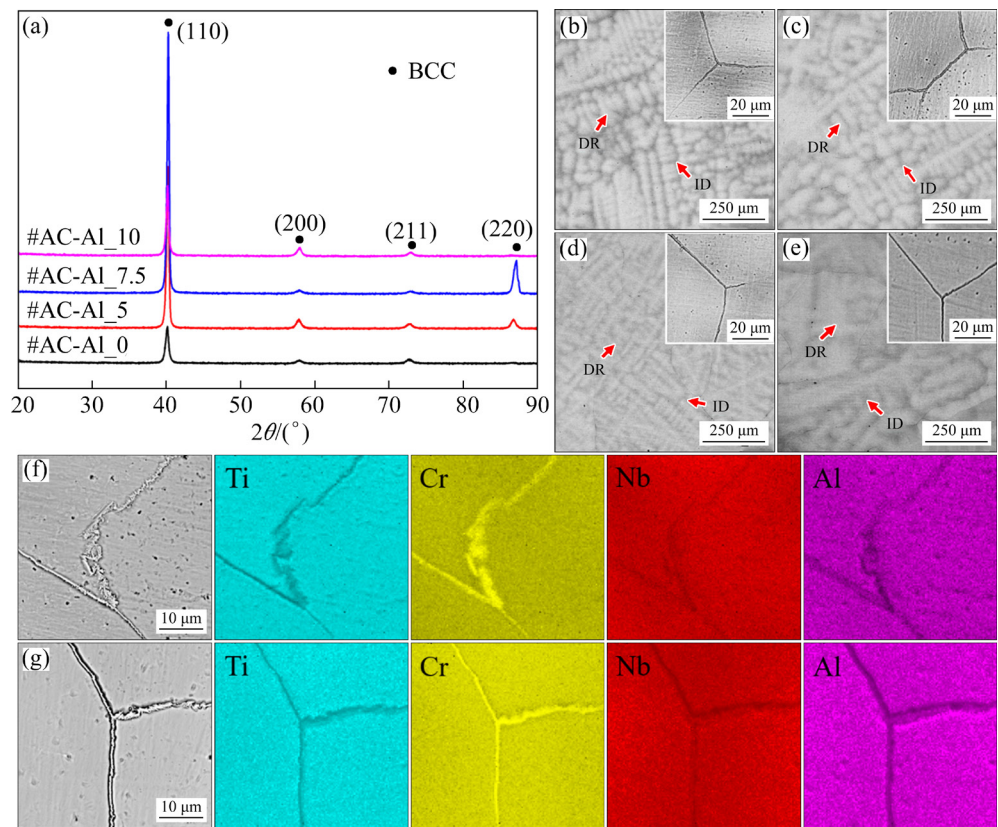


Fig. 1 XRD patterns (a) and BSE-SEM images (b–e) of as-cast $(\text{Ti}_{60}\text{Cr}_{30}\text{Nb}_{10})_{100-x}\text{Al}_x$ MEAs, and EDS mappings of #AC-Al_5 (f) and #AC-Al_10 (g) (DR–Dendrite region; ID–Inter-dendrite region)

alloy, suggesting that the addition of Al has insignificant influence on microstructure. Additionally, the average grain sizes of these four alloys are estimated to be ~ 202 , ~ 269 , ~ 386 and $\sim 353 \mu\text{m}$ using the linear intercept method. According to EDS/SEM results, as listed in Table 2, dendrite (DR) regions with light contrast in the studied alloys are enriched in Nb, due to high melting point of Nb. In contrast, inter-dendrite (ID) regions with dark contrast are enriched in Cr, attributed to the high content of Cr in the studied MEAs. Furthermore, some bright precipitates located at grain boundaries are visible in all these alloys, as shown in the zoom-in images in Figs. 1(b–e). According to the EDS maps of the #AC-Al_5 and #AC-Al_10 alloys (Figs. 1(f, g)), these bright precipitates are distinctly enriched in Cr, which are in line with the results obtained by CHEN et al [26]. And they found that the C15-structured TiCr_2 Laves phase was formed in Ti–Cr alloys with high content of Cr. However, due to the low volume fraction and small size of these precipitates, the TiCr_2 Laves phase in the as-cast $(\text{Ti}_{60}\text{Cr}_{30}\text{Nb}_{10})_{100-x}\text{Al}_x$ MEAs could not be identified by XRD.

Table 2 Chemical compositions of dendrite (DR) and inter-dendrite (ID) regions of as-cast $(\text{Ti}_{60}\text{Cr}_{30}\text{Nb}_{10})_{100-x}\text{Al}_x$ MEAs measured by EDS (at.%)

Alloy	Region	Ti	Cr	Nb	Al
#AC-Al_0	DR	61.25	24.54	14.21	–
	ID	61.45	29.81	8.74	–
#AC-Al_5	DR	55.85	23.09	13.98	7.08
	ID	57.72	27.11	9.47	5.70
#AC-Al_7.5	DR	55.95	22.72	11.72	9.61
	ID	55.15	26.24	9.41	9.20
#AC-Al_10	DR	54.33	21.93	11.20	12.54
	ID	53.69	26.78	8.42	11.11

Figure 2(a) shows the XRD patterns of the homogenized $(\text{Ti}_{60}\text{Cr}_{30}\text{Nb}_{10})_{100-x}\text{Al}_x$ MEAs. After the homogenization at 1200°C for 12 h, all the $(\text{Ti}_{60}\text{Cr}_{30}\text{Nb}_{10})_{100-x}\text{Al}_x$ MEAs exhibit a single BCC phase according to the XRD results. In the meantime, as shown in Figs. 2(b–d), remarkable grain growth can also be observed along with the disappearance of dendritic segregations and Cr-rich precipitates located at grain boundaries, indicating

that these homogenized alloys should be composed of a single BCC phase with homogeneous microstructure, which is in accordance with the XRD results (Fig. 2(a)). The average grain sizes of homogenized MEAs are estimated to be ~623, ~608, ~631 and ~617 μm , respectively. In addition, insert images in Figs. 2(b–d) reveal serrated structure at the grain boundaries in the homogenized alloys, which has also been reported in the nickel-based

superalloys and Ni-rich high entropy alloys [27,28]. It can be inferred that the serrated structure of the homogenized $(\text{Ti}_{60}\text{Cr}_{30}\text{Nb}_{10})_{100-x}\text{Al}_x$ ought to ascribe to the cooling process after homogenization and the pinning effect of precipitates at grain boundaries during dissolution.

TEM images of the #AC-Al_5 and #HT-Al_5 alloys are illustrated in Fig. 3. Figure 3(a) shows a bright-field (BF) TEM image of the #AC-Al_5

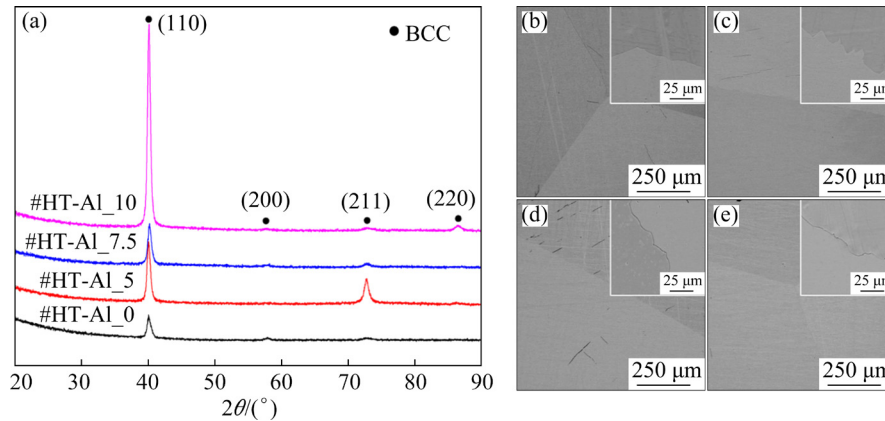


Fig. 2 XRD patterns (a) and BSE-SEM images (b–e) of homogenized $(\text{Ti}_{60}\text{Cr}_{30}\text{Nb}_{10})_{100-x}\text{Al}_x$ MEAs: (b) #HT-Al_0; (c) #HT-Al_5; (d) #HT-Al_7.5; (e) #HT-Al_10

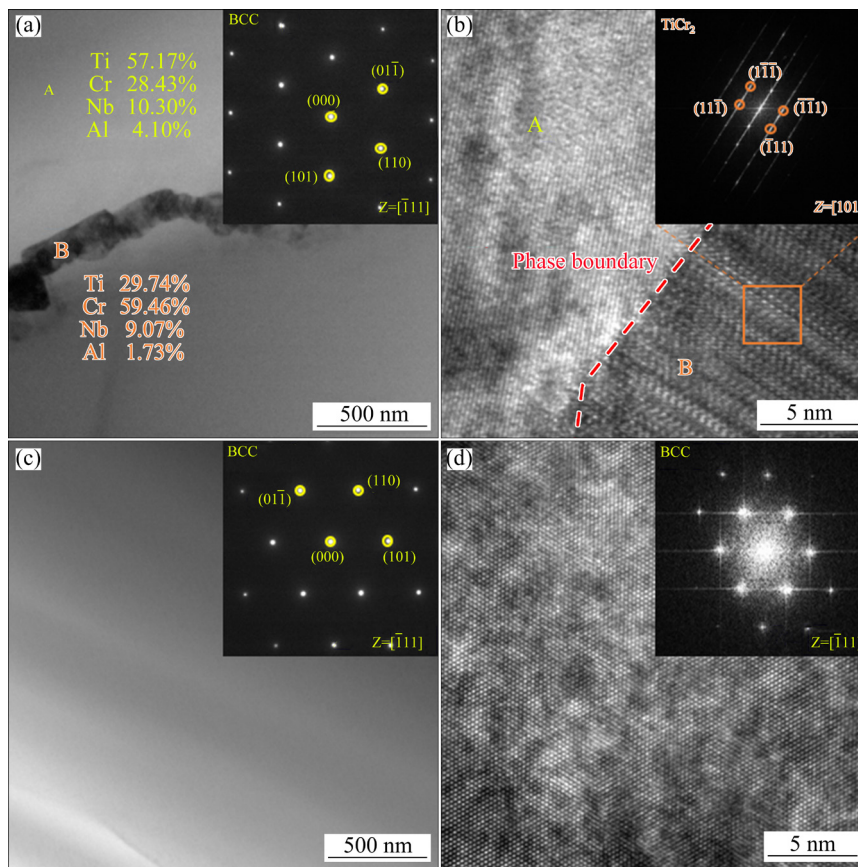


Fig. 3 TEM images of #AC-Al_5 and #HT-Al_5 alloys: (a) BF TEM image of #AC-Al_5 alloy with SAED pattern corresponding to Grain A showing BCC structure; (b) HRTEM image containing phase boundary between Grain A and TiCr_2 , with FFT of TiCr_2 phase presented in inset; (c) BF TEM image of #HT-Al_5 alloy with corresponding SAED pattern presented in inset; (d) HRTEM image of (c) with corresponding FFT presented in inset

alloy with SAED pattern corresponding to Grain A revealing a BCC structure. Based on the EDS/TEM results, the average composition of the precipitates at grain boundary (Grain B) is ~ 59.46 at.% Cr and ~ 29.74 at.% Ti with some Nb and Al. Accordingly, these precipitates can be preliminarily described as TiCr_2 -type phase. Figure 3(b) displays a high-resolution transmission electron microscope (HRTEM) image of the phase boundary between the BCC phase and precipitates. Based on the corresponding SAED diffraction of Grain A and the fast Fourier transform (FFT) image of Grain B, the precipitates are the C15-structured TiCr_2 -type Laves phase and the matrix is the BCC phase. Figure 3(c) shows a BF TEM image of the #HT-Al_5 alloy with the corresponding SAED pattern taken along the $[\bar{1}11]$ zone axis presented in the inset, suggesting a BCC structure. Figure 3(d) shows the HRTEM image of Fig. 3(c), revealing a BCC structure. Hence, the #HT-Al_5 alloy is composed of a single BCC phase, which is consistent with the XRD and BSE-SEM results.

3.2 Mechanical properties

Figures 4(a, b) show the compressive true stress–strain curves of the as-cast and homogenized

$(\text{Ti}_{60}\text{Cr}_{30}\text{Nb}_{10})_{100-x}\text{Al}_x$ MEAs conducted at room temperature, and the detailed values of yield strength (σ_y), compressive strength ($\sigma_{\max}$) and strain-to-failure (ε_f) are summarized in Table 3. In the as-cast condition (Fig. 4(a)), it can be seen that the #AC-Al_0 alloy exhibits relatively high yield strength of ~ 965 MPa and excellent plasticity (strain-to-failure higher than 50%). The addition of 5 at.% and 7.5 at.% of Al results in slight increments in yield strength. The yield strengths of the #AC-Al_5 and #AC-Al_7.5 alloys are ~ 1048 and ~ 1162 MPa, respectively. It is noted that these three alloys do not fracture when their compressive strain exceeds 50% during compression testing. However, adding 10 at.% Al into the $\text{Ti}_{60}\text{Cr}_{30}\text{Nb}_{10}$ alloy leads to an evident loss of plasticity along with a slightly increased yield strength, and in detail, the yield strength and strain-to-failure of #AC-Al_10 are ~ 1191 MPa and $\sim 42\%$, respectively. As discussed above, all the as-cast $(\text{Ti}_{60}\text{Cr}_{30}\text{Nb}_{10})_{100-x}\text{Al}_x$ MEAs have good combinations of strength and plasticity.

As shown in Fig. 4(b), after homogenization, all homogenized alloys show pronounced decrease in yield strength in comparison with their as-cast counterparts. In particular, the #HT-Al_7.5 and

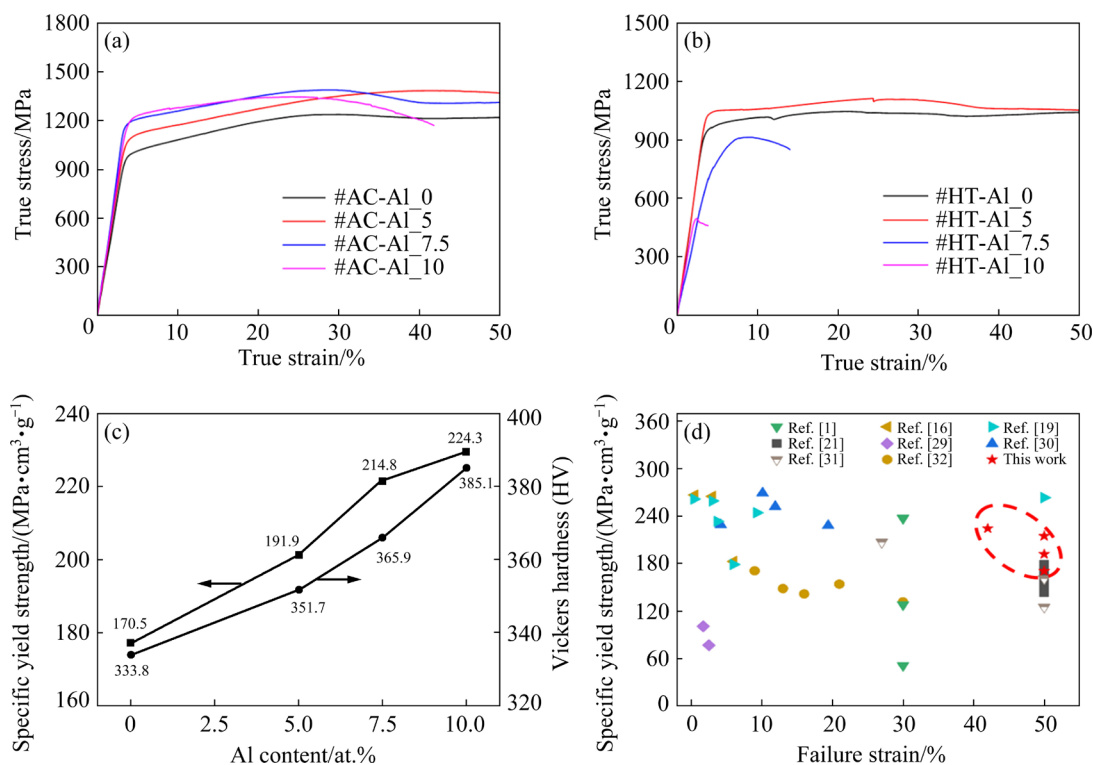


Fig. 4 Compressive true stress–strain curves of as-cast (a) and homogenized (b) $(\text{Ti}_{60}\text{Cr}_{30}\text{Nb}_{10})_{100-x}\text{Al}_x$ alloys; specific yield strength (SYS) and Vickers hardness curves of as-cast $(\text{Ti}_{60}\text{Cr}_{30}\text{Nb}_{10})_{100-x}\text{Al}_x$ alloys (c); Comparison of SYS and failure strain of several typical HEAs/MEAs and as-cast $(\text{Ti}_{60}\text{Cr}_{30}\text{Nb}_{10})_{100-x}\text{Al}_x$ MEAs (d)

Table 3 Mechanical properties of as-cast and homogenized $(\text{Ti}_{60}\text{Cr}_{30}\text{Nb}_{10})_{100-x}\text{Al}_x$ alloys at room temperature

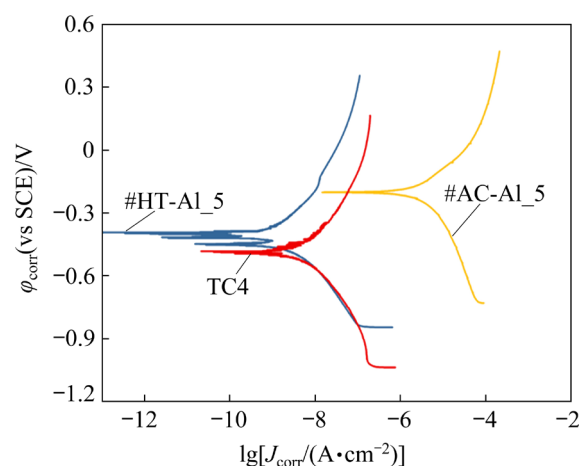
Alloy	σ_y/MPa	$\sigma_{\max}/\text{MPa}$	$\varepsilon_f/\%$
#AC-Al_0	965	—	≥ 50
#AC-Al_5	1048	1384	≥ 50
#AC-Al_7.5	1162	1388	≥ 50
#AC-Al_10	1191	1345	42
#HT-Al_0	917	1045	≥ 50
#HT-Al_5	1017	1113	≥ 50
#HT-Al_7.5	703	914	14
#HT-Al_10	492	494	4

#HT-Al_10 alloys display decreases of $\sim 39.5\%$ and $\sim 58.7\%$ in yield strength as composed to their as-cast counterparts, which may be attributed to the remarkable grain growth of these alloys. Early fracture also occurs in the #HT-Al_7.5 and #HT-Al_10 alloys during compression testing, and therefore these two alloys show strain-to-failure of $\sim 14\%$ and $\sim 4\%$, respectively. Figure 4(c) presents the specific yield strength (SYS) and microhardness curves of the as-cast $(\text{Ti}_{60}\text{Cr}_{30}\text{Nb}_{10})_{100-x}\text{Al}_x$ MEAs. It could be seen that both SYS and microhardness show an increasing trend with the increasing of Al addition. Specifically, as the addition of Al increases from 0 to 10 at.%, the value of SYS increases from ~ 170.5 to $\sim 224.3 \text{ MPa}\cdot\text{cm}^3/\text{g}$, whereas the value of Vickers hardness (HV) increases from HV ~ 333.8 to ~ 385.1 . As shown in Fig. 4(d), compared with some typical FCC and BCC structured HEAs/MEAs [1,16,19,21,29–32], the as-cast $(\text{Ti}_{60}\text{Cr}_{30}\text{Nb}_{10})_{100-x}\text{Al}_x$ MEAs exhibit outstanding mechanical properties, and in particular, the #AC-Al_10 alloy displays SYS of $\sim 224.3 \text{ MPa}\cdot\text{cm}^3/\text{g}$ and strain-to-failure of $\sim 42\%$. This suggests that the as-cast $(\text{Ti}_{60}\text{Cr}_{30}\text{Nb}_{10})_{100-x}\text{Al}_x$ MEAs have promising application prospect in the field of structural materials.

3.3 Corrosion behavior

Since the as-cast and homogenized $(\text{Ti}_{60}\text{Cr}_{30}\text{Nb}_{10})_{95}\text{Al}_5$ alloys exhibit the best comprehensive mechanical properties, the potentiodynamic polarization (PDP) tests of the #AC-Al_5 and #HT-Al_5 alloys were carried out in 3.5 wt.% NaCl solution at room temperature. To assess the electrochemical performance of these two alloys,

TC4 alloy was used as contrast material. Figure 5 shows the dynamic potential polarization curves of the #AC-Al_5, #HT-Al_5 and TC4 alloys in 3.5 wt.% NaCl solution, and their values of the corrosion potential (ϕ_{corr}), corrosion current density (J_{corr}) and initial passivation potential (ϕ_{ip}) are listed in Table 4. According to the PDP results, the #HT-Al_5 alloy shows an extremely low corrosion current density ($9.41 \times 10^{-10} \text{ A/cm}^2$), which is one order of magnitude lower than that of TC4 alloy ($3.23 \times 10^{-9} \text{ A/cm}^2$) and noticeably lower than that of #AC-Al_5 alloy ($2.07 \times 10^{-6} \text{ A/cm}^2$). Accordingly, #HT-Al_5 alloy is the most passive in 3.5 wt.% NaCl solution. In addition, ϕ_{corr} of #HT-Al_5 alloy (-0.39 V (vs SCE)) is slightly higher than that of TC4 alloy (-0.49 V (vs SCE)) and lower than that of the #AC-Al_5 alloy (-0.19 V (vs SCE)), suggesting that the #HT-Al_5 alloy has a lower susceptibility to corrosion compared with TC4 alloy. One should know that the polarization process is mainly controlled by dynamics rather than thermodynamics. In other words, corrosion performance of the alloys is principally determined by the corrosion current density, revealing that the #HT-Al_5 alloy may display the best corrosion resistance among the studied three alloys.

**Fig. 5** Potentiodynamic polarization curves of #AC-Al_5, #HT-Al_5 and TC4 alloys in 3.5 wt.% NaCl solution**Table 4** Corrosion kinetic parameters of potentiodynamic polarization for #AC-Al_5, #HT-Al_5 and TC4 alloys tested in 3.5 wt.% NaCl solution

Alloy	$\phi_{\text{corr}}(\text{vs SCE})/\text{V}$	$\phi_{\text{ip}}(\text{vs SCE})/\text{V}$	$J_{\text{corr}}/(\text{A}\cdot\text{cm}^{-2})$
#AC-Al_5	-0.19	0.10	2.07×10^{-6}
#HT-Al_5	-0.39	-0.24	9.41×10^{-10}
TC4	-0.49	-0.28	3.23×10^{-9}

Figure 6 presents the EIS results for these three alloys in order to gain further insights of corrosion properties. As shown in Fig. 6(a), the corresponding Nyquist curves of the #HT-Al_5 and TC4 alloys illustrate a single capacitive semicircle over the whole frequency area. But for the #AC-Al_5 alloy, the low-frequency region of EIS responses can be fitted by a straight line, as shown in the inset image in Fig. 6(a). The Nyquist curve of the #HT-Al_5 alloy exhibits larger diameter than that of the #AC-Al_5 and TC4 alloys, indicating that the #HT-Al_5 alloy possesses the best corrosion resistance. Figure 6(b) illustrates the Bode plots of the examined specimens. The solution resistance (R_s) and the polarization resistance (R_p) are estimated using the values of $\lg |Z|$ at high frequency and low frequency, respectively. Values of R_s and R_p for the #HT-Al_5 alloy are noble than those of the #AC-Al_5 and TC4 alloys, further confirming that the #HT-Al_5 alloy has the best corrosion resistance. Moreover, the maximum phase angle approaching -80° of the #HT-Al_5 alloy indicates higher capacity and more stabilized passive films compared with the #AC-Al_5 and

TC4 alloys, agreeing well with the results of potentiodynamic polarization curves in Fig. 5. The Bode-phase angle plots in the frequency range from 10^{-2} to 10^5 Hz suggest that only one time constant for the #AC-Al_5 alloy and two-time constants for the #HT-Al_5 and TC4 alloys. Figure 6(c) displays a $R_s(\text{CPE}_{dl}(R_{ct}Z_w))$ equivalent electrical circuit (EEC) used to interpret the EIS data of #AC-Al_5 alloy, including the solution resistance (R_s), the constant phase element (CPE_{dl}), the charge-transfer resistance (R_{ct}) and the Warburg impedance of electrode (Z_w). In contrast, the $R_s(\text{CPE}_f(R_f(\text{CPE}_{dl}R_{ct})))$ EEC is used to fit the EIS results of the #HT-Al_5 and TC4 alloys, as shown in Fig. 6(d). CPE_f and R_f are defined as capacitance and resistance of the compact passive film, respectively. The simulation parameters of EEC are listed in Table 5. Generally, the R_{ct} value is correlated to the stability of the passive film. According to fitting results, the R_{ct} value of the #HT-Al_5 alloy ($3.32 \times 10^7 \Omega \cdot \text{cm}^2$) is larger than that of the #AC-Al_5 ($4025 \Omega \cdot \text{cm}^2$) and TC4 ($1.36 \times 10^7 \Omega \cdot \text{cm}^2$) alloys, revealing a better resistance to the pitting corrosion.

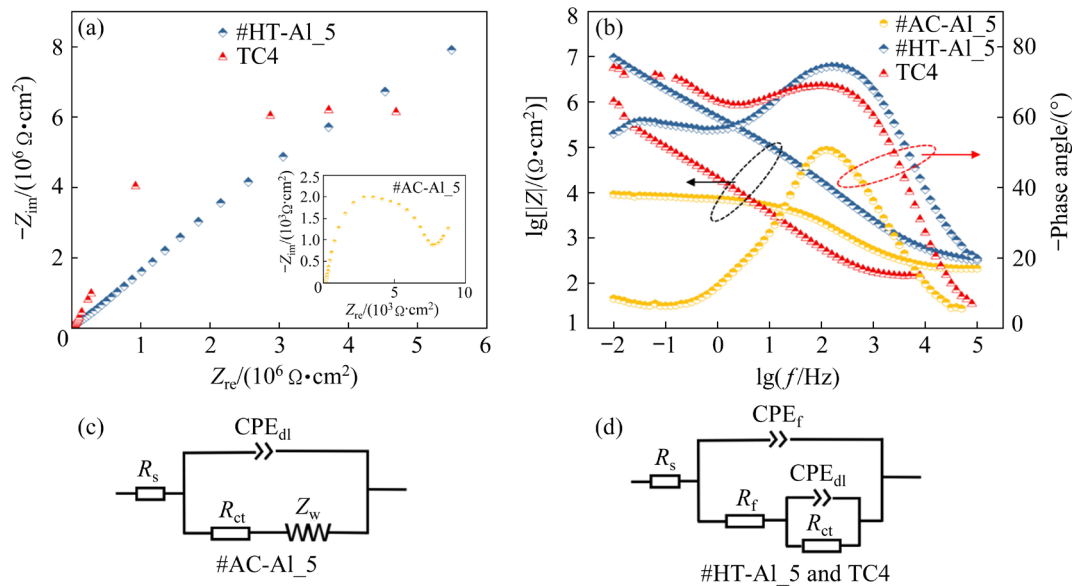


Fig. 6 Nyquist (a) and Bode (b) plots for #AC-Al_5, #HT-Al_5 and TC4 alloys in 3.5 wt.% NaCl solution, and equivalent circuit used to interpret EIS data (c, d)

Table 5 Typical electrochemical parameters of #AC-Al_5, #HT-Al_5 and TC4 alloys simulated by EIS data tested in 3.5 wt.% NaCl solution

Alloy	R_s/Ω	$\text{CPE}_{dl}/(\Omega \cdot \text{s}^n \cdot \text{cm}^{-2})$	n_1	$\text{CPE}_f/(\Omega \cdot \text{s}^n \cdot \text{cm}^{-2})$	n_2	$R_{ct}/(\Omega \cdot \text{cm}^2)$	$R_f/(\Omega \cdot \text{cm}^2)$	$Z_w/(\Omega \cdot \text{s}^5 \cdot \text{cm}^{-2})$
#AC-Al_5	110	7.50×10^{-6}	0.7136	—	—	4025	—	2.2×10^{-3}
#HT-Al_5	178	5.00×10^{-7}	0.8286	8.76×10^{-7}	0.6770	3.32×10^7	2.83×10^5	—
TC4	153	3.03×10^{-7}	0.9613	2.41×10^{-7}	1	1.36×10^7	6576	—

The morphologies of the specimens after the potentiodynamic polarization tests in 3.5 wt.% NaCl solution are shown in Fig. 7. It can be seen that the dissolution pits are visible at grain boundaries in the #AC-Al₅ alloy, as shown in Fig. 7(a). Figure 7(b) illustrates that the #HT-Al₅ alloy only shows some circled casting defects without any distinct corrosion products. It can be inferred that the uniform microstructure and chemical composition account for the outstanding corrosion resistance of the #HT-Al₅ alloy [33].

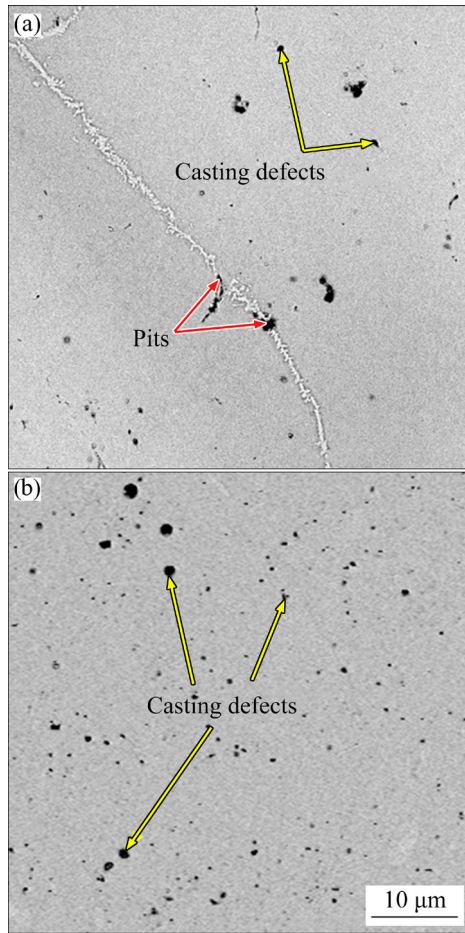


Fig. 7 BSE-SEM images of samples after potentiodynamic polarization tests: (a) #AC-Al₅ alloy; (b) #HT-Al₅ alloy

4 Discussion

4.1 Phase formation mechanisms

As discussed above, all the as-cast (Ti₆₀Cr₃₀-Nb₁₀)_{100-x}Al_x MEAs show a major BCC phase with a small amount of C15-structured TiCr₂-type Laves phase. To date, many empirical parameters have been used for predicting the phase formation of HEAs/MEAs, including the mixing entropy (ΔH_{mix}),

mixing enthalpy (ΔS_{mix}), atomic size difference (δ), solid solution formability (Ω), valence electron concentration (VEC) and average Allen electronegativity ($\Delta\chi_{\text{Allen}}$), and these parameters are defined by the following equations [34–37]:

$$\Delta H_{\text{mix}} = \sum_{i=1, i \neq j}^n 4\Delta H_{ij}^{\text{mix}} c_i c_j \quad (1)$$

$$\Delta S_{\text{mix}} = -R \sum_{i=1}^n (c_i \ln c_i) \quad (2)$$

$$\delta = \sqrt{\sum_{i=1}^n c_i \left[1 - r_i / \left(\sum_{i=1}^n c_i r_i \right) \right]^2} \quad (3)$$

$$\Omega = \frac{T_m \Delta S_{\text{mix}}}{|\Delta H_{\text{mix}}|} \quad (4)$$

$$\text{VEC} = \sum c_i \text{VEC}_i \quad (5)$$

$$\Delta\chi_{\text{Allen}} = \sqrt{\sum_{i=1}^n c_i \left(1 - \chi_i^{\text{Allen}} / \left(\sum_{i=1}^n c_i \chi_i^{\text{Allen}} \right) \right)^2} \quad (6)$$

where n , c_i , c_j , $\Delta H_{ij}^{\text{mix}}$ and R represent the number of the components, molar fractions of the i th and j th elements, enthalpy of mixing between the i th and j th elements and the molar gas constant, respectively. r_i , T_m , VEC_i and χ_i^{Allen} are the atomic radius of the i th element, average melting temperature, VEC of the i th element and Allen electronegativity of the i th element, respectively.

The calculated results of these parameters are listed in Table 6. It was revealed that a disorder single solution phase would be formed when the value of ΔH_{mix} is in the range from −15 to 5 kJ/mol and the value of δ is lower than 5%, whereas a mixture of disorder and order phases would be formed when the value of ΔH_{mix} is in the range from −20 to 0 kJ/mol and the value of the value of δ ranges from 5% to 6.6% [34]. Evidently, all the values of the studied as-cast MEAs meet the latter criterion. Taking the relationship of mixing entropy and mixing enthalpy into account, YANG and ZHANG [35] proposed that solid-solution phases could be formed in MEAs/HEAs when $\Omega \geq 1.1$ and $\delta \leq 6.6\%$. It is evident that the values of Ω for the (Ti₆₀Cr₃₀Nb₁₀)_{100-x}Al_x MEAs are notably higher than 1.1. Interestingly, a small amount of TiCr₂ Laves phase is detected in as-cast (Ti₆₀Cr₃₀Nb₁₀)_{100-x}Al_x, but these secondary phases are absent in their homogenized counterparts, suggesting that these two empirical parameters are more suitable for

the homogenized MEAs/HEAs. GUO et al [36] proposed that single BCC phases would be formed if the value of VEC is lower than 6.87. Similarly, this criterion is not in accordance with the as-cast alloys but fits the homogenized alloys, attributed to the phase decomposition during the solidification process. Recently, the formation criterion of Laves phase for HEAs/MEAs has also been studied. YURCHENKO et al [37] pointed out that Laves phase would be formed when $\delta > 5.0\%$ and $\Delta\chi_{\text{Allen}} > 7.0\%$, and our experimental results of the as-cast MEAs are in consistence with these two empirical criteria.

The CALPHAD (CALculation of PHase Diagrams) approach is an effective computational tool for predicting phase stability and phase transformations in target alloys. In this work, the equilibrium and non-equilibrium simulations for the as-cast $(\text{Ti}_{60}\text{Cr}_{30}\text{Nb}_{10})_{100-x}\text{Al}_x$ MEAs have been carried out and compared with the experimental data on the phase stability to demonstrate the validity of this approach. The equilibrium step diagrams (Fig. 8) show that several phases are formed, including BCC, C15-structured Laves and HCP phases. However, only two phases (disordered BCC phase and TiCr_2 -type C15-type Laves phase)

Table 6 Values of parameters (ΔH_{mix} , ΔS_{mix} , T_{m} , δ , Ω , VEC, and $\Delta\chi_{\text{Allen}}$) for $(\text{Ti}_{60}\text{Cr}_{30}\text{Nb}_{10})_{100-x}\text{Al}_x$ MEAs

Alloy	$\Delta H_{\text{mix}}/(\text{kJ}\cdot\text{mol}^{-1})$	$\Delta S_{\text{mix}}/(\text{kJ}\cdot\text{mol}^{-1})$	T_{m}/K	δ	Ω	VEC	$\Delta\chi_{\text{Allen}}$
$\text{Ti}_{60}\text{Cr}_{30}\text{Nb}_{10}$	−5.40	7.47	2077	5.77	2.87	4.70	8.34
$(\text{Ti}_{60}\text{Cr}_{30}\text{Nb}_{10})_{95}\text{Al}_5$	−9.21	8.74	2020	5.64	1.92	4.62	8.38
$(\text{Ti}_{60}\text{Cr}_{30}\text{Nb}_{10})_{92.5}\text{Al}_{7.5}$	−10.95	9.12	1991	5.57	1.66	4.57	8.39
$(\text{Ti}_{60}\text{Cr}_{30}\text{Nb}_{10})_{90}\text{Al}_{10}$	−12.58	9.42	1963	5.50	1.47	4.53	8.39

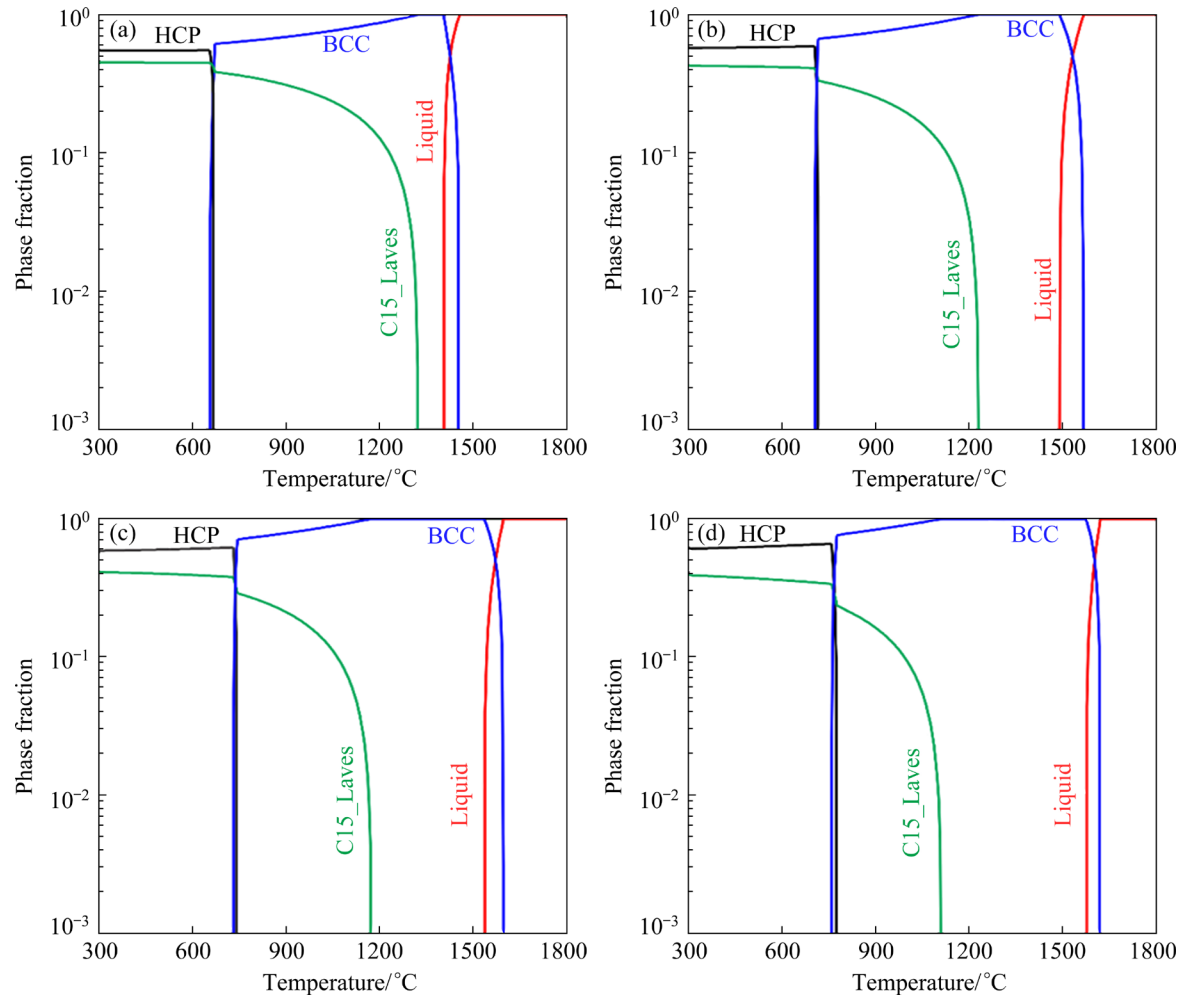


Fig. 8 Equilibrium step diagrams of #AC-Al_0 (a), #AC-Al_5 (b), #AC-Al_7.5 (c) and #AC-Al_10 (d) alloys

are observed experimentally in all as-cast alloys. This difference can be attributed to the non-equilibrium nature of the casting process. In most cases, Al acts as a strong α -Ti (HCP structure) stabilizer while Cr and Nb are strong β -Ti (BCC structure) stabilizers, which means that the phase in the alloys may transform to HCP when Al atoms are added to the matrix. In fact, some research studies have demonstrated that Al unexpectedly acts as a strong BCC stabilizer in many HEAs/MEAs, such as $\text{Al}_x\text{CoCrFeNi}$ [38,39] and $\text{Ti}_{60}\text{Al}_x(\text{VCrNb})_{40-x}$ [40]. In addition, the step diagrams of #AC-Al_0 and #AC-Al_5 suggest that the C15-type Laves phase remains after heat treatment at 1200 °C. On the contrary, experimental data show that all alloys exhibit a single BCC phase and no secondary phases can be identified after homogenization through XRD, SEM and TEM analysis. Unless the secondary phase formation requires a much longer time, the experimental results in this work can be used to further improve thermodynamic databases

and calibrate the CALPHAD-model prediction.

Furthermore, Scheil solidification modeling (Fig. 9) can be used successfully predicted the formation of the BCC and Laves phases. Generally speaking, TiCr_2 -type Laves phase possesses two typical structures, including a low-temperature cubic (C15) structure and a high-temperature hexagonal (C14) structure. According to Ref. [41], the C14-structured TiCr_2 Laves phase can be retained at room temperature; however, this metastable structure can transform to stable C15 structure after high temperature annealing. In this work, the stable C15_Laves phase is observed in all as-cast alloys.

Overall, the empirical parameters and the CALPHAD method both achieve relatively good agreement with the experimental data in the current work. Meanwhile, in spite of some discrepancies, the CALPHAD method still offers valuable guidance when understanding the phase stability in HEAs/MEAs.

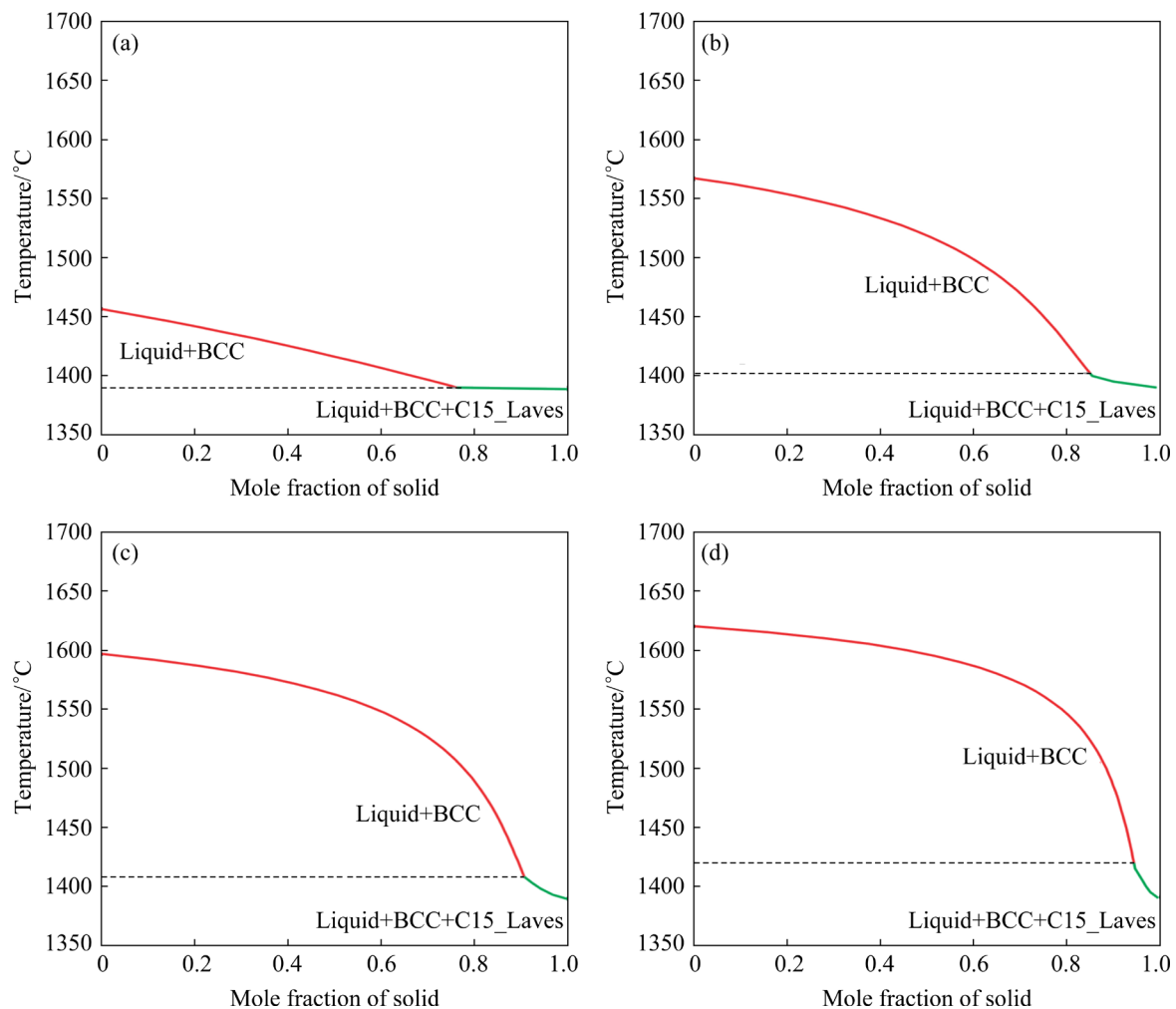


Fig. 9 Scheil solidification curves of #AC-Al_0 (a), #AC-Al_5 (b), #AC-Al_7.5 (c) and #AC-Al_10 (d) alloys

4.2 Strengthening mechanisms

The strengthening mechanisms of metallic materials mainly include solid-solution hardening, grain-boundary hardening, second-phase hardening, and dislocation hardening, and therefore the yield strength (σ_y) can be calculated by the following equation:

$$\sigma_y = \sigma_0 + \sigma_{ss} + \sigma_{gb} + \sigma_{sp} + \sigma_{ds} \quad (7)$$

where σ_0 , σ_{ss} , σ_{gb} , σ_{sp} and σ_{ds} represent the lattice friction strength, solid-solution strengthening, grain-boundary strengthening, second-phase strengthening and dislocation strengthening, respectively. In Section 3.2, it is illustrated that with more Al atoms dissolved into the matrix, the yield strength of the #AC-Al_10 alloy is increased by ~226 MPa compared with the #AC-Al_0 alloy (Table 3), indicating that Al atom has a strong solid solution strengthening effect. Meanwhile, due to the formation of TiCr₂-structured Laves phase observed in the studied alloys, the second-phase strengthening should be taken into consideration.

(1) Lattice friction strength and grain boundary strengthening

In the case of the studied MEAs, the lattice friction strength σ_0 can be replaced by the Peierls–Nabarro (P–N) stress σ_p , which is defined as follows [42]:

$$\sigma_p = G \exp\left(-\frac{2\pi d}{b}\right) \quad (8)$$

where G is the shear modulus of the alloys and can be calculated by the rule of mixture, which is ~64.7, ~62.8, ~61.8 and ~60.8 GPa for the #AC-Al_0, #AC-Al_5, #AC-Al_7.5 and #AC-Al_10 alloys, respectively. b is the magnitude of Burgers vector, and d is the dislocation core width depending on the temperature. It can be expressed as follows [43,44]:

$$d = d_0(1 + \mu T) \quad (9)$$

where d_0 is the dislocation core width at 0 K, μ is a constant which can be estimated by $\mu = 1/T_m$. T_m and T mean the melting temperature and the examination temperature of the alloys, respectively. The σ_0 values of the #AC-Al_0, #AC-Al_5, #AC-Al_7.5 and #AC-Al_10 alloys are ~50, ~47, ~46 and ~44 MPa, respectively.

The influence of grain boundary on yield strength can be estimated by using the classical Hall–Petch equation [45,46]:

$$\sigma_{gb} = kD^{-1/2} \quad (10)$$

where k and D denote the Hall–Petch slope and mean grain size, respectively. The values of k can be calculated as ~388, ~373, ~366 and ~358 MPa· $\mu\text{m}^{1/2}$ by applying the rule of mixture. The grain boundary strengthening values of the #AC-Al_0, #AC-Al_5, #AC-Al_7.5 and #AC-Al_10 alloys are evaluated to be ~77, ~69, ~64 and ~63 MPa, respectively.

(2) Solid solution strengthening

Compared with the dilute solution alloys, the solid solution strengthening of HEAs/MEAs is different due to the severe lattice distortion [42]. To investigate the solid solution strengthening effect of HEAs/MEAs, several models have been established. According to Refs. [47,48], combined with interaction force (F), the solid solution strengthening originated from the i th component can be calculated from

$$\Delta\sigma_i = AGf_i^{4/3}c_i^{2/3} \quad (11)$$

where A denotes a dimensionless constant of the order of 0.02 [49]. And f_i can be estimated by

$$f_i = \sqrt{\delta_{G_i}^2 + \alpha^2 \delta_{r_i}^2} \quad (12)$$

where δ_{G_i} and δ_{r_i} are the shear modulus and the atomic radius mismatches, respectively. The value of α depends on the type of the mobile dislocations, which can be defined as 9 for a random mixture of edge and screw dislocations [49]. For BCC structure HEAs/MEAs, the δ_{G_i} and δ_{r_i} can be acquired as follows [31, 49]:

$$\delta_{G_i} = \frac{9}{8} \sum c_j \delta_{G_{ij}} \quad (13)$$

$$\delta_{r_i} = \frac{9}{8} \sum c_j \delta_{r_{ij}} \quad (14)$$

where $\delta_{G_{ij}} = 2(G_i - G_j)/(G_i + G_j)$, and $\delta_{r_{ij}} = 2(r_i - r_j)/(r_i + r_j)$. Table 7 gives the calculated values of $\delta_{G_{ij}}$ and $\delta_{r_{ij}}$ for different atom pairs. The solid solution strengthening σ_{ss} is obtained by summation of $\Delta\sigma_i$ of the alloying elements:

$$\sigma_{ss} = \left(\sum \Delta\sigma_i^{3/2}\right)^{2/3} \quad (15)$$

The calculated solid solution strengthening values for the #AC-Al_0, #AC-Al_5, #AC-Al_7.5 and #AC-Al_10 alloys are ~756, ~741, ~732 and ~722 MPa, respectively. The largest σ_{ss} of the #AC-Al_0 alloy indicates its severest lattice distortion.

Thus, the estimated values of $\sigma_0 + \sigma_{ss} + \sigma_{gb}$ for the #AC-Al_0, #AC-Al_5, #AC-Al_7.5 and #AC-Al_10 alloys are ~883, ~857, ~842 and ~829 MPa, showing deviations of ~8.5%, ~18.2%, ~27.5% and ~30.4%, as compared to the measured yield strengths, respectively.

Table 7 Calculated values of atomic size difference $\delta_{r_{ij}}$ and shear modulus difference $\delta_{G_{ij}}$ for different atom pairs

Element i/j	$\delta_{r_{ij}}$	$\delta_{G_{ij}}$	Element i/j	$\delta_{r_{ij}}$	$\delta_{G_{ij}}$
Ti/Ti	0	0	Cr/Nb	-0.143	-1.007
Ti/Cr	0.088	0.893	Cr/Al	-0.081	-1.262
Ti/Nb	-0.055	-0.146	Nb/Nb	0	0
Ti/Al	0.007	-0.514	Nb/Al	0.062	-0.372
Cr/Cr	0	0	Al/Al	0	0

(3) Second-phase strengthening

Commonly, the second-phase strengthening is mediated by either a dislocation by-pass mechanism (Orowan-type) or particle shearing mechanism [50,51]. Orowan mechanism occurs when the radius of particles exceeds a critical value or is incoherent with the matrix. As presented in Fig. 3(b), the HRTEM image of the phase boundary confirms the incoherent relationship between the BCC phase and TiCr₂ phase. Hence, for TiCr₂ phase, the Orowan by-pass mechanism contribution to yield strength increment can be calculated as [51–53]

$$\Delta\sigma_{\text{Orowan}} = M \frac{0.4Gb \ln(2\bar{r}/b)}{\pi\sqrt{1-\nu} \lambda_p} \quad (16)$$

where M is the Taylor factor, and it is generally taken as 3.1 for BCC-structured metals [52]. ν is the Poisson ratio, and the value of 0.23 is commonly used for BCC phase alloys [52]. $\bar{r}$ is the mean radius of a circular cross-section in a random plane for the precipitate, $\bar{r} = \sqrt{2/3}r$, r is the mean radius of the precipitates. λ_p is the mean edge-to-edge interprecipitate spacing, $\lambda_p = 2\bar{r}(\sqrt{\pi/(4V)} - 1)$, V is the volume fraction of the TiCr₂ phase, and can be estimated by the proportion of the area occupied by the TiCr₂ phase of the matrix in SEM/TEM images. In detail, the fractions of the Laves phase for the #AC-Al_0, #AC-Al_5, #AC-Al_7.5 and #AC-Al_10 alloys are ~4.4%, ~4.6%, ~3.9% and ~4.6%, respectively. The calculated strength increments from Orowan by-pass mechanism are

estimated to be ~25, ~26, ~21 and ~24 MPa, respectively. Overall, the increment of yield strength for the current MEAs is mainly attributed to the solid-solution strengthening.

Furthermore, in Section 3.1, serrated-grain-boundaries (SEGBs) are observed in the homogenized alloys. Previous studies have revealed that the SEGBs have some unusual effects on the mechanical behavior. TANG et al [54] demonstrated that the SEGBs could enhance intergranular cracking resistance in the high-temperature/low-stress regimes because the growth of cavities greatly is suppressed by serration. WU et al [28] proposed that the SEGBs architecture could overcome the intergranular cracking issue and improve the strength and ductility simultaneously at high temperature, which is attributed to the enhanced resistance of the SEGBs to the intergranular crack nucleation and propagation. However, the SEGBs show negligible influence on the mechanical properties of the studied alloys, which may be due to the small amount of the SEGBs ascribing to the large grain size and the room-temperature testing condition. In addition, the strength and plasticity of the #HT-Al_7.5 and #HT-Al_10 alloys decrease evidently after the homogenization. This is mainly because high content of Al dissolved into the BCC matrix can give rise to markedly severe lattice distortion, and therefore early fracture occurs during the compression tests.

4.3 Influence of homogenization on corrosion behavior

Corrosion is likely to occur at defects, grain boundaries and large segregation of elements, suggesting that the more uniform the microstructure and chemical composition are, the better the corrosion resistance will be achieved [55]. It was revealed that microstructure simplification and chemical-segregation reduction could decrease the work function variations, and accordingly the improvement of corrosion resistance of Al_xCoCr-FeNi HEAs occurred after homogenization [56]. The method of friction stir processing was adopted to refine and homogenize the surface microstructure of AZ91 magnesium alloy, which significantly enhanced its corrosion resistance [57]. In this work, the homogenized MEA shows the best corrosion resistance among the tested alloys according to the

potentiodynamic polarization and EIS results. The difference in corrosion behavior between the as-cast and homogenized (Ti₆₀Cr₃₀Nb₁₀)₉₅Al₅ MEAs can be attributed to the existence of dendritic segregations and grain boundary precipitates which could act as the origin of corrosion. Accordingly, the localized pitting corrosion is likely to take place in the as-cast MEA.

Additionally, the corrosion products also have significant effects on the corrosion behavior. It is well known that Cr element is beneficial to the corrosion properties due to the formation of the Cr₂O₃ passive oxide film and then preventing further corrosion. It was reported that when the Cr-rich BCC phase disappears and transforms to other phases, the corrosion resistance of the alloys is degraded [33]. It was also demonstrated that Cr-depleted BCC phase would be readily attacked by Cl⁻ ions, thereby leading to deteriorative corrosion resistance [56]. Clearly, Cr element should be uniformly distributed in the #HT-Al_5 alloy. This plays an important role in resisting Cl⁻ ions erosion, and therefore pitting corrosion would hardly occur in the #HT-Al_5 alloy. In conclusion, the uniform microstructure and chemical composition are conducive to achieving outstanding corrosion resistance for the #HT-Al_5 alloy.

5 Conclusions

(1) All the studied MEAs show relatively low densities (lower than 5.8 g/cm³), and accordingly they can be regarded as lightweight alloys. With the increasing of Al content, the phase compositions of the as-cast MEAs remain unchanged, consisting of a BCC phase with a small amount of TiCr₂ Laves phase. After the homogenization, all the MEAs show a single BCC phase.

(2) As the content of Al increases, the compressive yield strength of the as-cast MEAs increases from 965 to 1191 MPa, displaying failure strain higher than 50%. However, both of the strength and plasticity can be decreased after the homogenization, especially for #HT-Al_7.5 and #HT-Al_10 alloys, arising from the early fracture during compression testing caused by severe lattice distortion induced by the dissolving of high content of Al.

(3) Compared with the TC4 and the #AC-Al_5 alloys, the #HT-Al_5 alloy shows a superior

corrosion resistance, primarily attributed to the uniform microstructure and chemical composition.

CRediT authorship contribution statement

Hao WANG: Conceptualization, Methodology, Investigation, Writing – Original draft, Writing – Review & editing; **Wei-ping CHEN:** Supervision, Conceptualization, Funding acquisition; **Liang-yan HAO:** Experiment, Investigation; **Chen-liang CHU:** Experiment, Investigation; **Wei XIONG:** Writing – Review & editing; **Zhi-qiang FU:** Supervision, Conceptualization, Writing – Review & editing.

Declaration of competing interest

The authors declare that they have no known competing financial interests or personal relationships that could have appeared to influence the work reported in this paper.

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$(\text{Ti}_{60}\text{Cr}_{30}\text{Nb}_{10})_{100-x}\text{Al}_x$ 轻质中熵合金的力学与腐蚀性能

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摘要: 研究轻质 $(\text{Ti}_{60}\text{Cr}_{30}\text{Nb}_{10})_{100-x}\text{Al}_x$ ($x=0, 5, 7.5, 10$, 摩尔分数, %)中熵合金在铸态和均匀化态下的显微组织、力学性能和腐蚀行为。结果表明, 随着 Al 含量从 0 增加到 10%(摩尔分数), 铸态合金均由相同的相组成, 即体心立方(BCC)相和少量 TiCr_2 型 Laves 相组成。均匀化后, 所有合金均由单相 BCC 组成。此外, $(\text{Ti}_{60}\text{Cr}_{30}\text{Nb}_{10})_{95}\text{Al}_5$ 中熵合金表现出在铸态和均匀态条件下最优异的力学性能, 屈服强度分别约为 1048 MPa 和 1017 MPa, 同时保持超 50%压缩应变。与 TC4 和铸态 $(\text{Ti}_{60}\text{Cr}_{30}\text{Nb}_{10})_{95}\text{Al}_5$ 合金相比, 均匀态 $(\text{Ti}_{60}\text{Cr}_{30}\text{Nb}_{10})_{95}\text{Al}_5$ 合金表现出更优异的耐腐蚀性能, 这主要归因于均匀的显微组织与化学成分。

关键词: 中熵合金; 轻质合金; 显微组织; 力学性能; 腐蚀行为

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