Microstructures and mechanical properties of $\text{Al}_x\text{CrFeNiTi}_{0.25}$ alloys

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Aiming to lower the cost and improve mechanical properties of AlCoCrFeNiTi_x high-entropy alloys that were studied previously, the present research investigated the effect of removing Co and lowering Ti contents at various Al contents, namely $\text{Al}_x\text{CrFeNiTi}_{0.25}$. The microstructures were investigated using optical microscopy, scanning electron microscopy, energy disperse spectroscopy, x-ray diffraction and transmission electron microscopy. Compression tests were conducted at room temperature. The present study showed that with increasing Al contents, the phase structures of the alloys changed from FCC + BCC to double BCC as the main phases. Among the alloys studied, all of them exhibit distinguished workhardening especially the $\text{Al}_{0.5}\text{CrFeNiTi}_{0.25}$ alloy has the highest fracture strength and plastic-strain limit of

3.47 GPa and 40 %, respectively with a yield strength of 1.88 GPa. The observed microstructure is analyzed using CALPHAD calculations.

1. INTRODUCTION

Recently, high-entropy alloys (HEAs)¹⁻³, which is based on the concept of the tremendously high configurational entropy of mixing due to the formation of solid solutions of multi-principal elements, have drawn rising interest. HEAs have been reported to possess a series of unique properties, for instance, high hardness⁴⁻⁸, outstanding wear resistance^{9,10}, good fatigue resistance characteristics¹¹, excellent high-temperature strength^{12,13}, good thermal stability¹⁴, good oxidation⁹, and corrosion resistance^{15,16}. Research efforts have been undertaken to gain the fundamental understanding of the HEA formation and explore new high-performance HEAs for various applications¹⁷⁻²³. Ranganathan²⁴ regarded HEAs, bulk metallic glasses, and rubber metal as the most three breakthroughs alloying theory during the recent decade.

At present, the research on HEAs is still in the early exploratory stage. Synthesis of new HEAs by traditional trial and error methods has dominated the literature. The effect of addition of the Al content in various HEAs based on 3d-transition metals on their structures, thermodynamics, and mechanical properties has been reviewed and analyzed by Tang et al.²⁵ A previous study⁵ showed that the AlCoCrFeNiTi_x alloy system (x = 0, 0.5, 1.0, and 1.5 in molar ratio) has promising mechanical properties.

The dominant phase in the system is the body-centered cubic (BCC) solid solution, and the system possesses excellent room-temperature compression mechanical properties. In particular, AlCoCrFeNiTi_{0.5} has a yield strength, fracture strength, and plastic strain limit as high as 2.26 GPa, 3.14 GPa, and 23.3% respectively. Following that study, we have developed a series of new Al_xCrFeNiTi_{0.25} alloys based on the valence electron concentration (VEC) scheme proposed by Guo et al.²⁶. The new alloys do not contain expensive Co in order to reduce the alloy cost and has a much lower Ti content to minimize the formation of brittle Laves phase. The objective of the present study is to identify alloys with the optimal combination of strength and ductility in the Al_xCrFeNiTi_{0.25} system.

2. EXPERIMENTAL PROCEDURES

Alloy ingots with nominal composition of Al_xCrFeNiTi_{0.25} (x: molar ratio, x = 0, 0.25, 0.5, 0.75, and 1.0 denoted by Al0, Al0.25, Al0.5, Al0.75, and Al1, respectively) were prepared by arc melting in a Ti-gettered high-purity argon atmosphere. The pure elements have a purity higher than 99.6 weight percent. The alloys were remelted at least four times to improve homogeneity. The ingots were then remelted under high vacuum (10⁻³ Pa) and injection cast into a water-cooled copper mould to obtain cylindrical rods of $\Phi 5 \times 70$ mm.

The microstructure of as-cast samples was characterized by X-ray diffraction (XRD). Samples of $\Phi 5 \times 10$ mm for compressive tests were prepared and investigated

at room temperature with a strain rate of 2×10^{-4} /s. The fracture surface and the morphology of the alloys were examined with scanning electron microscopy (SEM) equipped with an energy-dispersive X-ray (EDX) spectrometer. TEM samples were first mechanically ground to 40 μ m thickness and then thinned using ion milling till perforations. The TEM observations were carried out with a 300 keV Tecnai F30 microscope at the Electron Microscopy Laboratory of Peking University.

3. RESULTS

3.1 Microstructures

The XRD patterns of $\text{Al}_x\text{CrFeNiTi}_{0.25}$ alloys with various Al contents in the as-cast condition are shown in Fig. 1(a). The main phases identified are face-centered cubic (FCC) and BCC structures. There are minor unidentified peaks at small angles in Al0, and the structure corresponding to those peaks are desirable for the TEM study as the future work. The FCC phase disappears in Al0.5, while two BCC-type phases appear in Al0.5, Al0.75 and Al1. The overlapping peaks of the BCC phases indicate that their lattice parameters are close to each other. The TEM work showed that they are disordered BCC (designated as A2) and NiAl-type ordered BCC (designated as B2), and the details are provided on the TEM paragraph. As indicated in Fig. 1(b), the lattice parameter of FCC and BCC phases increases with increasing Al contents.

The SEM micrographs of the as-cast $\text{Al}_x\text{CrFeNiTi}_{0.25}$ alloys are shown in Fig. 2. Typical cast dendrite and interdendrite (defined as DR and ID, respectively) structures

were observed. The morphology of dendrites and interdendrites is diverse in the $\text{Al}_x\text{CrFeNiTi}_{0.25}$ alloy system, which suggests that the Al content also affects the growth behavior of dendrites. This trend is expected because the equilibrium phases and their compositions strongly depend on the Al content in this alloy system. The chemical composition (atomic %) in the dendrite and interdendrite regions of the Al0.5 alloy are listed in Table I. Iron and chromium are segregated to the interdendrite region, and the dendrite is enriched in aluminum, nickel, and titanium.

The TEM examination of the microstructure of Al0.5 has been conducted, and the results are shown in Fig. 3. Two types of microstructures are observed: inter-weaved short rods align at about 90 degree angles (Fig. 3a), and nano-particles of irregular shape distribute in a matrix (Fig. 3b). The selected area electron diffraction (SAED) pattern [the insert in Fig. 3 (a)] indicate that those rods are ordered BCC phases (about 30 nm in width, 100 nm in length.) with a lattice constant of 0.303 nm. The SAED indicate that the nano-particle is the disordered BCC phase with a lattice constant of 0.297 nm (the insert in Fig. 3b). According to Orowan mechanism, these particles with a disordered BCC structure would provide effective dispersing-strengthening effects. The chemical compositions of both structures are analyzed by EDX, as shown in Table II. It can be observed that the matrix is enriched in Cr and Fe, while the particles and tiny long precipitates are enriched in Al and Ni. Those two different structures provide both effective precipitation strengthening and nano-composite strengthening effects.

3.2 Mechanical Properties

Figure 4 shows the compression stress-strain curves of $\text{Al}_x\text{CrFeNiTi}_{0.25}$ at different Al contents, and the values of the fracture strength, σ_b , yield strength, $\sigma_{0.2}$, and plastic strain limit, ε_p are listed in Table III. All $\text{Al}_x\text{CrFeNiTi}_{0.25}$ alloys exhibit fairly high fracture strength ($> 1,800$ MPa) and yield strength ($> 1,300$ MPa). When the x value is lower than 0.5, the fracture strength, yield strength and plastic strain increase with increasing the Al content. When $x = 0.5$, the fracture strength and plastic strain achieved the maximum, namely 3,474 MPa and 40%, respectively. With further increasing the Al content, both the fracture strength and strain drop rapidly. Note that Al1 shows little ductility with the lowest strength. Furthermore, a comparison with other typical HEAs²⁷⁻³² and bulk metallic glasses (BMGs)²⁹ are summarized in Fig. 5 and the data are listed in Table IV. It seems that $\text{Al}_{0.5}\text{CrFeNiTi}_{0.25}$ has the most desirable combination of strength and ductility among HEAs materials listed.

Figure 6 indicates the fracture surfaces of $\text{Al}_x\text{CrFeNiTi}_{0.25}$ alloys. The fracture surface exhibits both apparently featureless regions (a) and regions with a “rock candy” appearance (b). Rock-candy features are a characteristic sign of the intergranular brittle fracture, where the crack path follows along grain boundaries. Many dimple-like features with a size of 3 – 5 μm are observed in (c), which are typically correlated with good plasticity. With a further addition of Al, the dimple-like features gradually decrease, as shown in (d) for Al0.75, and a typical river pattern is observed. As for Al1, the fracture surface shows high-density short tear edges and cleavage steps with little ductility in (e), and these features indicate a cleavage crack in the Al1 alloy.

4. DISCUSSION

4.1 Effect of removing Co and lowering Ti contents

New materials design need to consider not only the property enhancement but also the decrease of the alloy cost. In the case of developing HEAs based on CoCrFeMnNi, lowering or removing expensive Co but without negatively impacting the materials properties is particularly useful because it will make these new materials competitive against commercial stainless or even Ni-based superalloys. Removing Co can largely reduce the cost of the alloy, by adjusting the Al contents, the alloy $\text{Al}_{0.5}\text{CrFeNiTi}_{0.25}$, seems to possess the best combination of the high fracture strength and high plastic strain among a dozen different alloy systems, as shown in Fig. 5. Lowering the Ti content is also shown to be very effective in the depressing formation of the brittle Laves phase.

4.2 Effect of Al contents

The present work shows the addition of Al promotes the BCC-phase formation in the $\text{Al}_x\text{CrFeNiTi}_{0.25}$ alloy system. This trend is in agreement with the work by Tang et al.²⁵ who reviewed $\text{Al}_x\text{CoCrCuFeNi}$, $\text{Al}_x\text{CoCrFeNi}$, $\text{Al}_x\text{CrFe}_{1.5}\text{MnNi}_{0.5}$, $\text{Al}_x\text{CoCrFeNiTi}$, and $\text{Al}_x\text{CrCuFeNi}_2$ alloy systems. Although Al has an FCC structure, it has an atomic radius (1.43 Å) that is much larger than Cr, Fe, and Ni (whose radii are about 1.25 Å). Subsequently increasing the lattice constant and large lattice strain results when Al atoms are introduced to the lattice site of the FCC Ni or Fe. The excessive addition of Al eventually leads to the formation of a BCC structure to minimize the lattice-strain energy. Among the Al-3d transition metal interactions

analyzed by Tang et al.²⁵ using the acronym of CALculation of PHase Diagram (CALPHAD) method, the B2 type Al-Ni bond is the strongest, which has an enthalpy value of - 43 kJ/mole of atoms at 800 °C, while the enthalpy in the Cr-Al and Fe-Al solid solution in the BCC or FCC structures are positive. Accordingly, preference of formation B2-type AlNi is expected in $\text{Al}_x\text{CrFeNiTi}_{0.25}$ alloys with higher Al contents, and such preference introduces alloying element partitioning among B2, A2, and A1 phases during solidification. This trend is exactly why the dendrites are enriched in Al and Ni, and the interdendrites are enriched in Cr and Fe. In addition, theoretical studies using *ab initio* molecular dynamics simulations^{1,23} (AIMD) predict the preferred interatomic interaction of Al-Ni and Cr-Fe pairs even in the liquid state. In short, the results from the published CALPHAD analysis²⁵ and AIMD^{1,23} simulations are consistent, and are confirmed in the present experiments and others.

4.3 Phase analysis using CALPHAD modeling

To assist analyzing the observed trend in the microstructure and mechanical property of $\text{Al}_x\text{CrFeNiTi}_{0.25}$, the CALPHAD calculations were carried out using the TCNI6 database supplied with the ThermoCalc software. An isopleth is shown in Fig. 7. Only phases stable at high temperatures are shown, since they are most relevant to the processing condition of the as-cast ingots. The liquidus temperature increases steadily, as the Al content is increased, suggesting the increased interatomic bonding pertaining to Al (primarily due to the formation of the B2 type structure). During further cooling after primary crystallization, FCC-A1 starts to form in Al0 and Al0.25, and B2 in Al0.5, Al0.75, and Al1. As the temperature is lowered further, $\text{Ni}_3\text{Ti}_{0.25}\text{D}_{0.24}$

forms in Al₀, and H_L2₁ in Al_{0.25}, Al_{0.5}, Al_{0.75}, and Al₁. Here Ni₃Ti_D0₂₄ refers to the stable binary hexagonal Ni₃Ti phase (Pearson symbol hP16, Space group P6₃/mmc) with very small solubilities of Fe, Cr, and Ni, and H_L2₁ refers to the Heusler phase, AlNi₂Ti (Pearson symbol cF16, Space group $Fm\bar{3}m$). At 950 °C, all alloys are in the solid state, and the equilibrium phases with their mole fractions are listed in Table V.

Although the predicted isopleth is of equilibrium in nature, it at least qualitatively explains the observed microstructure. For example, the A2 phase was observed in all 5 alloys as the primary phase. At a fixed temperature of 1,150 °C, the phase diagram shows that A1 is stable in Al₀ and Al_{0.25}, and not stable in Al_{0.5} and above. This is in excellent agreement with experiments (see Fig. 1).

In terms of the phase composition in Al_{0.5}, equilibrium calculations at 1,050 °C and 1,000 °C were carried out, which are slightly below the solidus temperature. Since both FCC and H_L2₁ phases were not observed in the experiments, they were set "dormant" in the calculations to maintain a microstructure of A2 + B2. The phase compositions calculated at 1,000 °C are listed in Table II. The excellent agreement for both phases between experiments and modeling demonstrates that the CALPHAD modeling based on reliable thermodynamic databases can be used to predict the phase transition and element partitioning among phases even though the equilibrium solidification is usually not possible during arc/induction/suction melting in a laboratory. Such modeling can be applied to arbitrary virtual alloys to search for ideal compositions that may have the widest phase field of the disordered solid-solution

phase of interest (e.g., A1, A2, or A3 for the HEA formation), and, thus, it can accelerate the HEA development (examples are shown in reference ²³).

4.4 Mechanical properties

The present study showed that the optimal combination of strength and ductility occurs in the alloy with the intermediate Al content, namely $\text{Al}_{0.5}\text{CrFeNiTi}_{0.25}$. Alloys with lower or higher Al contents have both lower strength and ductility. Such observation is counterintuitive, since the B2-type NiAl bond that is promoted by the Al addition is much stronger than the A2-type CrFe bond. But the same trend has also been observed in other alloy systems, such as $\text{FeCoNi}(\text{AlSi})_x$ ³¹, CoCrFeNiTiAl_x ³³, $\text{Al}_x\text{NiFeCuCo}$ ³⁴, $\text{Al}_x\text{CoCrFeNiTi}_{0.5}$ ³⁵, and $\text{Al}_x\text{NbTiMoV}$ ³⁶. Taking CoCrFeNiTiAl_x ³³ for example, the maximum strength of 2.28 GPa was reached when $x = 1.0$ while the compressive strength decrease gradually to 2.11 GPa for $x = 1.5$ and 1.03 GPa for $x = 2.0$.

The presence of the FCC in a large amount in Al0 and Al0.25 is responsible for the impressive ductility. Al0.25 is stronger because B2 appears and B2 phase is very strong. The presence of cast porosity about 20% percent in Al0 also reduces the strength, because the porosity can not carry load. If the porosity is avoided, then the yield strength can be more than 1500 MPa. The high-density dimple-shaped structures can be clearly observed in Fig. 6(c), which suggests large local deformation in Al0.5 alloy. CALPHAD modeling shows that the B2 phase dominates the microstructure at high temperatures in the solid state (not shown) in Al0.75 and Al1. Brittle fracture along grain boundaries in Al0.75 (see Fig. 6) is apparent. Planar fracture combined

with tear dominates in Al1. For Al0.75 alloys [as shown in Fig. 6(d)], the rough facets and river-like patterns indicate the fracture morphology of alloys exhibit to be a typical cleavage fracture, and the flat fracture of the Al1.0 alloy in Fig. 6(e) indicates its poor ductility. In both Al0.75 and Al1, transgranular fracture did not occur, rather fracture proceeds along paths where atomic bonds are weak, e.g., high-energy defect sites. As a result, despite the dominance of the high-strength B2-NiAl phase, both alloys show lower strength.

It is amazing to note that distinguished work hardening was found during compression test. Due to the large work hardening capacity, these compositions exhibit excellent ultimate strengths. Especially for Al0.5, the ultimate strength is as high as 3474 MPa, and the maximum fracture strain is as large as 40%. The large plasticity and distinguished work hardening are accommodated by the inter-weave short rods align at about 90 degree angles (Fig. 3a), and nano-particles of irregular shape distribute in a matrix (Fig. 3b). The nano-particles uniformly distributed in matrix taking up 61% percent can fix the dislocations effectively, and strengthen the work hardening property at room temperature. Both inter-weaved short rod and nano-particles are responsible for the strain improvement and distinguished work hardening capacity.

It is worth noting that the L2₁ phase is stable in Al0.25 and above, as predicted by CALPHAD modeling. The presence of the nano-precipitation of L2₁ in these alloys by proper thermo-mechanical processing is possible. Future work in

manipulating the microstructure between A1/A2/B2/L2₁ and, thus, property will be useful to design new alloys with possibly enhanced properties.

5. SUMMARY

The microstructure and mechanical properties of Al_xCrFeNiTi_{0.25} (x = 0 to 1.0) alloys were studied in the present work, and the following conclusions were reached:

- 1) The Al addition promotes the BCC formation.
- 2) Al_{0.5}CrFeNiTi_{0.25} possesses the optimal combination of strength and ductility, and the precipitation of nano-scale BCC particles in a BCC matrix is presumably responsible for its superior mechanical properties.
- 3) The CALPHAD calculations were carried out to explain the observed microstructure with success. Excellent agreement in the phase composition between modeling and experiments was observed.

ACKNOWLEDGEMENTS

This research is supported by the Natural Science Foundation (NSF) of China, granted No.NSFC51210105006. M.C.G. acknowledges the support of the Innovative Processing and Technologies Program of the National Energy Technology Laboratory's (NETL) Strategic Center for Coal under the RES contract DE-FE-0004000.

DISCLAIMER

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References:

1. Y. Zhang, T. T. Zuo, Z. Tang, M. C. Gao, K. A. Dahmen, P. K. Liaw, Z. P. Lu, Prog. Mater Sci, **61**, 1-93 (2014).
2. J. Yeh, Ann. Chim, Sci. Mat, **2006**, 633-648 (2006).
3. B. Cantor, I. Chang, P. Knight, A. Vincent, Mater. Sci. Eng. A, **375**, 213-218 (2004).
4. C. Tong, M. Chen, J. Yeh, S. Lin, S. Chen, T. Shun, S. Chang, Metall. Mater. Trans. A, **36**, 1263-1271 (2005).
5. Y. J. Zhou, Y. Zhang, Y. L. Wang, G. L. Chen, Appl. Phys. Lett, **90**, 181904 (2007).
6. L. H. Wen, H. C. Kou, J. S. Li, H. Chang, X. Y. Xue, L. Zhou, Intermetallics, **17**, 266-269 (2009).
7. J. M. Zhu, H. M. Fu, H. F. Zhang, A. M. Wang, H. Li, Z. Q. Hu, Mater. Sci. Eng. A, **527**, 7210-7214 (2010).
8. S. Chen, W. Tang, Y. Kuo, S. Chen, C. Tsau, T. Shun, J. Yeh, Mater. Sci. Eng, **527**, 5818-5825 (2010).
9. M. Chuang, M. Tsai, W. Wang, S. Lin, J. Yeh, Acta Mater, **59**, 6308-6317 (2011).
10. M. Chuang, M. Tsai, C. Tsai, N. Yang, S. Chang, J. Yeh, S. Chen, S. J. Lin, Alloys Compd, **551**, 12-18 (2013).
11. M. A. Hemphill, T. Yuan, G. Y. Wang, J. W. Yeh, C. W. Tsai, A. Chuang, P. K. Liaw, Acta Mater, **60**, 5723-5734 (2012).

12. C. Hsu, C. Juan, W. Wang, T. Sheu, J. Yeh, S. Chen, Mater. Sci. Eng. A, **528**, 3581-3588 (2011).
13. O. N. Senkov, G. B. Wilks, J. M. Scott, D. B. Miracle, Intermetallics, **19**, 698-706 (2011).
14. M. Tsai, C. Wang, C. Tsai, W. Shen, J. Yeh, J. Gan, W. J. Wu, Electrochem. Soc, **158**, H1161-H1165 (2011).
15. Y. L. Chou, Y. C. Wang, J. W. Yeh, H. C. Shih, Corros. Sci, **52**, 3481-3491 (2010).
16. Y. Kao, T. Lee, S. Chen, Y. Chang, Corros. Sci, **52**, 1026-1034 (2010).
17. Y. Huang, L. Chen, H. Lui, M. Cai, J. Yeh, Mater. Sci. Eng. A, **457**, 77-83 (2007).
18. Y. J. Zhou, Y. Zhang, Y. L. Wang, G. L. Chen, Mater. Sci. Eng. A, **454**, 260-265 (2007).
19. L. H. Wen, H. C. Kou, J. S. Li, H. Chang, X. Y. Xue, L. Zhou, Intermetallics, **17**, 266-269 (2009).
20. Y. Y. Chen, U. T. Hong, H. C. Shih, J. W. Yeh, T. Duval, Corros. Sci, **47**, 2679-2699 (2005).
21. C. Tsai, M. Tsai, J. Yeh, C. Yang, J. Alloys Compd, **490**, 160-165 (2010).
22. T. Shun, Y. Du, J. Alloys Compd, **478**, 269-272 (2009).
23. M. C. Gao, D. E. Alman, Entropy, **15**, 4504-4519 (2013).
24. S. Ranganathan, Current Science, **85**, 1404-1406 (2003).

25. Z. Tang, M. C. Gao, H. Diao, T. Yang, J.P. Liu, T.T. Zuo, Y. Zhang, Z.P. Lu, Y.Q. Cheng, Y.W. Zhang, K.A. Dahmen, P.K. Liaw, T. Egami, JOM, **65**, 1848-1858 (2013).
26. S. Guo, C. Ng, J. Lu, C. T. Liu, J. Appl. Phys, **109**, 103505 (2011).
27. F. J. Wang, H. A. Davies, Y. Zhang, G. L. Chen, J. Eng. Mater. Technol, **131**, 34501 (2009).
28. X. F. Wang, Y. Zhang, Y. Qiao, G. L. Chen, Intermetallics, **15**, 357-362 (2007).
29. P. Matteis, J. Alloys Compd, **509**, S99-S104 (2011).
30. B. S. Li, Y. P. Wang, M. X. Ren, C. Yang, H. Z. Fu, Mater. Sci. Eng. A, **498**, 482-486 (2008).
31. Y. Zhang, T. Zuo, Y. Cheng, P. K. Liaw, Scientific reports, **3** (2013).
32. J. W. Qiao, Y. F. Wang, R. Q. Wang, J. Y. Shi, S. B. Sang, H. J. Yang, Mater. Sci. Forum, **2013**, 768-774 (2013).
33. K.B. Zhang et al, Mater. Sci. Eng. A, **508**, 214–219 (2009).
34. T. Zhang, Y. Zhang, H. Zhou, Advanced Materials and Information Technology Processing, **87**, 47 (2014).
35. Y. Dong, Y. P. Lu, J. J. Zhang, T. J. Li, Materials Science Forum, **745-746**, 775-780, (2013).
36. S. Y. Chen, X. Yang, K. A. Dahmen, P. K. Liaw, Y. Zhang, Entropy, **16**, 870-884, (2014).

FIGURE CAPTIONS

Table I. SEM-EDX results of dendrites and interdendrites in Al_{0.5}Cr FeNiTi_{0.25} alloy.

	Al	Ti	Cr	Fe	Ni
Nominal	13.3	6.6	26.7	26.7	26.7
ID	6.6	1.7	39.8	41.4	10.5
DR	23.8	7.5	13.9	19.5	35.3

Table II. TEM-EDX results of particle A2 and tiny long precipitate B2 and matrix in

$\text{Al}_{0.5}\text{CrFeNiTi}_{0.25}$ alloy at room temperature. Phase compositions calculated

from CALPHAD at 1,000 °C are also provided (for details see the text).

phase	Al	Al	Ti	Cr	Fe	Ni
B2	TEM	22.9	9.7	7.5	13.7	46.2
	CALPHAD	22.9	12.3	8.3	11.5	45.0
A2	TEM	2.7	0.6	48.9	39.4	8.4
	CALPHAD	4.3	1.4	43.9	40.9	9.5

Table III. Room-temperature mechanical properties of $\text{Al}_x\text{CrFeNiTi}_{0.25}$ alloys.

Alloy	$\sigma_{0.2}$ (MPa)	σ_b (MPa)	ε_p (%)
Al0	1309.9	2621.4	20.4
Al0.25	1390.8	3412.3	33.4
Al0.5	1879.8	3474.7	40
Al0.75	1726.1	2811.1	20
Al1.0	1444.7	1829.1	7.2

Table IV. Room-temperature mechanical properties of HEAs.

NO.	Alloy	$\sigma_{0.2}$ (Mpa)	σ_b (Mpa)	ε_p (%)	Ref
1	$\text{Al}_{0.5}\text{CrFeNiTi}_{0.25}$	1880	3475	40	This work
2	AlCoCrFeNi	1423	2852	22.7	27
3	$\text{AlCoCrFeNiTi}_{0.5}$	2260	3140	23.3	5
4	AlCoCrFeNiTi_0	1500	2830	26.9	5
5	$\text{CuCoFeNiCrTi}_{0.5}$	700	1650	21.6	28
6	$\text{Cu}_{48}\text{Zr}_{48}\text{Al}_4$	1200	1880	5.3	29
7	CuCoFeNiCrAlV	1469	1970	16	30
8	CuCoFeNiCrAl	1303	2081	24	30
9	NbMoTaW	1058	1211	1.5	13
10	$\text{FeCoNi}(\text{AlSi})_{0.3}$	938	2857	33	31
11	$\text{FeCoNi}(\text{AlSi})_{0.2}$	342	2500	47	31
12	$\text{AlCoCrFeNiTi}_{0.3}$	1980	3600	12	32

Table V . Calculated equilibrium mole fraction of phases at 950 °C

Alloy	Phases	Mole fraction
Al0	A1	0.34
	A2	0.38
	Ni ₃ Ti_D0 ₂₄	0.28
Al0.25	A1	0.37
	A2	0.39
	H_L2 ₁	0.20
	Ni ₃ Ti_D0 ₂₄	0.04
Al0.5	A1	0.11
	A2	0.52
	B2	0.06
	H_L2 ₁	0.31
Al0.75	A2	0.52
	B2	0.18
	H_L2 ₁	0.30
Al1.0	A2	0.48
	B2	0.29
	H_L2 ₁	0.23

Figure 1. XRD patterns (a) and magnified main peaks of $\text{Al}_x\text{CrFeNiTi}_{0.25}$ alloys (b).

Figure 2. SEM pictures of $\text{Al}_x\text{Cr FeNiTi}_{0.25}$ alloys. (a) Al_0 ; (b) $\text{Al}_{0.25}$; (c) $\text{Al}_{0.5}$; (d) $\text{Al}_{0.75}$; (e) $\text{Al}_{1.0}$.

Fig 3. TEM micrograph of as-cast $\text{Al}_{0.5}\text{CrFeNiTi}_{0.25}$ alloy: (a) bright-field image with tiny long precipitates in the matrix; (b) bright-field image with a number of particles precipitate in the matrix.

Figure 4. Stress-strain curves of $\text{Al}_x\text{CrFeNiTi}_{0.25}$ alloys.

Fig 5. Compressive behavior of $\text{Al}_{0.5}\text{CrFeNiTi}_{0.25}$ and other HEAs.

Figure 6. SEM fracture surface of (a) $\text{Al}_0\text{CrFeNiTi}_{0.25}$; (b) $\text{Al}_{0.25}\text{CrFeNiTi}_{0.25}$; (c) $\text{Al}_{0.5}\text{CrFeNiTi}_{0.25}$; (d) $\text{Al}_{0.75}\text{CrFeNiTi}_{0.25}$; and (e) $\text{Al}_{1.0}\text{CrFeNiTi}_{0.25}$.

Fig 7. Calculated isopleth of $\text{Al}_x\text{CrFeNiTi}_{0.25}$ ($x = 0 - 1$) using the TCNI6 database.

The short arrows on the top mark the composition of the five alloys studied.