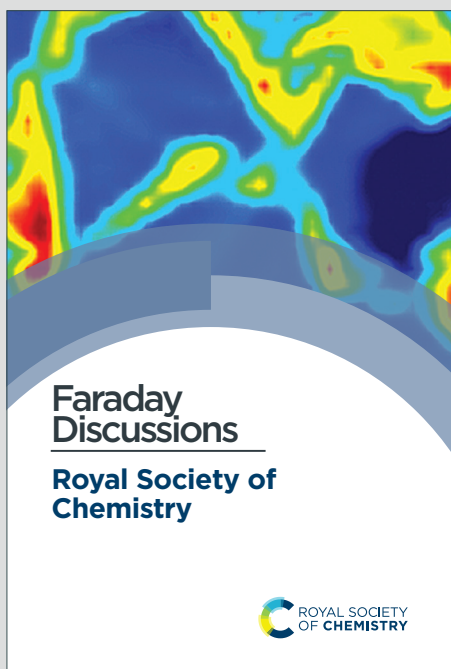


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Processing-Driven Chemical Ordering and Its Effect on Magnetic Properties in a High Entropy Alloy

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Abstract

The influence of chemical ordering and its length scale on the magnetic behaviour of the $\text{Al}_{0.2}\text{Ti}_{0.3}\text{Co}_{1.5}\text{CrFeNi}_{1.5}$ high entropy alloy (HEA) was investigated across three distinct microstructural conditions, including a solution annealed state, an annealed condition at 750 °C, and a cold rolled plus annealed condition. These processing routes produced changes in the volume fraction, size, and morphology of L_{12} ordered precipitates along with the formation of a minor L_{21} phase in the annealed states. Magnetic measurements performed between 2 K and 300 K revealed clear differences in saturation magnetization, coercivity, and magnetic transition temperatures across the three conditions. The solution annealed condition exhibited a single magnetic transition near 48 K and low coercivity, consistent with a fine dispersion of coherent L_{12} precipitates. In the annealed and cold rolled conditions, coarsening and morphological evolution of the L_{12} phase led to the emergence of a second magnetic transition at lower temperatures, attributed to partial magnetic decoupling of the precipitates. Coercivity increased significantly in these conditions due to enhanced domain wall pinning, while residual hysteresis at 300 K is attributed to a combination of minor L_{21} phase at grain boundaries and microstructural features in the cold-worked condition. These results demonstrate that chemical ordering and its structural evolution play a central role in governing low temperature magnetic behaviour in this alloy system. The findings contribute to the broader understanding of multifunctional HEA that combine tunable magnetic properties with excellent mechanical performance.



1. Introduction

High entropy alloys (HEA), composed of multiple principal elements in near-equiatom proportions, have gained attention for their promising structural and functional properties[1-5]. Among these, FCC-based CoCrFeNi-derived alloys are particularly attractive due to their good thermal stability, corrosion resistance, and mechanical strength[6, 7]. Alloying with Al and Ti has been shown to promote chemical ordering, leading to the formation of $L1_2$ type precipitates, which improve strength by precipitation hardening. In some cases, minor amounts of ordered intermetallic phases such as $L2_1$ Heusler-like structures can also form, particularly at grain boundaries. While mechanical properties of such precipitation strengthened HEA have been well studied[7], their magnetic behavior, especially how it evolves with phase formation and ordering length scale, has been rarely explored. This is important because magnetic properties in these systems are closely tied to composition, atomic scale ordering, and microstructure. The presence of ordered phases such as $L1_2$ and $L2_1$, and their interaction with the FCC matrix, can significantly affect magnetization, coercivity, and magnetic transition temperatures.

Nanoscale phase engineering in high-entropy alloys and compositionally complex alloys enables the simultaneous optimization of magnetic and mechanical performance[8-13]. For example, a Fe-CoNi-Ta-Al alloy with coherent nanoprecipitates uniformly dispersed in an FCC matrix exhibited enhanced ultimate tensile strength (≈ 1526 MPa) and good ductility ($\sim 15\%$), while maintaining a coercivity close to 10 Oe[11]. In a related system, the $\text{Fe}_{32.6}\text{Co}_{27.7}\text{Ni}_{27.7}\text{Ta}_5\text{Al}_7$ HEA exhibited a dual-phase microstructure containing $L1_2$ precipitates within a ferromagnetic FCC matrix. By tailoring the precipitate size, this alloy achieved a low coercivity (~ 1 Oe), moderate saturation magnetization (~ 100 emu/g), and high ductility ($\sim 54\%$ elongation)[12]. These properties resulted from the combined effects of precipitation strengthening, solid-solution hardening, and microband refinement, with minimal magnetic pinning due to the full coherence and uniform distribution of the precipitates. Advanced electron microscopy reveals that magnetization reversal in $\text{Al}_{0.3}\text{CoFeNi}$ evolves from a nucleation-controlled mechanism in the FCC + $L1_2$ state (domain wall width ~ 171 nm) to a pinning-dominated process in a nanorod-containing BCC/B2 matrix, where the domain wall width narrows to ~ 35 nm[9].

Magnetic materials are key components in everyday life[14, 15], and they are not only important for room temperature or elevated temperature applications but are also increasingly



relevant in the low temperature range[5, 16]. In the 20 to 80 K range, they are being explored for magnetic refrigeration, particularly in pre-cooling stages for hydrogen liquefaction as well as in cryogenic actuators and control systems where both mechanical reliability and magnetic responsiveness are required. At lower temperatures, around 2 to 10 K, such materials are used in superconducting magnet systems, magnetic shielding, and low temperature sensing, including emerging quantum devices and precision instrumentation[17, 18]. In these settings, materials with well-defined magnetic transitions, stable coercivity, and reliable microstructural stability are essential for functional integration and performance.

In this context, the $\text{Al}_{0.2}\text{Ti}_{0.3}\text{Co}_{1.5}\text{CrFeNi}_{1.5}$ HEA is a valuable model system for studying the relationship between chemical ordering and magnetic behavior. Although it is not designed to exhibit strong ferromagnetism at room temperature, it develops multiple ordered phases including L_{12} precipitates of different morphology and size scale and, in some conditions, a minor L_{21} phase. The size, volume fraction, and morphology of these phases can be tuned through thermal and thermomechanical processing. These features enable a systematic investigation of how ordering at different length scales influences saturation magnetization, coercivity, and the emergence of distinct magnetic transitions. The insights gained from this system are relevant for the broader design of multiphase alloys with balanced magnetic and mechanical performance in low temperature environments.

2. Experimental Methods

The $\text{Al}_{0.2}\text{Ti}_{0.3}\text{Co}_{1.5}\text{CrFeNi}_{1.5}$ alloy was synthesized by arc melting a mixture of elemental metals under an argon atmosphere. To ensure compositional homogeneity, the alloy was re-melted multiple times. The ingot was then homogenized at 1150 °C for 6 hours and water quenched to room temperature. Four processing conditions were investigated. The solution-annealed (SA) condition corresponds to the homogenized and quenched state. In the second condition (SA750), the SA sample was annealed at 750 °C for 50 hours. The third and fourth conditions (60CRA and 90CRA) involved cold rolling the SA material to 60% and 90% thickness reductions, respectively, followed by annealing at 750 °C for 50 hours. All heat treatments were performed in sealed quartz tubes under argon and were subsequently quenched in water.

Microstructural characterization was carried out using scanning electron microscopy (SEM) equipped with energy-dispersive X-ray spectroscopy (EDS) and electron backscatter diffraction (EBSD), as well as transmission electron microscopy (TEM). Site-specific TEM



lamellae were prepared using focused ion beam (FIB) lift-out techniques. Atom probe tomography (APT) was performed in laser mode to analyse nanoscale chemical partitioning.

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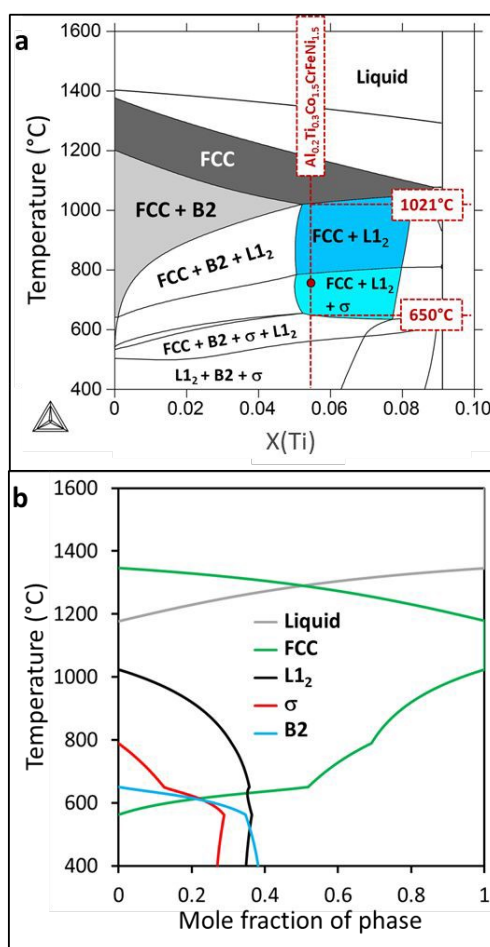
Magnetic measurements were carried out using a Physical Property Measurement System (PPMS EverCool, Quantum Design, USA). Temperature-dependent magnetization was measured under zero-field-cooled (ZFC) and field-cooled (FC) conditions with an applied magnetic field of 1000 Oe, over the temperature range of 5 K to 400 K. Isothermal magnetization (M–H) curves were recorded at 2 K, 10 K, 50 K, 100 K, and 300 K using applied magnetic fields of ± 5000 Oe. The term “saturation magnetization” (M_s) in this study refers to the magnetization measured at an applied field of 5000 Oe. This may not correspond to true saturation under conditions above the magnetic transition temperature, where the material is paramagnetic or exhibits a mixed ferromagnetic and paramagnetic response. Nevertheless, it is used as a comparative measure of magnetic behaviour across different conditions.

3. Results and discussions

3.1. Thermodynamic modelling

The alloy composition was selected to enable the precipitation of nanoscale $L1_2$ in an FCC matrix, thereby enhancing its strength. Solution thermodynamic modelling in **Figure 1(a)** shows an isopleth constructed by varying the Al/Ti ratio in $Al_{0.5-x}Ti_xCo_{1.5}CrFeNi_{1.5}$ alloy. The left side of the isopleth corresponds to the composition $Al_{0.5}Co_{1.5}CrFeNi_{1.5}$, while the right side corresponds to $Ti_{0.5}Co_{1.5}CrFeNi_{1.5}$. The current alloy $Al_{0.2}Ti_{0.3}Co_{1.5}CrFeNi_{1.5}$, as indicated by the red dashed line in this figure, exhibits a large FCC+ $L1_2$ phase field. The phase fraction vs temperature plot in **Figure 1(b)** shows that $\sim 35\%$ $L1_2$ can be precipitated at 750 °C.





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Figure 1. (a) Pseudo-binary isopleth for $\text{Al}_{0.5-x}\text{Ti}_x\text{Co}_{1.5}\text{CrFeNi}_{1.5}$ ($x = 0$ to 0.5 in relative mole fraction) (b) Phase fraction vs temperature diagram for $\text{Al}_{0.2}\text{Ti}_{0.3}\text{Co}_{1.5}\text{CrFeNi}_{1.5}$.

3.2. Microstructural characterization

The microstructural characterization results from SA and SA750 conditions are shown in **Figure 2**. The SEM backscattered images in **Figure 2(a-b)** show the coarse-grain microstructures of SA and SA750 conditions, respectively. They are expected to be single-phase FCC, as predicted by the thermodynamic modelling in Figure 1. The inset in **Figure 2(b)** shows a high-magnification image of the precipitates in the SA750 condition. This microstructure likely consisting of L_{12} nanoprecipitates in an FCC matrix, is also consistent with the prediction (**Figure 1**).

APT analysis confirms the presence of nanoscale L_{12} precipitates in both SA and SA750 conditions. Representative APT reconstructions from the SA condition are shown in **Figure 2(c-d)**. The superimposed ion map in **Figure 2(c)** highlights a uniform distribution of ~ 5 nm spherical precipitates with an estimated volume fraction of $\sim 9\%$, which appear as Ni-rich and



Fe-depleted features in the Fe ion map in **Figure 2(d)**. Compositional partitioning between matrix and precipitate phases is summarized in table given in **Figure 2(e)**, showing Ti and Al enrichment in the $L1_2$ phase. While the APT data correspond to the SA condition, prior characterization[7] indicates that in SA750, the $L1_2$ precipitates coarsen to ~ 25 nm and grow in volume fraction to $\sim 46\%$.

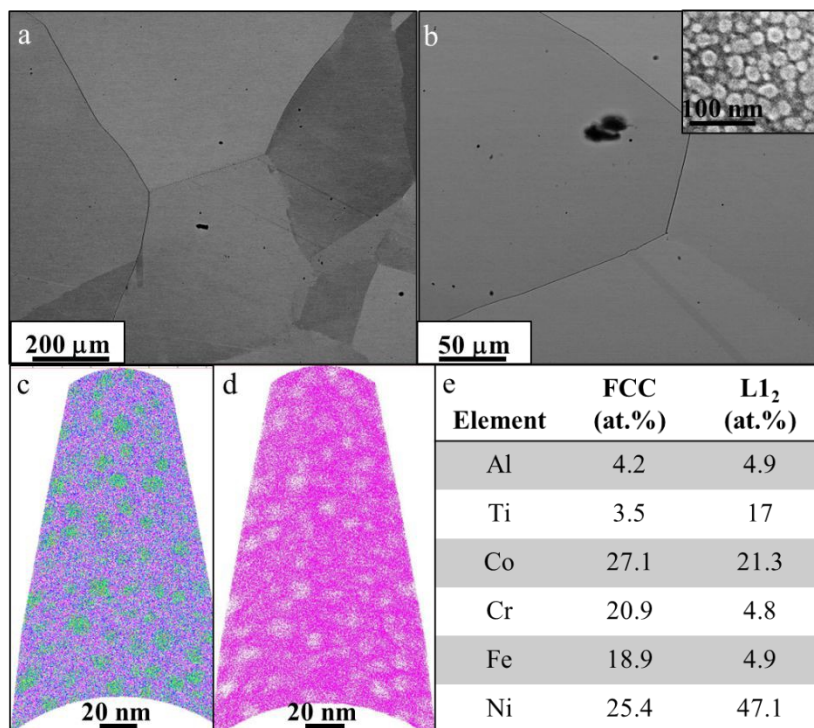


Figure 2. Backscattered SEM images of (a) SA and (b) SA750 conditions. The inset in (b) shows a high magnification image of the precipitates in the SA750 condition. APT results from the SA condition are shown in (c-e). (c) A 2D section of the superimposition of all raw ion maps (Al=blue, Ti=dark green, Co=blue, Cr=light pink, Fe=dark pink, Ni-green) (d) Fe raw ion map section (e) FCC and $L1_2$ compositions in the SA condition.

Figure 3(a-b) show the microstructures of 60CRA and 90CRA conditions respectively. Partial recrystallization can be observed in the 60CRA condition as indicated by the red boundary between recrystallized and non-recrystallized regions in **Figure 3(a)**. On the other hand, a fully recrystallized fine-grained structure is visible in the 90CRA condition in **Figure 3(b)**. Additionally, dark polygonal or dot-like precipitates are visible in both 60CRA and 90CRA conditions. The degree of recrystallization in the two conditions was further studied using EBSD. **Figure 3(c)** shows the inverse pole figure (IPF) map from the 60CRA condition. Here,



coarse non-recrystallized grains, as indicated by the orientation contrast, i.e., variations in color within the same grain, and fine recrystallized grains, can be seen to co-exist. Islands of recrystallized grains can also be observed in non-recrystallized grains, which result from higher strain in those regions. The IPF map from the 90CRA condition in **Figure 3(d)** shows a homogeneously recrystallized grain structure.

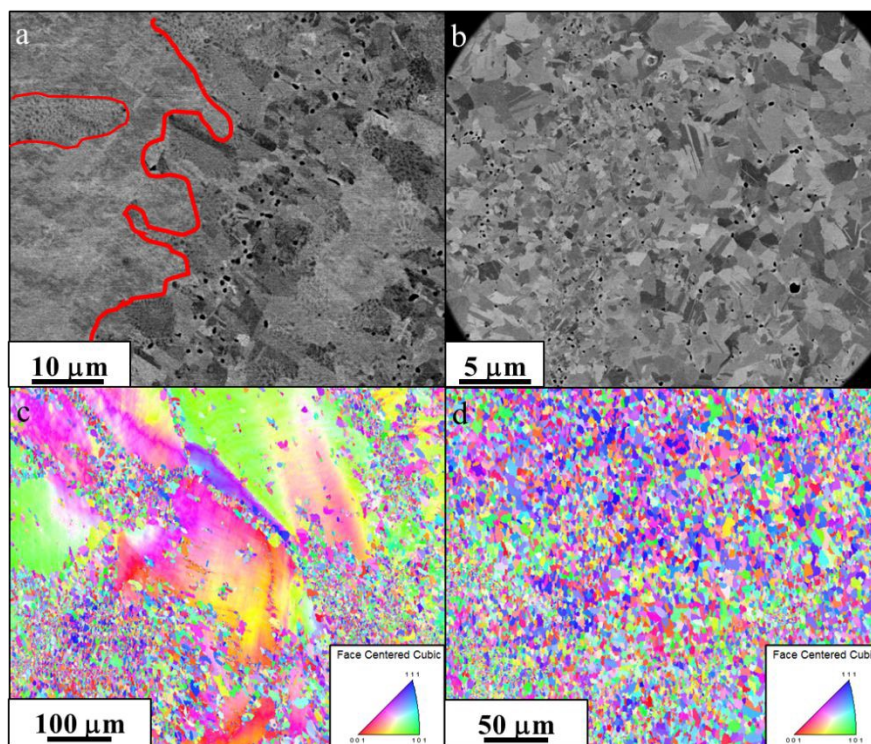


Figure 3. Backscattered SEM images of (a) 60CRA and (b) 90CRA conditions. EBSD IPF maps show (c) a partially recrystallized microstructure in 60CRA and (d) a fully recrystallized microstructure in 90CRA.

The nanoscale microstructures in 60CRA and 90CRA conditions were investigated using TEM to reveal $L1_2$ precipitation, as expected from annealing at 750°C (refer to **Figure 1**). While detailed microstructural characterization of all conditions was reported elsewhere[7], two important TEM results are shown in **Figure 4**. The TEM results from a non-recrystallized region in the 60CRA condition are shown in **Figure 4(a-c)**. The HAADF-STEM image in **Figure 4(a)** and the Ti and Ni STEM-EDS maps in Figure 4(b-c) show blocky $L1_2$ precipitates within this region. On the other hand, the recrystallized grains in 90CRA conditions show a rod-like or lamellar $L1_2$ precipitation in Figure 4(d-f).

The lamellar morphology is much clearer from the APT results from the 90CRA condition shown in **Figure 5**. **Figure 5(a-c)** show 2D sections of maps with all ions, Ti, and Fe,



respectively. The compositions of the L1₂ and FCC phases are also listed in the table in **Figure 5(d)**. The two precipitate morphologies, i.e., spherical or blocky and lamellar, arise from two distinct transformation pathways: continuous precipitation and discontinuous precipitation. In the case of SA750 and the non-recrystallized regions of 60CRA condition, L1₂ precipitates nucleate homogenously within the grains and coarsen with annealing time. This leads to spherical or blocky precipitates, as seen in **Figure 4(b-c)**. However, when there is a driving force for recrystallization, the precipitation occurs concurrently with grain boundary migration leading to rod-like L1₂ precipitation. This reaction, known as discontinuous precipitation[19, 20] occurs in the recrystallized regions of 60CRA and 90CRA conditions.

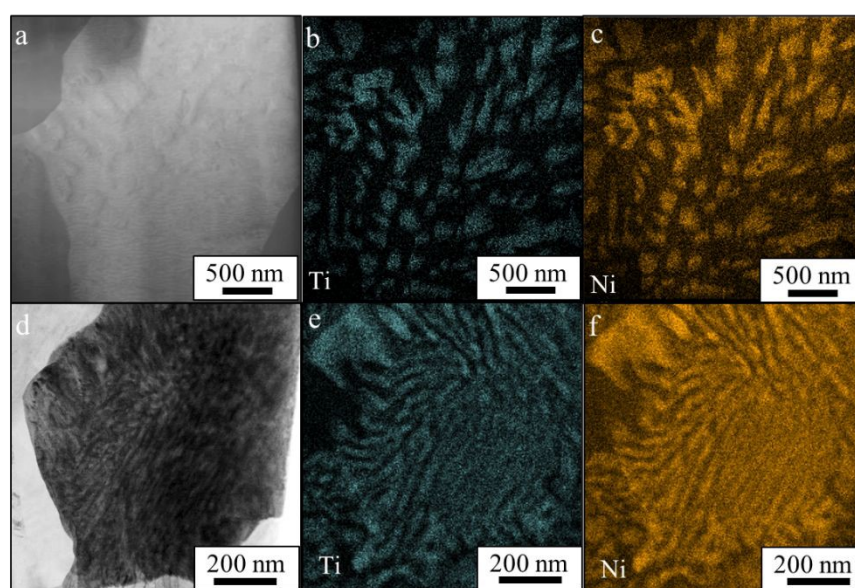


Figure 4. (a) HAADF-STEM image and (b) Ti and (c) Ni STEM-EDS maps from a non-recrystallized grain in 60CRA condition show blocky L1₂ precipitation. (d) HAADF-STEM image and (e) Ti and (f) Ni STEM-EDS maps from a recrystallized grain in 90CRA condition show rod-like L1₂ precipitation.



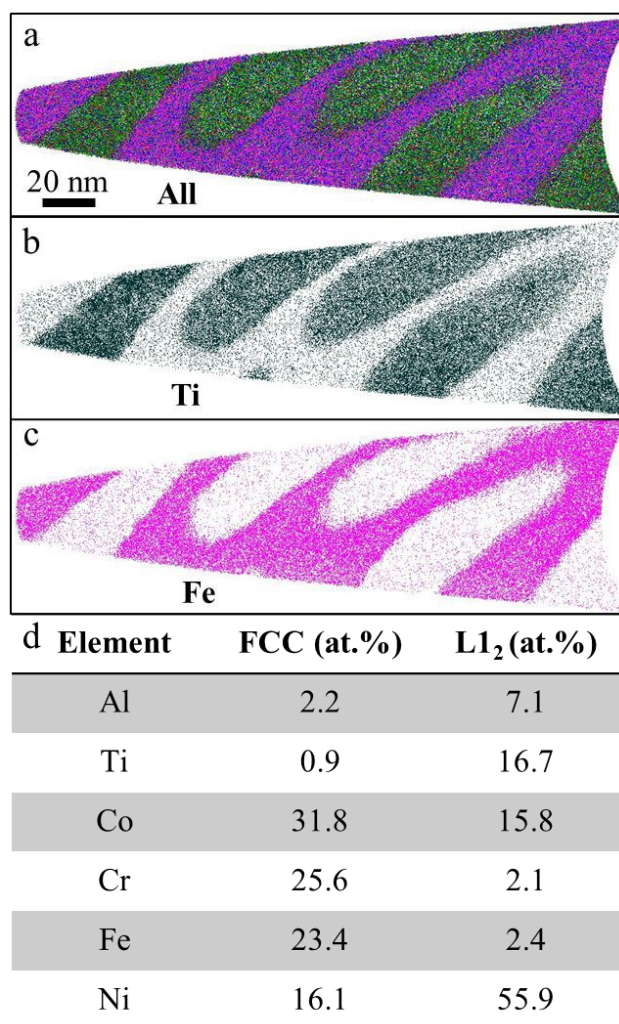


Figure 5. APT results from 90CRA condition. (a) A 2D section of the superimposition of all raw ion maps (Al=blue, Ti=dark green, Co=blue, Cr=light pink, Fe=dark pink, Ni-green) (b) Ti raw ion map section, (c) Fe raw ion map section, and (d) FCC and L1₂ compositions in the 90CRA condition.

The summary of the microstructures of the four processing conditions is as follows. The SA condition has a coarse grain structure with ~5 nm diameter L1₂ precipitates. These non-equilibrium precipitates formed during water quenching, as reported in similar alloys previously[21]. SA750 also exhibited a coarse grain structure but with a uniform distribution of equilibrium L1₂ precipitates. The average precipitate diameter in SA750 is 25 nm. The 60CRA condition exhibited a partially recrystallized microstructure. The non-recrystallized grains contained blocky L1₂ precipitates (formed via continuous precipitation) while the recrystallized grains contained rod-like L1₂ precipitates (formed via discontinuous precipitation). 90CRA condition has a fully recrystallized grain structure with rod-like L1₂



precipitates. The approximate length and diameter of nano-rod L1₂ precipitates in 90CRA condition are ~330 nm and ~30 nm, respectively. It is worth noting that the length may span up to the grain size (1.8 microns in 90CRA), but it is usually smaller due to processes such as branching and coarsening. A minor fraction of L2₁ precipitates also exist in both 60CRA and 90CRA conditions, which appear as dark polygonal or dot-like precipitates in **Figure 2**.

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3.3. Magnetic properties

Four distinct microstructural conditions, SA, SA750, 60CRA, and 90CRA, achieved via four distinct thermo-mechanical processing routes, have been discussed in the previous section. Since the SA, SA750, and 90CRA conditions exhibited a nominally homogeneous microstructure, while the 60CRA condition revealed a partially recrystallized inhomogeneous condition, this condition was not analysed further in terms of magnetic properties. The following section describes the magnetic behaviour of the SA, SA750, and 90CRA conditions.

3.3.1. Magnetic behaviour of the SA condition

The SA condition, consisting of a fully FCC matrix with a fine dispersion of L1₂ precipitates (~5 nm, ~9% volume fraction), shows a clear temperature dependence in both saturation magnetization (M_s) and coercivity (H_c) (**Figure 6 a and b**). M_s decreases gradually from 27.7 emu/g at 2 K to 26.9 emu/g at 10 K, 17.5 emu/g at 50 K, and 5.4 emu/g at 100 K, reflecting the loss of long-range magnetic order as the system approaches and exceeds its Curie temperature (T_c) (**Figure 6 b**). H_c increases from 1.6 Oe at 2 K to 4.8 Oe at 10 K, reaching a maximum of 8.5 Oe around 48 K, and then slightly decreases to 8.2 Oe at 100 K (**Figure 6 b**). The relatively high coercivity near the T_c is attributed to the increase in magnetic anisotropy and exchange stiffness, which enhances resistance to domain wall motion as the system approaches magnetic ordering. Similar behaviour has been reported in nanocrystalline soft magnetic alloys, where exchange-decoupled grains lead to independent switching and domain wall formation near the decoupling temperature[22].

The M – T curve, measured under zero-field-cooled and field-cooled protocols with a 1000 Oe applied field (**Figure 6 c**), reveals a single magnetic transition near 48 K, as indicated by the minimum in the dM/dT curve (**Figure 6 d**).



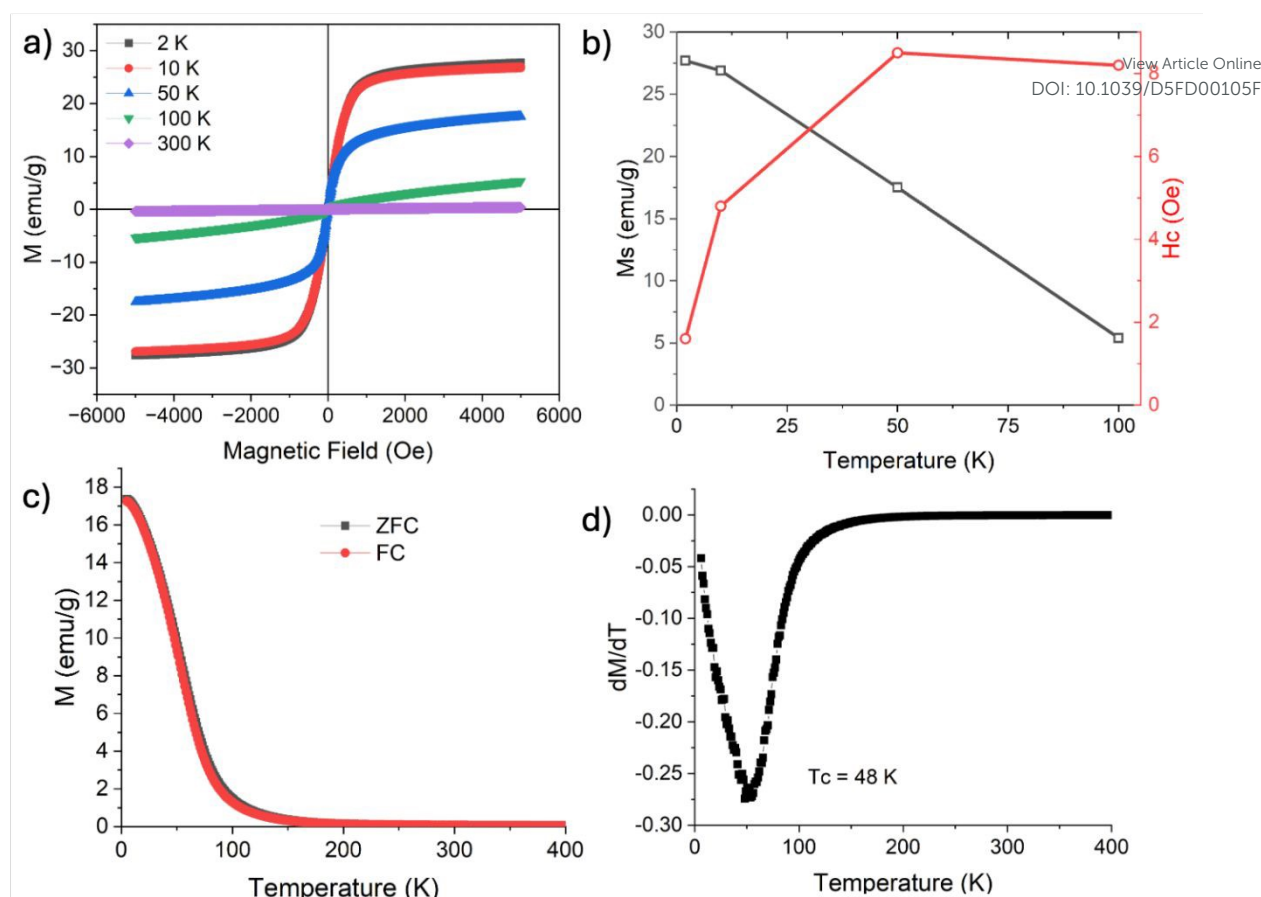


Figure 6 Magnetic behaviour of the solution-annealed (SA) condition. (a) Isothermal magnetization (M - H) loops measured at multiple temperature, showing ferromagnetic behaviour at low temperature and a paramagnetic behaviour at room temperature. (b) Temperature dependence of saturation magnetization (M_s , left axis, black squares) and coercivity (H_c , right axis, red circles) (c) Zero-field-cooled (ZFC) and field-cooled (FC) magnetization curves measured in a 1000 Oe applied field, revealing a single magnetic transition. (d) Derivative of magnetization (dM/dT) used to determine the Curie temperature, $T_c \approx 48$ K.

3.3.2. Magnetic behaviour of the SA750 condition

The SA750 condition, obtained by homogenization at 1150 °C followed by annealing at 750 °C for 50 hours, reflects the influence of a multiphase microstructure on magnetic response. M_s decreases with increasing temperature, from 17.1 emu/g at 2 K to 16.5 at 10 K, 10.1 at 50 K, and 3.7 at 100 K (**Figure 7 a and b**). These values are lower than those of the SA condition, likely due to Ni depletion and Cr enrichment in the FCC matrix after partitioning into $L1_2$. H_c is highest at 2 K (19.1 Oe), decreasing to 10.7 Oe at 10 K and reaching a minimum of 4.4 Oe at 50 K, before rising to 7.6 Oe at 100 K (**Figure 7a**). This coercivity trend differs from the SA



condition and reflects stronger domain wall pinning at low temperatures due to coarser and higher-volume-fraction $L1_2$ precipitates (~ 25 nm, $\sim 46\%$). The minimum in H_c occurs between the T_c of the $L1_2$ phase (~ 36 K) and the FCC matrix (~ 90 K), and is attributed to thermally assisted depinning as magnetic stiffness and interphase coupling are progressively weakened with increasing temperature.

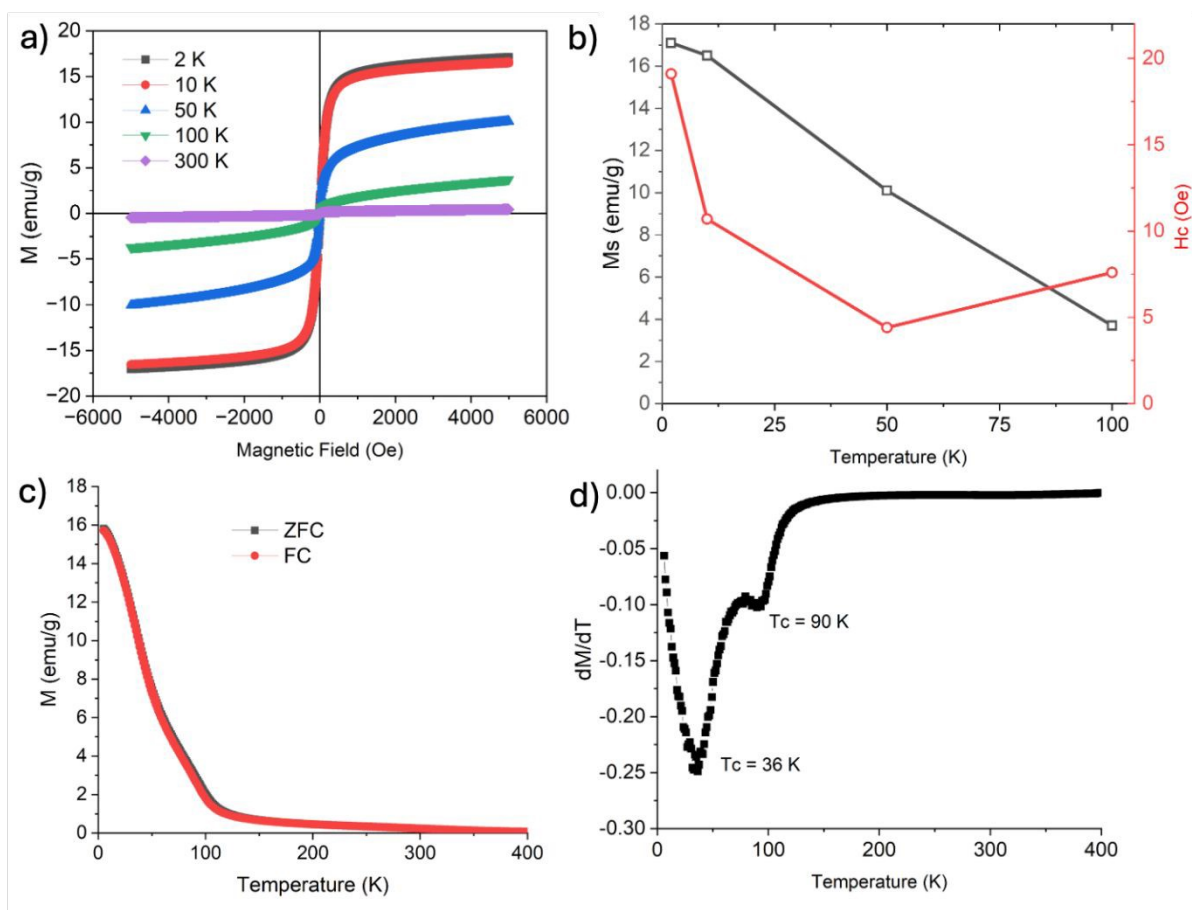


Figure 7 Magnetic behaviour of the SA750 condition. (a) M–H loops measured at different temperatures, showing ferromagnetic behaviour at low temperatures with decreasing magnetization at higher temperatures. (b) Temperature dependence of saturation magnetization (M_s , left axis, black squares) and coercivity (H_c , right axis, red circles). (c) ZFC and FC magnetization curves recorded under a 1000 Oe applied field, showing a gradual decrease in magnetization with increasing temperature. (d) First derivative of magnetization (dM/dT), indicating two distinct magnetic transitions with Curie temperatures (T_c) at ~ 36 K and ~ 90 K.

The M–T curve recorded under ZFC and FC protocols in an applied field of 1000 Oe shows two magnetic transitions (**Figure 7c**). The higher T_c (90 K), identified from the dM/dT curve



(**Figure 7d**), corresponds to the FCC matrix, which becomes enriched in Fe and Co due to Ni, Al and Ti partitioning into the $L1_2$ phase. The lower transition at ~ 36 K likely arises from magnetic ordering within the $L1_2$ precipitates, which become partially decoupled from the matrix as their size and volume increase. Their composition, enriched in Al and Ti but still containing ferromagnetic Ni and Co, supports localized magnetic ordering distinct from the matrix.

3.3.3. Magnetic behaviour of the 90CRA condition

The 90CRA condition, produced by 90 % cold rolling followed by annealing at 750 °C for 50 hours, exhibits a distinct magnetic response governed by its fine-grained, recrystallized microstructure and discontinuous precipitation of the $L1_2$ phase. The saturation magnetization decreases with increasing temperature, from 14.9 emu/g at 2 K to 12.3 at 10 K, 8.3 at 50 K, and 3.1 at 100 K (**Figure 8 a and b**). These values are lower than those observed in SA and SA750, consistent with differences in FCC matrix composition and phase connectivity arising from processing history.

Among the three conditions studied, 90CRA exhibits the highest coercivity values. At 2 K, H_c reaches 41.7 Oe, significantly higher than both SA750 and SA, indicating strong resistance to domain wall motion. This is attributed to effective pinning by rod-shaped $L1_2$ precipitates that form along grain boundaries during annealing. With increasing temperature, H_c decreases steadily to 28.8 Oe at 10 K, 23.8 Oe at 50 K, and 2.5 Oe at 100 K (**Figure 8 b**). H_c decreases steadily with temperature, in contrast to the nonmonotonic trend observed in SA750. This behaviour is attributed to strong domain wall pinning by rod-shaped $L1_2$ precipitates, which becomes progressively less effective with increasing temperature[23]. The continuous decline in H_c across the two magnetic transitions (~ 29 K for $L1_2$ and ~ 84 K for the FCC matrix) reflects a gradual reduction in pinning strength and magnetic coupling in a more heterogeneous microstructural environment.

The magnetization versus temperature curves measured under ZFC and FC protocols in a 1000 Oe field reveal two magnetic transitions (**Figure 8 c**). The higher T_c around 84 K corresponds to the FCC matrix, while the lower transition near 29 K is attributed to magnetic ordering in the $L1_2$ precipitates (**Figure 8 d**). These precipitates, enriched in Al and Ti while retaining Ni



and Co, become magnetically decoupled due to their size and boundary-localized distribution, giving rise to a distinct low temperature transition.

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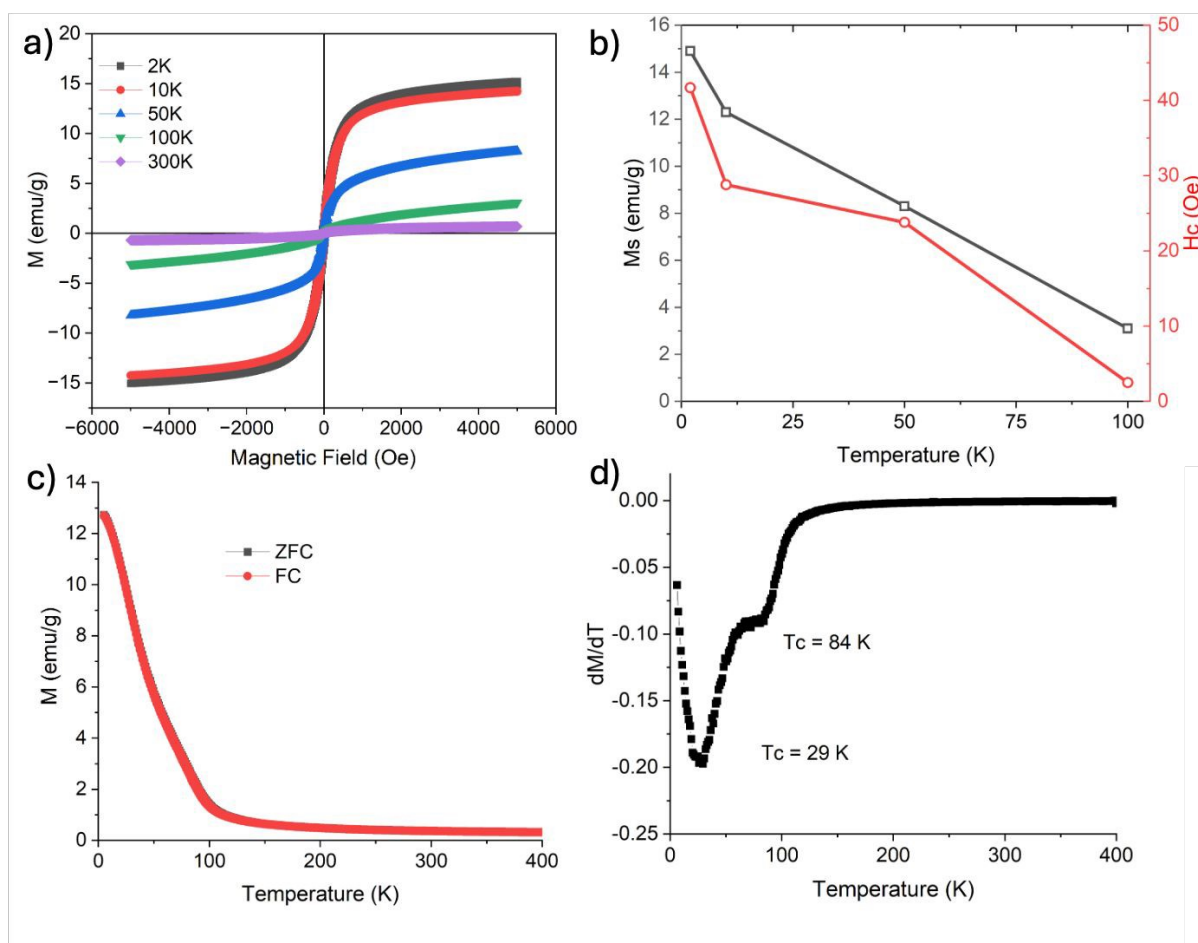


Figure 8 Magnetic behaviour of the 90CRA condition. (a) M–H loops measured at 2, 10, 50, 100, and 300 K. (b) Temperature dependence of saturation magnetization (M_s) and coercivity (H_c), indicating a decrease in M_s and a decreasing trend in H_c with increasing temperature. (c) ZFC and FC magnetization curves recorded under a 1000 Oe field, showing a continuous decrease in magnetization with temperature. (d) dM/dT plot highlighting two magnetic transitions, with Curie temperatures identified at ~ 84 K and 29 K.

3.3.4. Room temperature magnetic behavior

The M–H loops measured at 300 K for all three conditions are shown in **Figure 9**. Although the overall magnetization remains low, clear differences in coercivity are evident (inset of **Figure 9**). The SA condition displays an almost linear response with negligible hysteresis, confirming its paramagnetic nature at room temperature. In contrast, both SA750 and 90CRA exhibit measurable coercivity, with 90CRA showing the largest hysteresis. The weak hysteresis



in SA750, despite the absence of any identified secondary magnetic phase, may arise from residual short-range magnetic interactions or frozen magnetic clusters in the FCC or L_{12} phases. In 90CRA, however, the higher coercivity is consistent with the presence of a minor L_{21} Heusler-like phase at grain boundaries, previously identified by electron diffraction and EDS[7, 24], as well as possible discontinuities or magnetic inhomogeneities in the FCC matrix introduced by the rod-shaped precipitates.

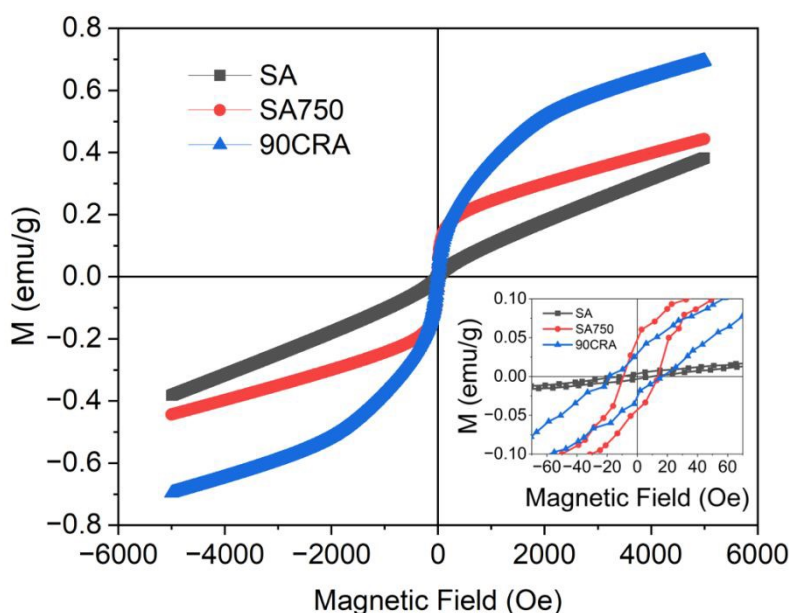


Figure 9 Room temperature M–H loops for SA, SA750, and 90CRA. (a) Full loops showing low magnetization across all conditions. (b) Zoomed view of the central region highlights negligible hysteresis in SA and finite coercivity in both SA750 and 90CRA.

Conclusion

The magnetic properties of the $\text{Al}_{0.2}\text{Ti}_{0.3}\text{Co}_{1.5}\text{CrFeNi}_{1.5}$ alloy were examined in three processing conditions: solution-annealed (SA), annealed at 750 °C (SA750), and cold-rolled plus annealed (90CRA). The SA condition exhibited a single magnetic transition near 48 K and low coercivity, consistent with a fully FCC matrix containing fine, coherent L_{12} precipitates. In SA750 and 90CRA, the growth and redistribution of the L_{12} phase led to magnetic decoupling, reflected in the emergence of two magnetic transitions and an increase in coercivity. The highest low temperature coercivity was observed in 90CRA, where rod-shaped L_{12} precipitates formed along grain boundaries. At 300 K, weak hysteresis was present in both SA750 and 90CRA. In the latter, this behaviour is attributed to a minor L_{21} Heusler-like phase located at grain boundaries. These results demonstrate that the magnetic response in this alloy system can



be tuned through processing, by controlling phase morphology and chemical partitioning. More broadly, the findings highlight how nanoscale chemical ordering and microstructural architecture can be exploited to design magnetic materials with tailored coercivity and transition behaviour for cryogenic and multifunctional applications.

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References

- [1] Cantor B, Chang ITH, Knight P, Vincent AJB. Microstructural development in equiatomic multicomponent alloys. *Materials Science and Engineering: A*. 2004;375-377:213-8.
- [2] Yeh J-W, Chen S-K, Lin S-J, Gan J-Y, Chin T-S, Shun T-T, et al. Nanostructured High-Entropy Alloys with Multiple Principal Elements: Novel Alloy Design Concepts and Outcomes. *Advanced Engineering Materials*. 2004;6:299-303.
- [3] Miracle DB, Senkov ON. A critical review of high entropy alloys and related concepts. *Acta Materialia*. 2017;122:448-511.
- [4] George EP, Raabe D, Ritchie RO. High-entropy alloys. *Nature Reviews Materials*. 2019;4:515-34.
- [5] Chaudhary V, Chaudhary R, Banerjee R, Ramanujan RV. Accelerated and conventional development of magnetic high entropy alloys. *Materials Today*. 2021;49:231-52.
- [6] Liu WH, Yang T, Liu CT. Precipitation hardening in CoCrFeNi-based high entropy alloys. *Materials Chemistry and Physics*. 2018;210:2-11.
- [7] Dasari S, Chang YJ, Jagetia A, Soni V, Sharma A, Gwalani B, et al. Discontinuous precipitation leading to nano-rod intermetallic precipitates in an Al_{0.2}Ti_{0.3}Co_{1.5}CrFeNi_{1.5} high entropy alloy results in an excellent strength-ductility combination. *Materials Science and Engineering: A*. 2021;805:140551.



[8] Xu X, Davidson KP, Tan LP, Gorsse S, Chaudhary V, Ramanujan RV. Improvement in mechanical as well as magnetic properties of a $(\text{FeCoNi})_{90}\text{Ti}_{10-x}\text{Al}_x$ complex concentrated alloy series by tuning the chemical order. *Scripta Materialia*. 2025;254:116333.

View Article Online
DOI: 10.1016/j.scriptamat.2025.116333

[9] Kovács A, Venkataraman NB, Chaudhary V, Dasari S, Denneulin T, Ramanujan RV, et al. Role of heterophase interfaces on local coercivity mechanisms in the magnetic $\text{Al}_{0.3}\text{CoFeNi}$ complex concentrated alloy. *Acta Materialia*. 2023;246:118672.

[10] Chaudhary V, Nartu MSKKY, Dasari S, Varahabhatla SM, Sharma A, Radhakrishnan M, et al. Magnetic and mechanical properties of additively manufactured $\text{Al}_x(\text{CoFeNi})$ complex concentrated alloys. *Scripta Materialia*. 2023;224:115149.

[11] Han L, Rao Z, Souza Filho IR, Maccari F, Wei Y, Wu G, et al. Ultrastrong and Ductile Soft Magnetic High-Entropy Alloys via Coherent Ordered Nanoprecipitates. *Advanced Materials*. 2021;33:2102139.

[12] Han L, Maccari F, Souza Filho IR, Peter NJ, Wei Y, Gault B, et al. A mechanically strong and ductile soft magnet with extremely low coercivity. *Nature*. 2022;608:310-6.

[13] Dasari S, Chaudhary V, Gwalani B, Jagetia A, Soni V, Gorsse S, et al. Highly tunable magnetic and mechanical properties in an $\text{Al}_{0.3}\text{CoFeNi}$ complex concentrated alloy. *Materialia*. 2020;12:100755.

[14] Chaudhary V, Nirmala R, Huang Y. Magnetic materials and devices. *Materials Research Express*. 2025;12:010201.

[15] Teh WH, Tan LP, Chen S, Wei F, Lee JJ, Padhy SP, et al. Breaking conventional limits of silicon content in Fe-xSi magnetic alloys through additive manufacturing. *Journal of Alloys and Compounds*. 2024;983:173829.

[16] Chaudhary V, Soni V, Gwalani B, Ramanujan RV, Banerjee R. Influence of non-magnetic Cu on enhancing the low temperature magnetic properties and Curie temperature of $\text{FeCoNiCrCu}(x)$ high entropy alloys. *Scripta Materialia*. 2020;182:99-103.

[17] Zhou Z-Z, Xu L-J, Liu Y-T, Hao L-J, Zhang Y-H, Luo R, et al. Magnetic Shield Design and Simulation Optimization of Metallic Magnetic Calorimeter in Ultra-Low Temperature Working State. *Journal of Low Temperature Physics*. 2024;215:46-63.

[18] Pelliccione M, Jenkins A, Ovartchaiyapong P, Reetz C, Emmanouilidou E, Ni N, et al. Scanned probe imaging of nanoscale magnetism at cryogenic temperatures with a single-spin quantum sensor. *Nature Nanotechnology*. 2016;11:700-5.

[19] Williams DB, and Butler EP. Grain boundary discontinuous precipitation reactions. *International Metals Reviews*. 1981;26:153-83.



[20] Manna I, K. PS, and Gust W. Discontinuous reactions in solids. *International Materials Reviews*. 2001;46:53-91.

View Article Online
DOI: 10.1039/D5FD00105F

[21] Dasari S, Jagetia A, Sharma A, Nartu MSKKY, Soni V, Gwalani B, et al. Tuning the degree of chemical ordering in the solid solution of a complex concentrated alloy and its impact on mechanical properties. *Acta Materialia*. 2021;212:116938.

[22] Škorvánek I, O'Handley RC. Fine-particle magnetism in nanocrystalline Fe-Cu-Nb-Si-B at elevated temperatures. *Journal of Magnetism and Magnetic Materials*. 1995;140-144:467-8.

[23] Herzer G. Modern soft magnets: Amorphous and nanocrystalline materials. *Acta Materialia*. 2013;61:718-34.

[24] Ghosh S, Sangwan S, Mandal S, Datta S, Kar M, Singh P, et al. Room temperature giant magneto-caloric effect in Ni₄₅Mn₄₄Sn₁₁-XInX (X = 1, 3) disordered Heusler alloy: The role of martensite transition. *Journal of Magnetism and Magnetic Materials*. 2022;562:169797.



Data availability

The authors confirm that the raw data supporting the findings of this study are available from the corresponding author, upon reasonable request.

