

# Effects of heat treatment on microstructure and damping capacity of cast $\text{Fe}_{42.75}\text{Mn}_{19}\text{Co}_{19}\text{Cr}_{14.25}\text{Al}_5$ high-entropy alloy

Peng Gao<sup>1</sup>, Hao-yang Song<sup>2</sup>, Jia-hui Zhu<sup>1</sup>, Jian Shi<sup>1</sup>, Jing-chang Cheng<sup>1</sup>, Zhong-han Liu<sup>1</sup>, \*Bo Yu<sup>1</sup>, and \*\*Ke-qiang Qiu<sup>2</sup>

1. State Key Laboratory of Advanced Casting Technologies, Shenyang Research Institute of Foundry Co., Ltd. CAM, Shenyang 110022, China

2. School of Materials Science and Engineering, Shenyang University of Technology, Shenyang 110870, China

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**Abstract:** Alloys with good damping capacity are urgently needed to eliminate vibrations and noise in various situations. To explore the damping potential of the  $\text{Fe}_{42.75}\text{Mn}_{19}\text{Co}_{19}\text{Cr}_{14.25}\text{Al}_5$  high-entropy alloy, the effects of annealing (from 800 to 1,100 °C for 1 h) on the phase constitution, microstructure, damping capacity ( $Q^{-1}$ ), and tensile properties of the  $\text{Fe}_{42.75}\text{Mn}_{19}\text{Co}_{19}\text{Cr}_{14.25}\text{Al}_5$  high-entropy alloy were investigated. Results indicate that annealing at 800 and 900 °C removes the minor unstable body-centered cubic (BCC) constituent, yielding a single-phase face-centered cubic (FCC) matrix. Raising the temperature to 1,000 or 1,100 °C promotes the precipitation of a Cr-rich Laves phase within the FCC matrix. The  $Q^{-1}$  peaks of 0.0636 at 900 °C, attributable to energy dissipation associated with twin boundaries and dislocation structures. Compared with the as-cast state, the alloy annealed at 900 °C shows a decrease in yield strength from  $270 \pm 6$  to  $146 \pm 4$  MPa and in ultimate tensile strength from  $560 \pm 3$  to  $520 \pm 3$  MPa, while its elongation increases from  $(33.2 \pm 0.4)\%$  to  $(37.1 \pm 0.4)\%$ . This combination of high damping and balanced mechanical properties indicates high potential for engineering applications.

**Keywords:** high-entropy alloy; damping capacity; dislocations; twinning

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## 1 Introduction

High-entropy alloys (HEAs), typically composed of five or more principal metallic elements with atomic concentrations ranging from 5at.% to 35at.%, generally favor the formation of single-phase solid solution structures<sup>[1-3]</sup>. Compared to traditional alloys, the high-entropy design expands the compositional space, delivering superior strength-ductility synergy<sup>[4-8]</sup>, high-temperature stability<sup>[9-11]</sup>, and wear and corrosion resistance<sup>[12, 13]</sup>. These exceptional properties stem from core HEA effects: high entropy, severe lattice distortion, and sluggish diffusion.

Recently, considerable attention has been directed toward the damping performance of HEAs<sup>[14-17]</sup>. The damping capacity of these alloys, i.e., their ability to suppress vibrations by effectively absorbing and dissipating vibrational energy, can significantly mitigate noise and potential structural damage. Consequently, high-entropy damping alloys hold great promise for applications in vibration- and noise-sensitive fields, including aerospace, automotive manufacturing, and mechanical engineering.

The damping performance of HEAs can be tailored via microstructural design<sup>[14, 15, 18]</sup> or by eliminating crystal defects through heat treatment<sup>[19]</sup>. For example, research has shown that the damping capacity of a  $\text{Fe}_{30}\text{Cr}_{25}\text{Ni}_{25}\text{Mn}_{15}\text{Co}_5$  alloy produced via spark plasma sintering was significantly enhanced following annealing<sup>[18]</sup>. Although casting is the most prevalent method for preparing HEAs, few studies have focused on improving the damping properties of cast HEAs by directly manipulating microstructural defects via heat treatment. A critical question remains: how do heat treatments alter the phase constituents and

### \*Bo Yu

Born in 1963, Researcher. His research interests primarily focus on advanced functional materials, high-entropy alloys, damping alloys, and advanced forming technologies.

E-mail: Yub@chinasrif.com

### \*\*Ke-qiang Qiu

E-mail: kqiu@sut.edu.cn

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microstructural defects in cast HEAs, and how do these alterations, in turn, govern their damping behavior?

To address this, a  $\text{Fe}_{42.75}\text{Mn}_{19}\text{Co}_{19}\text{Cr}_{14.25}\text{Al}_5$  HEA was designed and subjected to annealing across a range of temperatures (800 °C–1,100 °C) in this work. Comprehensive characterization techniques, including X-ray diffractometry, scanning electron microscopy, electron backscatter diffraction, and transmission electron microscopy were employed to systematically compared the phase constitution, microstructural evolution, damping capacity, and tensile properties between the as-cast and annealed states. Through this systematic investigation, the underlying mechanisms governing the synergistic enhancement of damping capacity and plasticity are revealed, aiming to provide a viable pathway for tailoring the multifunctional performance of Fe-Mn-Co-Cr-Al based HEA alloys.

## 2 Experimental procedures

The HEA samples with a nominal composition of  $\text{Fe}_{42.75}\text{Mn}_{19}\text{Co}_{19}\text{Cr}_{14.25}\text{Al}_5$  (at.%) were prepared. The fabrication process was as follows: Master alloy ingots were produced by melting high-purity ( $\geq 99.9\text{wt.}\%$ ) elemental blocks of Al, Fe, Co, and Cr, alongside an Fe-Mn intermediate alloy (containing 45.75wt.% Mn) in a non-consumable vacuum arc furnace under a high-purity argon atmosphere. To ensure compositional homogeneity, each ingot was flipped and remelted at least four times under electromagnetic stirring. Subsequently, the crushed master alloy chunks were remelted in a vacuum induction furnace under an argon atmosphere and cast into rectangular plates (50 mm $\times$ 14 mm $\times$ 2 mm) via copper mold injection casting. The as-cast samples were then annealed in a vacuum resistance furnace at 800, 900, 1,000, and 1,100 °C, respectively for 1 h, followed by air cooling. This specific temperature range (800–1,100 °C) was designed to encompass the complete phase transformation process, from microstructural homogenization to second-phase precipitation.

Phase identification was performed using a SHIMADZU 7000 X-ray diffractometer (XRD) equipped with  $\text{Cu K}\alpha$  radiation ( $\lambda=0.15406$  Å) at a scanning rate of  $4^\circ\cdot\text{min}^{-1}$ . Microstructural characterization was conducted using a Gemini SEM-300 scanning electron microscope (SEM) equipped with an energy-dispersive spectrometer (EDS), an electron backscatter diffraction (EBSD) system, and a JEM-2100 transmission electron microscope (TEM). For TEM and EBSD analysis, the specimens were mechanically ground to a thickness of 50  $\mu\text{m}$ , punched into 3 mm diameter disks, and subsequently ion-milled using a Gatan 695 precision ion polishing system. SEM samples were mechanically polished and etched with aqua regia ( $\text{HCl}:\text{HNO}_3=3:1$  by volume).

The damping capacity was evaluated using a Q800 dynamic mechanical analyzer (DMA) on specimens

measuring 35 mm $\times$ 14 mm $\times$ 2 mm. The tests were performed in a single cantilever beam mode at a frequency of 1 Hz. Room-temperature tensile properties were measured using a WDW-100 universal testing machine at a constant strain rate of  $1\times 10^{-3}\text{ s}^{-1}$ , using dog-bone shaped specimens with gauge dimensions of 4 mm $\times$ 2 mm $\times$ 7 mm ( $a\times b\times L_0$ ).

## 3 Results and discussion

### 3.1 Microstructure evolution and phase constitution

Figure 1 presents the XRD patterns of the  $\text{Fe}_{42.75}\text{Mn}_{19}\text{Co}_{19}\text{Cr}_{14.25}\text{Al}_5$  alloy in both the as-cast and annealed states. The XRD patterns of the as-cast alloy reveal characteristic peaks corresponding to a face-centered cubic (FCC) phase. Following annealing at 800 and 900 °C, the intensity of the FCC diffraction peaks initially increases and then decreases, implying a temperature-dependent change in the phase constitution. However, upon annealing at 1,000 °C, a distinct diffraction peak corresponding to the (004) crystal plane emerges, signifying the precipitation of a new phase. When the temperature is further increased to 1,100 °C, the (004) peak intensifies, and an additional peak corresponding to the (103) plane appears. Comparison with the Joint Committee on Powder Diffraction Standards (JCPDS) database confirms that these new peaks are attributed to the Laves phase. These findings demonstrate a distinct evolution in the phase constitution, driven by the high-temperature formation of the Laves phase.

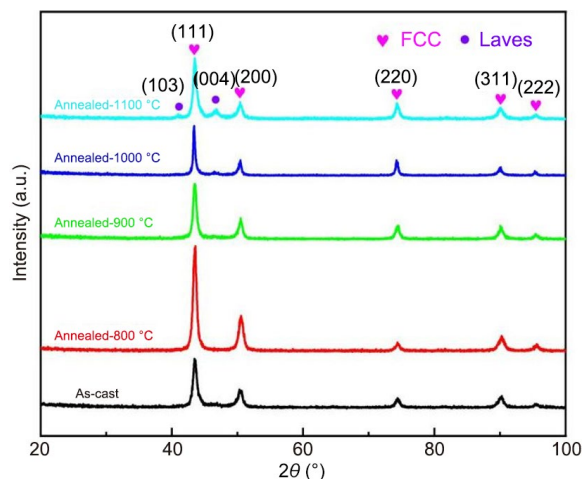


Fig. 1: XRD patterns of  $\text{Fe}_{42.75}\text{Mn}_{19}\text{Co}_{19}\text{Cr}_{14.25}\text{Al}_5$  HEA in as-cast and annealed states

Figure 2 presents the EBSD phase maps of the  $\text{Fe}_{42.75}\text{Mn}_{19}\text{Co}_{19}\text{Cr}_{14.25}\text{Al}_5$  HEA in both the as-cast and annealed states. As depicted in Fig. 2(a), the as-cast HEA is primarily composed of an FCC matrix phase (99.7vol.%) with a minor fraction of a body-centered cubic (BCC) second phase. The discrepancy in phase identification between the XRD results (indicating a single FCC phase, Fig. 1) and the EBSD maps [revealing an FCC+BCC dual-phase structure, Fig. 2(a)] is likely attributable to the limited detection resolution of

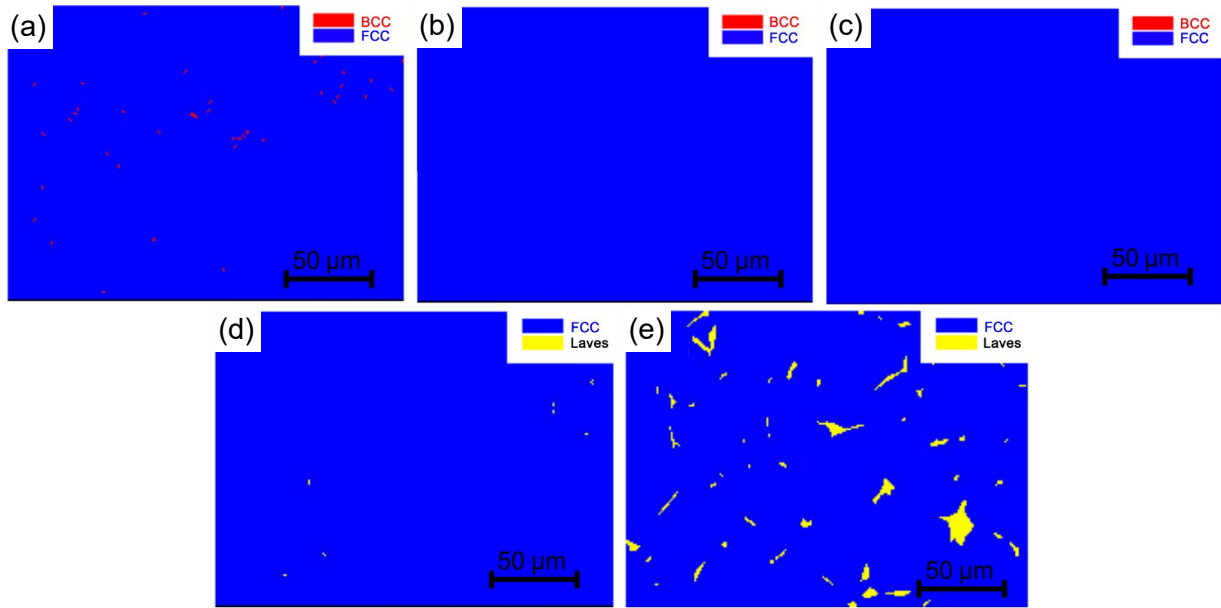


Fig. 2: EBSD maps of  $\text{Fe}_{42.75}\text{Mn}_{19}\text{Co}_{19}\text{Cr}_{14.25}\text{Al}_5$  HEA as-cast (a) and annealed at 800 °C (b), 900 °C (c), 1,000 °C (d), and 1,100 °C (e), respectively for 1 h

conventional XRD for trace phases. Following annealing at 800 °C or 900 °C, this minor BCC phase completely transforms into the FCC phase, resulting in a single-phase FCC microstructure [Figs. 2(b) and (c)]. The absence of BCC reprecipitation during subsequent cooling further validates its thermodynamic instability within the as-cast alloy. When the annealing temperature is elevated to 1,000 °C, a small amount of Laves phase begins to precipitate, as illustrated in Fig. 2(d). Upon further annealing at 1,100 °C, the volume fraction of the Laves phase significantly increases to ~7vol.%, consistent with the enhanced Laves phase diffraction peaks observed in the XRD analysis. Overall, the EBSD results align well with the XRD data for all conditions, providing robust evidence for the phase transformation sequence in this HEA.

Figure 3 illustrates the microstructural evolution of the

$\text{Fe}_{42.75}\text{Mn}_{19}\text{Co}_{19}\text{Cr}_{14.25}\text{Al}_5$  alloy in both the as-cast and annealed states. As observed in the EBSD phase map [Fig. 2(a)], the as-cast alloy predominantly consists of a striated FCC phase. The distinct orientations of these striations correspond to different grains. After annealing at 800 °C, the microstructure remains entirely composed of the striated FCC phase, with no discernible grain growth observed [Fig. 3(b)].

Upon annealing at 900 and 1,000 °C, the microstructure transitions to an equiaxed grain structure containing a small fraction of intragranular twin boundaries [Figs. 3(c) and (d)]. However, as the annealing temperature is further increased to 1,100 °C, distinct second-phase precipitates within the alloy. Combined EBSD (Fig. 2) and energy-dispersive X-ray spectroscopy (EDS) analysis (Fig. 4) verify that this newly formed second phase is a Cr-rich Laves phase. The elemental

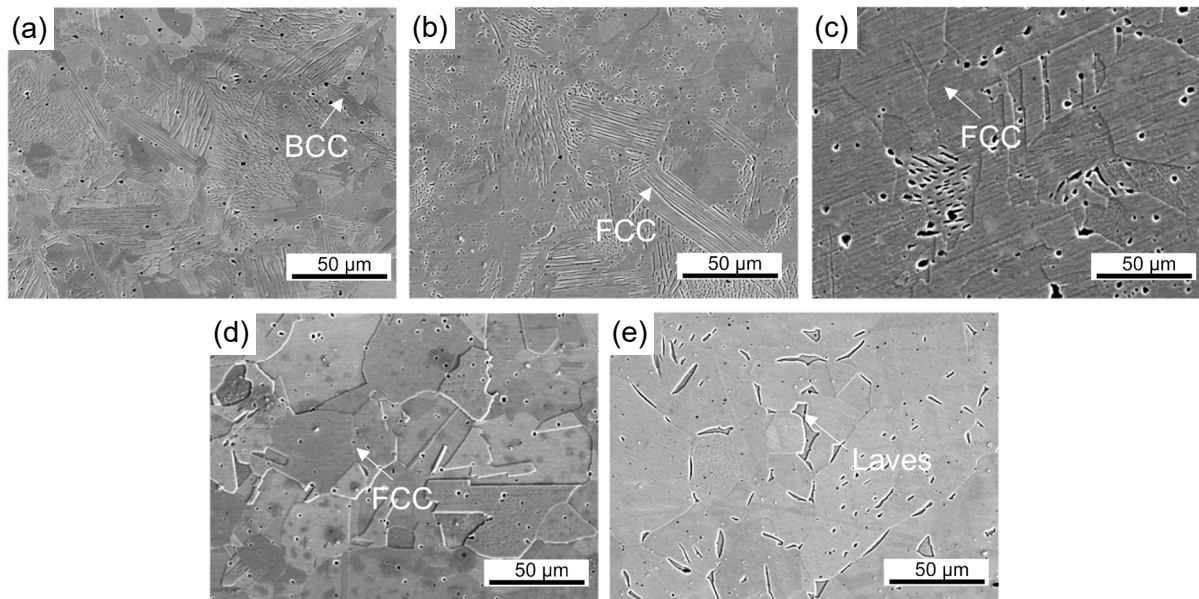


Fig. 3: SEM images of  $\text{Fe}_{42.75}\text{Mn}_{19}\text{Co}_{19}\text{Cr}_{14.25}\text{Al}_5$  HEA as-cast (a) and annealed at 800 °C (b), 900 °C (c), 1,000 °C (d), and 1,100 °C (e), respectively for 1 h

distribution maps (Fig. 4) demonstrate a uniform distribution of Fe, Mn, Co, and Al within the matrix, whereas Cr exhibits distinct enrichment regions corresponding to the Laves phase. Additionally, localized Al-rich spots (observed as red regions in the Al map, Fig. 4) correspond to the black particulate features in Fig. 3(e). These are likely  $\text{Al}_2\text{O}_3$  inclusions formed by the reaction between Al and residual oxygen in the melt during the preparation of the master alloy ingots.

Figure 5(a) shows a bright-field TEM micrograph of the as-cast alloy. The as-cast microstructure primarily consists of an FCC matrix containing twin boundaries, which are marked by blue dotted lines. Figure 6 presents selected diffraction spots corresponding to the BCC phase, FCC phase, twin structure, and Laves phase. The red dashed square in Fig. 5(a) highlights the FCC matrix region selected for SAED analysis, and the corresponding SAED pattern is displayed in Fig. 6(b). Diffraction features associated with twins are illustrated in Fig. 6(c). Furthermore, a small amount of the BCC phase, marked by the yellow dashed square in Fig. 5(a), is observed within the FCC matrix; its relevant SAED pattern is provided in Fig. 6(a).

The microstructural evolution of the annealed alloys is

shown in Figs. 5(b–e). After annealing at 800 °C [Fig. 5(b)], the microstructure mainly consists of a single FCC phase, as shown in the red dashed square, which confirmed by the SAED pattern in Fig. 6(b). Local dislocation-cell or dislocation-network features can be observed near the region marked by the yellow arrow. Upon increasing the annealing temperature to 900 °C [Fig. 5(c)], distinguishable twin boundaries are observed in the FCC matrix, indicating the formation of annealing twins during heat treatment. It is generally accepted that twinning in FCC alloys is favored when the stacking fault energy (SFE) falls within a suitable range, typically  $18\text{--}45\text{ mJ}\cdot\text{m}^{-2[20]}$ , and similar twinning behavior has also been reported in other Co-containing alloys<sup>[21–25]</sup>. Further annealing at 1,000 °C [Fig. 5(d)] retains distinct twin lamellae in the FCC matrix. The orange dashed square marks the local secondary-phase or FCC/Laves interfacial region for SAED analysis, and the diffraction pattern in Fig. 6(d) confirms the Laves-phase formation. This observation is consistent with the XRD and EBSD results, which indicate the initial precipitation of the Laves phase at 1,000 °C. At 1,100 °C [Fig. 5(e)], no obvious twin lamellae are observed in the examined TEM field. Instead,

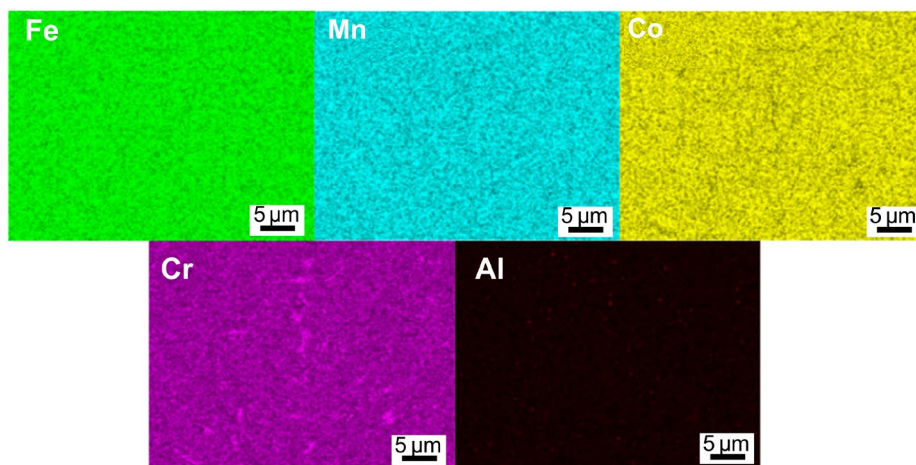


Fig. 4: Elemental distribution in  $\text{Fe}_{42.75}\text{Mn}_{19}\text{Co}_{19}\text{Cr}_{14.25}\text{Al}_5$  alloy after annealing at 1,100 °C

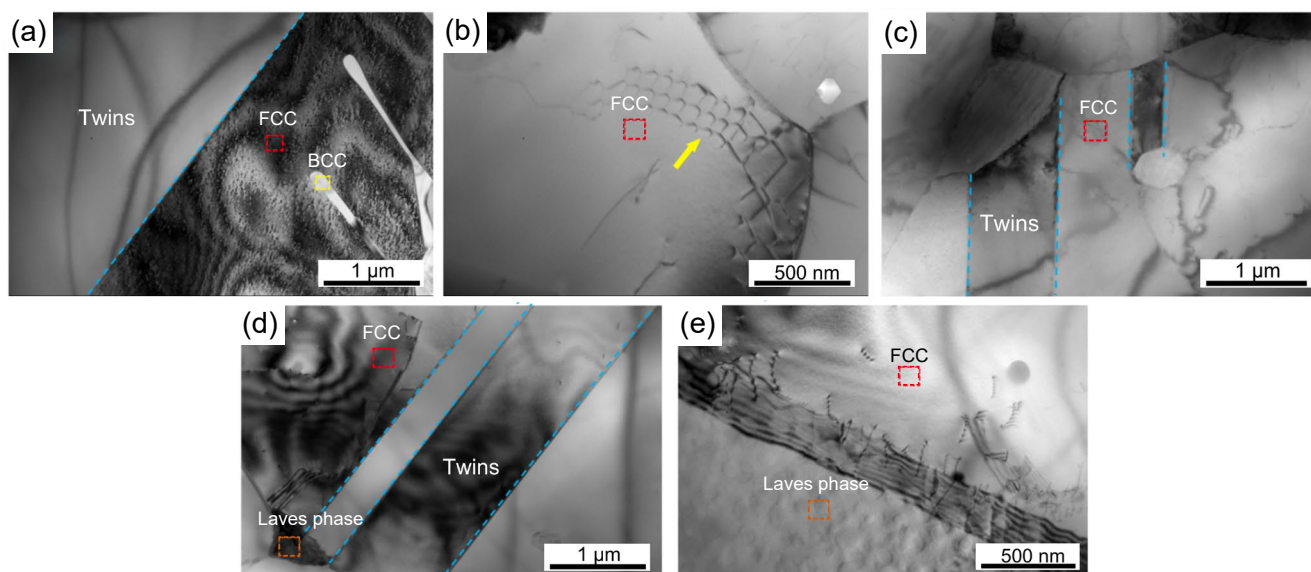


Fig. 5: TEM bright-field images of  $\text{Fe}_{42.75}\text{Mn}_{19}\text{Co}_{19}\text{Cr}_{14.25}\text{Al}_5$  HEA as-cast (a) and annealed at 800 °C (b), 900 °C (c), 1,000 °C (d), and 1,100 °C (e)

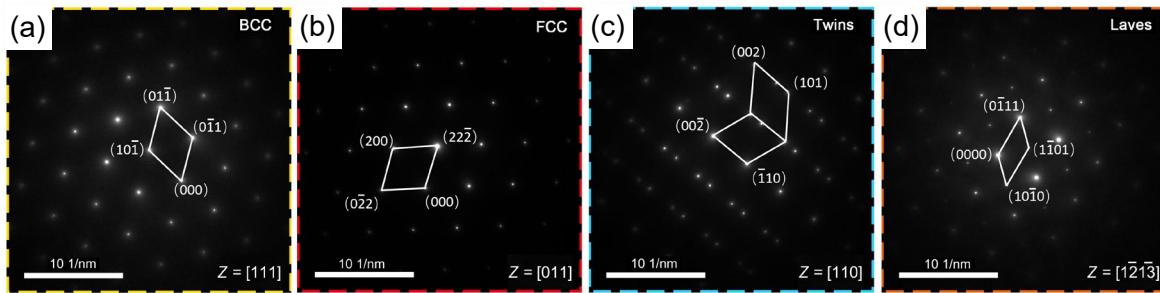


Fig. 6: Selected diffraction spots of BCC phase (a), FCC phase (b), twin crystal structure (c), and Laves phase (d)

dislocation walls or dislocation tangles are observed near the interface regions associated with the FCC matrix and the precipitated Laves phase. Overall, the TEM observations are consistent with the preceding SEM, EBSD, and XRD analysis.

### 3.2 Damping property

Figure 7 illustrates the variations of damping capacity ( $Q^{-1}$ ) of the alloy in different states with respect to the strain amplitude ( $\gamma$ ). It is evident that the damping capacity of the as-cast alloy shows a monotonically increasing trend with the increasing strain amplitude. The damping capacity increases rapidly when the strain amplitude is less than  $2 \times 10^{-4}$ , yet the rate of increase decelerates as the strain amplitude exceeds  $2 \times 10^{-4}$ . At a strain amplitude of  $7.5 \times 10^{-4}$ , the damping capacity for the as-cast alloy reaches its peak value of 0.04850. In comparison with the as-cast alloy, the annealed alloys exhibit significantly improved damping properties. The crucial disparity lies in the strain amplitude within the range of less than  $3 \times 10^{-4}$ , resulting in a more pronounced enhancement in damping capacity. When all the damping capacity values attain their maxima within a small strain amplitude range, they subsequently exhibit a gradual decline. Specifically, the damping capacity of the sample annealed at 900 °C for 1 h reaches the maximum value of 0.06364 at strain amplitude of  $3.299 \times 10^{-4}$ . While at 800; 1,000; and 1,100 °C, the maximum damping capacity values are 0.06188, 0.05774, and 0.05586, respectively, with corresponding strain amplitudes of  $2.706 \times 10^{-4}$ ,  $4.106 \times 10^{-4}$ , and  $3.089 \times 10^{-4}$ . The variation in the strain amplitude corresponding to the maximum damping capacity implies

that the microstructural defects are sensitive to the annealing temperature.

Although SEM images reveal minor micropores, the identical fabrication parameters ensured a consistent baseline porosity across all samples. Since the subsequent heat treatment exerts a negligible effect on pore distribution, the observed variations in damping capacity are primarily attributed to the evolution of phase constituents (FCC/BCC/Laves) and defect structures (dislocations/twins). Thus, the influence of porosity on the observed trends is considered limited.

Under a low strain amplitude of less than  $2 \times 10^{-4}$ , damping capacity exhibits a rapid increase, with the hysteresis effect being the dominant factor. This phenomenon is consistent with the internal stress theory model, known as the S-B model<sup>[26]</sup>:

$$Q^{-1} = \frac{4KE^2\lambda\gamma}{3\pi\sigma_i} \quad (1)$$

where  $K$  represents a dimensionless constant,  $E$  denotes the Young's modulus,  $\lambda$  signifies the magnetostrictive coefficient, and  $\sigma_i$  stands for the average internal stress. Based on Eq. (1), the damping capacity is directly proportional to the strain amplitude and inversely proportional to the stress. This implies that lower average internal stress and greater strain amplitude result in better damping behavior. Compared to the as-cast alloy, the slope of the damping capacity curves in the low strain amplitude region for the annealed alloys in Fig. 7 increases significantly. This is mainly due to the growth in grain size resulted from the annealing treatment, which decreases the corresponding surface area of the grain boundary and significantly reduces the internal stresses in the alloy. These combined effects reduce the energy barrier for magnetic-domain motion and enhance domain mobility. Annealing further yields better-ordered, more finely spaced magnetic domains, lowering domain-movement resistance, improving migration efficiency, and providing more domain-damping sources to dissipate vibration energy more effectively<sup>[16]</sup>. Therefore, annealing plays a crucial role in improving the damping performance of alloys.

Following annealing, the damping peak first increases and then decreases with increasing annealing temperature (Fig. 7). This behavior is associated with the presence of defects, such as vacancies and twins, which act as sources of magnetic domain damping and thereby reduce the migration rate of magnetic domains. At higher annealing temperatures, atoms in the alloy move more vigorously, allowing for more vacancies

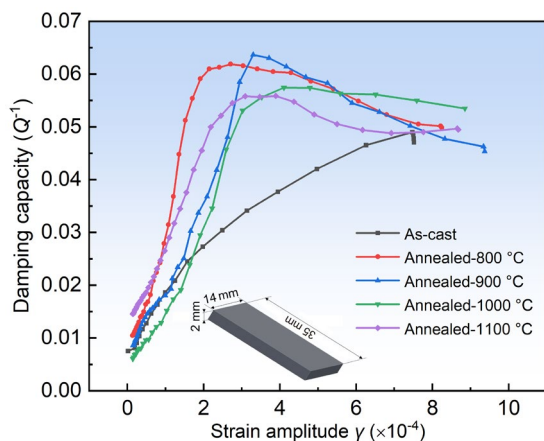


Fig. 7: Damping-strain amplitude curves of  $\text{Fe}_{42.75}\text{Mn}_{19}\text{Co}_{19}\text{Cr}_{14.25}\text{Al}_5$  HEA in as-cast and annealed states at a test frequency of 1 Hz

to be eliminated by migrating to the surface or binding with interstitial atoms. Therefore, a higher strain is required to enhance the motion of the damping source when the annealing temperature is lower. When annealing at 1,000 °C, Laves phase enriched in Al and Cr elements forms. Upon further increasing the temperature to 1,100 °C, a significant amount of Laves phase precipitates, hindering the movement of magnetic domains, dislocations, and twinning defects within the grains. Therefore, the damping performance of the sample annealed at 900 °C for 1 h reaches its optimal level.

When the strain is at a high level exceeding  $2 \times 10^{-4}$ , ferromagnetic damping is not the only damping mechanism. The energy dissipation arising from the synergy between dislocation motion and twin boundary movement maintains high damping performance within the medium-to-high strain amplitudes range. The damping performance,  $Q^{-1}$ , under high strain is described by Eq. (2)<sup>[27]</sup>:

$$Q^{-1} = Q_{\text{dis}}^{-1} + Q_{\text{twin}}^{-1} \quad (2)$$

where  $Q_{\text{dis}}^{-1}$  is related to the dislocation pinning theory, or Granato-Lücke theory<sup>[28]</sup>, which suggests that the dissipation of vibrational energy is primarily caused by the stress required to overcome the movement of dislocation lines pinned at atomic or vacancy sites.  $Q_{\text{twin}}^{-1}$  represents the damping contribution associated with twin-boundary motion. The damping characteristics expressed by a logarithmic decay function are determined by the following equation:

$$Q_{\text{dis}}^{-1} = \frac{C_1}{\varepsilon} \exp\left(-\frac{C_2}{\varepsilon}\right) \quad (3)$$

where  $C_1$  and  $C_2$  can be expressed as follows:

$$C_1 = \left(\frac{\rho L_n^3}{L_c^2}\right) \quad (4)$$

$$C_2 = \left(\frac{1}{L_c}\right) \quad (5)$$

where  $\rho$  represents dislocation density,  $L_n$  and  $L_c$  signify the average distances between strong pinning points and weak pinning points, respectively, and  $\varepsilon$  represents the strain amplitude. The magnitude of damping not only depends on dislocations but also on the number of anchoring points, as the quantity of pinning points directly affects the values of  $C_1$  and  $C_2$ <sup>[29]</sup>. When a dislocation moves within the lattice, it encounters pinning points that act as obstacles. Additional stress is required to overcome this pinning force for the dislocation to continue its advancement. Consequently, dislocation motion is impeded, leading to a reduction in velocity and additional energy dissipation, thereby demonstrating high damping performance. Furthermore, under cyclic stress, the movement of twin boundaries can effectively absorb vibrational energy and enhance the damping capacity of alloys<sup>[30, 31]</sup>. According to Xu et al.<sup>[32]</sup>, the movement of twinning directly influences the damping capacity. As depicted in Fig. 7, the  $Q^{-1}$  value of the as-cast alloy increases with the increasing strain amplitude in the high strain range, which is associated with the movement of twin boundaries within the alloy. During the annealing treatment, the alloy undergoes a continuous growth

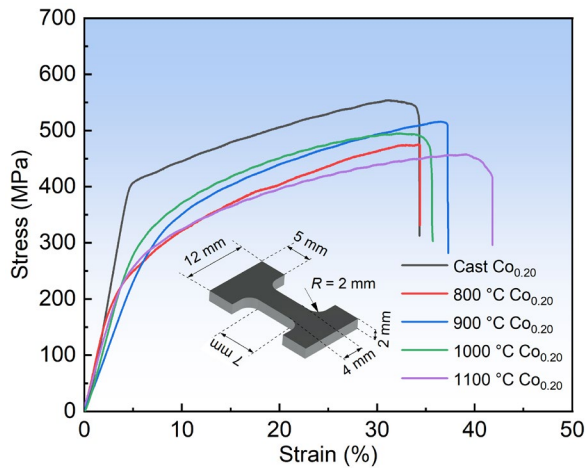
in grain size as the annealing temperature increases. This phenomenon results in enhanced mobility of dislocations and twins within larger grains. The migration resistance of twin boundaries within the same grain is relatively low, resulting in greater energy dissipation during their migration and further enhancement of damping properties. Compared with the BCC structure, the FCC structure is more prone to dislocations and stacking faults, enabling FCC-based HEAs to exhibit higher resistance to deformation. The motion of twin boundaries also contributes to the high strain amplitude damping capacity. At room temperature, point defects diffuse slowly, and essentially remain stationary at their original positions. Under stress, point defects can detach from the interface temporarily. Once the external force is removed, the interface quickly returns to its original position due to the attraction of the point defects, exhibiting a rubber-like behavior that dissipates energy. The motion of twin boundary is influenced by twin boundaries, precipitates, and segregation, which leads to changes in the damping properties of the alloy during annealing<sup>[33]</sup>.

### 3.3 Tensile properties

Figure 8 illustrates the room-temperature engineering stress-strain curves of the HEA in both the as-cast and annealed states. The inset in the figure presents the dimensions of the dog-bone tensile specimen. The quantitative tensile properties for all conditions (mean±standard deviation) are summarized in Table 1. As shown in Table 1, the as-cast alloy exhibits the highest yield strength (YS) and ultimate tensile strength (UTS), accompanied by a baseline plastic elongation (EL) of approximately 33.2%. In contrast, the annealed alloys generally exhibit enhanced ductility at the expense of YS and UTS. Notably, the specimen annealed at 900 °C exhibits the maximum damping capacity while retaining a reasonable combination of strength and ductility, indicating a favorable balance between functional and mechanical properties.

The tensile deformation behavior of these specimens is closely associated with their underlying microstructures. As shown in Figs. 5(c) and (d), annealing twins actively develop in the alloys heat-treated at 900 and 1,000 °C. During plastic deformation, these newly formed twin boundaries act as effective obstacles that hinder dislocation slip, triggering a shift in the deformation mechanism toward twinning-induced plasticity (TWIP). Conversely, the absence of annealing twins in the alloys treated at 800 and 1,100 °C [Figs. 5(b) and (e)] suggests that their deformation relies primarily on conventional dislocation slip.

Generally, metallic materials exhibit a significant mutual constraint, or trade-off, between damping capacity and mechanical strength. Conventional strengthening mechanisms, such as second-phase precipitation or high-density dislocation pinning, inherently restrict the mobility of interfaces and magnetic domains, inevitably leading to a deterioration in damping performance. However, this work demonstrates that a synergistic optimization of both properties can be achieved by regulating the stacking fault energy (SFE) via targeted heat treatments. In the specimen annealed at 900 °C, the high density of annealing twins and stacking faults exhibits a dual



**Fig. 8: Stress-strain curves of  $\text{Fe}_{42.75}\text{Mn}_{19}\text{Co}_{19}\text{Cr}_{14.25}\text{Al}_5$  HEA in as-cast and annealed states measured at room temperature**

**Table 1: Mechanical properties of the alloy in as-cast state and after annealed at 800–1,100 °C for 1 h**

| Condition     | YS (MPa) | UTS (MPa) | EL (%)   |
|---------------|----------|-----------|----------|
| As-cast       | 270±6    | 560±3     | 33.2±0.4 |
| 800 °C, 1 h   | 201±4    | 483±4     | 33.3±0.3 |
| 900 °C, 1 h   | 146±4    | 520±3     | 37.1±0.4 |
| 1,000 °C, 1 h | 180±7    | 502±2     | 35.2±0.3 |
| 1,100 °C, 1 h | 192±3    | 463±3     | 41.2±0.5 |

functionality: They serve as effective sources for vibrational energy dissipation (damping) and simultaneously act as key carriers for plastic deformation via the TWIP effect. This microstructural duality enables the alloy to acquire superior functional damping properties while maintaining a substantial structural load-bearing capacity. These findings suggest that tuning defect substructures via an optimized heat treatment window offers a promising pathway to balance damping and mechanical performance, providing a valuable paradigm for damping-oriented microstructural design.

## 4 Conclusions

In this work, the trade-off relationship among microstructural stability, damping capacity, and mechanical properties of the cast FeMnCoCrAl high-entropy alloy was systematically investigated via annealing heat treatment. The main findings and mechanisms are summarized as follows:

(1) Microstructural evolution mechanism: Annealing temperature serves as a critical variable for regulating the phase stability and defect substructures of the as-cast FeMnCoCrAl high-entropy alloy. Low-temperature annealing effectively eliminates the unstable BCC phase, achieving homogenization of the FCC matrix, whereas annealing above 1,000 °C induces the precipitation of the brittle Cr-rich Laves phase. Furthermore, annealing at 900 °C induces a unique microstructural state characterized by a high density of annealing twins and stacking

faults, which lays the structural foundation for the synergistic enhancement of damping and plasticity.

(2) Competition and synergy of damping mechanisms: The damping performance is governed by the competition between internal stress relief and defect evolution. The peak damping capacity ( $Q^{-1}=0.0636$ ) obtained at 900 °C is primarily attributed to two factors. First, the reduction in internal stress and grain growth induced by annealing minimize the pinning effect on magnetic domain walls, facilitating their irreversible movement and thereby enhancing ferromagnetic damping. Second, a moderate density of stacking faults and twins provides effective energy dissipation without obstructing domain wall and dislocation motion, unlike the Laves phase that precipitated at higher temperatures.

(3) Structural-functional synergy based on the TWIP effect: This work proposes an effective strategy to balance the damping and mechanical properties of high-entropy alloys. Although annealing inevitably leads to a reduction in strength, annealing at 900 °C successfully activates the twinning-induced plasticity (TWIP) deformation mechanism. This mechanism enables the alloy to maintain good work-hardening capability and sufficient ductility reserve even within a microstructure rich in crystalline defects, thereby avoiding the brittleness risks common in functional materials. This finding suggests that precisely regulating stacking fault energy via heat treatment to trigger the TWIP effect is a universal pathway for designing high-damping high-entropy alloys, under the premise of ensuring adequate structural load-bearing capacity.

## Acknowledgments

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## Conflict of interest

The authors declare that they have no conflict of interest.

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