

Cite this article as: Qiao Junwei, Wang Zeming, Zhang Yong, et al. Dynamic Tensile Deformation Mechanisms of $\text{Fe}_{40}\text{Mn}_{20}\text{Cr}_{20}\text{Ni}_{20}$ High-Entropy Alloys at Cryogenic Temperatures[J]. Rare Metal Materials and Engineering, 2026, 55(09): 2149-2164. DOI: <https://doi.org/10.12442/j.issn.1002-185X.20250358>.

ARTICLE

Dynamic Tensile Deformation Mechanisms of $\text{Fe}_{40}\text{Mn}_{20}\text{Cr}_{20}\text{Ni}_{20}$ High-Entropy Alloys at Cryogenic Temperatures

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Abstract: The mechanical behavior and microscopic mechanism of the $\text{Fe}_{40}\text{Mn}_{20}\text{Cr}_{20}\text{Ni}_{20}$ high-entropy alloy (HEA) at high/cryogenic temperatures and high strain rates (split Hopkinson bar) were investigated, and relevant theoretical models were used to fit the yield strength and flow stress of HEA. The tensile test results show that HEA exhibits excellent strength-plasticity synergy and excellent work-hardening ability by reducing the temperature or increasing the strain rate. During the dynamic tensile process, the interactions between different forms of dislocations and deformation twins jointly improve the strength and work-hardening ability of HEA. The Zerilli-Armstrong constitutive model was used to predict the temperature sensitivity and strain-rate sensitivity of the yield strength of HEA. At the same time, the Taylor model was used to predict the flow stress change of HEA with the strain rate under dynamic tension, and the model was applied to low-temperature dynamic experiments. The fitting results are consistent with the experimental results, providing a theoretical basis for the subsequent prediction of HEA strength.

Key words: high-entropy alloy; dynamic tension; cryogenic temperature; deformation mechanism; work hardening

1 Introduction

High-entropy alloys (HEAs) are a unique class of alloys with a high mixing entropy and a near-equimolar ratio^[1-3]. Their high configuration entropy and topologically ordered lattice structure permit HEAs many excellent properties, including excellent combination of strength and ductility, excellent low-temperature plasticity, excellent corrosion resistance, and wear resistance^[4-8]. These properties indicate that HEAs have great application potential under high strain rates and high/ cryogenic (low) temperatures. With the development of aerospace, nuclear, and defense industries, greater requirements have been placed on the performance of materials with a good balance of strength and plasticity in extreme environments^[9], which has attracted widespread attention^[10-11].

Face-centered-cubic (fcc) HEAs have excellent deformability,

and their most prominent mechanical properties include low strength but good plasticity^[3]. For example, the CoCrFeMnNi HEA (Cantor alloy) has a good combination of strength and plasticity^[1]. Researchers have developed several related alloys based on Cantor alloy, such as CoCrNi^[12-14], FeMnCoCr^[15-16], and CoCrFeNi^[17-19] HEAs. Gali et al^[19] investigated the tensile behavior of CoCrFeNiMn and CoCrFeNi HEAs in the temperature range of 77–1273 K and the strain rate range of 10^{-3} – 10^{-1} s⁻¹. It was found that the strength and plasticity of the alloys were increased simultaneously with the decrease in temperature. This result might be caused by the deformation-induced nanoscale twins, which delayed the occurrence of necking and further increased the plasticity. At the same time, the appearance of twins leads to continuous and stable work hardening^[20]. Subsequently, considering that the element Co is relatively costly, some Co-free FeCrNi^[21-22] and FeMnCrNi^[23] HEAs are gradually derived. These alloys still maintain

Received date: September 07, 2025

Foundation item: National Natural Science Foundation of China (52271110); Taiyuan Key Core Technology Tackling Project (2024TYJB0113)

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excellent work-hardening ability and a good combination of strength and plasticity at low temperatures or high strain rates. However, the service environment of most metals and alloys is not quasi-static but involves high-speed dynamic loads. To further explore the application of HEAs under extreme conditions, the study of dynamic deformation behavior is very important. Generally, fcc HEAs are more likely to activate deformation twins under high strain-rate loading^[24]. In recent years, researchers have conducted extensive research on the properties of CrMnFeCoNi HEA, including adiabatic shear localization^[25], strain-rate sensitivity^[26], and impact loading^[5]. It has been found that deformation twinning is a common microstructure mechanism of this HEA under extreme loading conditions, such as low-temperature or high-speed loading^[27–28]. In addition, the CoCrFeNi HEA with a single-phase fcc structure is mainly deformed by dislocation slip under quasi-static tension, while deformation twins are easily activated under high-speed dynamic loading^[29]. The enhancement in mechanical properties of the Fe₄₀Mn₂₀Cr₂₀Ni₂₀ HEA under dynamic loading is due to the activation of deformation twins^[30]. Therefore, different deformation mechanisms determine various mechanical responses. The deformation mechanism of HEAs depends on not only the intrinsic parameters of the material, such as stacking fault energy (SFE), but also the deformation conditions, such as temperature and strain rate^[31–33].

In this study, fcc Fe₄₀Mn₂₀Cr₂₀Ni₂₀ HEA of 500 kg was prepared by industrial vacuum induction melting. The effects of heat treatment on the microstructures and properties of Fe₄₀Mn₂₀Cr₂₀Ni₂₀ HEA were discussed in Ref. [30, 34]. Therefore, the microstructure evolution of Fe₄₀Mn₂₀Cr₂₀Ni₂₀ HEA under different strain rates became the research focus of this research. Relevant theoretical models were used to analyze and fit its strength and stress, providing a theoretical basis for subsequent alloy strength predictions. Under the condition of reduced temperature or increased strain rate, Fe₄₀Mn₂₀Cr₂₀Ni₂₀ HEA has excellent strength-plastic synergy and excellent work-hardening ability. Under conditions of 77 K and 10⁻³ s⁻¹, the yield strength and ultimate tensile strength of the alloy are 405 and 900 MPa, which increase by about 78% and 83%, respectively, and the plasticity is as high as 97.5%. With the increase in strain rate, the yield strength of HEA can reach up to 481 MPa, and the plasticity is as high as 95.0%. Compared with the quasi-static tensile experiment, the yield strength of the HEA is 447 MPa at room temperature, which increases by about 97%, and the elongation is at a high level. With the increase in strain rate, there is no contradiction between the strength and plasticity inversion. At low temperatures, the yield strength of the alloy is 1120 MPa, and the plasticity is 45.7%. During the dynamic tensile process, the interactions between different forms of dislocations and the deformation twins jointly improve the strength and work-hardening ability of the Fe₄₀Mn₂₀Cr₂₀Ni₂₀ HEA. Finally, the temperature sensitivity and strain-rate sensitivity of the yield strength of Fe₄₀Mn₂₀Cr₂₀Ni₂₀ HEA were predicted using the Zerilli-Armstrong (Z-A) constitutive model. At the same time,

the Taylor model was employed to predict the flow stress change of the alloy under dynamic tension. In the low-temperature dynamic loading simulation, the fitting results were consistent with the experimental results. These excellent mechanical properties under dynamic loading provide great potential for Fe₄₀Mn₂₀Cr₂₀Ni₂₀ HEA as a candidate for impact-resistant structural material.

2 Experiment

Under industrial production conditions, Fe₄₀Mn₂₀Cr₂₀Ni₂₀ HEA ingots with a nominal composition of Fe₄₀Mn₂₀Cr₂₀Ni₂₀ were prepared by vacuum induction-melting equipment using pure metal iron, manganese, chromium, and nickel (purity > 99.95wt%). The total mass was about 500 kg. The ingot was homogenized at 1200 °C for 1.5 h, cooled to room temperature in air, and finally hot-rolled into a plate with a thickness of 20 mm.

The quasi-static uniaxial tensile test was conducted using an Instron 5969 tensile tester at different temperatures and strain rates. The size of the dog-bone-shaped tensile sample was 14 mm×3 mm×1 mm, the test temperatures were 77–1073 K, and the strain rates were 10⁻³–10⁻¹ s⁻¹. To study the microstructure evolution with the increase in strain rate, the split Hopkinson tensile bar was used for dynamic tensile tests. The size of the dynamic tensile sample was 10 mm×4 mm×1 mm, the test temperatures were 77 and 298 K, and the strain rates were 1000, 1600, and 2000 s⁻¹. According to the one-dimensional stress-wave theory, the strain rate $\dot{\varepsilon}$, strain ε , and stress σ under dynamic tension can be calculated by Eq.(1–3), respectively:

$$\dot{\varepsilon} = \frac{d\varepsilon}{dt} = \frac{2C_b\varepsilon_r}{L_s} \quad (1)$$

$$\varepsilon = \frac{2C_b}{L_s} \int_0^t \varepsilon_r(t) dt \quad (2)$$

$$\sigma = \frac{EA_b}{A_s} \varepsilon_t(t) \quad (3)$$

where ε_r and ε_t are the strains on the incident rod and the transmission rod, respectively; C_b is the longitudinal wave velocity in the tensile rod; L_s is the cross-sectional length of the sample; A_b is the cross-sectional area of the sample; A_s is the cross-sectional area of the incident rod and the transmission rod; E is the elastic modulus of the incident rod and the transmission rod; t is time. The tensile rod is made of a maraging steel with a yield strength of 1.8 GPa. The rod must be in the elastic stage during the whole loading process. To ensure the data reliability, at least three tensile tests were repeated under each condition.

The initial microstructure of the sample was observed by the Leica DM 6000 metallographic microscope. The phase identification was performed by an X-ray diffractometer (XRD, PANalytical AERIS) with Co K α radiation. The grain size of the initial sample was roughly estimated by Image-Pro Plus software. The sheet sample for transmission electron microscope (TEM) observation was extracted from the central area of the fracture section of tensile sample with a sampling

thickness of 1 mm. The sheet size was about 5 mm×5 mm×1 mm, and it was mechanically ground to about 30 μm in thickness along the thickness direction. Electrochemical thinning was performed using a double-jet polishing instrument at −10 °C and 15 V (operating voltage). The selected etchant was a mixed solution of perchloric acid (10vol%) and ethanol (90vol%). Field emission scanning electron microscope (FEI Talos F200X G2) and high-resolution TEM were used to characterize the samples after dynamic tension.

3 Results

3.1 Initial microstructure

Fig. 1a shows XRD pattern and metallographic structure of the hot-rolled $\text{Fe}_{40}\text{Mn}_{20}\text{Cr}_{20}\text{Ni}_{20}$ HEA. It can be seen that $\text{Fe}_{40}\text{Mn}_{20}\text{Cr}_{20}\text{Ni}_{20}$ HEA has single-phase fcc solid solution structure. The metallographic structure exhibits typical equiaxed grains and a small number of annealing twins inside the grains. The grain size of $\text{Fe}_{40}\text{Mn}_{20}\text{Cr}_{20}\text{Ni}_{20}$ HEA is about 21.8 μm. Fig. 1b shows XRD patterns of the $\text{Fe}_{40}\text{Mn}_{20}\text{Cr}_{20}\text{Ni}_{20}$ HEA alloy after stretching at low (cryogenic) temperature and different strain rates. The diffraction peaks in Fig. 1b are all related to the single-phase fcc structure, indicating that $\text{Fe}_{40}\text{Mn}_{20}\text{Cr}_{20}\text{Ni}_{20}$ HEA does not undergo a phase structure transformation after stretching at low temperature and different strain rates.

3.2 Tensile performance at different temperatures

Fig. 2 shows the engineering stress-engineering strain curves of the $\text{Fe}_{40}\text{Mn}_{20}\text{Cr}_{20}\text{Ni}_{20}$ HEA at different temperatures (298, 473, 673, 873, and 1073 K), revealing the changes in strength and plasticity with temperature. At the same time, Table 1 shows the specific values of strength and plasticity of $\text{Fe}_{40}\text{Mn}_{20}\text{Cr}_{20}\text{Ni}_{20}$ HEA at different temperatures. At room temperature (298 K), the yield strength and ultimate tensile strength of $\text{Fe}_{40}\text{Mn}_{20}\text{Cr}_{20}\text{Ni}_{20}$ HEA are 227 and 491 MPa, respectively, and the uniform elongation is 43.4%. This result is similar to that of most HEAs with fcc structure^[11,29], showing a good combination of strength and plasticity. According to Fig.2, with the increase in temperature, the alloy is softened, which is similar to the situation of most alloys at high temperatures^[35–37]. When the temperature is 1073 K, the

alloy shows obvious softening phenomenon.

Fig. 3 shows the engineering stress-engineering strain curves of the $\text{Fe}_{40}\text{Mn}_{20}\text{Cr}_{20}\text{Ni}_{20}$ HEA at low temperature (77 K) under quasi-static (0.001, 0.01, and 0.1 s^{−1}) conditions. Table 2 lists the strength and plasticity of $\text{Fe}_{40}\text{Mn}_{20}\text{Cr}_{20}\text{Ni}_{20}$ HEA at low temperature (77 K) and different strain rates. Compared with the results at quasi-static state at 298 K (room temperature), the strength and plasticity of $\text{Fe}_{40}\text{Mn}_{20}\text{Cr}_{20}\text{Ni}_{20}$ HEA at low temperature are significantly improved. It is known that HEAs with fcc structures have the characteristics of high strength and high plasticity at low temperatures^[38–39]. The inset in Fig.3 shows the fracture diagram of $\text{Fe}_{40}\text{Mn}_{20}\text{Cr}_{20}\text{Ni}_{20}$ HEA after tensile test at strain rate of 0.001 s^{−1}, showing a large number of dimples. The dimples are small, indicating that $\text{Fe}_{40}\text{Mn}_{20}\text{Cr}_{20}\text{Ni}_{20}$ HEA undergoes significant plastic deformation during the tensile process.

3.3 Mechanical properties under dynamic tension at room temperature and low temperature

Fig. 4 shows the engineering stress-engineering strain curves and true stress-true strain curves of $\text{Fe}_{40}\text{Mn}_{20}\text{Cr}_{20}\text{Ni}_{20}$ HEA under room-temperature dynamic loading (1000, 1600, and 2000 s^{−1}), and the specific location of the yield strength ($\sigma_{0.2}$) is also marked in Fig.4. Compared with the quasi-static state, the dynamic tension shows a typical strength-plasticity inversion phenomenon^[11]. When the strain rate is 1000 s^{−1}, the yield strength and ultimate tensile strength of the alloy are improved, and the uniform elongation is as high as 25.3%. At the same time, obvious secondary work-hardening behavior occurs at 1000 s^{−1}, resulting in a higher ultimate tensile strength of the alloy. The promotion of secondary work-hardening behavior at high strain rates leads to the complexity of work-hardening behavior during the dynamic tensile process. When the strain rate is 2000 s^{−1}, the yield strength and plasticity of the alloy increase significantly. The plasticity is comparable to that of HEA under quasi-static condition, but the strength increases significantly. Qiao et al^[29] reported that for the fcc-structured HEA, the faster the deformation speed, the stronger the toughness.

Fig. 5 shows the engineering stress-engineering strain curves and true stress-true strain curves of $\text{Fe}_{40}\text{Mn}_{20}\text{Cr}_{20}\text{Ni}_{20}$ HEA under low-temperature dynamic loading (1000, 1600,

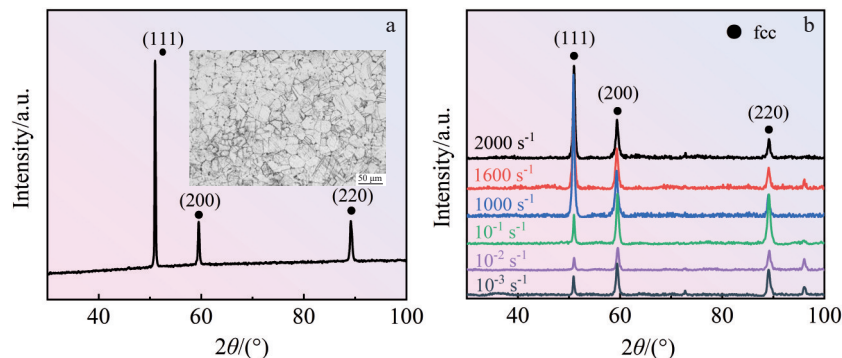


Fig.1 XRD patterns of $\text{Fe}_{40}\text{Mn}_{20}\text{Cr}_{20}\text{Ni}_{20}$ HEA under different conditions: (a) hot-rolled sample with inset of corresponding metallographic diagram; (b) stretched samples at low temperature and different strain rates

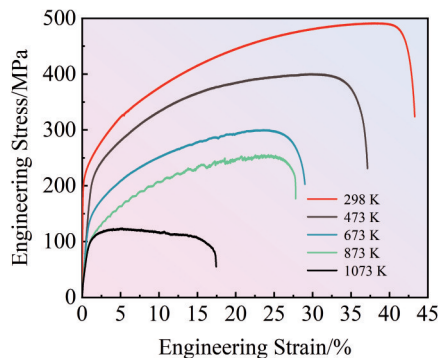


Fig.2 Engineering stress-engineering strain curves of Fe₄₀Mn₂₀Cr₂₀Ni₂₀ HEA at 298–1073 K under quasi-static conditions

Table 1 Strength and plasticity of Fe₄₀Mn₂₀Cr₂₀Ni₂₀ HEA at 298–1073 K under quasi-static conditions

Temperature, T/K	Yield strength/MPa	Ultimate tensile strength/MPa	Uniform elongation/%
298	227	491	43.4
473	207	400	37.2
673	135	301	29.2
873	101	253	27.8
1073	95	123	17.5

and 2000 s⁻¹), and the specific location of yield strength ($\sigma_{0.2}$) point is also presented. Compared with those under the room-temperature dynamic loading, the yield strength of the alloy at low-temperature (77 K) dynamic loading is obviously improved, the ultimate tensile strength slightly increases, and the elongation is basically unchanged. At the same time, compared with the quasi-static results at low temperature, the strength of Fe₄₀Mn₂₀Cr₂₀Ni₂₀ HEA at low temperature increases by about one time, but the plasticity decreases in the same proportion. The phenomenon of the strength-plasticity inversion also appears. Finally, the specific values of strength and plasticity of the alloy under dynamic tension at room temperature (298 K) and low temperature (77 K) are listed in Table 3. The length of the gauge section of the tensile sample after the test is basically consistent with the uniform elongation obtained from the experimental results. With the increase in strain rate, the yield strength of Fe₄₀Mn₂₀Cr₂₀Ni₂₀ HEA can reach 995 MPa, and the plasticity is as high as 44.5%. At low temperature, the yield strength is 1120 MPa, and the plasticity is 45.7%.

This study summarizes the yield strength of various metals and alloys under dynamic deformation at room temperature and low temperature, and the strain rate is mainly 10³ s⁻¹. As shown in Fig.6, the yield strength of Fe₄₀Mn₂₀Cr₂₀Ni₂₀ HEA is in the middle and upper positions of the figure at similar strain rates. Compared with titanium alloy, the alloy in this research shows a lower yield strength. However, compared with that of the magnesium alloy and aluminum alloy, the yield strength of HEA is significantly improved. Under low-temperature

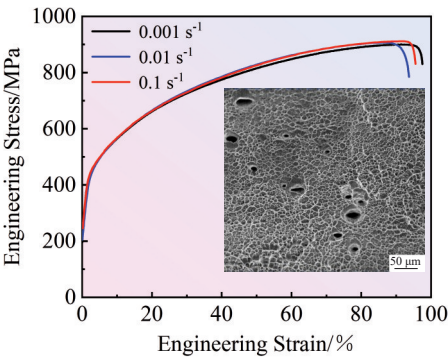


Fig.3 Engineering stress-engineering strain curves of Fe₄₀Mn₂₀Cr₂₀Ni₂₀ HEA at low temperature (77 K) and different strain rates

Table 2 Strength and plasticity of Fe₄₀Mn₂₀Cr₂₀Ni₂₀ HEA at low temperature (77 K) and different strain rates

Strain rate, $\dot{\epsilon}/s^{-1}$	Yield strength/MPa	Ultimate tensile strength/MPa	Uniform elongation/%
10 ⁻³	405	900	97.5
10 ⁻²	418	907	93.6
10 ⁻¹	481	912	95.0

dynamic loading, the current alloy exhibits a yield strength very close to that of the CoCrFeMnNi HEA. Therefore, the Fe₄₀Mn₂₀Cr₂₀Ni₂₀ HEA exhibits excellent dynamic-mechanical properties compared to most metal materials, and it is expected to become a candidate material for super-strength structural materials.

3.4 Microstructure after dynamic tensile loading

3.4.1 Characterization after dynamic tension

As shown in Fig.7–Fig.8, the microstructures of Fe₄₀Mn₂₀Cr₂₀Ni₂₀ HEA after dynamic tensile loading are characterized by electron back-scattered diffractometer (EBSD), and the changes in the recrystallization fraction, grain-boundary type, and dislocation density of Fe₄₀Mn₂₀Cr₂₀Ni₂₀ HEA after deformation are analyzed. Inverse pole figures (IPFs) are also used for analysis. As shown in Fig. 7, the grain morphology, deformation degree, volume fractions of different types of grain boundaries, such as low-angle grain boundaries (LAGBs), high-angle grain boundaries (HAGBs), and Σ3 twin boundaries, as well as kernel average misorientation (KAM) values of Fe₄₀Mn₂₀Cr₂₀Ni₂₀ HEA change significantly after dynamic tensile loading at room temperature, indicating that this HEA has obvious plastic deformation^[40]. With the increase in strain rates, the proportions of deformed grains, LAGBs, and KAM values in Fe₄₀Mn₂₀Cr₂₀Ni₂₀ HEA are increased significantly, while the proportions of HAGBs and Σ3 twin boundaries are decreased.

As shown in Fig.8, more severe plastic deformation occurs in Fe₄₀Mn₂₀Cr₂₀Ni₂₀ HEA after low-temperature dynamic tensile loading. At the same strain rate, Fe₄₀Mn₂₀Cr₂₀Ni₂₀ HEA after low-temperature dynamic tensile loading shows a higher proportion of deformed grains, a smaller proportion of

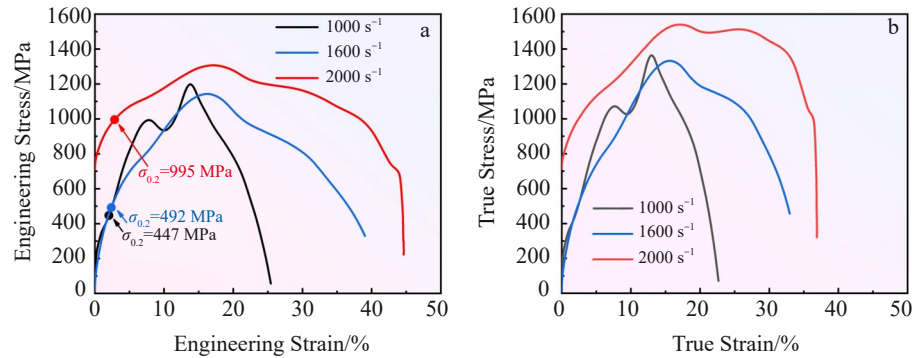


Fig.4 Tensile test results of Fe₄₀Mn₂₀Cr₂₀Ni₂₀ HEA at room temperature and different strain rates: (a) engineering stress-engineering strain curves; (b) true stress-true strain curves

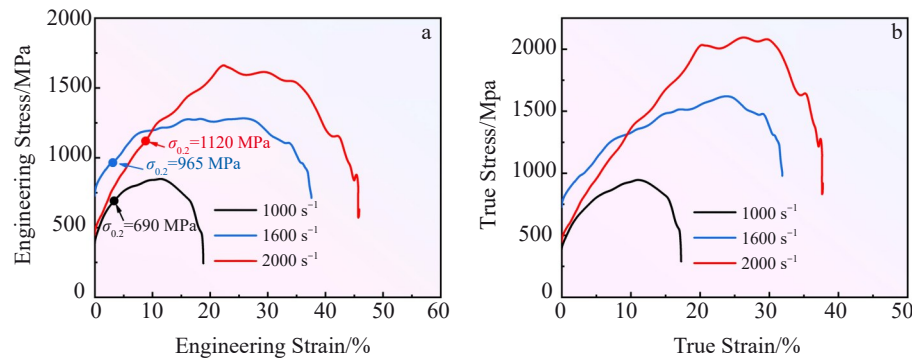


Fig.5 Tensile test results of Fe₄₀Mn₂₀Cr₂₀Ni₂₀ HEA at low temperature (77 K) and different strain rates: (a) engineering stress-engineering strain curves; (b) true stress-true strain curves

Table 3 Strength and plasticity of Fe ₄₀ Mn ₂₀ Cr ₂₀ Ni ₂₀ HEA at room temperature (298 K)/low temperature (77 K) and different strain rates				
Temperature, T/K	Strain rate, $\dot{\epsilon}/s^{-1}$	Yield strength/MPa	Ultimate tensile strength/MPa	Uniform elongation/%
298	1000	447	1198	25.3
	1600	492	1143	39.0
	2000	995	1307	44.5
77	1000	690	848	18.8
	1600	965	1275	37.5
	2000	1120	1660	45.7

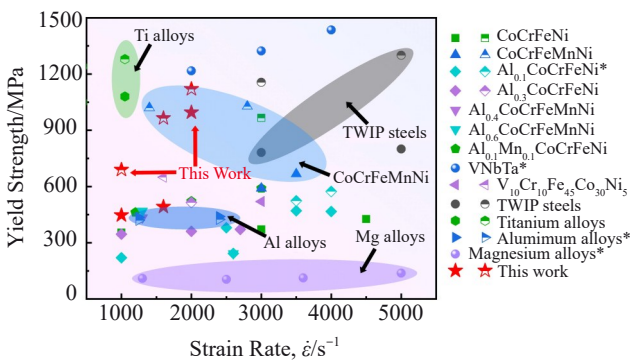


Fig.6 Comparison of mechanical properties of common metals, alloys, and HEAs under dynamic deformation (* represents dynamic-compression deformation)

HAGBs, and a higher KAM value. At the same time, the proportion of $\Sigma 3$ twin boundaries is greater, indicating an increase in the number of deformation twins^[34]. With the increase in strain rates, the change in grain structures in Fe₄₀Mn₂₀Cr₂₀Ni₂₀ HEA is similar to that at room temperature, which is manifested as the increase in proportion of deformed grains, LAGBs, and KAM value as well as the decrease in proportion of HAGBs. The corresponding KAM diagram shows that KAM values are higher under low-temperature dynamic tensile loading, and this highly deformed region appears not only on the grain boundary but also inside the grains. It is shown that almost all grains in Fe₄₀Mn₂₀Cr₂₀Ni₂₀ HEA are involved in plastic deformation after low-temperature dynamic tensile loading.

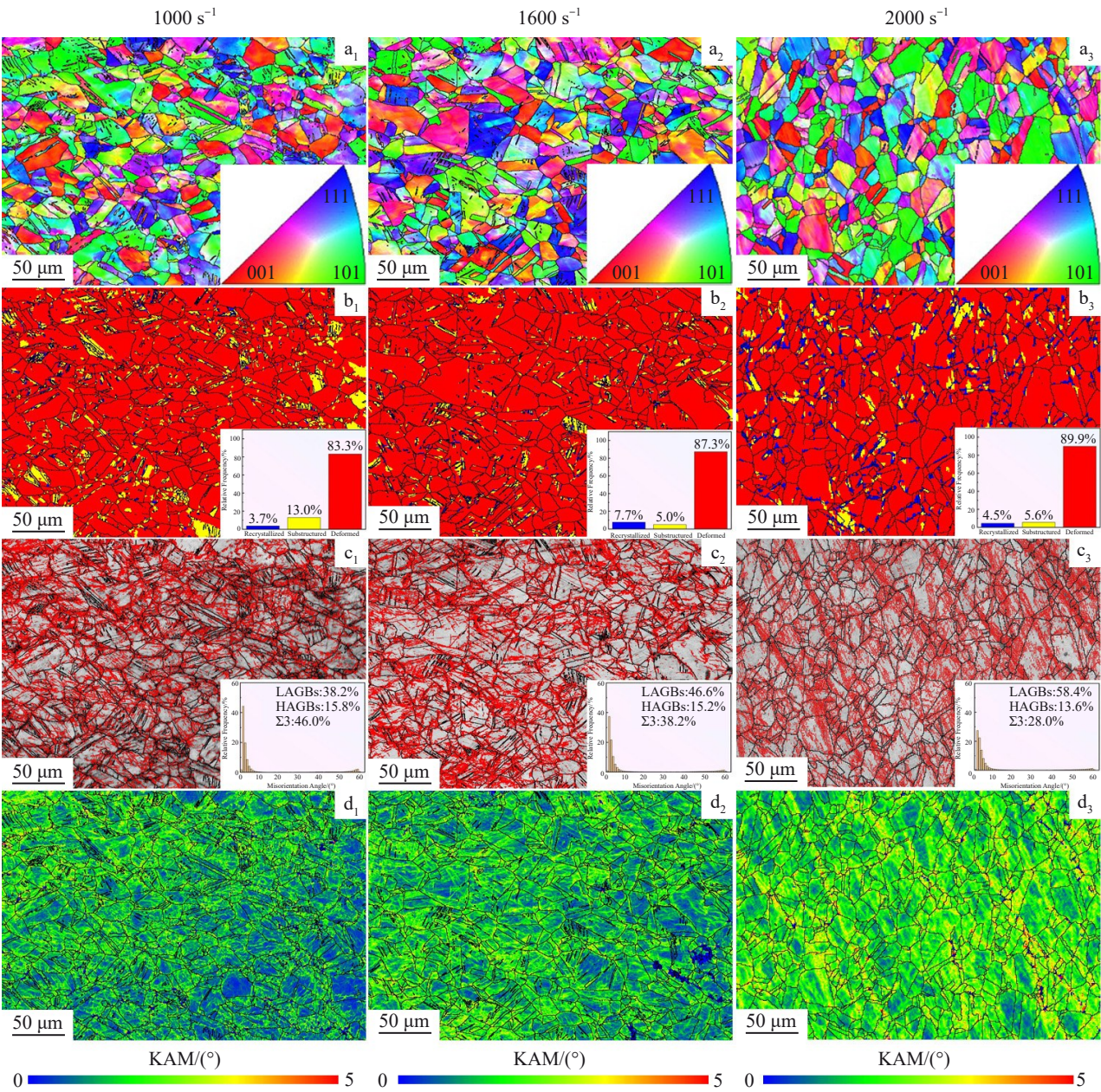


Fig.7 EBSD analysis of $\text{Fe}_{40}\text{Mn}_{20}\text{Cr}_{20}\text{Ni}_{20}$ HEA after dynamic tensile loading at room temperature and strain rates of 1000 s^{-1} (a_1-d_1), 1600 s^{-1} (a_2-d_2), and 2000 s^{-1} (a_3-d_3): (a_1-a_3) IPFs; (b_1-b_3) recrystallization-distribution maps and statistical results of recrystallization fractions; (c_1-c_3) LAGB and HAGB distribution maps with statistical results; (d_1-d_3) KAM diagrams

3.4.2 Dislocation configuration after dynamic tension

Fig. 9 and Fig. 10 present TEM images of the dislocation structures of $\text{Fe}_{40}\text{Mn}_{20}\text{Cr}_{20}\text{Ni}_{20}$ HEA after dynamic tension at strain rates of 1000, 1600, and 2000 s^{-1} at room temperature. Different forms of dislocation structures can be observed, such as dislocation lines, dislocation cells, dislocation tangles, and dislocation walls. In Fig. 9–Fig. 10, dislocation cells are indicated by white circles, dislocation lines are indicated by yellow arrows, dislocation wall is indicated by white dotted line, and dislocation tangles are indicated by red circles. At the strain rate of 1000 s^{-1} , the plastic deformation is still dominated by dislocation slips. With the increase in strain rate, the overall dislocation density is increased. In addition,

the smaller dislocation cells in the tensile sample provide a higher flow stress for the alloy^[41]. At the same time, the dislocation structure changes from the planar slip to wavy slip (i. e., dislocation cell structure). It is reported that this slip-transition phenomenon also occurs in $\text{Fe}_{40}\text{Mn}_{40}\text{Co}_{10}\text{Cr}_{10}$ and NiCoCrFeMn HEAs^[39,42].

Fig. 11 shows TEM images of the dislocation structures of $\text{Fe}_{40}\text{Mn}_{20}\text{Cr}_{20}\text{Ni}_{20}$ HEA after dynamic tension at low temperature (77 K) and different strain rates. At low temperature, the dislocation density is obviously increased with the increase in strain rate. Compared with that after dynamic tension at room temperature, the dislocation density at low temperature is also significantly improved. The

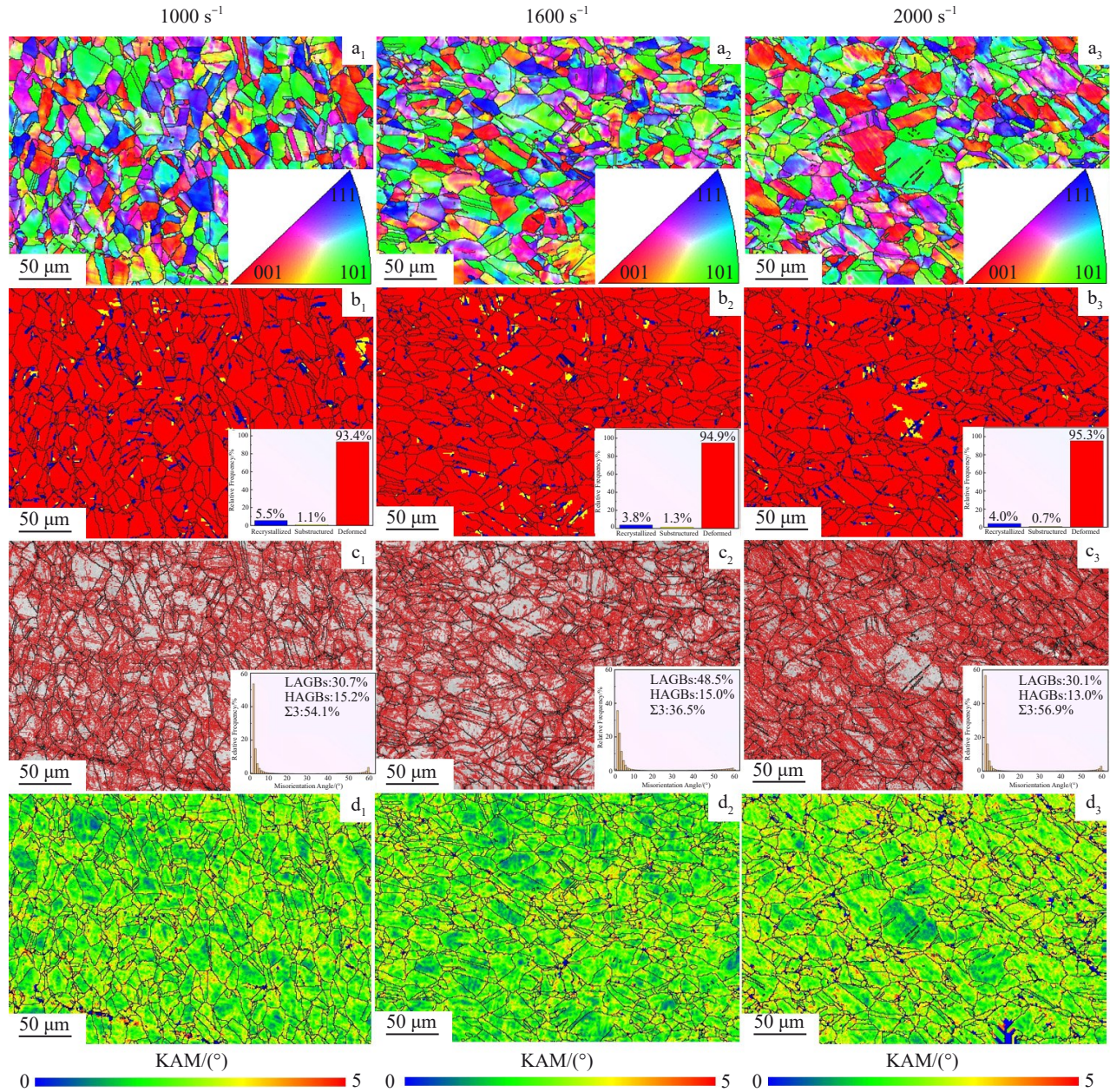


Fig.8 EBSD analysis of $\text{Fe}_{40}\text{Mn}_{20}\text{Cr}_{20}\text{Ni}_{20}$ HEA after dynamic tensile loading at low temperature and strain rates of 1000 s^{-1} (a₁–d₁), 1600 s^{-1} (a₂–d₂), and 2000 s^{-1} (a₃–d₃): (a₁–a₃) IPFs; (b₁–b₃) recrystallization-distribution maps and statistical results of recrystallization fractions; (c₁–c₃) LAGB and HAGB distribution maps with statistical results; (d₁–d₃) KAM diagrams

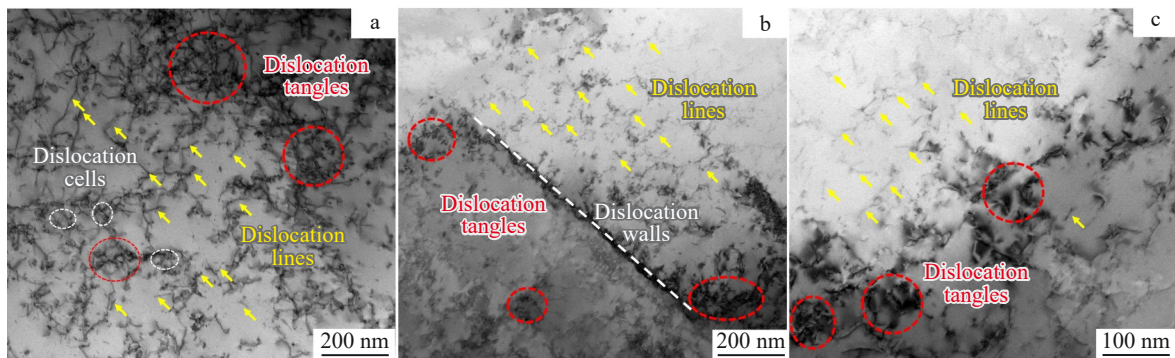


Fig.9 TEM images of dislocation structures of $\text{Fe}_{40}\text{Mn}_{20}\text{Cr}_{20}\text{Ni}_{20}$ HEA after dynamic tension at room temperature and strain rate of 1000 s^{-1}

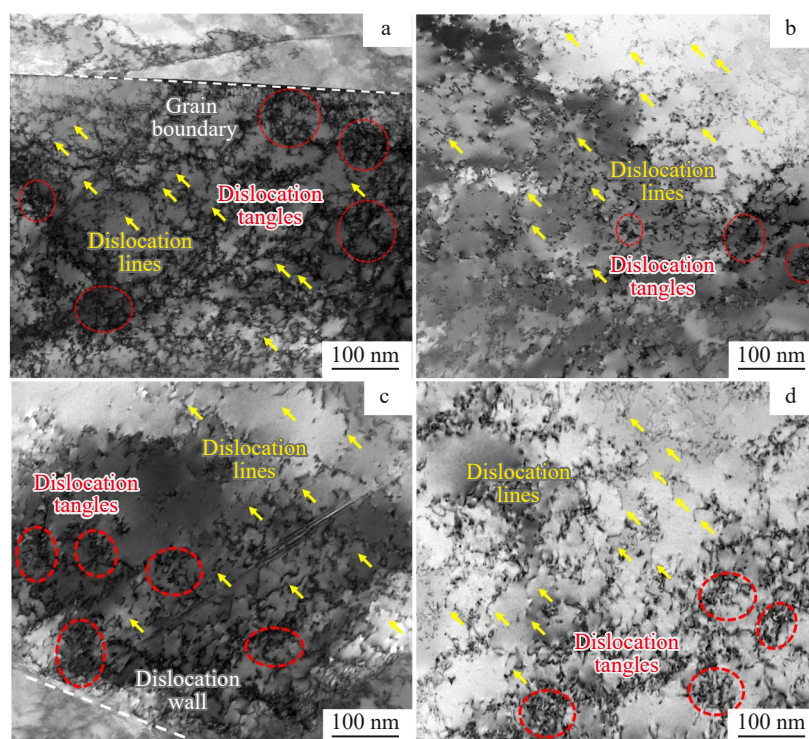


Fig.10 TEM images of dislocation structures of $\text{Fe}_{40}\text{Mn}_{20}\text{Cr}_{20}\text{Ni}_{20}$ HEA after dynamic tension at room temperature and strain rates of 1600 s^{-1} (a–b) and 2000 s^{-1} (c–d)

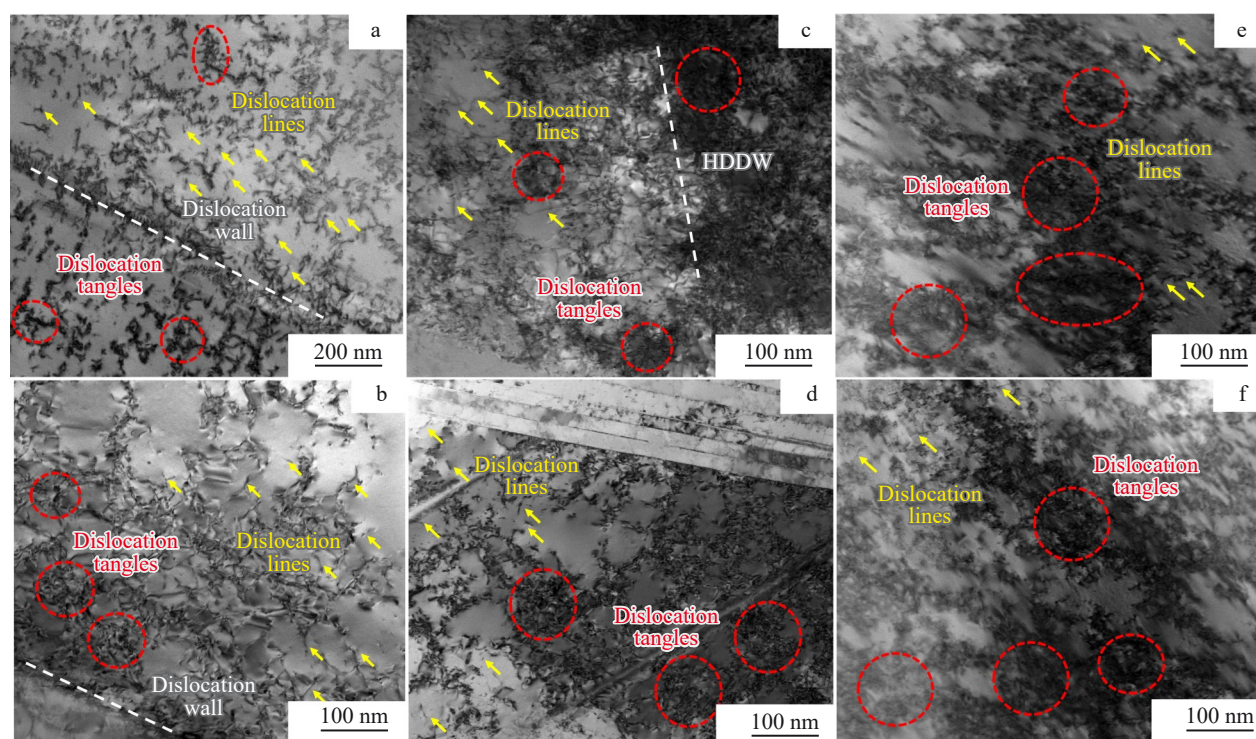


Fig.11 TEM images of dislocation structures of $\text{Fe}_{40}\text{Mn}_{20}\text{Cr}_{20}\text{Ni}_{20}$ HEA after dynamic tension at low temperature and different strain rates: (a–b) 1000 s^{-1} ; (c–d) 1600 s^{-1} ; (e–f) 2000 s^{-1}

interaction between dislocations and grain boundaries results in high-density dislocation tangles, which is also the result of dislocation cross-slip^[30]. At the same time, the appearance of high-density dislocation walls (HDDWs) indicates that the dislocation motion changes from planar slip to cross slip or

multiple slip after high-strain-rate deformation, as shown in Fig. 10c. HDDWs in HEAs will affect the slip of isolated dislocations and activate the accumulation, absorption, or transfer of dislocations. Besides, the movement of HDDWs will lead to the release of local stress, enhance the work-

hardening ability of HEAs, and improve the strength and plasticity^[43]. Therefore, the high-density dislocations become one of the factors that increase the flow stress of Fe₄₀Mn₂₀Cr₂₀Ni₂₀ HEA under dynamic loading.

3.4.3 Twin configuration after dynamic tensile loading

At room temperature, Fe₄₀Mn₂₀Cr₂₀Ni₂₀ HEA exhibits a deformation twin structure after dynamic tension at strain rates of 1600 and 2000 s⁻¹, as indicated by the yellow dotted lines in Fig. 12. Additionally, in Fig. 12, t is the twin width, and the red circle represents the partial dislocation structure. After dynamic tension, nano-scale deformation twins along the [011] zone axis appear, and the corresponding selected area electron diffraction (SAED) pattern shows diffraction spots belonging to the fcc twins and matrix. As the strain rate increases, the number of twins increases, and the twin width decreases from 31.2 nm to 19.5 nm. Deformation twins are usually observed after severe plastic deformation in medium/low SFE alloys, which is an important deformation mechanism to improve the work-hardening ability and mechanical properties of HEA^[44].

As shown in Fig. 13, a large number of close-packed deformation twins also appear in Fe₄₀Mn₂₀Cr₂₀Ni₂₀ HEA after dynamic tension at low temperature and strain rates of 1000, 1600, and 2000 s⁻¹. In Fig. 13, the yellow dotted line represents the deformation twins, t is the twin width, and the red circle represents the partial dislocation structure. The corresponding SAED pattern shows diffraction spots belonging to fcc twins and matrix. Compared with that after room-temperature dynamic loading, the number of deformation twins increases significantly under low-temperature dynamic loading, and deformation twins appear in two directions at an angle of

about 70°. A large number of dislocations pile up at the junction of twins, thereby inhibiting the dislocation slips. At the same time, the corresponding high-resolution TEM images show clear twin boundaries. The appearance of nano-twins will destroy the slip trajectory of dislocations and make dislocations concentrate near the twin boundaries, thus indirectly weakening the characteristics of dislocation cells^[23]. This process not only improves the strength of HEA but also improves the plasticity. At low temperature, with the increase in strain rate, the twin width is also decreased. The twin width is 32.4, 25.5, and 16.1 nm when the strain rate is 1000, 1600, and 2000 s⁻¹, respectively. Therefore, deformation twinning is a common microstructure mechanism of HEAs under extreme conditions (low temperature and high-speed loading).

4 Discussion

4.1 Yield-strength model of thermal-activation mechanisms

In the process of plastic deformation, the deformation mechanism can be divided into thermally activated dislocation motion, dislocation drag, and relativistic effect mechanisms based on different dislocation velocities^[11]. The thermal-activation mechanism is a dislocation-motion mechanism based on the dislocation-motion law and dislocation dynamics. It usually dominates plastic deformation at slow strain rates^[45]. Dislocation drag, namely phonon drag, refers to the interaction between dislocation and lattice vibration under high-strain-rate deformation, which affects dislocation motion and improves properties of HEAs. For example, the dislocation drag effect of CoCrFeNi HEA is significantly improved when the strain rate is greater than 3000 s⁻¹^[29]. The relativistic effect is similar to the dislocation drag effect. In extreme

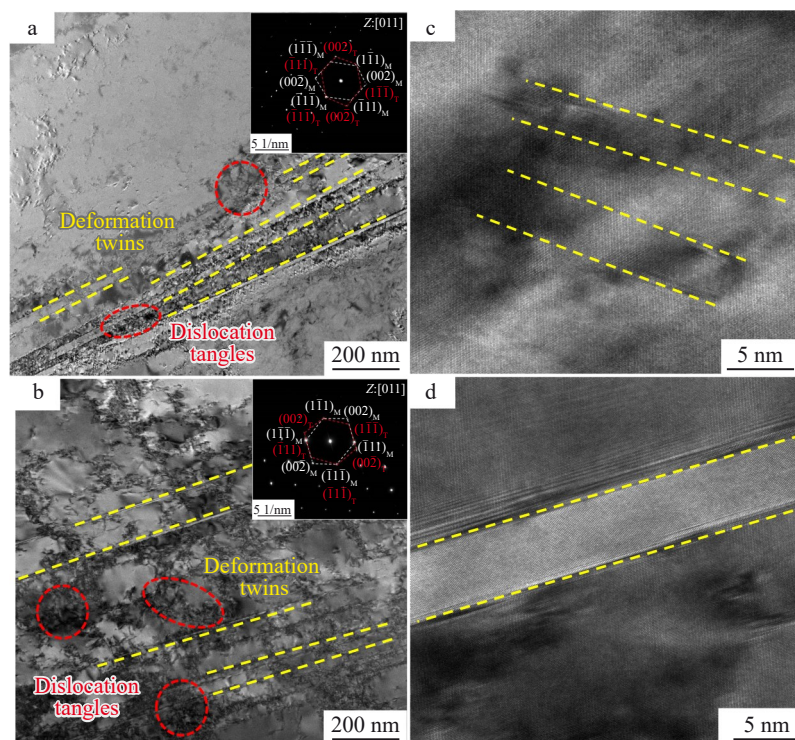


Fig. 12 Twin structures of Fe₄₀Mn₂₀Cr₂₀Ni₂₀ HEA after dynamic tension at room temperature and different strain rates: (a–b) 1600 s⁻¹; (c–d) 2000 s⁻¹

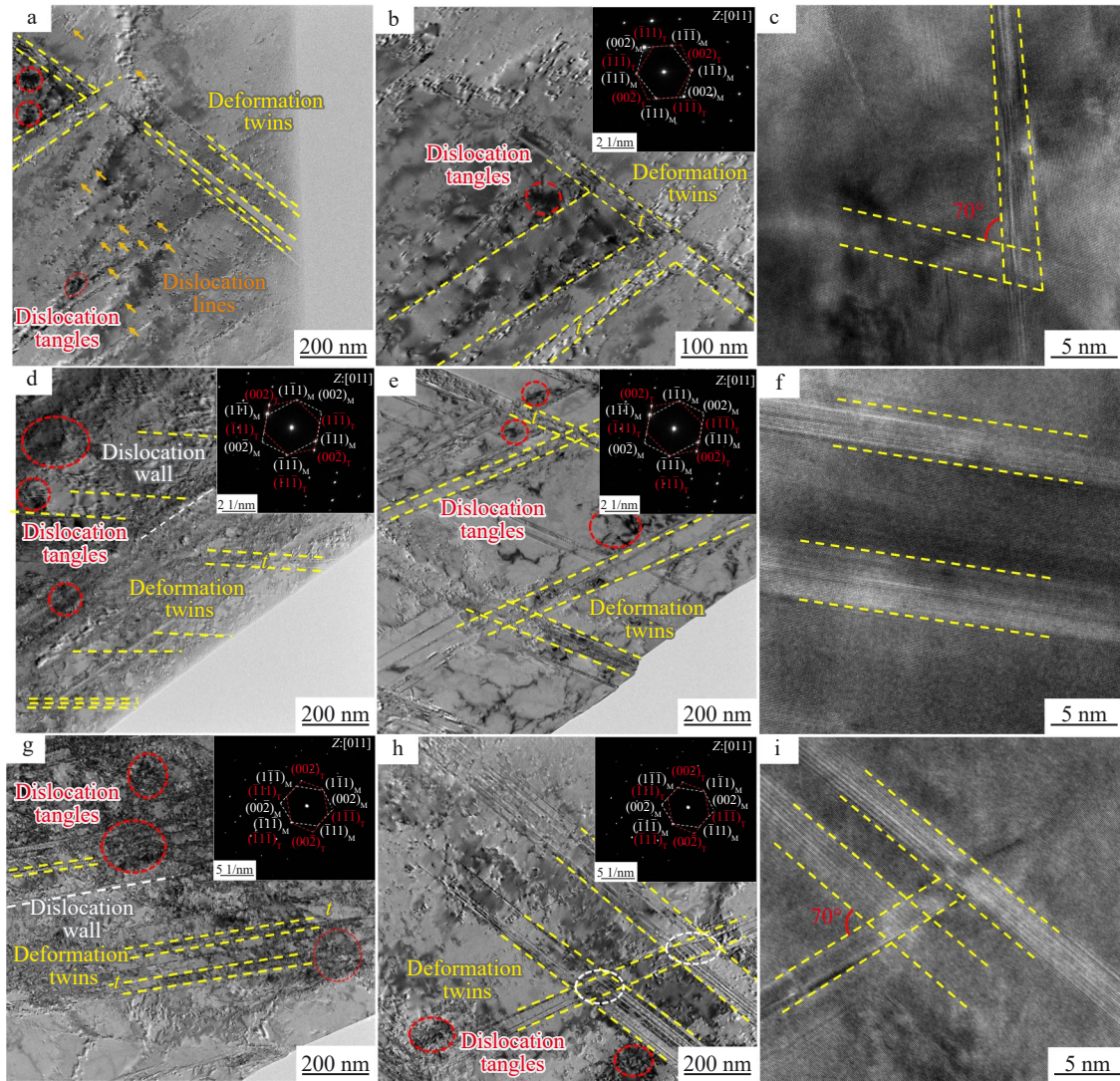


Fig.13 Twinning structures of $\text{Fe}_{40}\text{Mn}_{20}\text{Cr}_{20}\text{Ni}_{20}$ HEA after dynamic tension at low temperature and different strain rates: (a–c) 1000 s^{-1} ; (d–f) 1600 s^{-1} ; (g–i) 2000 s^{-1}

conditions (such as high-speed dislocation motion or high-energy particle irradiation), it may have an effect on dislocation motion. This effect mechanism believes that dislocations encounter many obstacles during their motion. These obstacles can be divided into short-range thermal obstacles and long-range thermal obstacles^[45]. At present, the thermally activated constitutive model used to describe the temperatures and strain rates of materials during deformation is Z-A model^[46], and the expression is described as follows:

$$\sigma_y = \sigma_0^* \exp(-\beta_0 T + \beta_1 T \ln \dot{\epsilon}) + \sigma_g \quad (4)$$

where β_0 and β_1 are material constants; T is the absolute temperature; σ_0^* is the thermal stress at $T=0\text{ K}$; σ_y is yield stress; σ_g is the grain-boundary-strengthening term. Since the $\text{Fe}_{40}\text{Mn}_{20}\text{Cr}_{20}\text{Ni}_{20}$ HEA has the single-phase fcc structure, and the initial dislocation density is extremely low, only the grain boundary strengthening term σ_g is the non-thermal stress in this study.

The Hall-Petch coefficient of the FeMnCrNi HEA has been reported to be $494\text{ MPa} \cdot \mu\text{m}^{1/2}$ ^[47]. Combined with the related

grain size ($21.8\text{ }\mu\text{m}$), the grain-boundary-strengthening term σ_g of the FeMnCrNi HEA is calculated to be 106 MPa . The experimental results of wide temperatures and strain rates of the FeMnCrNi HEA are described by the constitutive model based on thermal-activation mechanisms. The fitting results are shown in Fig. 14a. It is found that the thermal-activation mechanism can well reflect the temperature- and strain rate-related yield strength of HEA, and the fitting parameters are self-consistent. The obtained σ_0^* , β_0 , and β_1 are 396 MPa , $2.2 \times 10^{-3}\text{ K}^{-1}$, and $2.0 \times 10^{-4}\text{ K}^{-1}$, respectively. Substituting the parameters into the yield strength constitutive model based on the thermal activation mechanism, the predicted yield strength of HEA under the conditions of $77\text{ K}/0.001\text{ s}^{-1}$ and $1073\text{ K}/0.001\text{ s}^{-1}$ is 408 and 90 MPa , close to the experimental values of 405 and 95 MPa , respectively. In conclusion, at a strain rate of 10^{-3} s^{-1} , compared with those at quasi-static state at room temperature, the yield strength and plasticity at low temperature increase by about 78% and 125% , which can reach 405 MPa and 97.5% , respectively. With the increase in

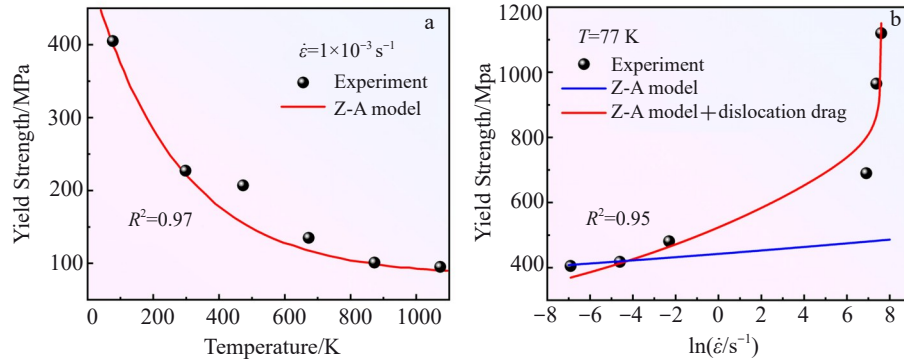


Fig.14 Experimental data and Z-A model fitting results of Fe₄₀Mn₂₀Cr₂₀Ni₂₀ HEA after dynamic tension at low temperature: (a) temperature sensitivity to yield strength; (b) strain rate sensitivity to yield strength

strain rate, the yield strength of the alloy can reach 481 MPa, and the plasticity is still as high as 95.0%. The abovementioned trend is sufficient to illustrate the validity of the yield strength model, proving that it can be used for strength prediction.

However, at higher strain rates, the constitutive model of the thermal activation mechanism cannot evaluate the yield strength well, as presented by the blue line in Fig. 14b, indicating that the model needs to be modified. During high-strain-rate deformation, the interaction between the dislocation and lattice vibration (i.e., phonon viscosity) cannot be ignored, which affects dislocation motion and plays a strengthening role, and this phenomenon is called phonon traction. Therefore, at high strain rates, phonon drag must be considered in the calculation of strain rate sensitivity to yield strength. According to Ref.[48], the thermally activated strain rate analysis model, including the phonon drag mechanism, is simplified as follows:

$$\sigma_y = \sigma_0^* \exp(-\beta_0 T + \beta_1 T \ln \dot{\epsilon}) / (1 - c \dot{\epsilon})^{\beta_1 T} + \sigma_g \quad (5)$$

where c is the material constant. This modified model with previous fitting parameters is used to fit the strength of the alloy after high-strain-rate tensile deformation. It was found that the phonon drag began to play a role at a strain rate of about $1 \times 10^3 \text{ s}^{-1}$ and well reflected the yield strength sensitivity at high strain rates. The c value was $4.98 \times 10^{-4} \text{ s}$, and the fitting results are shown by the red curve in Fig. 14b. This trend is similar to the research results of the yield strength

sensitivity of the FeCoCrNi HEA at high strain rates^[29].

4.2 Work-hardening constitutive equation related to strain rates

Three mechanisms may exist in the dynamic loading process of materials: work hardening, strain rate hardening, and thermal softening^[45]. The former two mechanisms usually cause an increase in flow stress, while the thermal softening will reduce the flow stress. In the process of plastic deformation, plastic work is usually converted into heat, resulting in a thermal softening phenomenon. The change of an adiabatic temperature rise with true strain during thermal softening can be calculated by the following equation^[45]:

$$\Delta T = T - T_0 = \int_{T_0}^T dT = \frac{\beta}{\rho C_p} \int_0^{\epsilon_p} \sigma d\epsilon_p \quad (6)$$

where ΔT is the adiabatic temperature rise, T is the instantaneous temperature of tensile deformation, T_0 is the initial temperature of the test (77 or 298 K), β is the fraction of plastic energy converted into heat energy (0.9)^[49], ρ is the density of the Fe₄₀Mn₂₀Cr₂₀Ni₂₀ HEA (7.9 g/cm³), C_p is the constant-pressure molar heat capacity (as estimated by the mixing rule, $0.454 \text{ J} \cdot \text{mol}^{-1} \cdot \text{K}^{-1}$)^[50], σ is stress, and ϵ_p is the true strain. The fitting results are shown in Fig. 15. The temperature rise at room temperature and strain rates of 1000, 1600, and 2000 s^{-1} is 68.0, 94.5, and 120.9 K, respectively. The temperature rise at low temperature and strain rates of 1000, 1600, and 2000 s^{-1} is 37.7, 102.0, and 157.3 K, respectively.

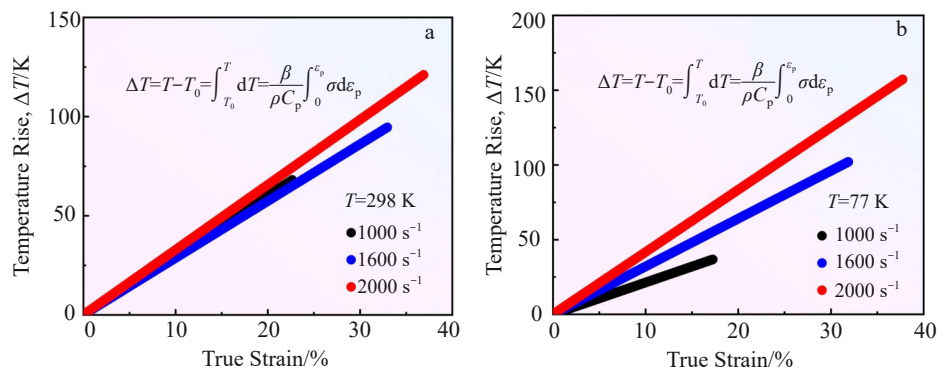


Fig.15 Temperature rise results of Fe₄₀Mn₂₀Cr₂₀Ni₂₀ HEA after dynamic tension at different temperatures and strain rates: (a) room temperature (298 K); (b) low temperature (77 K)

According to the experimental results in this research, adiabatic shear bands cannot be observed, and the calculated results show that the adiabatic temperature rise is too small to soften HEA. Hence, the effect of the adiabatic temperature rise on strain hardening is ignored.

In this study, microstructure analysis shows that both dislocation strengthening and twin strengthening contribute to the work hardening. Based on the classical Taylor model, a constitutive model based on the micro-deformation structure is developed. The stress induced by forest dislocation (σ_f) is calculated as follows^[51]:

$$\sigma_f = MaGb\sqrt{\rho} \quad (7)$$

where M is the average Taylor factor (M value of fcc metals is usually 3.06); G is the shear modulus (the shear modulus of $\text{Fe}_{40}\text{Mn}_{20}\text{Cr}_{20}\text{Ni}_{20}$ HEA at 298 K is 85 GPa^[23]); a is a constant related to the strength of dislocation interactions (0.4 at room temperature^[52]); b is the value of Burgers vector of the full dislocation; ρ is the total dislocation density. The evolution of dislocation density depends on the dynamic competition between dislocation generation and annihilation. The Kocks-Mecking equation describes the change of dislocation density with strain^[53]:

$$\frac{d\rho}{d\varepsilon_p} = M \left(\frac{1}{bA} - k\rho \right) \quad (8)$$

where A is the average free path of dislocation; k is a dynamic recovery factor related to temperature and strain rate. In this case, k can be calculated as follows^[53]:

$$k = k_0 \left(\frac{\dot{\varepsilon}}{\dot{\varepsilon}_0} \right)^{-\frac{KT}{A}} \quad (9)$$

where k_0 is the recovery factor at 0 K; K is a Boltzmann constant with a value of 1.38×10^{-23} J/K; $\dot{\varepsilon}_0$ is the reference strain rate (10^7 s^{-1}); T is the instantaneous temperature during the deformation process. Because the temperature rise is small, it is considered to be a constant value as 77 or 298 K in this research; A is a material parameter that depends on SFE. Besides, k_0 and A can be calculated by Eq. (10 – 11)^[54], respectively:

$$k_0 = \exp \left(\frac{79305\xi^{1.4213}}{1 + 15001\xi^{1.4213}} \right) \quad (10)$$

$$\frac{A}{Gb^3} = \frac{1}{\exp(1.44 + 27.55\xi - 390.54\xi^2)} \quad (11)$$

where ξ is the normalized SFE with $\xi = \gamma_{\text{SF}}/Gb$ (γ_{SF} is SFE); G is shear modulus. SFE of $\text{Fe}_{40}\text{Mn}_{20}\text{Cr}_{20}\text{Ni}_{20}$ HEA at 298 and 77 K is 42.6 and 14.6 mJ/m², respectively^[30,34]. Thus, at room temperature and low temperature, the values of k_0 are 36.8 and 4.6, and the values of A are 4.8×10^{-18} and 5.5×10^{-18} J, respectively.

The microstructure observation shows that the twins play a crucial role in the dynamic tensile process and have a little effect on the quasi-static tensile process. In the quasi-static tensile process, the only obstacle to dislocation motion is the dislocation structure^[11]. Therefore, nano-twin boundaries should be considered in the analysis of high-strain-rate tension, and the average free path A can be calculated by Eq.(12–13), as follows:

$$\frac{1}{A} = \frac{1}{d} + k_1\sqrt{\rho} \quad (12)$$

$$\frac{1}{A} = \frac{1}{d} + \frac{i_t}{t} + k_1\sqrt{\rho} \quad (13)$$

where d is the average grain size of the sample (21.8 μm); k_1 is a non-thermal hardening factor (0.018)^[54]; i_t is the ratio constant of twin spacing (0.14)^[54]. At room temperature, the average twin thickness at strain rates of 1600 and 2000 s^{-1} is 31.2 and 19.5 nm, respectively. At low temperature, the average twin thickness at strain rates of 1000, 1600, and 2000 s^{-1} is 32.4, 25.5, and 16.1 nm, respectively.

By substituting the corresponding parameters into the abovementioned formula, the variation of the flow stress with strain at room temperature can be fitted. The fitting results are shown in Fig. 16a. At the same time, the change of the flow stress with strain at 77 K is predicted. It is worth noting that many parameters in the abovementioned formula will vary with temperature. For example, the shear modulus G is a temperature-dependent constant that is increased with the decrease in temperature. Because $\text{Fe}_{40}\text{Mn}_{20}\text{Cr}_{20}\text{Ni}_{20}$ HEA is rarely discussed, there is a lack of research on the temperature sensitivity to shear modulus G . The shear modulus G of the $\text{Fe}_{40}\text{Mn}_{20}\text{Cr}_{20}\text{Ni}_{20}$ HEA at 77 K is considered to be 1.1 times that at 298 K, namely 93.5 GPa^[55]. Obviously, the current constitutive model can well predict the change of the flow stress with strain at low temperature (77 K), and the fitting results are shown in Fig. 16b. Table 4 lists the values of the relevant parameters in the Z-A model and the Taylor model.

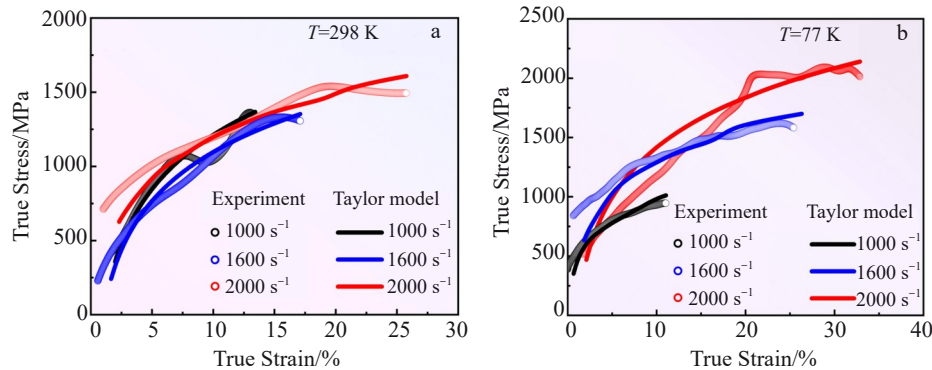


Fig.16 Experimental data and Taylor model fitting results of $\text{Fe}_{40}\text{Mn}_{20}\text{Cr}_{20}\text{Ni}_{20}$ HEA after dynamic tension at different temperatures and strain rates: (a) room temperature; (b) low temperature

Table 4 Material parameters used in Z-A thermal activation model and Taylor model

Parameter	Value	
	Room temperature	Low temperature
Thermal stress, σ_0^*/MPa	396	-
Grain boundary strengthening, σ_g/MPa	106	-
Material constant, $\beta_0/\times 10^{-3} \text{ K}^{-1}$	2.2	-
Material constant, $\beta_1/\times 10^{-4} \text{ K}^{-1}$	2.0	-
Material constant, $c/\times 10^{-4} \text{ s}$	4.98	-
Taylor factor, M	3.06	3.06
Shear modulus, G/GPa	85.0	93.5
Value of Burgers vector, b/nm	0.255	0.255
Heat capacity, $C_v/\text{J}\cdot\text{g}^{-1}\cdot\text{K}^{-1}$	0.454	0.454
Plastic energy-thermal energy conversion fraction, β	0.9	0.9
Density, $\rho/\text{g}\cdot\text{cm}^{-3}$	7.90	7.90
SFE, $\gamma_{\text{SF}}/\text{mJ}\cdot\text{m}^{-2}$	42.8	14.6
Boltzmann constant, $K/\times 10^{-23}$	1.38	1.38
Material constant, $A/\times 10^{-18} \text{ J}$	4.9	5.6
Reference strain rate, $\dot{\epsilon}_0/\text{s}^{-1}$	10^7	10^7
Adiabatic recovery factor, k_0	36.88	4.60
No thermal hardening factor, k_1	0.018	0.018
Proportional constant of twin spacing, i_t	0.14	0.14

4.3 Effect of dislocations and twins on mechanical properties of HEA

During the dynamic tensile process, a large number of dislocations and deformation twins with different morphologies are generated in $\text{Fe}_{40}\text{Mn}_{20}\text{Cr}_{20}\text{Ni}_{20}$ HEA. With the increase in strain rate, the dislocation distribution is more uniform, and the proportion of high-density dislocations is higher, as shown in Fig. 10 and Fig. 11. During the impact loading process, the dislocation lines cross each other to form high-density dislocations. At the same time, this trend will also lead to an increase in the Peierls-Nabarro force, which is the shear stress required to induce slip, thereby increasing the critical shear stress of slip and producing a strengthening effect on HEAs^[56]. With the increase in strain rate, the dislocations do not have enough time to slip to grain boundaries and disappear, which promotes the accumulation of dislocations^[57].

The main deformation mechanisms of $\text{Fe}_{40}\text{Mn}_{20}\text{Cr}_{20}\text{Ni}_{20}$ HEA at high strain rates are related to dislocations and twins. During the deformation process, high-density dislocations promote the work-hardening ability of HEA. However, a high dislocation density usually restricts the movement of

dislocations and induces dislocation accumulation, resulting in a serious decrease in plasticity of HEA at high strain rates, yielding a strength-ductility trade-off dilemma^[41,58]. Besides, by forming additional twin boundaries, deformation twins hinder the dislocation movement, increase the dislocation storage capacity, improve the alloy strength, and delay the occurrence of necking without significantly sacrificing plasticity^[59]. At the same time, at high strain rates, due to the activation of instantaneous high flow stresses, deformation twins are more easily activated than those at quasi-static state, reducing the average free path of deformation and enhancing the strengthening ability of dislocations^[60]. Generally, deformation twins are determined by the generation of SFE and grain size. Lower SFE and larger grain size will reduce the twin stress^[61]. SFE mainly triggers the formation of deformation twins, resulting in twin-induced plasticity^[60].

Compared with room-temperature dynamic tension, the $\text{Fe}_{40}\text{Mn}_{20}\text{Cr}_{20}\text{Ni}_{20}$ HEA under low-temperature dynamic tension produces dislocations and twins of higher density in different directions, as presented in Fig. 13. Generally, dislocation slip is difficult to start at low temperature, and high strain rates cannot provide sufficient time for dislocation slip, resulting in a local concentration of dislocations and decreased plasticity^[62-63]. However, SFE is decreased with the decrease in temperature, which increases the flow stress. At the same strain rate, more deformation twins are formed to improve the strength and plasticity of HEAs^[64]. Besides, a large number of deformation twin boundaries refine the grains and provide significant grain-boundary strengthening^[65]. It is reported that the samples under low-temperature dynamic loading show higher energy absorption than the samples under room-temperature dynamic loading^[56]. The synergistic effect of deformation twins and dislocation slip is the key to the simultaneous improvement of strength and plasticity of HEA under dynamic tension.

In order to more intuitively express the microstructure evolution mechanism of the $\text{Fe}_{40}\text{Mn}_{20}\text{Cr}_{20}\text{Ni}_{20}$ HEA after dynamic tension, a schematic diagram is shown in Fig. 17. The microstructure of the deformed HEA is mainly composed of deformation twins and dense dislocations, including dislocation lines, dislocation cells, dislocation tangles, and HDDWs. At low strain rates, the deformation mechanism of HEA after dynamic tension is still dominated by dislocation slip. With the increase in strain rate, the deformation twins are activated, becoming the main mechanism of plastic deformation of HEA at high strain rates. At low temperature, some deformation twins are also produced in HEA after deformation at a strain rate of 1000 s^{-1} . With the increase in strain rate, there are two kinds of deformation twins in different directions, and a large number of dislocations are piled up at the intersection of twins, which improves the work-hardening ability of HEA. Therefore, under dynamic tension, the interaction of different forms of dislocations and twins makes the strength and plasticity of the alloy increase simultaneously.

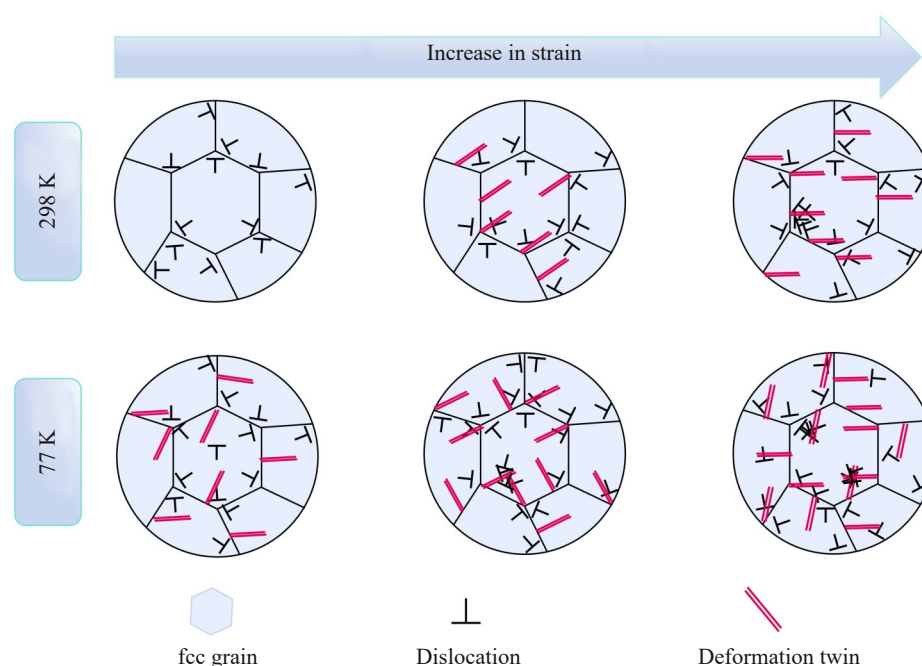


Fig.17 Schematic diagram of microstructure evolution of $\text{Fe}_{40}\text{Mn}_{20}\text{Cr}_{20}\text{Ni}_{20}$ HEA after dynamic tension

5 Conclusions

1) Under high-temperature tension, $\text{Fe}_{40}\text{Mn}_{20}\text{Cr}_{20}\text{Ni}_{20}$ HEA shows a particularly obvious softening phenomenon at 873 K, and the strength and plasticity are significantly reduced. Under low-temperature quasi-static tension, $\text{Fe}_{40}\text{Mn}_{20}\text{Cr}_{20}\text{Ni}_{20}$ HEA exhibits an excellent combination of strength and plasticity. At a strain rate of 10^{-3} s^{-1} , compared with those at the quasi-static state at room temperature, the yield strength and plasticity at low temperature increase by about 78% and 125%, which can reach 405 MPa and 97.5%, respectively. With the increase in strain rate, the yield strength of $\text{Fe}_{40}\text{Mn}_{20}\text{Cr}_{20}\text{Ni}_{20}$ HEA can reach 481 MPa, and the plasticity is still as high as 95.0%.

2) At room temperature, the yield strength of $\text{Fe}_{40}\text{Mn}_{20}\text{Cr}_{20}\text{Ni}_{20}$ HEA increases by about 97%, and the elongation is maintained at a high level (25.3%) when the strain rate is 1000 s^{-1} , compared with those after quasi-static tensile loading. With the increase in strain rate, the yield strength of $\text{Fe}_{40}\text{Mn}_{20}\text{Cr}_{20}\text{Ni}_{20}$ HEA can reach 995 MPa, and the plasticity is as high as 44.5%. At low temperature, the yield strength is 1120 MPa, and the plasticity is 45.7%.

3) During the dynamic tensile process, the interactions between different forms of dislocations and deformation twins together improve the strength and work-hardening ability of $\text{Fe}_{40}\text{Mn}_{20}\text{Cr}_{20}\text{Ni}_{20}$ HEA. At room temperature and a strain rate of 1000 s^{-1} , the deformation mechanism is still dominated by dislocation slip. With the increase in strain rate and the decrease in temperature, a large number of deformation twins are activated, which seriously hinders the dislocation motion and improves the work-hardening ability.

4) The temperature sensitivity and strain rate sensitivity to the yield strength of $\text{Fe}_{40}\text{Mn}_{20}\text{Cr}_{20}\text{Ni}_{20}$ HEA can be predicted

by the classical Z-A thermal-activation constitutive model, and the fitting results are in good agreement with the experimental results. At the same time, the Taylor model related to the flow stress can be used to predict the change of flow stress under dynamic tension. This model can be applied to the low-temperature dynamic loading, and the fitting results are basically consistent with the experimental results. Finally, the $\text{Fe}_{40}\text{Mn}_{20}\text{Cr}_{20}\text{Ni}_{20}$ HEA with excellent mechanical properties under dynamic loading is expected to be a candidate material for impact-resistant structural materials.

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Fe₄₀Mn₂₀Cr₂₀Ni₂₀高熵合金的低温动态拉伸变形机制

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摘 要: 探讨了Fe₄₀Mn₂₀Cr₂₀Ni₂₀高熵合金 (HEA) 在高/低温以及高应变率 (分离霍普金森杆) 条件下的力学行为和微观机制, 并利用相关理论模型来拟合该HEA的屈服强度和流动应力。拉伸测试结果表明, 在降低温度或增加应变率的条件下, 该HEA表现出优异的强度-塑性协同效应和出色的变形硬化能力。在动态拉伸过程中, 不同形式的位错和变形孪晶之间的相互作用共同提高了HEA的强度和变形硬化能力。Zerilli-Armstrong 本构模型被用于预测HEA屈服强度的温度敏感性和应变率敏感性。同时, 泰勒模型被用于预测HEA在动态拉伸作用下随应变率变化的流动应力, 并将该模型应用于低温动态实验中。拟合结果与实验结果一致, 为后续对HEA的强度预测提供了理论依据。

关键词: 高熵合金; 动态拉伸; 低温; 变形机制; 加工硬化

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