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ARTICLE

Effect of Hot Rolling on Microstructure and Mechanical Properties of N08120 Ni-Based Superalloy

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Abstract: The microstructure evolution and mechanical properties of the N08120 superalloy rolled at 780 °C and deformed through the cumulative reduction of 40%, 60%, and 80% with the single pass reduction of 10% were investigated. The experimental results show that MC and $M_{23}C_6$ carbides precipitate at grain boundaries and within the grains during deformation. These finely dispersed carbides effectively hinder dislocation motion, thereby enhancing the strength of the material. Dynamic recovery is identified as the predominant softening mechanism during hot rolling, where dislocations progressively reorganize into dislocation cell structures through slip and recombination. Owing to the strong pinning effect of grain boundaries on dislocations and their high interfacial energy, dislocations preferentially accumulate near grain boundaries, inducing localized recrystallization and the formation of fine recrystallized grains. With increasing deformation amount, the alloy texture transitions progressively from Rotated Goss $\{110\}<110>$ to Rotated Cube $\{110\}<112>$ and finally to the Copper $\{112\}<111>$ type. After 40% deformation, work hardening is significant, with the yield strength increasing from 322 MPa to 867 MPa and the microhardness rising from 196 HV_{0.5} to 337 HV_{0.5}. Beyond 40% deformation, both yield strength and microhardness tend to stabilize. This is mainly attributed to the evolution of carbides, dislocation density, substructure, and recrystallization in N08120 superalloy during hot rolling.

Key words: N08120 superalloy; hot rolling; microstructure; work hardening; mechanical properties

1 Introduction

The N08120 nickel-based superalloy is a high-performance material with extensive application potential in aerospace, petrochemistry, photovoltaic energy, and other fields, owing to its exceptional high-temperature mechanical properties and corrosion resistance^[1-6]. Nevertheless, its complex composition results in challenge to tailor the alloy performance, which significantly restricts its broader application. Hot rolling, as a well-established plastic deformation technique, offers the advantages of operational simplicity and a pronounced grain refinement effect^[7-9]. In the forming processes of nickel-based superalloy components, rolling is the most commonly used method for remarkably enhancing strength, ductility, and corrosion resistance^[10-11].

In recent years, deformation-induced twinning and dynamic

recrystallization behavior have primarily been focused on the microstructure control during the hot rolling process of nickel-based superalloys. For example, Soltani et al^[12] conducted rolling experiments on a new high-manganese austenitic steel at various temperatures and deformation levels, demonstrating that the introduction of dynamic recrystallization and twins synergistically improves the mechanical properties of the material. Li et al^[13] induced recrystallization and twins in AZ31B alloy through different deformation amounts, ultimately increasing the tensile strength by 90 MPa and achieving an elongation of 19.8%. Khalesian et al^[14] reported that during the hot rolling of steel with twinning-induced plasticity, deformation twins formed, promoting the nucleation of new recrystallized grains. Besides the twinning and recrystallization, the hot rolling process can also enhance

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mechanical properties in complex alloys by inducing precipitated phases. For instance, in the low-density Fe-21Mn-10Al-5Ni-C alloy, severe deformation under hot rolling leads to the formation of heterogeneous structures and high dislocation densities. Additionally, nanoscale B2-phase particles precipitate along grain boundaries and slip bands. These features effectively hinder dislocation motion, resulting in a simultaneous improvement in both ductility and hardness^[15]. During rolling deformation, the development of banded deformation structures, micro-twins, and crystallographic textures under high strain conditions significantly contributes to the increase in yield strength^[16-17]. Through the combined effects of work hardening induced by the secondary rolling and the precipitation of $\gamma'' + \gamma'$ strengthening phases, the yield strength of Inconel 718 can be increased to 954 MPa^[18]. Furthermore, short-term aging treatment of nickel-based alloys after rolling can promote the formation of nanoscale δ -phase precipitates, Lomer-Cottrell locks, and twin boundaries, leading to an increase in yield strength by more than 77% and an increase in tensile strength by over 26%^[19]. Therefore, hot-rolled nickel-based superalloys often develop micro-twins, preferred orientations, dislocation structures, and precipitates such as γ' , γ'' , and δ phases. Rational design and control of these microstructural features can significantly improve the overall mechanical performance of the alloy.

However, the evolution of microstructure and properties in the Ni-Cr-Fe-based N08120 alloy during hot rolling requires further investigation. Particularly, the recrystallization behavior, texture evolution, and strengthening mechanisms under high temperatures and varying deformation amounts remain obscure. In this research, rolling experiments were conducted at 780 °C with different deformation amounts (40%–80%) to examine the evolution of grain morphology, precipitate behavior, texture transformation, and dislocation structures in the N08120 alloy during hot deformation, along with the corresponding changes in the mechanical properties. This study represents an effort to control the microstructure and mechanical properties of Ni-Cr-Fe-based superalloys and provides valuable insights.

2 Experiment

The initial N08120 alloy was in the as-forged condition, and its chemical composition is listed in Table 1. The material was processed into blocks with dimensions of 50 mm×30 mm×20 mm for hot rolling. Fig.1 shows the phase diagram calculated using JMatPro software for the chemical composition. According to the phase diagram, only MC carbides are expected to precipitate at temperatures above 1017 °C. Below this temperature, however, brittle phases such as Laves phase, σ phase, G phase, and $M_{23}C_6$ carbides gradually form with decreasing temperature, which can

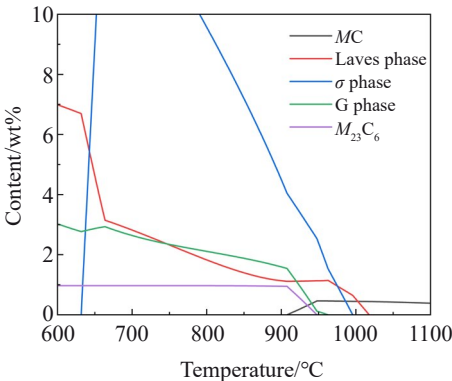


Fig.1 Local equilibrium phase diagram of N08120 superalloy

degrade the processability and mechanical properties of the materials. To dissolve these detrimental phases completely into the matrix and thereby to reduce material hardness prior to rolling, all samples underwent solution treatment at 1050 °C for 2 h before the rolling tests. Rolling was conducted on a Z-2-02 two-roll reversible mill for cold and hot rolling at 780 °C with the rolling speed of 0.5 m/s. Samples with deformation amount of 40%, 60%, and 80% were obtained by accumulative deformation with single pass reduction of 10%. Immediate water quenching was performed after rolling to preserve the high-temperature deformation microstructure.

The rolled samples were free of cracking or other macroscopic forming defects, as shown in Fig. 2a. The microstructures of the rolled samples were characterized using scanning electron microscopy (SEM), electron backscattered diffraction (EBSD), and transmission electron microscopy (TEM). For SEM analysis, the samples were first mechanically ground and polished, followed by electrolytic etching in a 10% oxalic acid aqueous solution at 5–6 V for 20–30 s. Prior to EBSD examination, the samples were electropolished in a 10% perchloric acid alcohol solution at 25 V for 25 s. The electrolyte for EBSD and TEM tests is a perchloric acid alcohol solution with volume ratio of perchloric acid to alcohol as 1 to 9. The acquired EBSD data were processed and analyzed using Channel 5 software. For TEM observation, the samples were first mechanically thinned to approximately 40 μ m in thickness, and they were further thinned via twin-jet electropolishing using a 10% perchloric acid solution. TEM observation was conducted immediately upon completion of the sample preparation.

The variation in hardness was measured using a Qness Q10A+microhardness tester under a load of 50 g and a dwell time of 15 s. Multiple indentations were randomly made on the surface of each sample to obtain an average value. Three tests were performed under every condition, and the average value was used as the final result. Tensile tests were conducted on an AGS-X 10KN universal testing machine at a constant crosshead speed of 0.5 mm/min. Small-sized tensile samples with a gauge length of 10 mm were used; the detailed dimensions are illustrated in Fig. 2b. Similarly, three tensile tests were conducted under each condition, and the average

Table 1 Chemical compositions of N08120 alloy (wt%)

Ni	Cr	C	Nb	W	Mn	Ti	Mo	Fe
39.29	25.32	1.56	0.39	0.42	0.11	0.15	0.80	Bal.

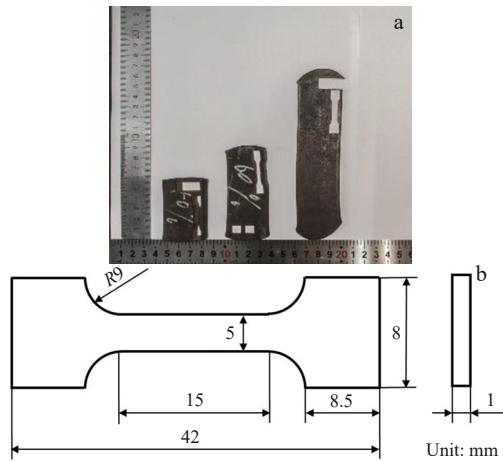


Fig.2 Appearances of rolled samples (a) and geometry of tensile sample (b)

value was adopted.

3 Results and Discussion

3.1 Precipitation

Fig. 3 presents the morphologies and corresponding composition analysis of the precipitates in as-received sample. As shown in Fig.3a and 3b, these finely dispersed precipitates distributed throughout the matrix exhibit sizes smaller than 10 μm . Their primary constituent elements are Nb, Ti, and Mo, identifying them as MC-type carbides^[20]. However, the element content varies among carbides of different sizes and morphologies, with the content of Nb being significantly higher than that of Ti and Mo. This composition distribution characteristic is closely related to the diffusion behavior of

each element in the Ni matrix. The activation energy for diffusion of Nb in the Ni matrix is higher than that of Ti (approximately 250 kJ/mol), and Nb exhibits a greater tendency for the secondary phase precipitation^[21–22]. Therefore, during the initial stages of carbide nucleation and growth, Nb preferentially combines with C to form stable carbide nuclei. As the precipitates continue to grow, Ti and Mo atoms gradually participate in the process, achieving partial substitution of Nb either by replacing some metal atoms in the MC carbides or through solid solution into the crystal lattice. Since the nucleation and growth processes of different carbides are not synchronized, their local composition is influenced by kinetic conditions, ultimately resulting in distinct elemental distribution characteristics for carbides with different morphologies.

Fig. 4 presents SEM images illustrating the grain morphology and distribution of precipitated phases in samples subjected to different deformation levels. With increasing deformation, the grains become progressively elongated along the rolling direction, and grain boundaries exhibit significant bending phenomenon. The maximum grain size and average grain size of the as-received sample are 265 and 38 μm , respectively. The maximum grain size of the sample after 40% deformation is 379 μm , while its average grain size decreases to 22 μm . The results indicate that the average grain size of N08120 alloy is reduced by rolling deformation, and the increase in maximum grain size may be the result of grain rotation and coalescence during high-temperature rolling. When the deformation amount reaches 60%, the deformed structures within the grains become notably more pronounced. With further deformation to 80%, a lamellar structure characteristic with face-centered cubic materials can be

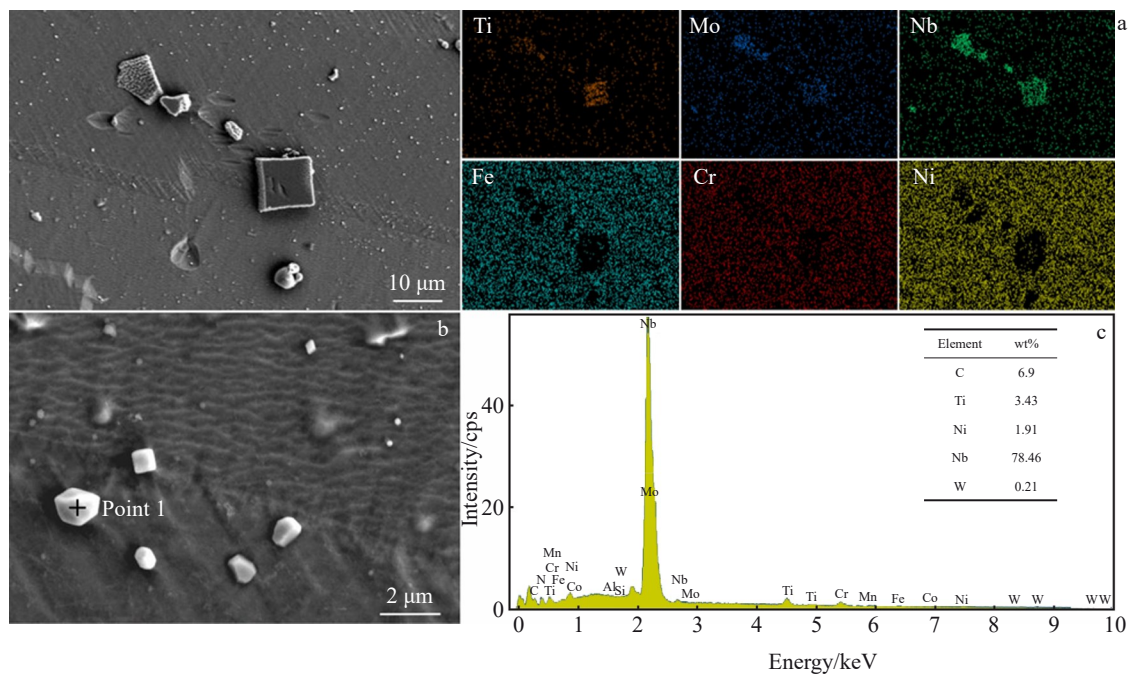


Fig.3 SEM microstructure and corresponding elemental mapping results of precipitates in as-received sample (a); high-magnification SEM image of precipitates (b); EDS analysis results of point 1 marked in Fig.3b (c)

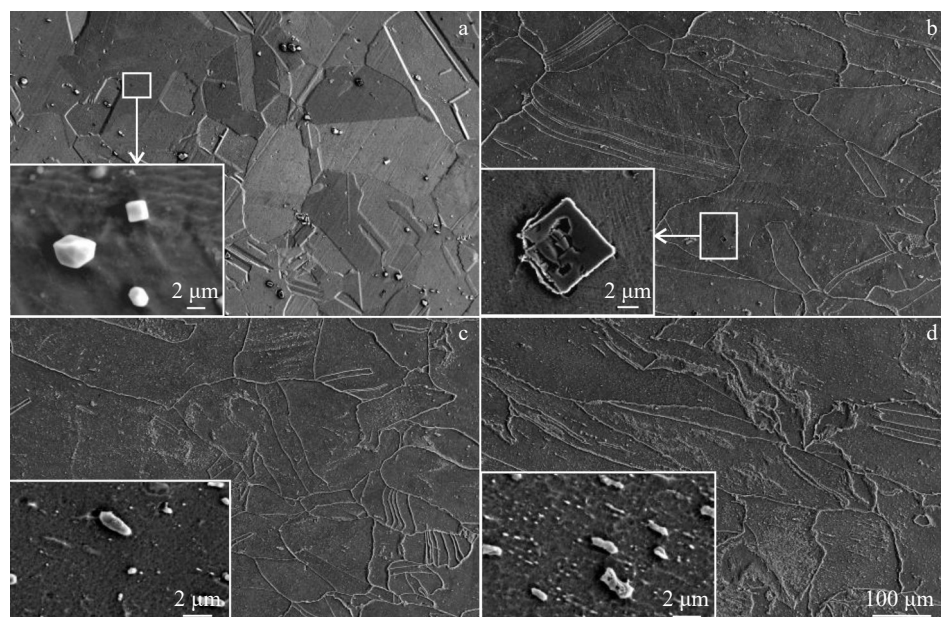


Fig.4 SEM microstructures of samples under different deformation levels: (a) as-received; (b) 40%; (c) 60%; (d) 80%

observed. Meanwhile, precipitated phases of different sizes demonstrate distinct evolutionary behavior during deformation. After 40% deformation, large precipitated phases with dimensions approaching or exceeding 10 μm exhibit visible cracking and begin to fracture, as shown in the inset of Fig. 4b. After 80% deformation, the fragmented precipitates undergo further separation along the deformation direction, ultimately forming dispersed particles with size smaller than 10 μm . The smaller precipitates with approximately 1 μm in size become gradually elongated with increasing deformation, and they show further growth in the sample after 60% deformation, eventually stabilizing at about 2 μm in length.

Fig. 5 presents TEM microstructures of grain boundary regions of samples under different deformation levels. The

initial microstructure (as-received sample) shows no carbide precipitation at grain boundaries. With increasing deformation, carbides progressively form along the grain boundaries. After 40% deformation, carbides precipitate in a discontinuous, strip-like morphology along the grain boundaries (Fig. 5b). When the deformation increases to 60%, the amount of carbide in the sample increases relatively (Fig. 5c). When the deformation is further increased to 80%, the carbides are obviously coarsened (Fig. 5d). The grain boundary precipitates consist of two types, namely $M_{23}C_6$ and MC carbides. Based on the relatively low content of strong carbide-forming elements such as Nb and Ti, combined with the high Cr content in the alloy, the formation of $M_{23}C_6$ carbides predominantly composed of Cr is favored during hot

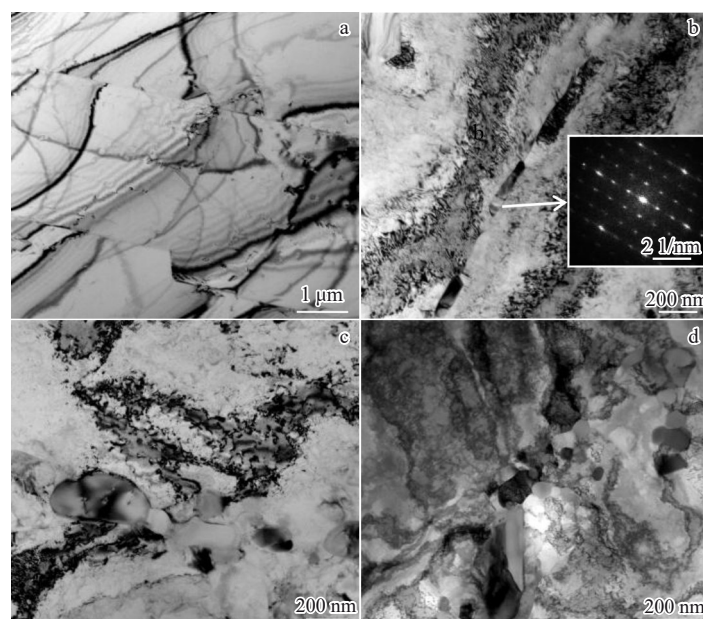


Fig.5 TEM images of the grain boundary regions of samples at different deformation levels: (a) as-received; (b) 40%; (c) 60%; (d) 80%

deformation. Consequently, $M_{23}C_6$ becomes the primary phase in the grain boundary precipitates.

Fig. 6a presents the elemental distribution characteristics of grain boundary precipitates of sample after 40% deformation. The results reveal the formation of MC and $M_{23}C_6$ carbides in a composite structure, characterized by MC phases attached to $M_{23}C_6$ surfaces or the core-shell structures with MC cores and $M_{23}C_6$ shells. Grain boundaries serve as fast diffusion channels for Nb, promoting preferential nucleation of MC carbides. The high-density dislocations generated during thermal deformation provide additional diffusion paths for Nb atoms, significantly enhancing their migration efficiency^[23]. However, owing to compositional constraints, MC carbides exhibit limited continuous growth, accompanied by the phase transformation of $MC + \gamma \rightarrow M_{23}C_6 + \gamma'$. The high Fe content in the matrix suppresses the stable existence of the γ' phase, causing the newly formed γ' phase to dissolve rapidly and eventually form the characteristic core-shell microstructure of $M_{23}C_6$ encapsulating MC . This phenomenon was also observed by Sun et al.^[24] in as-cast nickel-based alloys. Notably, the formation of $M_{23}C_6$ consumes surrounding Cr atoms, creating Cr-depleted zones without affecting Nb distribution. This configuration facilitates further MC nucleation at the $M_{23}C_6$ edges. Fig. 6b displays the distribution of intragranular precipitates of sample after 80% deformation. Intragranular carbides begin to precipitate after 40% deformation and undergo significant growth after 80% deformation. Their types, structures, and distribution characteristics show strong consistency with those at grain boundaries, indicating that carbide precipitation in both regions is governed by the same diffusion mechanism. The dislocation networks introduced by deformation provide effective pathways for the migration of

Nb and other elements, thereby guiding similar carbide precipitation behavior across different microstructural regions.

3.2 Texture

Fig. 7 presents the grain orientation distributions of the samples after different deformation processes, where red, blue, and green correspond to the $\langle 001 \rangle$, $\langle 111 \rangle$, and $\langle 101 \rangle$ orientations, respectively. In the as-received sample, the grain orientations are relatively random, and each individual grain exhibits nearly uniform orientation, indicating a low initial strain state (Fig. 7a). With the increase in deformation to 40%, regions with different orientations emerge within some grains, although no high-angle grain boundaries form between these regions. This phenomenon suggests the development of considerable orientation gradients within the grains, which are potential sites for grain subdivision and subsequent dynamic recrystallization (Fig. 7b). Concurrently, the grain orientations begin to intensify near the $\langle 111 \rangle$ direction. After 60% deformation, fine recrystallized grains are observed to distribute in a necklace-like arrangement along the high-angle serrated grain boundaries, forming a characteristic chain-shaped dynamic recrystallization structure (Fig. 7c). These serrated grain boundaries are characterized by high local orientation gradients and interfacial energy. Consequently, they act as favorable sites for recrystallization nucleation. Furthermore, Fig. 7c reveals the formation of fine grains along the original grain boundaries in the deformed sample, with evident grain boundary bulging, signifying the presence of a discontinuous dynamic recrystallization softening mechanism dominated by nucleation at bulged grain boundaries. As the deformation is further increased to 80%, the grain orientations become more aligned and predominantly stabilized along the $\langle 101 \rangle$ direction, indicating the gradual evolution

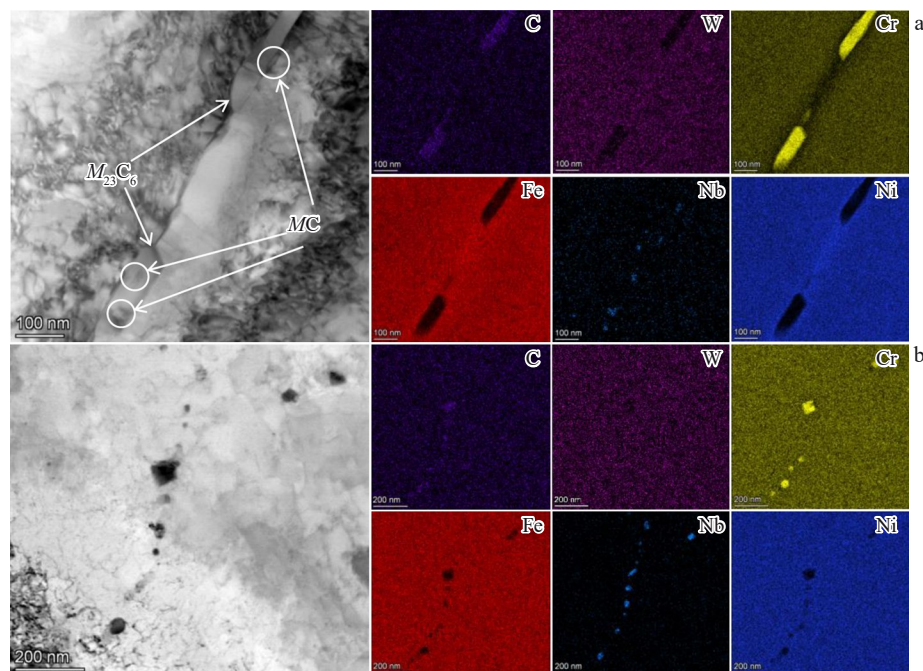


Fig. 6 SEM images and corresponding EDS mappings of grain boundary precipitates in sample after 40% deformation (a) and intragranular precipitates in sample after 80% deformation (b)

and stabilization of the materials' texture during the progressive deformation (Fig. 7d).

During the rolling deformation of metals, grains tend to rotate toward specific spatial orientations, thereby forming crystallographic textures. Fig. 8 presents the orientation distribution function (ODF) diagrams of the N08120 alloy under different deformation conditions, revealing significant texture transformations. For face-centered cubic metals, texture components are typically characterized in Euler space at $\varphi_2=0^\circ$, 45° , and 65° sections. Three texture components were identified in the as-received sample in the form of P {011} <112>, Rotated Goss {110} <110>, and Brass {110} <112>. Among them, the Rotated Goss texture exhibits the maximum intensity of 7.42 (Fig. 8a). When the deformation increases to 40%, the original P and Brass textures disappear, while Goss {011} <100> and Rotated Cube {110} <112> textures emerge. The intensity of the latter textures increases significantly to 5.19 (Fig. 8b). After 60% deformation, the Goss texture is further strengthened, and typical Copper {112} <111> and S {123} <634> textures are formed, with the Copper texture intensity reaching 5.19 (Fig. 8c). Previous studies indicate that the development of Goss, Copper, and S textures primarily results from plastic deformation mechanisms dominated by dislocation slip. The enhancement in these textures generally contributes to an increase in yield strength^[18]. With further increase in deformation to 80%, the Goss texture disappears, the Rotated Cube texture emerges, and the intensity of the Copper texture increases markedly to 11.1 (Fig. 8d). The continuous strengthening of the Copper texture is attributed to the consistent stress direction during unidirectional rolling, which promotes further alignment of grain orientations and leads to texture intensification. The formation and evolution of the Copper texture, a typical rolling texture in high-stacking-fault-energy face-centered cubic materials, confirm that dislocation

slip serves as the dominant plastic deformation mechanism in this alloy^[25-26].

3.3 Dislocation behavior

The kernel average misorientation (KAM) map effectively characterizes dislocation density distribution and plastic strain concentration by calculating the average orientation difference between each measurement point and its neighboring points. Fig. 9 presents KAM maps of samples under different deformation levels. The color gradient from blue to red corresponds to the transition from low to high dislocation density. The as-received sample appears predominantly blue, indicating a relatively low initial dislocation density in this sample. With increasing deformation, the blue regions gradually transition toward red (Fig. 9a). Fig. 9b reveals high dislocation density at high-angle grain boundaries, where distinct red zones appear, indicating significant strain concentration along with characteristic high dislocation density and stress concentration. In contrast, portions of the grain interiors remain blue, suggesting relatively lower dislocation density. This distribution results mainly from the preferential nucleation and accumulation of dislocations at grain boundaries, combined with limited intragranular dislocation sources and dynamic recovery. These factors lead to a slower increase in dislocation density within grains compared to grain boundaries, remaining below the average KAM value. Furthermore, numerous fine grains are observed along the original coarse grain boundaries in the sample after 40% deformation. These fine grains appear blue, indicating low internal dislocation density. When deformation increases to 60%, the number of such fine grains rises significantly (Fig. 9c). This phenomenon occurs because continued dislocation accumulation at grain boundaries raises the interfacial energy and induces local recrystallization at elevated temperatures during deformation, forming fine

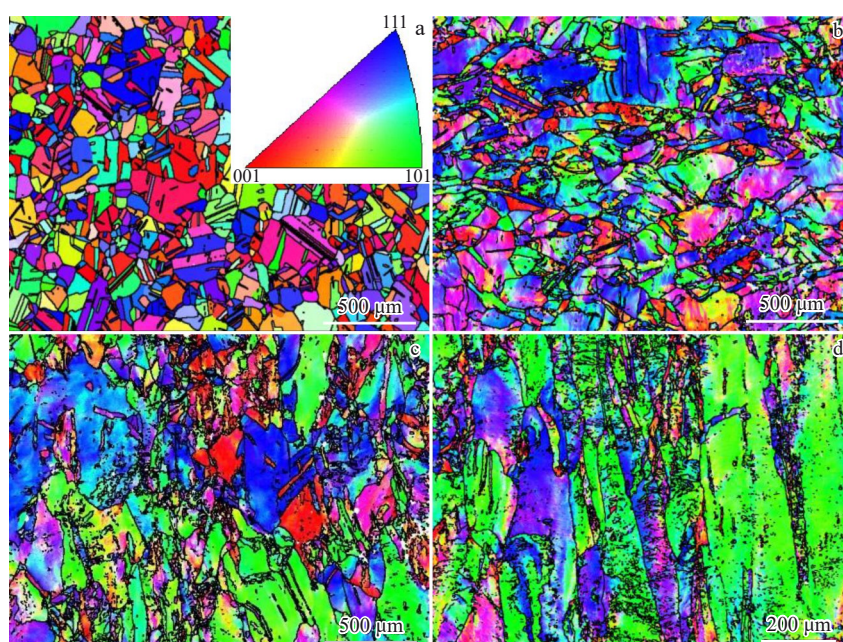


Fig.7 Inverse pole figures of samples at different deformation levels: (a) as-received; (b) 40%; (c) 60%; (d) 80%

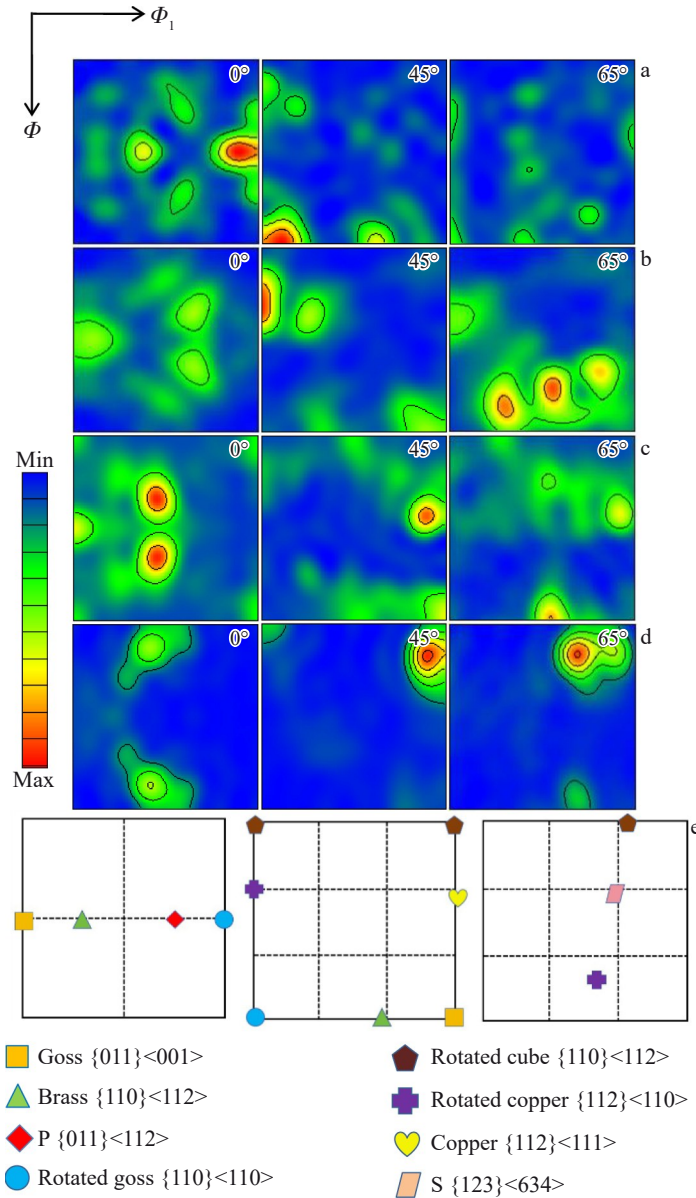


Fig.8 ODF sections at $\varphi_2=0^\circ$, 45° , and 65° of as-received sample (a) and samples after 40% (b), 60% (c), and 80% (d) deformation; schematic diagram of ODF sections for ideal orientations (e)

grains. Simultaneously, through recombination, entanglement, and accumulation, dislocations continuously increase the misorientation angle at original grain boundaries. When this angle exceeds 15° , low-angle grain boundaries are gradually transformed into high-angle grain boundaries, thereby promoting the formation of a refined grain structure.

KAM provides a quantitative measure of the local lattice distortion at the microscopic scale. During rolling deformation, this parameter reveals a significant increase in dislocation density. Based on strain gradient theory, the KAM value allows for the quantification of the geometrically necessary dislocation (GND) density, which is calculated using the following expression^[27–28]:

$$\rho_{\text{GND}} = \frac{2\text{KAM}_{\text{ave}}}{\mu b} \quad (1)$$

where KAM_{ave} represents the average KAM value within the

selected area, μ denotes the EBSD scanning step size, and b is the value of Burgers vector. The calculated GND densities of the as-received sample and the samples after 40%, 60%, and 80% deformation are 6.543×10^{12} , 1.127×10^{14} , 1.201×10^{14} , and $2.776 \times 10^{14} \text{ m}^{-2}$, respectively. The results demonstrate that the GND density increases progressively with rolling deformation, although it remains at a relatively low level overall. The growth rate of GND density exhibits distinct stages during rolling. The most substantial increase occurs in sample after 40% deformation, where extensive dislocation multiplication leads to significant work hardening. As deformation further increases from 40% to 80%, the growth rate of GND density slows down considerably. During this stage, the high cumulative dislocation density enhances the probability of dislocation recombination, while dislocations with opposite signs on the same slip plane undergo annihilation. These

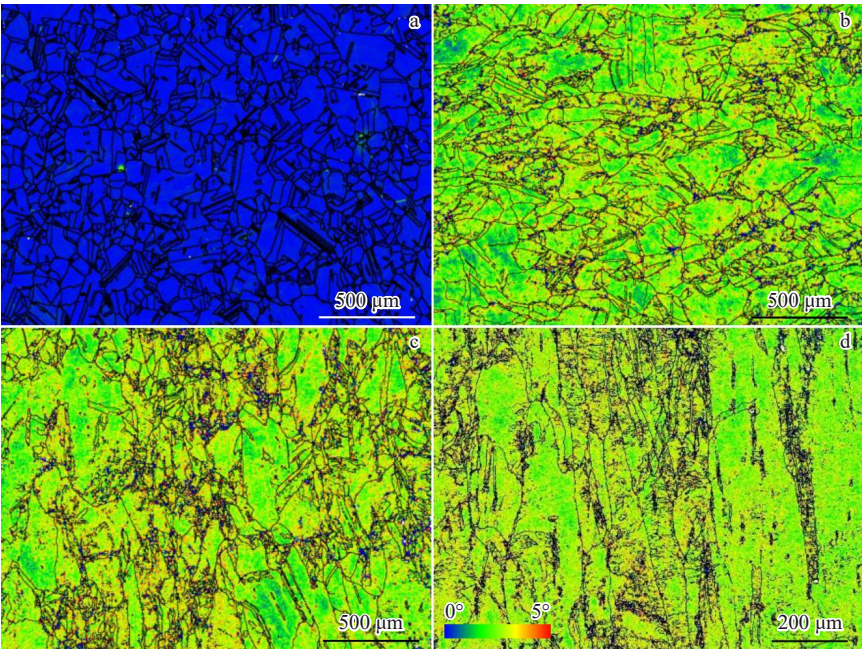


Fig.9 KAM maps of samples at different deformation levels: (a) as-received; (b) 40%; (c) 60%; (d) 80%

mechanisms promote an enhanced softening effect, ultimately establishing a dynamic balance between dislocation multiplication and the dislocation reduction caused by dynamic recovery.

Fig. 10 illustrates the evolution of dislocation structures at different deformation stages. Observation reveals that numerous dislocations become entangled and gradually assembled into dislocation walls, forming typical deformation substructures. With increasing strain, the dislocation density rises continuously, and the regions of dislocation entanglement expand and become more pronounced. A distinct dislocation wall gradually forms at the core of these

entangled regions, with its width progressively increasing with further deformation. As shown in Fig. 10a, dislocation cell structures have already developed, characterized by high-dislocation-density cell walls surrounding intracellular regions of low dislocation density. After 80% deformation, well-defined dislocation cells have evolved into stable subgrain structures. The evolution of these dislocation configurations provides strong evidence that dynamic recovery serves as the dominant softening mechanism during hot rolling. Through dislocation glide, recombination, and rearrangement, the material progressively forms organized dislocation walls and cellular structures, effectively releasing the stored energy.

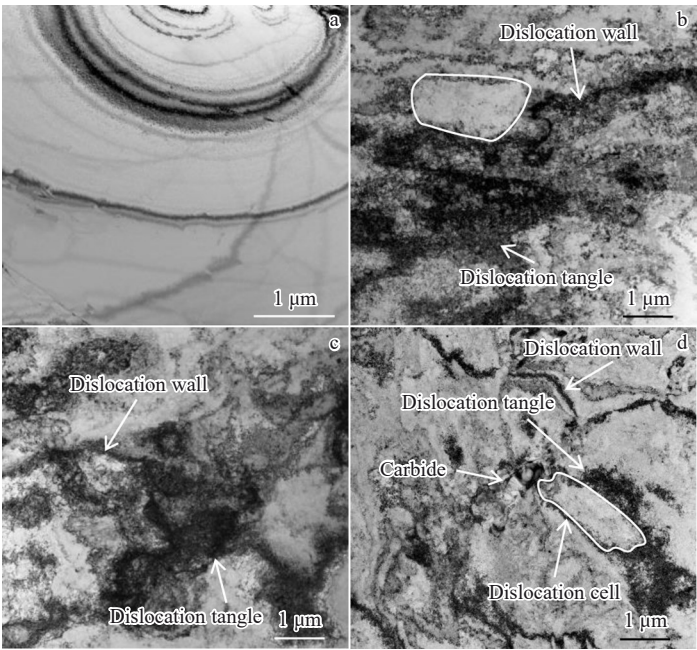


Fig.10 TEM micrographs of samples at different deformation levels: (a) as-received; (b) 40%; (c) 60%; (d) 80%

3.4 Mechanical properties

Fig. 11a presents the engineering stress-engineering strain curves of samples subjected to different deformation amounts. The results demonstrate that hot rolling deformation significantly enhances the material strength, along with a decrease in elongation. The yield strength and elongation of the as-received sample are 322 MPa and 57%, respectively, while the yield strength of the deformed sample remains around 860 MPa, and the elongation decreases to 11%–19%. The yield strength increases to 867 MPa after 40% deformation. The tensile properties of samples under various deformation conditions are summarized in Table 2. Notably, the work hardening effect is the most pronounced of sample after 40% deformation, the tensile strength increases by 34.29%, and the yield strength rises by 169.04%, compared to those of the as-received sample. As deformation amount further increases to 60% and 80%, the strength parameters reach a plateau. When the total rolling reduction is 40%, the yield strength and tensile strength increase significantly, while the elongation decreases due to the increase in dislocation density, average grain size, and precipitation of carbides. However, the yield strength and tensile strength remain substantially unchanged as the cumulative deformation increases to 60% and 80%. Although the GND density increases with the increase in deformation, the strengthening effect is weakened by the coarsening of carbides and the formation of substructure and recrystallization, which shows that the work hardening and dynamic recovery softening reach a dynamic equilibrium stage. Fig. 11b displays the average microhardness values of samples under different deformation conditions. In metallic materials, microhardness reflects the resistance of material to plastic deformation and deformation resistance in the localized region around the indentation. The initial

microhardness of the as-received sample is approximately 196 HV_{0.5}. After 40% deformation, the microhardness increases substantially to 337 HV_{0.5}, representing a 71.5% enhancement. With further deformation, the microhardness remains stable within the experimental error range, consistent with the observed trend in strength properties. The improvement in mechanical properties after hot rolling deformation is closely associated with the evolution of microstructural characteristics.

In conclusion, a microstructure containing recrystallized grains and high-density dislocations was formed in the sample deformed by 40%. Generally, high-density dislocations exert a strengthening effect on alloys^[29]. The formation of recrystallized grains, however, has a dual influence on the mechanical behavior of the material. On the one hand, the newly formed grain boundaries act as effective barriers to dislocation motion, enhancing the work hardening effect through pinning. On the other hand, the recrystallization process alleviates local stress concentrations by consuming stored energy in the high-strain regions near the original grain boundaries. This mechanism preserves the grain boundaries' ability to impede dislocations while improving the overall stress distribution in the material. Such a coordinated effect allows the material to retain the essential plastic compatibility while achieving enhanced strength. Furthermore, according to the Hall-Petch relationship, the fine recrystallized grains formed during rolling contribute to the strength improvement^[30]. When the cumulative deformation increases to 60%, the recrystallized grains and carbides in the alloy continue to increase. However, when the cumulative deformation amount is 80%, the carbide in the alloy is obviously coarsened, which weakens the strengthening effect. Both intergranular and intragranular precipitates effectively impede the dislocation glide and grain boundary migration,

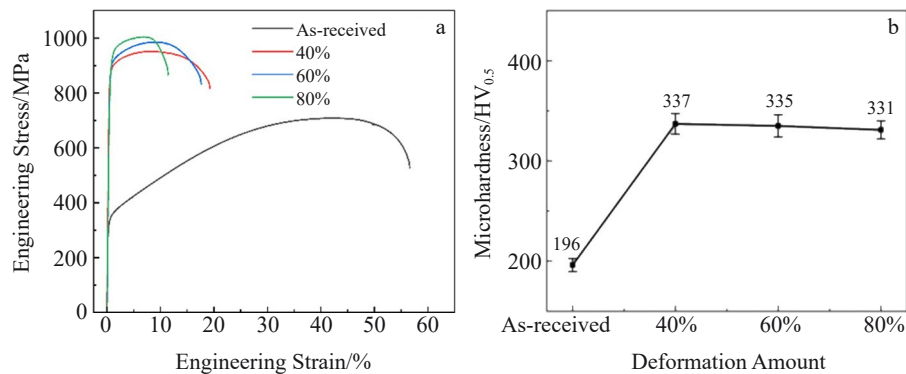


Fig.11 Engineering stress-engineering strain curves (a) and microhardness (b) of samples with different deformation amounts

Table 2 Mechanical properties of N08120 superalloy at different deformation levels

Deformation level	Yield strength, $\sigma_{0.2}$ /MPa	Tensile strength, σ_b /MPa	Elongation, δ /%	Microhardness/HV _{0.5}
As-received	322±14	708±21	57±3	196±7
40%	867±19	951±34	19±1	337±10
60%	862±26	985±39	18±3	335±11
80%	864±23	1004±33	11±2	331±9

thereby restricting the grain boundary mobility and enhancing the material strength. Concurrently, a Copper-type texture progressively develops since 60% deformation. Its evolution strengthens the resistance to dislocation slip through crystallographic-preferred orientation, further promoting the increase in yield strength^[25]. In this stage, the dislocation structure also evolves markedly. As shown in Fig. 10b, high-density dislocation tangles, and the widened dislocation walls are formed. Such configurations promote the dislocation recombination and annihilation, intensifying the dynamic recovery effect and enabling the material to maintain a certain softening capacity during continued deformation. Therefore, with the increase in cumulative deformation, although the GND density is increasing, the mechanical properties are basically unchanged in the later stage of the rolling deformation due to the dynamic softening and carbide coarsening caused by dislocation interaction.

4 Conclusions

1) With increasing deformation, the initial precipitates undergo progressive fragmentation and refinement. After 40% deformation, finely dispersed carbides are precipitated at the grain boundaries and within the grain interiors. These carbides subsequently experience coarsening during further deformation, effectively impeding the dislocation motion and thereby enhancing the material strength.

2) The crystallographic texture evolves from Rotated Goss $\{110\}<110>$ to Rotated Cube $\{110\}<112>$, eventually stabilizing as the Copper $\{112\}<111>$ texture.

3) Below 40% deformation, rapid dislocation multiplication results in significant work hardening. Beyond this threshold, the dislocation density growth rate is moderated considerably, where dynamic recovery emerges as the dominant softening mechanism, while the dynamic recrystallization remains confined to regions adjacent to original grain boundaries.

4) After 40% deformation, the material exhibits a substantial enhancement in yield strength and microhardness, though with a corresponding reduction in elongation. Further deformation beyond this level leads to stabilized mechanical properties. Overall, the condition of 40% deformation achieves a favorable strength-ductility synergy.

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热轧对N08120镍基高温合金微观组织与力学性能的影响

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摘 要: 研究了N08120高温合金在780 ℃和单次变形量为10%的热轧过程中, 通过累计变形获得40%、60%、80%变形量试样的微观组织演变和力学性能。结果表明, 变形过程中, 晶界及晶内均有MC和 $M_{23}C_6$ 碳化物析出, 这些弥散分布的碳化物能够有效阻碍位错运动, 提升材料强度。动态回复是热轧过程中的主要软化机制, 位错通过滑移和重组逐步形成位错胞结构。由于晶界对位错的强烈阻碍作用及较高的界面能, 位错优先在晶界处聚集, 进而诱发局部再结晶, 形成细小的再结晶晶粒。随着变形量增大, 合金织构由Rotated Goss {110}<110>逐渐转变为Rotated Cube {110}<112>, 并最终演化为Copper {112}<111>类型。在40%变形量下, 加工硬化效应显著, 屈服强度由322 MPa提升至867 MPa, 硬度由196 HV_{0.5}增至337 HV_{0.5}; 当变形量超过40%后, 材料的屈服强度与硬度趋于稳定。这主要归因于热轧过程中的N08120高温合金的碳化物、位错密度、亚结构以及再结晶演变。

关键词: N08120 高温合金; 热轧; 微观组织; 加工硬化; 力学性能

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