

Article

Development of a Flow-Stress Constitutive Model and Hot-Processing Window for a High-Nb, Ultra-High-Strength (205 ksi) Nickel-Based Corrosion-Resistant Alloy

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Abstract

The development of unconventional deep oil and gas resources requires 205 ksi high-Nb Ni-based corrosion-resistant alloys; however, the lack of hot-forming databases and processing maps for this grade has frequently resulted in hot cracking during ingot forging. To address this gap, this study investigated the hot-deformation behavior of this alloy and optimized its forming parameters. Isothermal-compression tests were performed using a Gleeble 3800-GTC simulator (Dynamic Systems Inc., Poestenkill, NY, USA) at 1173–1323 K and strain rates of 0.01–10 s⁻¹, followed by quantitative electron backscatter diffraction (EBSD) characterization of grain size, kernel average misorientation (KAM), and grain-boundary misorientation. A strain-compensated Arrhenius constitutive model was developed, giving a deformation activation energy of 540.33 kJ/mol and an average absolute relative error of only 4.77%. Dynamic materials model (DMM)-based power-dissipation and instability maps were constructed for this alloy for the first time, and an optimal hot-working window of 1260–1300 K and 0.01–0.1 s⁻¹ was identified. Under the representative condition of 1273 K and 0.1 s⁻¹, a uniform fine-grained microstructure with well-developed dynamic recrystallization was obtained without cracking, with an average grain size of 10.91 μm, a dynamic recrystallization (DRX) fraction of 48.4%, and an average KAM value of 0.476°. This work establishes the first complete hot-processing map system for the 205 ksi high-Nb Ni-based alloy, clarifies the coupled effects of dislocation evolution and dynamic recrystallization, and provides theoretical guidance for industrial hot forging and microstructural control.

Keywords: ultra-high-strength nickel-based corrosion-resistant alloy; high-temperature deformation; flow stress; processing map; dynamic recrystallization

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1. Introduction

With the continuous development of deep and ultra-deep oil and gas resources, downhole service environments have become increasingly severe, involving high pressure, high temperature, corrosive media, and complex mechanical loading. Under such conditions, conventional corrosion-resistant alloys and medium-strength nickel-based alloys may not simultaneously satisfy the requirements for ultra-high strength, toughness,

corrosion resistance, and long-term structural reliability. As a typical precipitation-strengthened nickel-based alloy, the 205 ksi high-Nb corrosion-resistant nickel-based alloy combines ultra-high strength, high toughness, and excellent corrosion resistance, making it a promising candidate material for the development of unconventional oil and gas resources in deep-water and deep subsurface environments [1–7]. However, during thermomechanical processing operations such as hot forging, billet breakdown, and cogging, this alloy is highly sensitive to deformation temperature, strain rate, and deformation degree. Improper processing parameters can readily induce cracking, localized flow instability, and grain coarsening, thereby significantly compromising the service performance and reliability of the final components.

Investigating high-temperature deformation behavior provides an essential theoretical basis for designing and optimizing hot-working processes. Because complex microstructural evolution occurs during hot deformation, constitutive modeling and processing-map construction are critical for understanding deformation mechanisms and optimizing processing parameters [8–10]. Based on the DMM proposed by Prasad et al. [11], processing maps can characterize power-dissipation efficiency and flow-instability domains under different deformation temperatures and strain rates, thereby enabling the identification of safe processing regions, instability zones, and optimal processing windows. Accordingly, processing maps play a key role in minimizing forming defects and improving both processing efficiency and microstructural homogeneity. Meanwhile, Arrhenius-type constitutive equations and the Zener–Hollomon parameter have been widely used to describe the high-temperature deformation behavior of superalloys. Accurate constitutive equations not only establish quantitative relationships between flow stress and deformation parameters but also provide fundamental models for numerical simulation and process prediction in hot-working applications [12–14].

Considerable progress has been made in studies of the hot-deformation behavior of nickel-based superalloys. Liao et al. [15] investigated dynamic recrystallization mechanisms in related alloys. Li et al. [16] systematically examined the effects of heat treatment on the microstructure and high-temperature tensile properties of GH4065A alloy. Zhong et al. [17] optimized the heat-treatment parameters of GH4738 superalloy using the Taguchi–Grey relational analysis method. Rui et al. [18] studied the rheological behavior of GH901 superalloy under different deformation temperatures and strain rates and established corresponding constitutive equations and processing maps. In addition, extensive investigations of the processing maps of GH4169 and GH4738 superalloys have provided valuable guidance for industrial hot-working practice.

However, GH4169 is generally classified as a 140 ksi strength-grade alloy, whereas GH4738 belongs to the 120 ksi strength grade; neither satisfies the requirements of the 205 ksi ultra-high-strength level. Moreover, most of these alloys were originally developed for aerospace, gas-turbine, or ultra-supercritical power-generation applications, where the primary design objectives are creep resistance, oxidation resistance, and high-temperature microstructural stability. Consequently, their alloy systems and service conditions differ substantially from those of high-strength corrosion-resistant nickel-based alloys designed for harsh oil and gas environments. Although constitutive modeling and processing-map analysis have been widely applied to superalloys such as GH4169 and GH4738, studies on the hot-deformation behavior and processing maps of high-Nb corrosion-resistant nickel-based alloys, especially those with a 205 ksi strength grade, remain limited.

In summary, compared with conventional nickel-based superalloys and corrosion-resistant alloys, the 205 ksi high-Nb corrosion-resistant nickel-based alloy investigated in this study exhibits a unique combination of ultra-high strength, elevated Nb content, precipitation-strengthening behavior, and excellent resistance to aggressive oil and gas envi-

ronments. Despite its promising application potential, systematic studies of its hot-deformation behavior, constitutive modeling, processing maps, and EBSD-based microstructural characterization remain insufficient, hindering the development and optimization of industrial thermomechanical processing routes. To address this research gap, the present study systematically investigates the high-temperature flow-stress behavior of the 205 ksi high-Nb alloy, establishes its constitutive equations and processing maps, and elucidates its dynamic recrystallization mechanisms through EBSD characterization. Correlations among flow-stress behavior, deformation activation energy, processing-stability domains, and dynamic recrystallization mechanisms are further elucidated, providing both theoretical guidance and experimental support for optimizing hot-working processes for this high-strength corrosion-resistant nickel-based alloy.

2. Materials and Methods

2.1. Materials

The material used in this study was a newly developed ultra-high-strength nickel-based corrosion-resistant alloy ingot with a high Nb content. The chemical composition of the alloy is listed in Table 1. The specimen surfaces were ground and polished to ensure uniform deformation during testing.

Table 1. Chemical composition of the ultra-high-strength nickel-based corrosion-resistant alloy (wt.%).

Element	Ni	Cr	Nb	Mo	Ti	Al	C	Mn	Fe
Wt./%	55	21.5	6	3.1	1	0.7	0.01	0.2	Bal.

Figure 1 shows the initial microstructure and EDS analysis of the ultra-high-strength nickel-based corrosion-resistant alloy after homogenization treatment. As shown in Figure 1a, the alloy exhibits a relatively uniform matrix, and no obvious dendritic segregation is observed. This indicates that homogenization effectively reduced the chemical segregation formed during solidification and provided a uniform initial microstructure for subsequent hot-compression tests. The elemental maps in Figure 1b show that Nb, Ti, Mo, and C are not distributed in the same manner. Mo is relatively uniformly distributed in the matrix, suggesting that it is mainly dissolved in the γ matrix and contributes to solid-solution strengthening and corrosion resistance. In contrast, Nb and Ti show obvious local enrichment in discrete particles. The C map also shows enrichment in similar regions, indicating that these particles are Nb/Ti-rich carbonitride-like particles. Compared with the surrounding matrix, the selected particles show pronounced enrichment of Nb and Ti, together with detectable C and N signals. This chemical feature is consistent with the possible presence of Nb/Ti-rich carbonitride-like particles. Therefore, the homogenized alloy can be described as consisting of a relatively uniform γ matrix containing a small amount of Nb/Ti-rich carbonitride-like particles. These particles may influence subsequent hot-deformation behavior by affecting dislocation motion, dynamic recrystallization nucleation, and grain-boundary migration. In particular, such Nb/Ti-rich particles may act as local stress-concentration sites during deformation, while finer particles may also restrict grain growth and contribute to microstructural refinement [19]. This initial microstructural feature provides an important basis for interpreting the later flow-stress behavior, dynamic recrystallization evolution, and processing-map results.

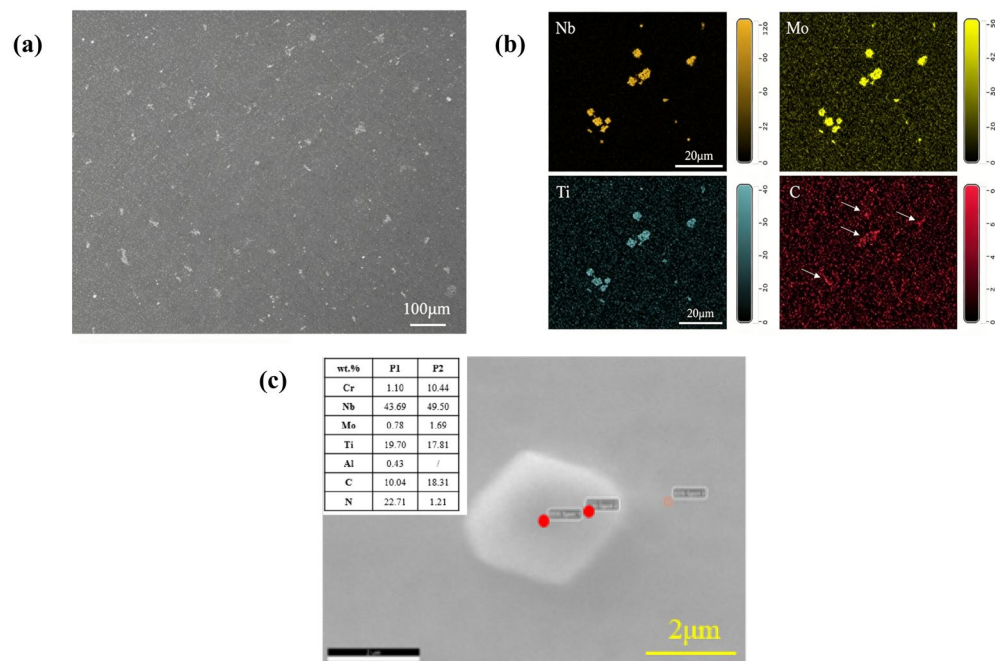


Figure 1. (a) Initial microstructure; (b) elemental mapping images of the ultra-high-strength nickel-based corrosion-resistant alloy; (c) EDS point-scan elemental analysis.

2.2. Methods

The thermal-compression procedure is summarized in Figure 2. Cylindrical hot-compression specimens with dimensions of $\Phi 8 \text{ mm} \times 12 \text{ mm}$ were machined along the radial direction of homogenized alloy ingots via wire electrical discharge machining (WEDM; model DK7732, Jiangsu Chuangwei CNC Machine Tool Co., Ltd., Taizhou, China). The cutting voltage was set as 36 V and wire feed speed was controlled at 2 mm/min to minimize thermal damage on specimen surfaces. After cutting, all specimens were ground sequentially by 400#, 800# and 1200# silicon carbide abrasive papers to eliminate surface cutting defects and guarantee strict parallelism of two compression end faces. A thin layer of graphite lubricant was evenly coated on both end surfaces of each specimen before thermal-compression tests, which effectively reduced friction between the sample and tungsten anvils and avoided uneven radial deformation under high-temperature loading.

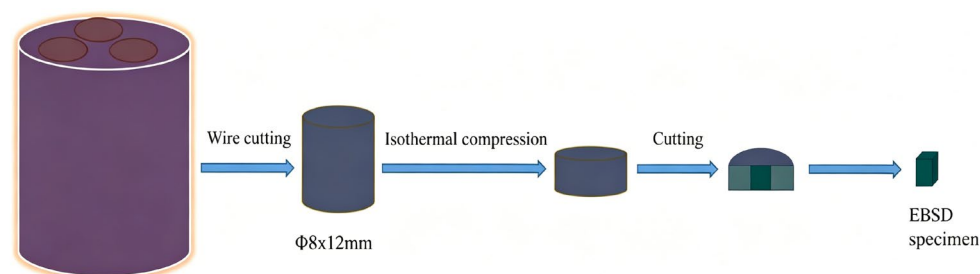


Figure 2. Flowchart of the thermal-compression test.

The deformation temperatures ranged from 1173 to 1323 K, and the strain rates ranged from 0.01 to 10 s^{-1} . An additional statement in the experimental section that the 1123 K/0.01 s^{-1} specimen was used only for fracture observation and was not included in the flow-curve dataset, constitutive-model fitting, or processing-map construction. Each specimen was heated in a Gleeble 3800-GTC simulator (Dynamic Systems Inc., Poestenkill, NY, USA) to the selected temperature at 274 K/s, held for 3 min to reduce temperature

gradients, and then compressed at the prescribed constant strain rate to a true strain of 0.92, corresponding to approximately 60% height reduction. A representative true strain of 0.5 was selected for the construction of the constitutive equation and processing maps. Water quenching was performed immediately after deformation to retain the high-temperature microstructure. The recorded load-displacement curves were converted into true stress-true strain data. After water quenching, all compressed cylindrical samples were longitudinally cut along the compression axis by WEDM to obtain central longitudinal cross-sections containing complete gradient of deformed microstructures. The cut surfaces were mechanically ground with 400#, 800#, 1200#, 2000# and 3000# SiC abrasive papers, followed by rough polishing with 2.5 μm diamond paste and fine polishing with 0.5 μm diamond suspension.

Electrolytic polishing was conducted to remove surface residual stress and obtain deformation-free surfaces suitable for EBSD testing. The electrolyte consisted of 10 vol.% perchloric acid and 90 vol.% anhydrous ethanol; polishing temperature was maintained between 243 K and 253 K, applied voltage was 22 V, and polishing time was 45 s. After electrolytic polishing, samples were ultrasonically cleaned in anhydrous ethanol for 5 min and dried with cold nitrogen flow to avoid surface oxidation. EBSD characterization was conducted on a Sigma 300 field emission scanning electron microscope (Carl Zeiss Microscopy GmbH, Jena, Germany) equipped with an Oxford NordlysNano EBSD detector (Oxford Instruments NanoAnalysis, High Wycombe, United Kingdom). The accelerating voltage was set to 20 kV, working distance was 15 mm, and EBSD scanning step size was 0.3 μm . For each deformation condition, at least three independent rectangular scanning regions of $120 \times 120 \mu\text{m}^2$ were measured to ensure statistical representativeness of microstructure data. EBSD was used to characterize grain size, local misorientation, and recrystallization state, thereby revealing the effects of deformation temperature and strain rate on microstructural evolution.

3. Results

3.1. Flow-Stress Characteristics

The flow stress-strain curves of the ultra-high-strength nickel-based corrosion-resistant alloy at different deformation temperatures and strain rates are shown in Figure 3 [20,21]. The flow stress is strongly dependent on deformation temperature and strain rate. In general, the curves first increase and then decrease: the peak stress decreases markedly with increasing temperature but increases with increasing strain rate. This behavior reflects the plastic-deformation mechanism of the alloy under different thermomechanical conditions. At the initial stage of deformation, pronounced work hardening occurs, and the flow stress increases rapidly with strain because the dislocation density rises quickly and dislocation interactions increase resistance to dislocation motion [22]. Owing to the high Nb content (~6 wt.%) and the presence of strong solid-solution-strengthening elements such as Cr and Mo, dislocation motion is significantly impeded, resulting in a pronounced work-hardening effect. The accumulated dislocations continuously increase the stored deformation energy, causing a rapid increase in flow stress. As deformation proceeds, dynamic recovery (DRV) and DRX gradually become active. Dislocation climb, cross-slip, and annihilation reduce the internal strain energy, thereby decreasing the work-hardening rate. When the rate of dynamic softening approaches that of work hardening, the flow stress reaches a peak. The occurrence of a distinct peak stress therefore indicates the initiation of extensive DRX during hot deformation. After the peak stress is reached, the flow stress gradually decreases or approaches a steady-state value, suggesting that a dynamic balance between dislocation generation and annihilation has been established. When dynamic softening balances work hardening, the flow stress approaches a steady state [23].

Temperature plays a dominant role in controlling flow behavior. At a constant strain rate, both the peak stress and the steady-state stress decrease significantly with increasing temperature. Higher temperatures promote atomic diffusion, grain-boundary migration, and dislocation climb, which accelerate the nucleation and growth of dynamically recrystallized grains. Consequently, deformation resistance is reduced and flow softening becomes more pronounced. In contrast, under low-temperature conditions, the mobility of dislocations and grain boundaries is limited, resulting in insufficient DRX and high flow stress. This observation is consistent with the instability region identified in the processing map, where cracking occurred under low-temperature deformation conditions.

The strain-rate dependence follows a different trend. Increasing the strain rate from 0.01 to 10 s^{−1} markedly raises the stress level because less time is available for dislocation annihilation and the formation of recrystallized grains. Rapid dislocation accumulation at high strain rates increases the stored energy and shifts the peak stress upward. This behavior also suggests that the hot workability of the alloy is controlled by the balance between the rate of dislocation generation and the kinetics of dynamic softening.

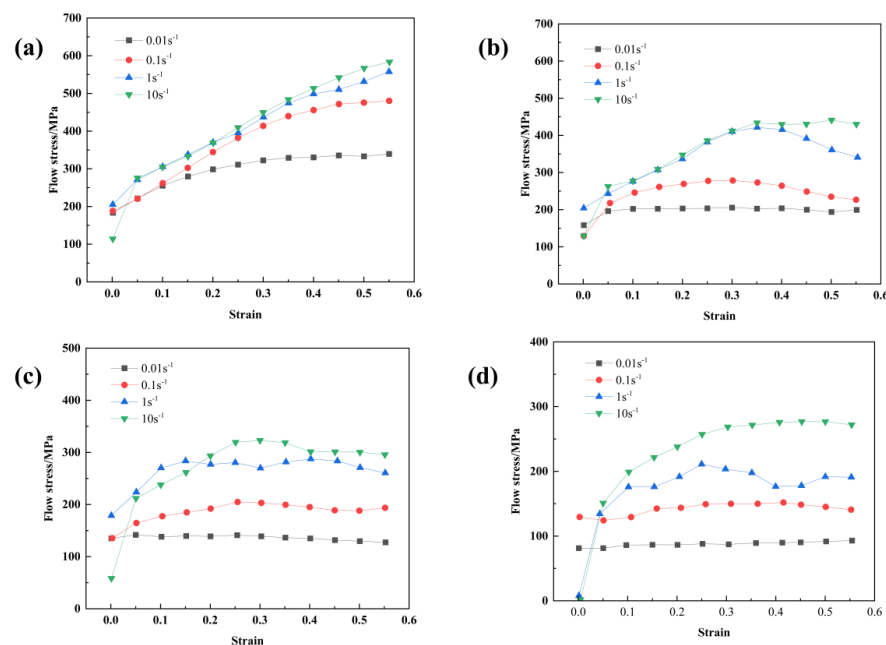


Figure 3. Flow stress–strain curves of the ultra-high-strength nickel-based corrosion-resistant alloy under isothermal compression at different deformation temperatures: (a) 1173 K; (b) 1223 K; (c) 1273 K; (d) 1323 K.

3.2. Construction of the Constitutive Equation

3.2.1. Arrhenius-Type Constitutive Equation

An Arrhenius-type hyperbolic-sine equation was selected to describe the relationship among flow stress, deformation temperature, and strain rate at a representative true strain of 0.5. This formulation was adopted because it can describe low-, medium-, and high-stress regimes within a single framework. The equations used for the calculation are listed below:

- (1) Low-stress region ($\alpha\sigma < 0.8$):

$$\dot{\epsilon} = A_1 \sigma^{n_1} \exp\left(-\frac{Q}{RT}\right) \quad (1)$$

- (2) High-stress region ($\alpha\sigma > 1.2$):

$$\dot{\varepsilon} = A_2 \exp(\beta\sigma) \exp\left(-\frac{Q}{RT}\right) \quad (2)$$

(3) Entire stress range (hyperbolic-sine equation):

$$\dot{\varepsilon} = A \left[\sinh(\alpha\sigma) \right]^n \exp\left(-\frac{Q}{RT}\right) \quad (3)$$

where ε is the strain rate (s^{-1}); σ is the flow stress (MPa); n and n_1 are stress exponents; A , A_1 , A_2 , β , and α are material constants, $\alpha = \beta/n_1$; R is the gas constant ($8.314 \text{ J}/(\text{mol K})$); T is the deformation temperature (K); and Q is the apparent deformation activation energy (kJ/mol).

After logarithmic transformation, Equations (1)–(3) can be rewritten as follows:

$$\ln \dot{\varepsilon} = n_1 \ln \sigma + \ln A_1 - \frac{Q}{RT} \quad (4)$$

$$\ln \dot{\varepsilon} = \beta\sigma + \ln A_2 - \frac{Q}{RT} \quad (5)$$

$$\ln \dot{\varepsilon} = \ln A + n \ln [\sinh(\alpha\sigma)] - \frac{Q}{RT} \quad (6)$$

The stresses measured under the selected test conditions were fitted according to Equations (4) and (5), as shown in Figure 4. The values of n_1 and β were obtained from the average slopes of the fitted lines in Figures 4a and 4b, respectively. The results were $n_1 = 6.6337$ and $\beta = 0.0202$; therefore, $\alpha = \beta/n_1 = 0.0030$.

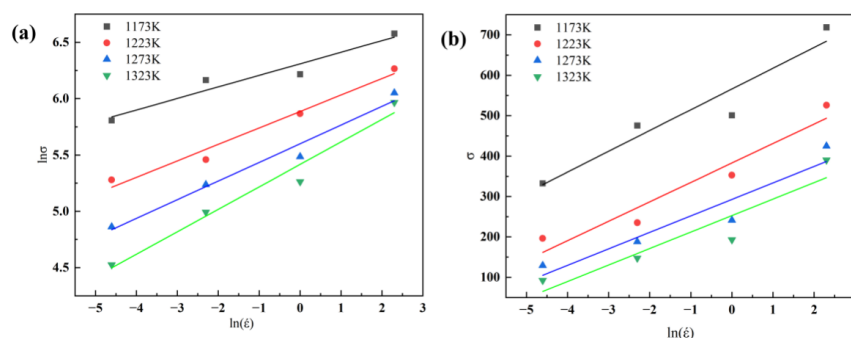


Figure 4. Fitted relationships for determining material constants: (a) Curve of $\ln(\sigma) - \ln(\dot{\varepsilon})$ and (b) Curve of $\sigma - \ln(\dot{\varepsilon})$.

The activation energy reflects the thermal barrier that must be overcome for dislocation motion and microstructural rearrangement during hot deformation. Because the present alloy contains a high Nb content and Nb/Ti-rich particles, this parameter is particularly useful for evaluating its resistance to high-temperature plastic flow.

Taking the logarithm of both sides of Equation (3), while assuming that the deformation activation energy is independent of temperature, gives:

$$\ln \dot{\varepsilon} = n \ln [\sinh(\alpha\sigma)] + \ln A - \frac{Q}{RT} \quad (7)$$

The determined α value was then combined with the experimental strain rates and flow stresses to construct the fitting plots shown in Figure 5.

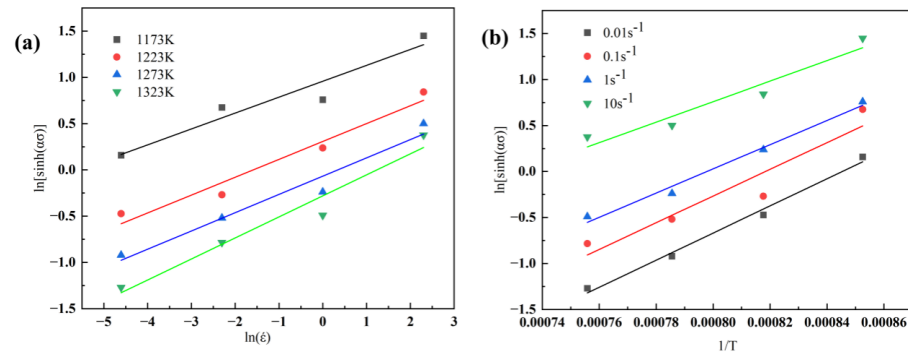


Figure 5. Fitted relationships for calculating hot-deformation parameters: (a) $\ln[\sinh(\alpha\sigma)] - \ln(\dot{\epsilon})$ fitted curve; (b) $\ln[\sinh(\alpha\sigma)] - \frac{1}{T}$ fitted curve.

From Figure 5a, n was determined from the reciprocal slopes, and Q/nR was calculated from the slopes in Figure 5b. The calculated values were $n = 4.8215$ and $Q = 540.33$ kJ/mol.

To combine the effects of temperature and strain rate into a single deformation parameter, the Zener–Hollomon parameter Z was introduced [24]:

$$Z = \dot{\epsilon} \exp\left(\frac{Q}{RT}\right) = A [\sinh(\alpha\sigma)]^n \quad (8)$$

Taking the natural logarithm of both sides of Equation (8) and rearranging it gives:

$$\ln Z = \ln \dot{\epsilon} + \frac{Q}{RT} = \ln A + n \ln [\sinh(\alpha\sigma)] \quad (9)$$

Linear fitting was used to obtain the relationship shown in Figure 6. The intercept was $\ln A = 50.9267$; therefore, $A = 1.3098 \times 10^{22}$.

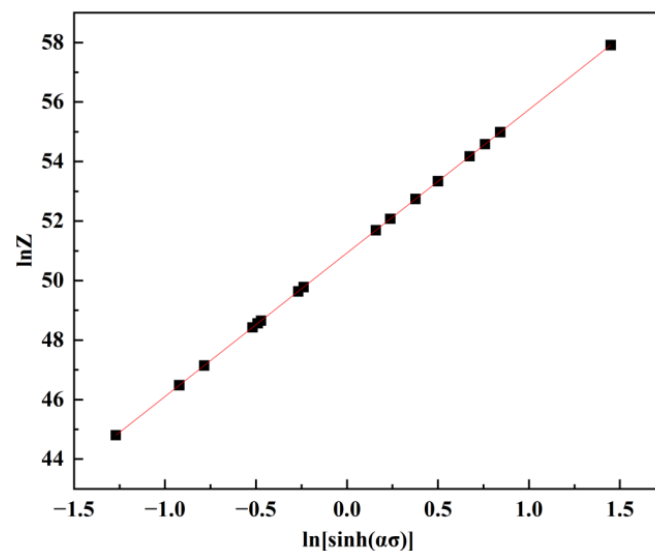


Figure 6. Variation of $\ln Z$ with $\ln[\sinh(\alpha\sigma)]$.

Based on the above calculations, the constitutive equation of the ultra-high-strength nickel-based corrosion-resistant alloy at deformation temperatures of 1173–1323 K and strain rates of 0.01 – 10 s^{-1} is given by:

$$\dot{\epsilon} = 1.3098 \times 10^{22} [\sinh(0.0030\sigma)]^{4.8215} \exp\left(-\frac{540330}{RT}\right) \quad (10)$$

The constitutive equation can also be expressed in terms of the Zener–Hollomon parameter Z :

$$\sigma = \frac{1}{0.0030} \ln \left\{ \left(\frac{Z}{1.3098 \times 10^{22}} \right)^{\frac{1}{4.8215}} + \left[\left(\frac{Z}{1.3098 \times 10^{22}} \right)^{\frac{2}{4.8215}} + 1 \right]^{\frac{1}{2}} \right\} \quad (11)$$

3.2.2. Strain-Compensated Constitutive Equation

Because the material constants vary with strain, the parameters α , n , Q , and $\ln A$ were recalculated at true strains from 0.1 to 0.5. Polynomial functions were then fitted to these strain-dependent values so that the constitutive model could account for the evolution of flow behavior during deformation.

$$\begin{aligned} \alpha &= B_0 + B_1 \varepsilon + B_2 \varepsilon^2 + B_3 \varepsilon^3 \\ n &= C_0 + C_1 \varepsilon + C_2 \varepsilon^2 + C_3 \varepsilon^3 \\ Q &= D_0 + D_1 \varepsilon + D_2 \varepsilon^2 + D_3 \varepsilon^3 \\ \ln A &= E_0 + E_1 \varepsilon + E_2 \varepsilon^2 + E_3 \varepsilon^3 \end{aligned} \quad (12)$$

The polynomial coefficients are summarized in Table 2. By substituting the strain-dependent parameters from Equation (12) into the Arrhenius equation, a strain-compensated constitutive model was obtained for the investigated alloy.

Table 2. Parameters of the strain-compensated equation at strains of 0.1–0.5.

Symbol	Value	Symbol	Value	Symbol	Value	Symbol	Value
B_0	4.81×10^{-3}	C_0	6.15	D_0	4.54×10^5	E_0	41.91
B_1	-8.19×10^{-3}	C_1	15.39	D_1	1.38×10^6	E_1	133.22
B_2	9.69×10^{-3}	C_2	−80.89	D_2	-4.56×10^6	E_2	−431.62
B_3	-1.16×10^{-3}	C_3	89.78	D_3	4.31×10^6	E_3	403.60

To verify the accuracy of the strain-compensated constitutive equation developed in this study, the relative errors between the experimental flow-stress measurements and theoretical values for the alloy at a strain of 0.5 under various deformation temperatures and strain rates are presented in Table 3. The predictive capability of the model was evaluated using the average absolute relative error (AARE), root mean square error (RMSE), and correlation coefficient (R). The obtained values were 4.77%, 19.49 MPa, and 0.9932 respectively, which were calculated using the experimental and predicted flow stresses at the representative true strain of $\varepsilon = 0.5$. The AARE of 4.77% falls within the excellent accuracy range below 5%. The RMSE of 19.49 MPa indicates a controllable overall deviation. The correlation coefficient of 0.9932 demonstrates an extremely strong linear correlation between the predicted and experimental stress values. The relative errors of individual data points under all test conditions range from 0.19% to 9.00%, with no obvious abnormal deviations. These results indicate excellent agreement between the predicted and experimental stresses, demonstrating that the developed model has high prediction accuracy and strong reliability for describing the hot-deformation behavior of the material. Thus, the proposed constitutive equation effectively describes the high-temperature plastic-deformation behavior of the ultra-high-strength nickel-based corrosion-resistant alloy and provides a theoretical foundation for its processing and forming applications.

Table 3. Comparison between experimental and predicted flow stresses at a true strain of 0.5.

Temperature/K	Strain Rate/s ^{−1}	Experimental Flow Stress/MPa	Predicted Flow Stresses/MPa	Relative Error/%
1173	0.01	302.61561	296.35	2.07%
	0.1	435.5952	411.16	5.61%
	1	500.91301	539.85	7.77%
	10	718.60432	674.02	6.20%

1223	0.01	196.35007	195.98	0.19%
	0.1	270.9879	289.47	6.82%
	1	390.03752	404.11	3.61%
	10	526.18254	532.09	1.12%
1273	0.01	129.34776	128.91	0.34%
	0.1	188.20728	197.01	4.68%
	1	269.11433	292.28	8.61%
	10	424.93527	407.07	4.20%
1323	0.01	92.4116	84.92	8.11%
	0.1	147.29163	134.03	9.00%
	1	193.07397	206.89	7.16%
	10	300.1939	302.73	0.84%

3.3. Construction of Processing Maps

Processing maps were constructed to correlate deformation parameters with hot workability and microstructural response. Within the DMM framework, the power supplied during deformation is divided into two parts: one is mainly consumed by plastic deformation and heat generation, and the other is associated with microstructural changes such as dynamic recovery, DRX, and possible grain growth [25–30]. The strain-rate sensitivity index m was calculated from the stress–strain-rate relationship as follows:

$$m = \frac{\partial \ln \sigma}{\partial \ln \dot{\epsilon}} \quad (13)$$

The power-dissipation efficiency η was then used to evaluate how effectively the input energy contributed to microstructural evolution. For an ideal linear dissipater, the dissipative co-content reaches its maximum value, and η is defined as:

$$\eta = \frac{2m}{m+1} \quad (14)$$

In this expression, m is the strain-rate sensitivity index. Although a high η value usually indicates active microstructural evolution, it does not necessarily correspond to a safe processing condition. Excessive grain growth or local flow localization may still occur; therefore, the power-dissipation map must be interpreted together with the instability map and the observed microstructures.

Flow instability was evaluated using the Prasad criterion. According to this criterion, deformation is considered unstable when the instability parameter ξ is negative:

$$\xi(\dot{\epsilon}) = \frac{\partial \left(\ln \frac{m}{m+1} \right)}{\partial (\ln \dot{\epsilon})} + m < 0 \quad (15)$$

Figure 7a shows the instability map of the ultra-high-strength nickel-based corrosion-resistant alloy at a true strain of 0.5 and strain rates of 0.01, 0.1, 1, and 10 s^{−1}. The power-dissipation efficiency η and instability parameter ξ were calculated for each condition using Equations (14) and (15). As shown in Figure 7a, the alloy approaches the instability boundary at low temperatures, especially below 1200 K at low strain rates, where localized defects may develop because of flow-stress concentration. Figure 7b shows severe hot cracking captured by an additional fracture-observation condition at 1123 K and 0.01 s^{−1}, which further verifies the high cracking tendency in low-temperature regions. In the high-temperature region near 1323 K, an extremely low instability-parameter region appears. Although the value is not negative, it represents a marginally unstable state. Figure 7c shows cracking at 1323 K and 0.1 s^{−1}. At such temperatures, atomic diffusion is strong; when the deformation time is relatively long, abnormal grain growth can occur. Coarse grains reduce grain-boundary cohesion and hinder further grain refinement by DRX, pro-

moting grain-boundary sliding or microvoid coalescence and ultimately causing flow instability [31]. The cracking at 1323 K results from abnormal grain growth, reduced grain-boundary cohesion, and local stress concentration; this failure mode is associated with grain coarsening. In contrast, the 1260–1300 K range exhibits positive ξ and relatively high η values, especially at 0.01–0.1 s^{−1}. This range is generally located within the safe hot-working domain, where the balance between work hardening and dynamic softening is favorable and the flow-stress distribution is relatively uniform.

Three standardized processing regimes are defined consistently throughout the manuscript as follows: classical DMM instability domain: $\xi < 0$, primarily corresponding to deformation conditions below 1200 K; Marginal processing degradation zone: $\xi > 0$ but accompanied by abnormal grain growth (typified by 1323 K/0.1 s^{−1}). This degradation cannot be captured by the Prasad DMM instability criterion. The associated cracking is governed by grain coarsening, reduced grain boundary cohesion, and localized stress concentration, rather than classical flow instability mechanisms; Safe hot-working window: 1260–1300 K and 0.01–0.1 s^{−1}, characterized by positive ξ and high power dissipation efficiency, stable deformation behavior, and low cracking susceptibility.

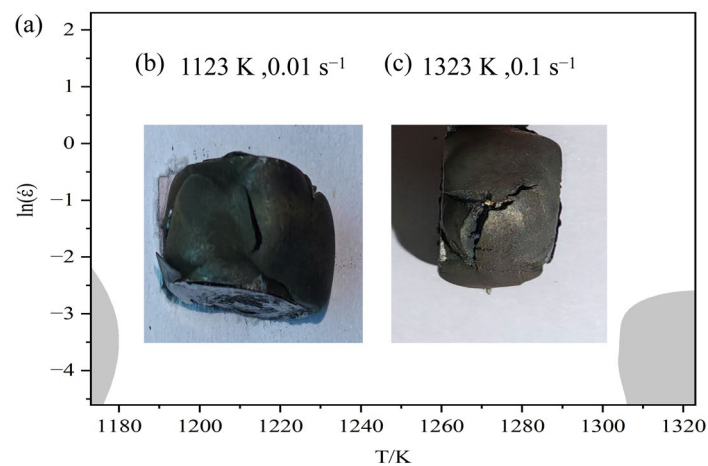


Figure 7. (a) Instability map of the ultra-high-strength nickel-based corrosion-resistant alloy; (b) cracking at 1123 K and 0.01 s^{−1}; (c) cracking induced by grain coarsening under marginal degradation condition: 1323 K, 0.1 s^{−1}.

Figure 8a shows the power-dissipation map of the ultra-high-strength nickel-based corrosion-resistant alloy at a true strain of 0.5. Figure 8b–e show the microstructures obtained under the conditions corresponding to points b–e in Figure 8a. The power-dissipation efficiency η varies significantly among the processing regions. A higher η value indicates that more input power is consumed by microstructural evolution rather than plastic dissipation, which favors DRX. The η value reaches 0.39–0.58 in the range of 1270–1320 K and 0.01–0.1 s^{−1}, with the highest value near 1323 K and 0.1 s^{−1}, indicating a potential high-energy-dissipation region. In contrast, η is generally lower than 0.27 in the range of 1180–1230 K, suggesting that dynamic recovery dominates and microstructural evolution is insufficient. Figure 8b, obtained at 1273 K and 0.01 s^{−1}, shows fine but nonuniform grains and localized banded deformation, indicating moderate plasticity and a relatively low cracking tendency. Figure 8c, obtained at 1273 K and 0.1 s^{−1}, shows fine and uniform grains with reduced banded deformation; both the power-dissipation efficiency and instability parameter are favorable, indicating stable deformation and a low cracking risk. This con-

dition represents a typical microstructure in the optimal processing zone. Figure 8d, obtained at 1323 K and 0.01 s^{-1} , shows coarsened grains, suggesting improved plasticity but a higher risk of local cracking because of grain coarsening. Figure 8e, obtained at 1323 K and 0.1 s^{-1} , shows the largest and most uniformly distributed grains with random orientations, corresponding to a fully dynamically recrystallized microstructure. Although the power-dissipation capacity is very high, excessive grain growth can induce local stress concentration and increase the likelihood of cracking. As shown in Figure 7c, cracking occurred under this condition; therefore, the 1323 K and 0.1 s^{-1} condition is not recommended from the standpoint of processing safety.

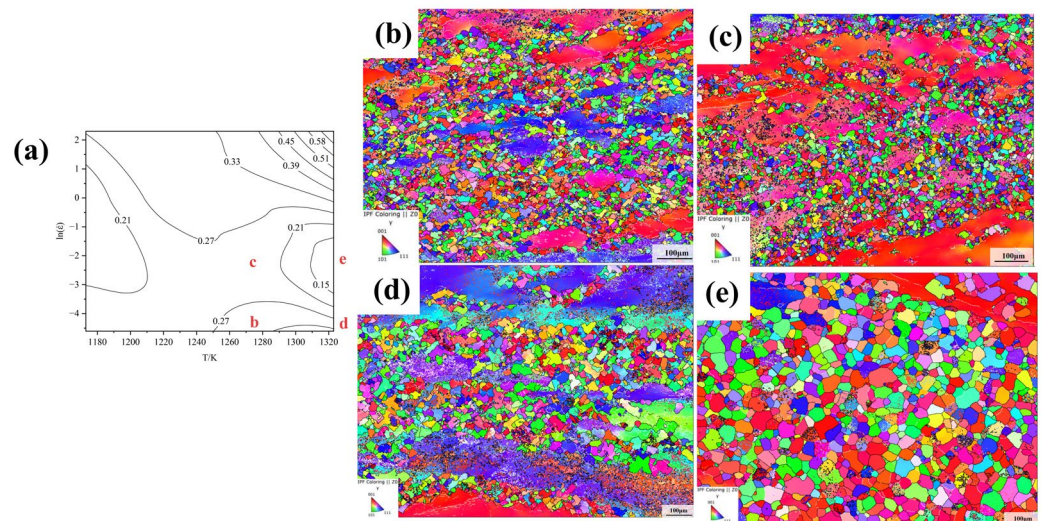


Figure 8. (a) Power-dissipation map of the ultra-high-strength nickel-based corrosion-resistant alloy at a true strain of 0.5. The lowercase red letters **b–e** in subfigure (a) indicate the representative deformation conditions selected for EBSD/IPF microstructural characterization. Microstructural IPF maps under four different conditions: (b) 1273 K, 0.01 s^{-1} ; (c) 1273 K, 0.1 s^{-1} ; (d) 1323 K, 0.01 s^{-1} ; (e) 1323 K, 0.1 s^{-1} .

As shown in Figure 9, the grain-size distributions of the alloy differ significantly under different hot-deformation conditions. At 1273 K and 0.01 s^{-1} (Figure 9a), the microstructure is mainly composed of fine recrystallized grains with an average grain size of $11.92\text{ }\mu\text{m}$. However, a small number of abnormally coarse grains are still observed, indicating that DRX has occurred but that microstructural homogeneity can be further improved. When the strain rate is increased to 0.1 s^{-1} (Figure 9b), the average grain size decreases to $10.91\text{ }\mu\text{m}$, and most grains are concentrated within a smaller size range with a more uniform distribution. This suggests that DRX proceeds more completely under this condition, producing a finer and more homogeneous microstructure. When the deformation temperature is increased to 1323 K, grain size increases noticeably. As shown in Figure 9c, the average grain size increases to $12.23\text{ }\mu\text{m}$, and some grains exhibit a tendency toward growth. At 1323 K and 0.1 s^{-1} (Figure 9d), the average grain size further increases to $19.11\text{ }\mu\text{m}$, accompanied by a significantly broader grain-size distribution. This indicates that the elevated temperature enhances grain-boundary mobility and promotes grain growth, leading to pronounced microstructural coarsening.

As shown in Figure 10, the degree of DRX of the alloy varies significantly under different hot-deformation conditions. At 1273 K and 0.01 s^{-1} (Figure 10a), the fractions of DRX grains and substructured grains are 43.9% and 51.5%, respectively, indicating that dynamic recrystallization has occurred to some extent, while microstructural evolution remains mainly dominated by dynamic recovery and substructure formation. When the

strain rate increases to 0.1 s^{-1} (Figure 10b), the fraction of DRX grains rises to 48.4%, whereas the fraction of deformed grains decreases to 2.51%. This suggests that a higher strain rate increases the stored deformation energy, promotes DRX nucleation, and facilitates the formation of a finer and more homogeneous microstructure. When the deformation temperature is increased to 1323 K, the microstructural evolution behavior changes markedly. At 1323 K and 0.01 s^{-1} (Figure 10c), the fraction of substructured grains increases to 62.32%, whereas the fraction of DRX grains decreases to 35.15%, indicating that the higher temperature and longer deformation time promote dynamic recovery and may be accompanied by grain growth. At 1323 K and 0.1 s^{-1} (Figure 10d), the fraction of DRX grains increases significantly to 84.12%, while deformed grains almost completely disappear, demonstrating that dynamic recrystallization is high under this condition. However, the grain-size distribution and processing-map analysis show that although the highest DRX fraction is achieved at 1323 K and 0.1 s^{-1} , obvious grain coarsening occurs and the risk of hot-processing instability increases. Therefore, the DRX fraction alone should not be used as the sole criterion for determining the optimal hot-working parameters.

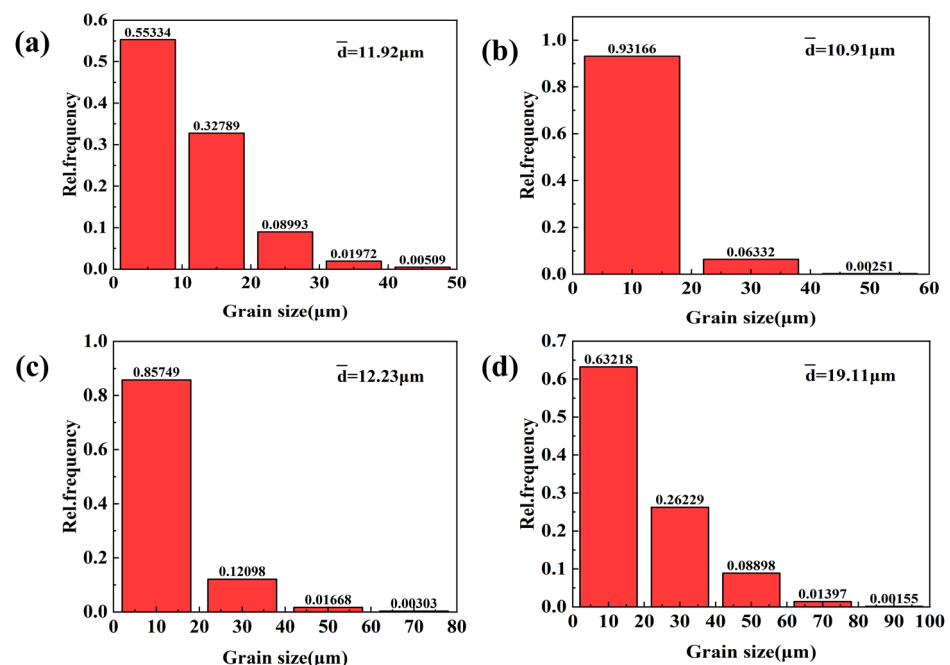


Figure 9. Grain-size distributions under four different conditions: (a) 1273 K, 0.01 s^{-1} ; (b) 1273 K, 0.1 s^{-1} ; (c) 1323 K, 0.01 s^{-1} ; (d) 1323 K, 0.1 s^{-1} .

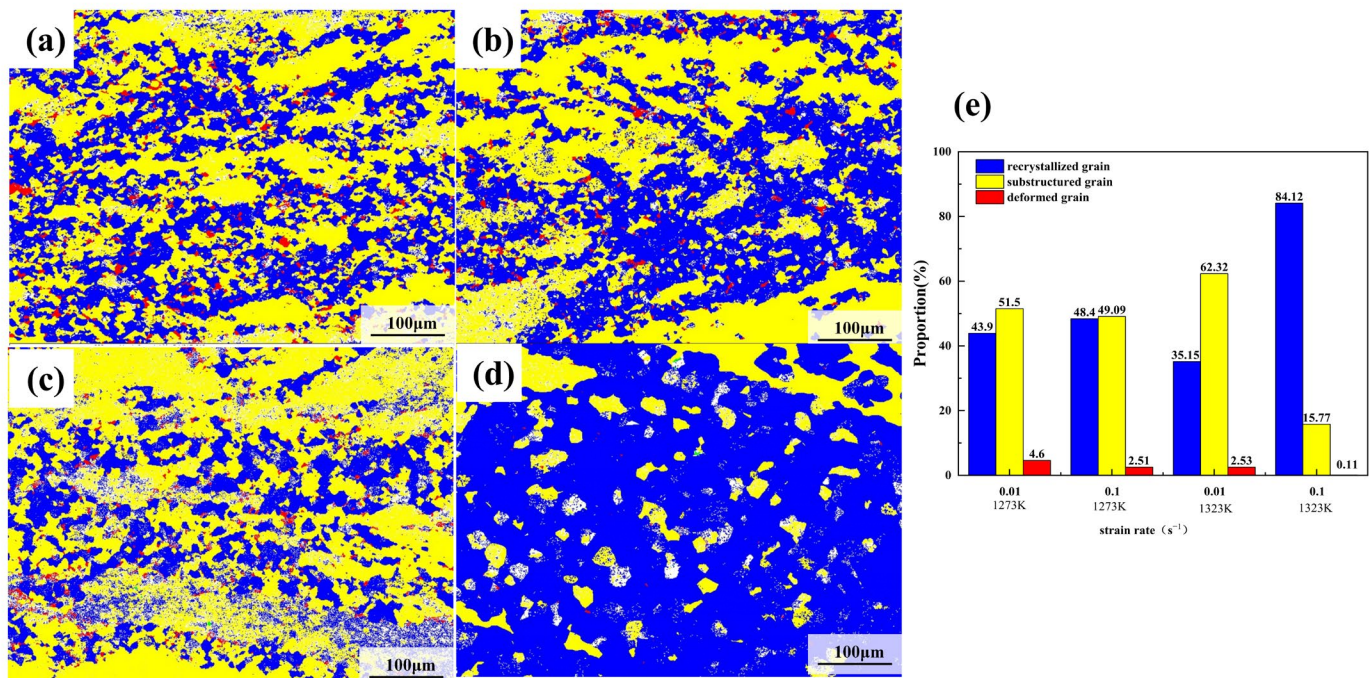


Figure 10. Distributions of dynamically recrystallized grains, subgrains, and deformed grains under four different conditions: (a) 1273 K, 0.01 s⁻¹; (b) 1273 K, 0.1 s⁻¹; (c) 1323 K, 0.01 s⁻¹; (d) 1323 K, 0.1 s⁻¹; (e) corresponding grain-area fractions.

Figure 11 presents the KAM maps and corresponding statistical distributions of the alloy under different hot-deformation conditions. KAM analysis is commonly used to characterize local misorientation, dislocation density, and residual deformation stored energy within a material. In general, a higher KAM value indicates greater intragranular orientation gradients, lattice distortion, and dislocation accumulation, whereas a lower KAM value suggests that DRV and/or DRX has proceeded more extensively, resulting in more effective release of stored deformation energy. At 1273 K and 0.01 s⁻¹ (Figure 11b), the average KAM value is 0.546°. A considerable number of regions with relatively high KAM values are still observed, indicating high dislocation density and residual strain within the grains and suggesting that dynamic recrystallization is not yet fully developed. When the strain rate is increased to 0.1 s⁻¹ (Figure 11d), the average KAM value decreases to 0.476°, implying that the higher strain rate enhances stored deformation energy, promotes DRX nucleation and dislocation annihilation, and consequently reduces local lattice distortion. At 1323 K and 0.01 s⁻¹ (Figure 11f), the average KAM value is 0.547°, comparable to that obtained at 1273 K and 0.01 s⁻¹. This indicates that a considerable amount of substructure and local misorientation is retained in the microstructure under low-strain-rate deformation. When the strain rate is further increased to 0.1 s⁻¹ (Figure 11h), the average KAM value decreases significantly to 0.180°, accompanied by a substantial increase in low-misorientation regions. This result demonstrates that dynamic recrystallization is high under this condition, leading to a marked reduction in dislocation density and residual strain.

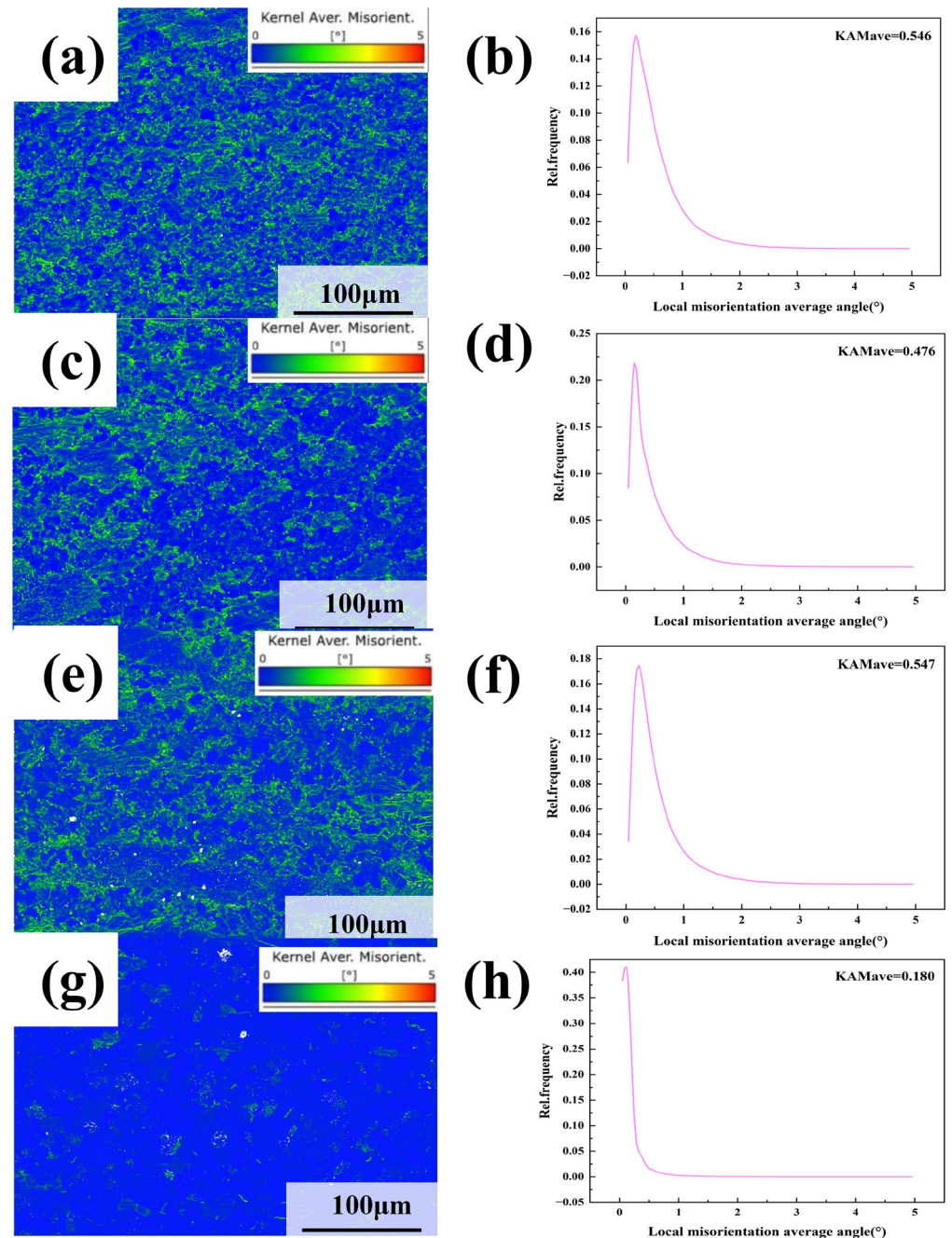


Figure 11. KAM maps under four different conditions: (a,b) 1273 K, 0.01 s⁻¹; (c,d) 1273 K, 0.1 s⁻¹; (e,f) 1323 K, 0.01 s⁻¹; (g,h) 1323 K, 0.1 s⁻¹.

The grain-boundary misorientation results in Figure 12 provide quantitative evidence for the evolution of subgrain boundaries and dynamically recrystallized boundaries under different hot-deformation conditions. Under all four conditions, the fraction of high-angle grain boundaries (HAGB) exceeds 90%, indicating that dynamic recrystallization occurred to varying degrees during hot compression. However, the variation in the low-angle grain boundary (LAGB) fraction further reveals the difference in residual substructures and the extent of boundary transformation. At 1273 K, increasing the strain rate from 0.01 to 0.1 s⁻¹ increases the HAGB fraction from 92.0% to 93.7%, while reducing the LAGB fraction from 8.0% to 6.3%. Correspondingly, the HAGB/LAGB ratio increases from 11.50 to 14.87, representing an increase of approximately 29.3%. This indicates that the higher strain rate enhances stored deformation energy and promotes the progressive transformation of low-angle subgrain boundaries into high-angle grain boundaries.

Therefore, the condition of 1273 K and 0.1 s^{-1} exhibits a more developed recrystallized boundary network than 1273 K and 0.01 s^{-1} . At 1323 K, the effect of strain rate is more pronounced. When the strain rate increases from 0.01 to 0.1 s^{-1} , the HAGB fraction increases from 90.1% to 94.1%, whereas the LAGB fraction decreases from 9.9% to 5.9%. The HAGB/LAGB ratio increases from 9.10 to 15.95, suggesting the most complete conversion of subgrain boundaries into recrystallized high-angle boundaries. This is consistent with the markedly reduced KAM value and the high DRX fraction observed under this condition. Nevertheless, the highest HAGB fraction should not be interpreted as the sole criterion for optimal hot workability, because the 1323 K and 0.1 s^{-1} condition also leads to significant grain coarsening and an increased risk of hot-processing instability.

Therefore, the grain-boundary statistics indicate that 1323 K and 0.1 s^{-1} produces the highest degree of recrystallized boundary formation, but this condition is accompanied by excessive grain growth. In contrast, 1273 K and 0.1 s^{-1} provides a better balance among high HAGB fraction, low LAGB fraction, fine grain size, reduced local misorientation, and stable deformation behavior. This confirms that 1273 K and 0.1 s^{-1} is the preferable hot-deformation condition for achieving a fine, homogeneous, and dynamically recrystallized microstructure in the investigated high-Nb nickel-based corrosion-resistant alloy.

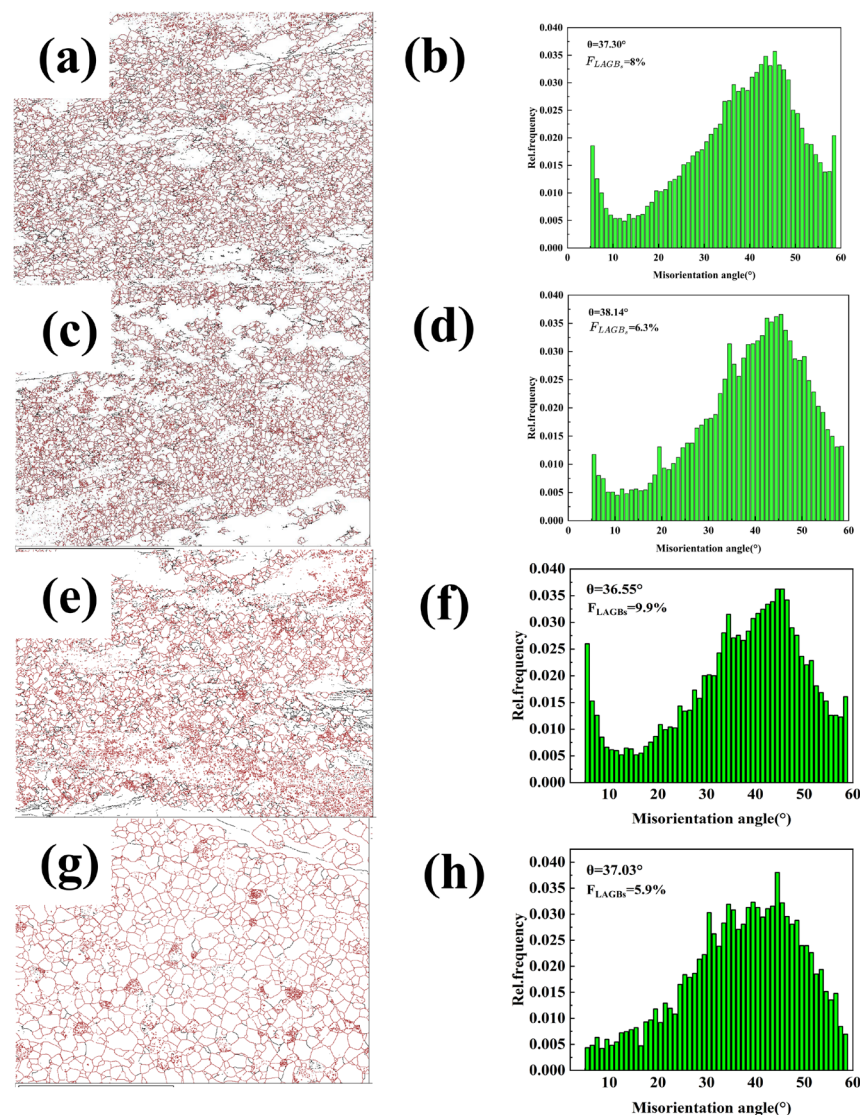


Figure 12. Local misorientation analysis and HAGB/LAGB fractions under four different conditions: (a,b) 1273 K, 0.01 s^{-1} , HAGB/LAGB = 92%/8% = 11.5%; (c,d) 1273 K, 0.1 s^{-1} , HAGB/LAGB = 93.7%/6.3%

= 14.87%; (e,f) 1323 K, 0.01 s⁻¹, HAGB/LAGB = 90.1%/9.9% = 9.1%; (g,h) 1323 K, 0.1 s⁻¹, HAGB/LAGB = 94.1%/5.9% = 15.95%.

Based on the above experimental results, the optimal hot-working window for the alloy is 1260–1300 K and 0.01–0.1 s⁻¹. This window is sufficiently far from the instability domain while maintaining high power-dissipation efficiency. It enables the formation of a uniform fine-grained microstructure through well-developed DRX while avoiding macroscopic cracking and local flow defects. This process window provides key parameter guidance for practical hot-working operations, such as cogging forging of as-cast microstructures, and can improve the microstructural uniformity and mechanical-property stability of components.

4. Conclusions

1. The flow-stress behavior of the alloy is governed by the competition between work hardening and dynamic softening. At the early stage of deformation, rapid dislocation multiplication leads to a sharp increase in flow stress. With increasing strain, dynamic recovery and dynamic recrystallization are activated, causing the flow stress to reach a peak and then decrease or approach a steady state. Flow stress decreases with increasing deformation temperature and increases with increasing strain rate.
2. A strain-compensated Arrhenius-type constitutive model was established for the alloy. The deformation activation energy was calculated to be 540.33 kJ/mol, indicating that the 205 ksi high-Nb alloy has strong resistance to hot deformation. The model shows high prediction accuracy, with AARE = 4.77%, RMSE = 19.49 MPa, and R = 0.9932, demonstrating that it can reliably describe the high-temperature flow behavior of this alloy.
3. Processing maps based on the dynamic materials model reveal that the optimal hot-working window is 1260–1300 K and 0.01–0.1 s⁻¹. In this range, the alloy exhibits stable deformation, high power-dissipation efficiency, and a low risk of flow instability. Low-temperature deformation below approximately 1200 K is prone to insufficient dynamic recrystallization and cracking, whereas excessive temperatures near 1323 K promote grain coarsening and hot-processing instability.
4. EBSD results further confirm the processing-map prediction. Deformation at 1273 K and 0.1 s⁻¹ produces the most favorable microstructure, characterized by fine and relatively homogeneous dynamically recrystallized grains. The average grain size is 10.91 µm, and the DRX fraction reaches 48.4%, indicating a good balance between recrystallization and grain refinement. In contrast, deformation at 1323 K leads to excessive grain growth, which increases local stress concentration and cracking risk.
5. The novelty of this work lies in establishing a complete relationship among flow-stress behavior, deformation activation energy, constitutive modeling, processing-map stability, and EBSD-based microstructural evolution for a 205 ksi high-Nb nickel-based corrosion-resistant alloy. These results provide a reliable theoretical basis and processing database for hot forging, billet breakdown, and microstructural control of ultra-high-strength nickel-based corrosion-resistant alloy components used in severe oil and gas environments.

Author Contributions: D.Z.: investigation, data curation, writing—original draft, formal analysis. G.M.: conceptualization, methodology, funding acquisition, writing—review and editing, supervision, project administration. W.J.: methodology, resources, validation, investigation. T.Z.: supervision, writing—review and editing, resources, conceptualization. Z.W.: supervision, methodology. All authors have read and agreed to the published version of the manuscript.

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