Hot Deformation Behavior and Processing Maps of 0.3C–15Cr–1Mo–0.5N High Nitrogen Martensitic Stainless Steel

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Abstract

Hot deformation behavior of 0.3C–15Cr–1Mo–0.5N high nitrogen martensitic stainless steel (HNMSS) was investigated in the temperature range of 1173–1473 K and at strain rates of 0.001–10 s^{−1} using a Gleeble 3500 thermal–mechanical simulator. The true stress–strain curves of the studied HNMSS were measured and corrected to eliminate the effect of friction on the flow stress. The relationship between the flow stress and Zener–Hollomon parameter for the studied HNMSS was analyzed in the Arrhenius hyperbolic sine constitutive model by the law of $Z = 3.76 \times 10^{15} \sinh(0.004979\sigma_p)^{7.5022}$. The processing maps at different strains of the studied HNMSS were plotted, and its flow instability regions in hot working were also confirmed in combination with the microstructure examination. Moreover, the optimal hot deformation parameters of the studied HNMSS could be suggested at $T = 1303\text{--}1423$ K and $\dot{\epsilon} = 5\text{--}10$ s^{−1} or $T = 1273\text{--}1473$ K and $\dot{\epsilon} = 0.005\text{--}0.04$ s^{−1}.

Keywords High nitrogen martensitic stainless steel (HNMSS) · Hot deformation · Processing map · Microstructure

1 Introduction

Alloying with nitrogen in martensitic stainless steels is becoming more and more attractive, in view of the fact that it has been considered to be an effective approach to reach the favorable combination of excellent mechanical properties and good corrosion resistance [1–4]. Nitrogen addition in martensitic stainless steels has been revealed to be beneficial to enhance the strength and improve the pitting resistance, but little affects the ductility and toughness [5–8]. The strength of 0.16C–5Cr–1Ni–Mo steel was significantly improved by 0.1 wt% nitrogen addition, attributing to solution strengthening effect of nitrogen atoms and ferrite suppression by nitrogen [6]. By hindering the austenite grain growth and refining the martensite lath, 0.022 wt%

nitrogen effectively increased the strength and hardness of 0.1–12Cr–Ni–Mo martensitic stainless steel [7]. A low content of nitrogen (less than 0.08 wt%) resulted in the more superior corrosion resistance of 0.3C–13Cr martensitic stainless steels, because of more protective passive films originating and Cr-rich M₂₃C₆ precipitates being reduced [8]. However, the thermomechanical properties of the martensitic stainless steel are severely deteriorated by the nitrogen atoms or nitride precipitates. The work hardening was raised due to the promotion of the planar glide of dislocations by adding 0.075 wt% nitrogen in the martensitic stainless steel of 0.2C–17Cr, which makes the hot deformation much more difficult [9]. As the interaction between nitrogen atoms or nitride precipitates and motional dislocations in the martensitic stainless steel of 0.10C–13Cr with 0.14 wt% nitrogen, the dynamic recovery (DRV) and dynamic recrystallization (DRX) were delayed to result in the increase of the flow stress at 1000 °C from 105 MPa to 144 MPa, which significantly increases the resistance to deformation [10]. The deformation activation energy of 0.30C–13Cr with 0.11 wt% nitrogen was enforced to 443.45 kJ mol^{−1} which is much higher than that of this steel without nitrogen [11, 12]. The DRX behavior of 0.30C–15Cr–1Mo alloyed with 0.45 wt% nitrogen was inhibited by M₂N precipitates at lower temperature and was promoted at higher temperature for the precipitate dissolution. Additionally, Carosi

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et al. [13] reported that cracks preferentially appeared in the 0.30C–15Cr–1Mo with 0.3 wt% nitrogen during wedge rolling at a temperature range of 850–1000 °C. It seems that the content of nitrogen along with deformation parameters plays an important role on the hot deformation behavior of martensitic stainless steels. But, martensitic stainless steels with high nitrogen content up to 0.5 wt% have been scantily noticed up to now. Especially, proper parameters for high nitrogen content martensitic stainless steels are essentially required to determine in detail to avoid damages in hot deformation.

In the present work, the hot deformation behavior of a novel high nitrogen martensitic stainless steel (HN MSS) with the nitrogen content over 0.5 wt% is systematically analyzed under the condition of the temperature in the range of 1173–1473 K and the strain rate of $0.001\text{--}10\text{ s}^{-1}$, using a Gleeble 3500 thermal–mechanical simulator and normalizing in several formulae, such as the constitutive model and DRX model. Microstructure evolution of this novel HN MSS has been clearly elucidated under different deformation conditions, and its processing maps have been experimentally determined. Moreover, optimal hot deformation parameters for this studied HN MSS have been finally proposed.

2 Materials and Methods

A 20-kg pilot ingot of the 0.3C–15Cr–1Mo–0.5N HN MSS with a diameter of 135 mm and a height of 275 mm was melted and cast in a vacuum pressurized induction furnace. The chemical composition of the ingot was examined as 0.30C–15.50Cr–0.55Mn–0.67Si–1.02Mo–0.55N-bal. Fe (wt%). After annealing at 1473 K, the ingot was forged into a round bar with a diameter of 45 mm. The forging bar was machined to cylindrical specimens with a diameter of 8 mm and a height of 12 mm for hot compression tests. The compressive deformation processes are shown in Fig. 1. The equilibrium phase diagram of the studied HN MSS was calculated by Thermo-Calc with TCFE8 database, and the equilibrium transformation temperatures of A_{e1} and A_{e3} are 1082 K and 1376 K, respectively. All the specimens were preheated to 1523 K at a heating rate of 10 K s^{-1} and held for 5 min to obtain a homogeneous austenite microstructure. The preheated specimens were then cooled to 1173, 1223, 1273, 1323, 1373, 1423 and 1473 K at a cooling rate of 10 K s^{-1} and held for 30 s to eliminate their thermal gradients. The specimens were compressed to a strain of 0.6 at a strain rate of 0.005, 0.05, 0.5, 1 and 10 s^{-1} , respectively, and then were subsequently quenched in water.

The aforementioned hot compression tests were carried out in a Gleeble 3500 thermal–mechanical simulator. The tantalum foils sprayed with graphite powders were used as the lubricant to minimize the friction between the specimens

Fig. 1 Schematic diagram of the hot compression test

and anvils of the thermal–mechanical simulator. After hot deformation, the specimens were sectioned from the center parallel to the compression axis to prepare samples for the microstructure characterization. The samples were mechanically ground with silica papers and polished in the diamond abrasive. And then, the cross sections of these samples were etched in a solution of ferric chloride hydrochloric acid and observed using an optical microscopy (OM, Axio Lab.A1).

3 Results and Discussion

3.1 Flow Stress Curves

The measured true stress–strain curves of the studied HN MSS deformed at different temperatures and strain rates are given with solid lines in Fig. 2. And the flow curves marked in dotted lines in Fig. 2 were corrected using the formula proposed by Ebrahimi and Najafizadeh [14] from the measured flow stress curves, considering the effect of the residue interfacial friction of the test samples. It seems that the DRX behavior of the studied HN MSS is obviously recognizable, when at higher deformation temperature with lower strain rates. Moreover, it is obvious that the stress increased with the decrease in deformation temperature and/or the increase in strain rate. For instance, at the strain rate of 0.005 s^{-1} with a temperature decreased from 1473 K to 1223 K, the peak stress σ_p of the studied HN MSS increased from 38.65 MPa to 234.28 MPa. With the further increase in the strain, the flow stress of the studied HN MSS turned to be steady due to the thermal softening and strain strengthening. The steady flow stress σ_{ss} of the studied HN MSS increased from 48.63 MPa to 245.37 MPa with the decrease in deformation temperature from 1473 K to

Fig. 2 True stress–strain curves of the studied HNMSS under different hot compressive deformation conditions: **a** 0.005 s^{-1} , **b** 0.05 s^{-1} , **c** 0.5 s^{-1} , **d** 1 s^{-1} , **e** 10 s^{-1}

1173 K at strain rate of 0.005 s^{-1} . In addition, at a given temperature of 1423 K with the strain rate increased from 0.005 to 10 s^{-1} , the σ_p increased from 46.43 to 163.78 MPa and the σ_{ss} increased from 33.64 to 149.5 MPa, respectively. In general, the restricted dislocation slip systems and limited recovery softening at low temperature resulted in a higher stress at low temperature to obtain further deformation [15]. With the increase in deformation temperature, the dynamic softening mechanism was promoted due to the improved grain boundaries mobility, vacancy diffusion, cross slip of screw dislocations and climb of edge dislocations [16], which is beneficial to the occurrence of DRX. In addition, the stress level decreased with the decrease in the strain rate, as the lower strain rates provided a longer time for the growth and nucleation of DRX grains.

3.2 Constitutive Analysis

The deformation behavior of studied HNMSS was analyzed by the Arrhenius equation to characterize the combination effects of the deformation temperature and strain rate [17].

$$Z = \dot{\epsilon} \exp \left(\frac{Q}{RT} \right), \quad (1)$$

$$\dot{\epsilon} = A \sinh (\alpha \sigma_p)^n \exp \left(-\frac{Q}{RT} \right), \quad (2)$$

where Z and Q are the Zener–Hollomon parameter and activation energy (kJ mol^{-1}) related to hot working, respectively. And $\dot{\epsilon}$, σ_p , R and T are corresponding to the strain rate (s^{-1}), peak stress (MPa), gas constant ($8.314 \text{ J mol}^{-1} \text{ K}^{-1}$) and absolute deformation temperature (K), respectively. Additionally, A and α are the material parameters dependent on the material and deformation conditions, respectively. And n is strain hardening exponent. After a logarithm of both sides of Eq. (2), the equation is written as:

$$\ln \dot{\epsilon} = n \ln [\sinh (\alpha \sigma_p)] + \left(\ln A - \frac{Q}{RT} \right). \quad (3)$$

In this work, α was firstly calculated using the equation of $\alpha = \beta / n_1$ [18]. n_1 and β at different deformation temperatures were obtained using the equations of $n_1 = [\partial \ln \dot{\epsilon} / \partial \ln \sigma_p]_T$ and $\beta = [\partial \ln \dot{\epsilon} / \partial \sigma_p]_T$ based on the experimental data, as shown in Figs. 3 and 4. Subsequently, the average values of n_1 and β were calculated as 10.6854 and 0.05321, respectively. And α was inferred to be 0.004979. Consequently, by plotting linear regression of $\ln \dot{\epsilon}$ against $\ln \sinh (\alpha \sigma_p)$, $\ln \sinh (\alpha \sigma_p)$ against $1/T$ and $\ln Z$ against $\ln \sinh (\alpha \sigma_p)$ (Figs. 5, 6, 7), the average values of n , Q and A were determined to be 7.5022, $481.74 \text{ kJ mol}^{-1}$ and 3.76×10^{15} , respectively. By filling in all the obtained material constants into Eqs. (1) and (2), the constitutive model of the studied HNMSS could be established as follows.

Fig. 3 Linear relationship of $\ln \dot{\epsilon}$ versus $\ln \sigma$

Fig. 6 Linear relationship of $\ln \sinh(\alpha \sigma_p)$ versus $1/T$

Fig. 4 Linear relationship of $\ln \dot{\epsilon}$ versus σ

Fig. 7 Regression analysis of the experimental data according to the hyperbolic sine constitutive equation as the function between flow stress and Z parameter

Fig. 5 Linear relationship of $\ln \dot{\epsilon}$ versus $\ln \sinh(\alpha \sigma)$

$$Z = \dot{\epsilon} \exp\left(\frac{481740}{8.314T}\right) = 3.76 \times 10^{15} \sinh(0.004979\sigma_p)^{7.5022}. \quad (4)$$

The obtained value of n in this study is higher than 410 martensitic stainless steel ($n=4.9$) [19] and 30Cr2Ni4MoV steel ($n=4.56$) [16] indicating that a worse workability of the studied HNMSS. On the other hand, in contrast to the reported values for hot deformation of conventional martensitic stainless steels such as $363.313 \text{ kJ mol}^{-1}$ for AISI 420 [20], $367.293 \text{ kJ mol}^{-1}$ for 403Nb alloy [21] and 413 kJ mol^{-1} for 13Cr% martensitic stainless steel [22], the activation energy for deformation obtained in this study was significantly higher. The reason for the higher Q might be attributed to the high content of nitrogen in the studied

HNMS. The one part of nitrogen in atoms could act as a strong strengthening solute; the other part of nitrogen in precipitates could conduct as obstacles to the dislocation movement. Both parts of the nitrogen take important roles to enhance Q , which results in the studied HNMS possessing higher hot deformation resistance.

3.3 DRX Analysis

3.3.1 Characteristic Parameters for DRX

The various parameters for DRX of the studied HNMS under different hot deformation conditions, including the critical strain (ε_c), the critical stress (σ_c), the peak strain (ε_p) and the peak stress (σ_p), were determined by the Avrami analysis [23, 24] and summarized in Table 1. Also, these parameters could be linearly regressed to the relationship with the natural logarithm of Zener–Hollomon parameter ($\ln Z$) [25], as described in Eq. (5) and shown in Fig. 8. It is

obviously seen that σ_c and ε_c corresponding to the DRX initiation decreased with an increase in deformation temperature and a decrease in strain rate, which indicates that DRX in the studied HNMS happened more easily at a higher deformation temperature and a lower strain rate.

$$\begin{cases} \sigma_p = 23.45751 \ln Z - 631.1699 \\ \sigma_c = 21.97981 \ln Z - 591.4062 \\ \varepsilon_p = 0.0229 \ln Z - 0.56066 \\ \varepsilon_c = 0.1102 \ln Z - 0.26951 \end{cases} \quad (5)$$

Moreover, the normalized critical stress ($\sigma_n = \sigma_c / \sigma_p$) and strain ($\varepsilon_n = \varepsilon_c / \varepsilon_p$) of the studied HNMS were also given in Table 1. It is found that the critical stress and strain are proportional to the peak stress and strain with the average values of σ_n and ε_n near to 0.94 and 0.48, respectively, which are independent on the deformation conditions. In general, the values of σ_n and ε_n can predict the occurrence of DRX. Therefore, the present investigation indicates that the DRX

Table 1 Values of the various parameters at different hot deformation conditions

Temperature (K)	$\dot{\varepsilon}$ (s ⁻¹)	ε_c	ε_p	σ_c (MPa)	σ_p (MPa)	$\sigma_n = \sigma_c / \sigma_p$	$\varepsilon_n = \varepsilon_c / \varepsilon_p$
1223	0.005	0.095	0.2	210	224	0.94	0.48
	0.05	0.16	0.315	255	267	0.96	0.51
	0.5	0.21	0.46	302	322	0.94	0.46
	1	0.18	0.365	305	327	0.93	0.49
	10	0.17	0.35	335	365	0.92	0.49
1273	0.005	0.085	0.18	185	196	0.94	0.47
	0.05	0.16	0.33	230	246	0.93	0.48
	0.5	0.19	0.42	255	272	0.94	0.45
	1	0.18	0.35	262	279	0.94	0.51
	10	0.18	0.42	310	330	0.94	0.43
1323	0.005	0.075	0.16	115	123	0.93	0.47
	0.05	0.125	0.26	173	185	0.94	0.48
	0.5	0.17	0.40	219	234	0.94	0.43
	1	0.14	0.285	235	247	0.95	0.49
	10	0.14	0.36	253	272	0.93	0.39
1373	0.005	0.073	0.15	89.5	96	0.93	0.49
	0.05	0.12	0.251	125	134	0.93	0.48
	0.5	0.17	0.375	157	170	0.92	0.45
	1	0.15	0.335	173	185	0.94	0.45
	10	0.15	0.35	227	244	0.93	0.43
1423	0.005	0.065	0.12	61	66	0.92	0.54
	0.05	0.115	0.24	104	110	0.95	0.48
	0.5	0.136	0.27	127	136	0.93	0.50
	1	0.15	0.354	146	154	0.95	0.42
	10	0.185	0.381	182	192	0.95	0.49
1473	0.005	0.055	0.108	41	44	0.93	0.51
	0.05	0.08	0.165	78	83	0.94	0.48
	0.5	0.101	0.21	115	122	0.94	0.48
	1	0.16	0.33	110	117	0.94	0.48
	10	0.16	0.37	154	164	0.94	0.43

Fig. 8 Relationship between characteristic parameters and $\ln Z$: **a** σ_p versus $\ln Z$, **b** σ_c versus $\ln Z$, **c** ε_p versus $\ln Z$, **d** ε_c versus $\ln Z$

starts when the normalized strain reaches to the value of 0.48. Similar proportional values have been reported for stainless steels [26, 27].

Additionally, the DRX fraction in volume X_{drx} of the studied HNMSS has been inferred in Avrami equation [27, 28]. And also, the relationship between X_{drx} and the strain could be linearly regressed, as shown in Fig. 9

$$\begin{cases} X_{\text{drx}} = 0 & (\varepsilon < \varepsilon_c) \\ X_{\text{drx}} = 1 - \exp \left[-0.20316 \left(\frac{\varepsilon - \varepsilon_c}{\varepsilon_p} \right)^{5.1999} \right] & (\varepsilon \geq \varepsilon_c) \end{cases} \quad (6)$$

3.3.2 Microstructural Evolution

Figure 10 shows the optical microstructure and XRD spectrum of the studied HNMSS annealed at 1523 K for 300 s and quenched in water. It seems that the initial microstructure of the studied HNMSS before the hot compression test

Fig. 9 Relationship between $\ln [-\ln (1 - X_{\text{drx}})]$ versus $\ln (\varepsilon - \varepsilon_c) / \varepsilon_p$

Fig. 10 **a** Optical micrograph and **b** X-ray diffraction pattern of the sample after preheating at 1523 K for 300 s and subsequently cooling in water

performance is the complete austenite phase with coarse equiaxed grains. Subsequently, the microstructures of the studied HNMS deformed at a small strain rate of 0.005 s^{-1} under the different temperatures are shown in Fig. 11. It can be seen that the DRX did not happen when the studied HNMS was deformed at 1223 K. The original austenite

grains were elongated, and a great many precipitates in lamellar zones which was identified as nitrides by EDS, alongside the original grain boundaries appeared, as shown in Fig. 11a. As the deformation temperature rose to 1273 K, a few precipitates were still observed alongside the elongated original austenite grain boundaries, and a small amount of

Fig. 11 Microstructures of HNMS deformed with a strain rate of 0.005 s^{-1} at temperature of 1223 K (**a–c**), 1273 K (**d, e**)

DRX zones brought out with several fine grains on the original austenite grain boundaries, as seen in Fig. 11d, e. When the deformation temperature was up to 1373 K and 1423 K, few precipitates and much finer equiaxed austenite grains were observed, as shown in Fig. 12. It is indicated that the DRX of the studied HNMSS occurred and completed to refine the grains, which is consistent with the conclusion above-mentioned that the critical strain of ϵ_c corresponding to the DRX initiation decreased with an increase in deformation temperature.

Figure 13 shows the optical microstructures of the studied HNMSS deformed at different strain rates under the temperature of 1423 K. Fine equiaxed grains corresponding to the complete DRX microstructure were observed when the studied HNMSS was deformed at the strain rate from 0.05 s^{-1} , 0.05 s^{-1} and 10 s^{-1} ; while smaller the strain rate was, coarser the grains turned to be. It is suggested that the critical strain to initiate the DRX at lower strain rate

is smaller than that one at higher strain rate, as mentioned above. The DRX of the studied HNMSS happened and completed earlier at lower strain rate, and the equiaxed grains of the DRX microstructure grew up during the subsequent high temperature duration.

3.4 Processing Maps

The processing maps of the studied HNMSS deformed at true strains of 0.1, 0.3, 0.6 and 0.8 under the different temperatures and strain rate conditions were determined by the dynamic material model (DMM) [29–31], as shown in Fig. 14. The power dissipation efficiency (η) and the flow instability of the studied HNMSS were delineated by the contour lines with different values and the hatched regions, respectively. The regions marked by A and B in these maps represent the two peak values of η for the tendency to DRX occurrence. When the studied

Fig. 12 Microstructures of HNMSS deformed with a strain rate of 0.005 s^{-1} at the temperature of 1373 K (a), 1423 K (b)

Fig. 13 Microstructures of the HNMSS deformed at 1423 K with a strain rate of a 0.05 s^{-1} , b 0.5 s^{-1} , c 10 s^{-1}

Fig. 14 Processing maps of the HNMSS at different true strains of 0.1 (a), 0.3 (b), 0.6 (c), 0.8 (d). A and B areas marked in each figure represent the flow stability regions with peak η values

HNMS was compressed to the strain of 0.1, these two regions shown in Fig. 13a happened under the deformation conditions of $T = 1323\text{--}1473\text{ K}$ and $\dot{\epsilon} = 5\text{--}10\text{ s}^{-1}$ as well as $T = 1423\text{--}1473\text{ K}$ and $\dot{\epsilon} = 0.001\text{ s}^{-1}$, respectively. Meanwhile, three flow instability regions with potential microstructure defects occurred. With the increase in strain, the peak values of η gradually increased, and its corresponding regions shifted to the lower temperatures. And the flow instability regions moved to the conditions at higher strain rates under lower temperatures. It is believed that the DRX of the studied HNMSS happened more easily at lower temperatures when deformed to a certain extent.

Figure 15 shows the optical microstructures of the studied HNMSS deformed at 0.5 s^{-1} under different temperatures in the flow instability regions. The inhomogeneous deformation microstructure of elongated original grains around with some fine necklace DRX grains occurred in all those regions, which is considered to be undesirable to mechanical properties. Therefore, the corresponding deformation parameters in these flow instability regions

for the studied HNMSS are suggested to avoid in the hot-working process.

The optical microstructures of the studied HNMSS in the region A at a strain rate of 10 s^{-1} with the lower η peak values are shown in Fig. 16. At 1223 K , small fraction DRX happened to result in the inhomogeneous microstructure (Fig. 16a). With the increase in the deformation temperature from 1323 K to 1473 K , complete DRX occurred and equiaxed grains appeared (Fig. 16b–d), while the optical microstructures of the studied HNMSS in the region B with the higher η peak values are shown in Fig. 17. The DRX completed and equiaxed austenite grains arose. And the grain size after the complete DRX of the studied HNMSS gradually increased with the increase in deformation temperature. It could be confirmed that much more homogeneous microstructure has been achieved by the complete DRX with a higher η .

As stated above, the optimal parameters for practical hot-working processes of the studied HNMSS are suggested to be $T = 1303\text{--}1423\text{ K}$ and $\dot{\epsilon} = 5\text{--}10\text{ s}^{-1}$ or

Fig. 15 Optical micrographs of the studied HNMSS deformed in the flow instability regions with the true strain of 0.6 at a strain rate of 0.5 s^{-1} different deformation temperatures of 1173 K (a), 1223 K (b), 1273 K (c), 1323 K (d)

$T = 1323\text{--}1473 \text{ K}$ and $\dot{\epsilon} = 0.005\text{--}0.04 \text{ s}^{-1}$, which is more precise than that reported earlier [13].

4 Conclusions

The hot deformation behavior of a novel HNMSS was systematically investigated. And the processing maps of this studied HNMSS were experimentally determined, accompanied with the microstructure analysis. The main conclusions are summarized as follows:

1. The hyperbolic sine constitutive equation of the studied HNMSS has been proposed as

$Z = 3.76 \times 10^{15} \sinh(0.004979\sigma_p)^{7.5022}$. And the activation energy is inferred as $481.74 \text{ kJ mol}^{-1}$.

2. The critical parameters for the initiation of DRX of the studied HNMSS have been determined as $\sigma_c/\sigma_p = 0.94$ and $\epsilon_c/\epsilon_p = 0.48$. And the DRX volume fraction model of this steel has been derived as follows:

$$\begin{cases} X_{\text{drx}} = 0 & (\epsilon < \epsilon_c) \\ X_{\text{drx}} = 1 - \exp \left[-0.70316 \left(\frac{\epsilon - \epsilon_c}{\epsilon_p} \right)^{5.1999} \right] & (\epsilon \geq \epsilon_c) \end{cases}$$

3. The optimal hot-working parameters of the studied HNMSS were suggested to be $T = 1303\text{--}1423 \text{ K}$ and $\dot{\epsilon} = 5\text{--}10 \text{ s}^{-1}$ or $T = 1323\text{--}1473 \text{ K}$ and $\dot{\epsilon} = 0.005\text{--}0.04 \text{ s}^{-1}$.

Fig. 16 Microstructures of the studied HNMSS deformed in the stability region A with the true strain of 0.8 at the strain rate of 10 s^{-1} and different forming temperatures of 1223 K (a), 1323 K (b), 1423 K (c), 1473 K (d)

Fig. 17 Microstructures of the studied HNMSS deformed in the stability region B with the true strain of 0.6 at the strain rate of 0.005 s^{-1} and different forming temperatures of 1323 K (a), 1373 K (b), 1423 K (c), 1473 K (d)

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