Hot Deformation Behavior and Processing Maps of a High Al-low Si Transformation-Induced Plasticity Steel: Microstructural Evolution and Flow Stress Behavior

H. Q. Huang¹ · H. S. Di¹ · N. Yan¹ · J. C. Zhang¹ · Y. G. Deng¹ · R. D. K. Misra² · J. P. Li¹

Received: 20 July 2017 / Revised: 18 September 2017 / Published online: 14 November 2017
© The Chinese Society for Metals and Springer-Verlag GmbH Germany 2017

Abstract Hot deformation behavior of a high Al-low Si transformation-induced plasticity (TRIP) steel was studied by an MMS-300 thermo-simulation machine at the temperature range of 1050–1200 °C and strain rate range of 0.01–10 s⁻¹. The constitutive equations of the TRIP steel were established at high temperature by fitting the strain factor with a sixth-order polynomial. The instability during hot rolling was discussed using processing maps. The results reveal that two types of flow stress curves (dynamic recrystallization and dynamic recovery) were observed during the hot compression of the high Al-low Si TRIP steel. Flow stress decreased with increasing deformation temperature and decreasing strain rate. The predicted flow stress of experimental TRIP steel is in agreement with the experimental values with an average absolute relative error of 4.49% and a coefficient of determination of 0.9952. According to the obtained processing maps, the TRIP steel exhibits a better workability at strain rate of 0.1 s⁻¹ and deformation temperature of 1200 °C as compared to other deformation conditions.

KEY WORDS: TRIP steel; Hot compression; Constitutive equation; Processing maps

1 Introduction

The demand is increasing for decreasing vehicle weight and ensuring passenger safety. TRIP steels are known as low cost, high strength, high ductility and excellent energy absorption materials that can be used for automotive applications, particularly for thin cross sections [1–4].

Two-stage heat treatment is typical in cold-rolled TRIP steel which employs an isothermal bainitic transformation (IBT) after intercritical annealing (IA). There is a feasible approach to combine continuous galvanizing thermal cycles with the IBT [5, 6]. Mn and Si are harmful selective oxidation alloying elements in galvanizing process [2, 7, 8]. One way to improve the reactive wetting property of galvanization is partially or completely replacing Si by Al in steels [9–12].

The flow stress evolution in steels and alloys was studied to optimize process parameters during hot working. Furthermore, constitutive equations with an Arrhenius term are widely used to accurately accomplish numerical simulation [13]. Constitutive modeling of alloys [14–20] has been accomplished recently to predict their hot deformation behavior. Based on the principles of dynamic material modeling [21], the processing maps have been successfully developed [22–25]. Researches show that the variety of thermal–mechanical condition in hot deformation process causes complex evolutions of

Available online at <http://link.springer.com/journal/40195>

✉ H. S. Di
hongshuangdi_ral@126.com

H. Q. Huang
huiqiang1220@163.com

¹ State Key Laboratory of Rolling and Automation, Northeastern University, Shenyang 110819, China

² Laboratory for Excellence in Advanced Steel Research, Department of Metallurgical, Materials and Biomedical Engineering, University of Texas at El Paso, El Paso, TX 79968, USA

dynamic recrystallization (DRX) content, dislocation density and grain size [26]. The DRX and instability domain are affected by pre-treatment of a Ni-based superalloy [27]. Recently, a developed DRX kinetics model was used to discuss the DRX behavior considering influences of precipitation phase [28]. With regards to low-carbon microalloyed TRIP steels, Wei *et al.* [29] analyzed flow stress evolution. Unfortunately, the strain effect was not considered while evaluating material constants, such as activation energy. Although the hot deformation behavior of TRIP steel has been focused in recent years, processing maps have not been developed.

In this study, the hot deformation behavior of a high Al-low Si TRIP steel was studied over a wide range of strain rate and temperature. Based on hot compression tests, the general constitutive equations were established and activation energies were analyzed while considering strain. To optimize processing parameters, the processing maps were obtained according to the dynamic material model. The dynamic recrystallization behavior of TRIP steel was also studied together with microstructure and flow stress.

2 Experimental

The experimental TRIP steel was melted in a vacuum induction furnace and then hot forged into ingots of 80 mm × 80 mm cross section. Chemical composition of the experimental steel is given in Table 1. Specimens for hot deformation experiment were machined into cylinders of diameter 8 mm and length 15 mm. Using an MMS-300 thermo-mechanical simulator, hot compression tests were performed at strain rates of 0.01, 0.1, 1 and 10 s⁻¹ and in the temperature range of 1050 to 1200 °C at an interval of 50 °C. The procedure for hot compression tests is shown in Fig. 1. Specimens were kept at 1200 °C for 180 s in order to homogenize the microstructure and then were homogenized at deformation temperature for 120 s before compression. The microstructure at the deformation temperature was used to analyze the hot deformation behavior. So the specimens were compressed to a true strain of 0.7 immediately followed by water quenching.

Table 1 Chemical composition of experimental TRIP steel (mass percent, %)

C	Si	Al	Mn	Ti	P	S	Fe
0.19	0.46	1.69	1.67	0.018	0.004	0.0029	Bal.

Fig. 1 Hot deformation procedure at different compression conditions. Deformation temperatures: 1050–1200 °C with 50 °C interval. Strain rates: 0.01, 0.1, 1, 10 s⁻¹. Deformation true strain: 0.7

The deformed specimens were sectioned parallel to the compression axis. Standard mechanical grinding and polishing techniques were used to prepare metallographic specimens. The polished specimens were etched with supersaturated picric acid at 70–80 °C to reveal the austenite grain boundaries of the compressed specimens. The microstructure was studied using standard metallographic technique with a Zeiss optical microscope.

3 Results and Discussion

3.1 Flow Stress Behavior

Representative flow stress plots presented in Fig. 2 were obtained at different strain rates and different deformation temperatures. It should be noted that deformation temperature and strain rate have a significant effect on flow stress. Flow stress decreases with increasing deformation temperature and decreasing strain rate. The underlying reason for the deformation temperature effect is that resistance to dislocation movement is lower and point defects such as vacancy and interstitial atom are more active with higher deformation temperature. As solute strengthening parameters in low-carbon steel, the addition of silicon had a more dominant effect than carbon on the flow behavior [30], and with the increase in Al the flow stress would be higher [18]. As strain rate increases, dislocation density increases and dislocations are entangled; therefore, there is an increase in the number of dislocations, substructures and flow stress increases along with increased work hardening [14, 26, 31]. As shown in Fig. 2b, at low strain rates true stress–true strain curves appeared to be dynamic recrystallization type. At low strain, stress increases rapidly with increasing strain

Fig. 2 Flow stress plots of experimental specimens

because of work hardening. Then, dynamic recrystallization takes place. The softening effect induced by dynamic recrystallization and dynamic recovery cannot offset the work-hardening effect, so stress increases continuously and reaches a peak. Beyond the peak value of stress, softening effect plays a leading role and stress decreases gradually. When dynamic recrystallization is complete, the softening effect and the work-hardening effect get balanced and stress attains a stable value. At high strain rates, the true stress–true strain curves are of the dynamic recovery type [28]. Because of interaction between dynamic softening and work hardening, the stress becomes stable and does not decrease when the strain increases beyond the peak value.

3.2 Constitutive Modeling of Flow Stress

In order to study hot deformation behavior of high Al-low Si TRIP steel, material constants in the constitutive equation should be determined using true stress–true strain data obtained from compression tests. Zener–Hollomon parameter, Z (Eq. (1)) [32], was widely applied to analyze the effects of deformation temperature and strain rate on flow stress under isothermal deformation condition. The correlation among strain rate, deformation temperature and flow stress can be generalized into three types of Arrhenius equations [13]. They are applicable at low stresses (Eq. (2)), high stresses (Eq. (3)) and all stress levels (Eq. (4)), respectively.

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

$$\dot{\epsilon} = A_1 \sigma^{n_1} \exp\left(-\frac{Q}{RT}\right), \quad \alpha\sigma < 0.8, \quad (2)$$

$$\dot{\epsilon} = A_2 \exp(\beta\sigma) \exp\left(-\frac{Q}{RT}\right), \quad \alpha\sigma > 1.2, \quad (3)$$

$$\dot{\epsilon} = A [\sinh(\alpha\sigma)]^n \exp\left(-\frac{Q}{RT}\right), \quad \text{for all } \sigma. \quad (4)$$

In the above equations, A_1 (s^{-1}), A_2 (s^{-1}), A (s^{-1}), n_1 , β (MPa^{-1}), α (MPa^{-1}) and n are material constants where $\alpha = \beta/n_1$, σ is the flow stress (MPa), $\dot{\epsilon}$ is the equivalent strain rates (s^{-1}), Q is the activation energy of hot deformation (J mol^{-1}), R is the universal gas constant ($8.314 \text{ J mol}^{-1} \text{ K}^{-1}$), and T is the absolute temperature (K).

Taken as an example, flow stresses at true strain of 0.1 were used to estimate the material constants. After taking the natural logarithms on both sides, Eqs. (2) and (3) are changed into

$$\ln \sigma = \frac{1}{n_1} \ln \dot{\epsilon} + \frac{1}{n_1} \left(\frac{Q}{RT} - \ln A_1 \right), \quad (5)$$

$$\sigma = \frac{1}{\beta} \ln \dot{\epsilon} + \frac{1}{\beta} \left(\frac{Q}{RT} - \ln A_2 \right). \quad (6)$$

Based on Eqs. (5) and (6), the value of n_1 and β can be obtained using curves of $\ln \sigma - \ln \dot{\epsilon}$ and $\sigma - \ln \dot{\epsilon}$ at different temperatures, respectively. As shown in Fig. 3a, b, it is evident that the line of each temperature is almost parallel. Therefore, the value of $1/n_1$ and $1/\beta$ at different temperatures can be obtained. The mean values of n_1 and β can be computed as 7.0081 and 0.0868 MPa^{-1} , respectively, and the material constant $\alpha = \beta/n_1 = 0.01234 \text{ MPa}^{-1}$.

At all flow stress levels, the following equation can be derived from Eq. (4):

$$\ln[\sinh(\alpha\sigma)] = \frac{1}{n} \ln \dot{\epsilon} - \frac{1}{n} \ln A + \frac{Q}{nRT}. \quad (7)$$

Based on Eq. (7) and the relationship curves of $\ln[\sinh(\alpha\sigma)] - \ln \dot{\epsilon}$ (Fig. 3c), the value of n was estimated as 5.1845 . As shown in Fig. 3d, the value of Q can be obtained by fitting the slope of plot of $\ln[\sinh(\alpha\sigma)] - 1/T$ and the value of Q was $367.6 \text{ kJ mol}^{-1}$. Then, material constant A can be obtained from the Y intercept via fitting of the relationship of $\ln[\sinh(\alpha\sigma)] - 1/T$ and the value of A was determined as $6.534 \times 10^{12} \text{ s}^{-1}$.

According to this research, the influence of strain on high-temperature flow behavior of high Al-low Si TRIP steel is not negligible. In order to accurately predict the flow stress, the values of material constants were evaluated at different strains (in the range of 0.05–0.60 at an interval

of 0.05). As shown in Fig. 4, the relationship between each material constant (α , n , Q and $\ln A$) and true strain is complicated and nonlinear. In general, the curves of material constants at different strains are fitted by polynomial equations. Different orders of the polynomial can be chosen depending on characters of steel or alloy. Fourth- [33, 34], fifth- [35], sixth [14, 23]-order polynomials were used for different metal materials in researches. In consideration of complex calculations of this metal, sixth-order polynomial was used to predict the data. Higher-order polynomial was observed to cause distortion for curve fitting [36]. As shown in Fig. 4, the relationship between material constants and true strain is obtained from Eq. (8) as follows:

$$\begin{cases} \alpha = 0.02135 - 0.1983\epsilon + 1.664\epsilon^2 - 7.139\epsilon^3 \\ \quad + 16.40\epsilon^4 - 19.06\epsilon^5 + 8.792\epsilon^6 \\ n = 8.323 - 65.10\epsilon + 535.6\epsilon^2 - 2473\epsilon^3 \\ \quad + 6086\epsilon^4 - 7513\epsilon^5 + 3661\epsilon^6 \\ Q = 384.3 - 1018\epsilon + 17190\epsilon^2 - 119500\epsilon^3 \\ \quad 377500\epsilon^4 - 554600\epsilon^5 + 307900\epsilon^6 \\ \ln A = 29.99 - 61.03\epsilon + 1196\epsilon^2 - 8836\epsilon^3 \\ \quad 28770\epsilon^4 - 43090\epsilon^5 + 24230\epsilon^6 \end{cases} \quad (8)$$

Once material constants have been evaluated, the flow true stress–true strain curves can be predicted. And the flow stress behavior can be described using Arrhenius equations and Zener–Hollomon parameter. It can be derived from Eqs. (1) and (4) that

$$\sigma = \frac{1}{\alpha} \ln \left\{ \left(\frac{Z}{A} \right)^{1/n} + \left[\left(\frac{Z}{A} \right)^{2/n} + 1 \right]^{1/2} \right\}. \quad (9)$$

After the values of Z were computed from Eq. (1), a comparison between the experimental data and predicted results was presented using Eq. (9). As shown in Fig. 5, it can be observed that predicted flow stress values from constitutive equation fit well with experimental data throughout the entire strain rate range.

In order to further evaluate the prediction accuracy of this constitutive equation, two standard statistical parameters were used in some researches. It was reported that the accuracy can be evaluated using correlation coefficient (R) and average absolute relative error (AARE) [37]. R (Eq. (10)) is used to analyze the linear relationship between predicted results and experimental data. Through a term comparison of the relative error, AARE is also calculated using Eq. (11).

Fig. 3 Relationships between **a** $\ln \sigma - \ln \dot{\epsilon}$, **b** $\sigma - \ln \dot{\epsilon}$, **c** $\ln[\sinh(\alpha\sigma)] - \ln \dot{\epsilon}$, **d** $\ln[\sinh(\alpha\sigma)] - 1/T$

$$R = \frac{\sum_{i=1}^N (E_i - \bar{E})(P_i - \bar{P})}{\sqrt{\sum_{i=1}^N (E_i - \bar{E})^2 (P_i - \bar{P})^2}}, \quad (10)$$

$$AARE = \frac{1}{N} \sum_{i=1}^N \left| \frac{E_i - P_i}{E_i} \right|, \quad (11)$$

where E_i is an experimental value of flow stress in the strains which was used for calculation, P_i is corresponding predicted value obtained from the modified constitutive equation. $\bar{E}$ and $\bar{P}$ are the mean values of E_i and P_i , respectively. N is the number of data employed in the investigation.

It can be seen from Fig. 6 that the obtained correlation between experimental and predicted data is good. The predicted curves can basically reflect shape of experimental

flow stress. The value of R was found to be 0.9976 (the coefficient of determination R^2 was 0.9952), and the value of AARE was 4.49%. The corresponding results confirm the accuracy and reliability of the modified constitutive equation. It can be also noted that there are bigger deviations at low temperatures or at large strain rates.

3.3 Processing Maps of Experimental TRIP Steel

Flow stress can be predicted using the constitutive equation. But processing maps should be applied to further study for the hot working properties and deformation parameters. And principles of dynamic material model (DMM) are widely used. According to DMM [21], processing and instability maps can be established. Microstructural changes are accompanied by power

Fig. 4 Relationships between **a** α , **b** n , **c** Q , **d** $\ln A$ and true strain

dissipation, and the efficiency of power dissipation η is given by Eq. (12).

$$\eta = \frac{2m}{m+1}, \quad (12)$$

where m is the strain rate sensitivity which is a function of temperature and strain rate. The efficiency η is concerned with the microstructural processes during hot working, and the value of η at different temperature and strain rate can be evaluated to obtain processing maps.

Processing maps are incomplete without the flow instability region being confirmed. And the processing property can be accurately predicted by studying power dissipation together with instability. Based on DMM and the theory of plastic flow [38], Kumar [39] has shown that flow instability occurs during hot deformation if

$$\zeta(\dot{\epsilon}) = \frac{\partial \ln\left(\frac{m}{m+1}\right)}{\partial \ln \dot{\epsilon}} + m < 0, \quad (13)$$

where $\dot{\epsilon}$ is strain rate. ζ is instability parameter which is used to obtain an instability map and varies with deformation temperature and strain rate. Combined with instability, processing maps can delineate the regimes of flow instability.

In order to ensure the validity of following calculation, the linear relationship between $\ln \sigma$ and $\ln \dot{\epsilon}$ was analyzed and the correlation coefficients are greater than 0.94 as shown in Fig. 7. It indicates that the constitutive behavior of experimental TRIP steel during hot working follows the exponential relationship (Eq. (3)). Therefore, processing maps of TRIP steel can be confirmed using principles of DMM and the theory of plastic flow.

Fig. 5 Comparison between the experimental and predicted flow stress from constitutive equation (considering the compensation of strain) at strain rates of **a** 0.01 s^{-1} , **b** 0.1 s^{-1} , **c** 1 s^{-1} , **d** 10 s^{-1}

Processing maps should take practical relevance into account. Furthermore, the processing maps were obtained corresponding to true strain of 0.3, 0.4, 0.5 and 0.6. A third-order polynomial was used to fit the relationship between $\ln \sigma$ and $\ln \dot{\epsilon}$ here (Eq. (14)).

$$\ln \sigma = a + b \ln \dot{\epsilon} + c(\ln \dot{\epsilon})^2 + d(\ln \dot{\epsilon})^3. \quad (14)$$

Equation (15) can be derived from Eq. (14) that

$$m = \frac{\partial(\ln \sigma)}{\partial(\ln \dot{\epsilon})} = b + 2c \ln \dot{\epsilon} + 3d(\ln \dot{\epsilon})^2. \quad (15)$$

So the strain rate sensitivity m was obtained after the values of b , c and d were fitted using Eq. (14). And once the value of m is confirmed, the efficiency of power dissipation η can be computed by Eq. (12). The instability

Fig. 6 Correlation between the experimental and predicted flow stress data from the modified constitutive equation

Fig. 7 Relationship between $\ln \sigma$ and $\ln \dot{\varepsilon}$ of chosen strains: **a** $\varepsilon = 0.3$; **b** $\varepsilon = 0.4$; **c** $\varepsilon = 0.5$; **d** $\varepsilon = 0.6$

parameter ξ was used to mark instability region. It can be derived from Eq. (13) that

$$\xi(\dot{\varepsilon}) = \frac{1}{m(m+1)} \frac{\partial m}{\partial \ln \dot{\varepsilon}} + m < 0. \quad (16)$$

Therefore, the processing maps were constituted the combining of the efficiency of power dissipation η from Eq. (12) with the instability parameter ξ from Eq. (16).

Figure 8 shows the processing maps at true strains of 0.3 (Fig. 8a), 0.4 (Fig. 8b), 0.5 (Fig. 8c) and 0.6 (Fig. 8d). The maps reveal three distinct domains at the temperature of 1200 °C: a valley domain occurring at strain rates greater than 1 s⁻¹, a valley domain occurring at strain rates less than 0.01 s⁻¹ and a peak domain occurring in the strain rate range of 0.1 to 1 s⁻¹. The peak domain is shown in

every figure that the value of η reveals as above 40%. Microstructural changes occur in the regimes where material dissipates more energy. The best deformation mechanism is considered to be DRX. As described by predecessors [40, 41], the DRX provides relatively stable flow stress and produces fewer defects. However, it should be noted that the unstable flow may occur in regimes with low η , which is usually manifested in the form of cracks or deformation bands [42]. As shown in Fig. 8, the unstable deformation domains can be observed at low temperatures and high strain rate [27]. According to the processing maps, there is a wide unstable flow domain at a strain rate of 10 s⁻¹ and in the temperature range of 1050 to 1100 °C. Figure 8d shows that there is an

Fig. 8 Processing maps for the experimental TRIP steel at true strains of **a** 0.3, **b** 0.4, **c** 0.5 and **d** 0.6. The numbers of the contour maps represent percent efficiency of power dissipation. The shaded corresponds to instability area

unstable deformation domain at the condition of 1200 °C, 0.01 s⁻¹ when the true strain is 0.6.

Figure 9 shows original austenite grain boundaries (OAGBs) of deformed high Al-low Si TRIP steel which obtained for six typical conditions: 1050 °C, 10 s⁻¹ (Fig. 9a); 1100 °C, 10 s⁻¹ (Fig. 9b); 1150 °C, 10 s⁻¹ (Fig. 9c); 1200 °C, 10 s⁻¹ (Fig. 9d); 1050 °C, 0.01 s⁻¹ (Fig. 9e) and 1200 °C, 0.01 s⁻¹ (Fig. 9f). It can be observed that (1) the deformed and non-uniform grains can be seen at 1050 °C, 10 s⁻¹ and 1100 °C, 10 s⁻¹, (2) the microstructure at deformation conditions of 1200 °C, 10 s⁻¹, 1200 °C, 0.1 s⁻¹, 1050 °C, 0.01 s⁻¹ and 1200 °C, 10 s⁻¹ consist of DRX grains, (3) the component DRX grains of 1200 °C, 0.1 s⁻¹ are finer and more uniform than other deformation conditions, and (4) the grains of 1200 °C, 0.01 s⁻¹ are coarse and non-uniform.

As shown in Fig. 9a, b, the representative microstructures of the instability domain were deformed at high strain rate conditions and this instability domain was present in the upper part of the processing map. Typically, the instability mechanisms are associated with cracking, localized plastic flow or adiabatic shear bands at low temperature and high strain rate conditions [43]. As for the

experimental steel, the original grains were elongated along the deformation direction at 1050 °C, 10 s⁻¹ and 1100 °C, 10 s⁻¹. It can be seen that the DRX grains became much coarse at deformation temperature of 1200 °C and at strain rate of 0.01 s⁻¹ (Fig. 9f). This condition is corresponding to the second instability domain of Fig. 8d at true strain of 0.6. In conclusion, the instability domains should be avoided during hot processing. In present study, DRX occurred, but was non-uniform at deformation conditions of 1200 °C, 10 s⁻¹ and 1050 °C, 0.01 s⁻¹ (Fig. 9c, e). As shown in Fig. 9d, at low strain rate, the characteristics agree well with the microstructure after DRX [44]. Combining with the processing maps, the condition of 1200 °C, 0.1 s⁻¹ is desirable and excellent for hot deformation of the experimental high Al-low Si steel, and conditions of low strain rates at temperature range of 1050–1150 °C can also be used for hot processing. As shown in Fig. 9, the volume fraction of DRX increases with increasing deformation temperature and decreasing strain rate.

Grain size has a significance effect on the mechanical property of experimental steel. Researches [29, 45] showed

Fig. 9 OAGBs of specimens deformed at: **a** 1050 °C, 10 s⁻¹; **b** 1100 °C, 10 s⁻¹; **c** 1200 °C, 10 s⁻¹; **d** 1200 °C, 0.1 s⁻¹; **e** 1050 °C, 0.01 s⁻¹, **f** 1200 °C, 0.01 s⁻¹

that the grain size of DRX D_s (μm) can be connected to Zener–Hollomon parameter Z (s⁻¹).

$$D_s = BZ^{-\kappa}, \quad (\ln D_s = -\kappa \ln Z + \ln B), \quad (17)$$

where B and κ are the Zener–Hollomon exponent.

The value of B and κ can be obtained from Fig. 10. Then, we can get the relationship between D_s and Z . As seen in Fig. 10, the grain size of DRX increased with decreasing Zener–Hollomon parameter.

4 Conclusions

Compression tests of high Al-low Si TRIP steel were studied at deformation temperature range of 1050–1200 °C and at strain rate range of 0.01–10 s⁻¹. The following are the conclusions.

1. Flow stress of experimental TRIP steel decreased with decreasing strain rate and increasing temperature. The

Fig. 10 Relationship between D_s and Z

true stress–true strain curves represented dynamic recovery and dynamic recrystallization at strain rate greater than 10 s^{-1} and strain rate less than 1 s^{-1} , respectively.

- The constitutive equation of experimental TRIP steel was established with sixth-order polynomial fitted material constants. Comparing with the experimental data, the predicted results fit well. The correlation coefficient (R) and average absolute relative error (AARE) were calculated to be 0.9976 and 4.49%, respectively.
- At $1200\text{ }^{\circ}\text{C}$, there are two valley domains in the processing maps with one being at strain rate greater than 1 s^{-1} and the other at strain rate less than 0.01 s^{-1} . And there was a peak at temperature of $1200\text{ }^{\circ}\text{C}$ and in strain rate range of 0.1 to 1 s^{-1} . Two obvious unstable flow domains were observed in the processing maps: One is in the temperature range of 1050 – $1100\text{ }^{\circ}\text{C}$ and at strain rate greater than 10 s^{-1} ; the other is at deformation temperature of $1200\text{ }^{\circ}\text{C}$ and strain rate of 0.01 s^{-1} .
- The flow instability domains should be avoided during hot processing. The flow instability domain was located in the high strain rate region and low temperature. The other is caused by coarse dynamic recrystallized grains. The desirable processing condition is at deformation temperature of $1200\text{ }^{\circ}\text{C}$ and at strain rate of 0.1 s^{-1} , where the DRX grains are uniform.

Acknowledgements This work was financially supported by the National Program on Key Basic Research Project (Grant No. 2011CB606306-2) and the National Natural Science Foundation of China (Grant No. 51775102).

References

- [1] T. Manabu, Y. Hiroshi, H. Shinji, Properties of TRIP type high strength steels. in *International Conference on TRIP-Aided High Strength Ferrous Alloys*. (Ghent, Belgium, 2002)
- [2] B. Mintz, *Int. Mater. Rev.* **46**(4), 169 (2001)
- [3] J. Van Slycken, P. Verleysen, J. Degrieck, J. Bouquereal, B.C. De Cooman, *Mater. Sci. Eng. A* **460–461**, 516 (2007)
- [4] B.C. De Cooman, *Curr. Opin. Solid State Mater. Sci.* **8**(3–4), 285 (2004)
- [5] E.M. Bellhouse, J.R. McDermid, *Metall. Mater. Trans. A* **41**(6), 1460 (2010)
- [6] J.R. McDermid, H.S. Zurob, Y. Bian, *Metall. Mater. Trans. A* **242**(12), 3627 (2011)
- [7] O. Matsumura, Y. Sakuma, H. Takechi, *Trans. Iron Steel Inst. Jpn.* **27**(7), 570 (1987)
- [8] A.R. Marder, *Prog. Mater. Sci.* **45**(3), 191 (2000)
- [9] M.D. Meyer, D. Vanderschueren, B.C. De Cooman, *ISIJ Int.* **39**(8), 813 (1999)
- [10] P.J. Jacques, E. Girault, A. Mertens, B. Verlinden, J. van Humbeeck, F. Delannay, *ISIJ Int.* **41**(9), 1068 (2001)
- [11] J. Mahieu, B.C. De Cooman, J. Maki, *Metall. Mater. Trans. A* **33**(8), 2573 (2002)
- [12] J. Maki, J. Mahieu, B.C. De Cooman, S. Claessens, *Mater. Sci. Technol.* **19**(1), 125 (2003)
- [13] C.M. Sellars, W.J. McTegart, *Acta Metall.* **14**(9), 1136 (1966)
- [14] C. Sun, J. Liu, R. Li, Q. Zhang, *Acta Metall. Sin.* **47**(2), 191 (2011). (in Chinese)
- [15] Y. Cao, H. Di, R.D.K. Misra, X. Yi, J. Zhang, T. Ma, *Mater. Sci. Eng. A* **593**, 111 (2014)
- [16] C. Sun, J. Luan, G. Liu, R. Li, Q. Zhang, *Acta Metall. Sin.* **48**(7), 853 (2012). (in Chinese)
- [17] N. Haghdadi, A. Zarei-Hanzaki, H.R. Abedi, *Mater. Sci. Eng. A* **535**, 252 (2012)
- [18] H. Wei, G. Liu, H. Zhao, R. Kang, *Mater. Des.* **50**, 484 (2013)
- [19] J. Zhang, H. Di, X. Wang, Y. Cao, J. Zhang, T. Ma, *Mater. Des.* **44**, 354 (2013)
- [20] X. Kai, C. Chen, X. Sun, C. Wang, Y. Zhao, *Mater. Des.* **90**, 1151 (2016)
- [21] Y.V.R.K. Prasad, H.L. Gegel, S.M. Doraivelu, J.C. Malas, J.T. Morgan, K.A. Lark, D.R. Barker, *Metall. Trans. A* **15**(10), 1883 (1984)
- [22] Y. Cao, H. Di, J. Zhang, T. Ma, J. Zhang, *Acta Metall. Sin.* **49**(07), 811 (2013). (in Chinese)
- [23] C. Sun, G. Liu, Q. Zhang, R. Li, L. Wang, *Mater. Sci. Eng. A* **595**, 92 (2014)
- [24] J. Zhang, H. Di, K. Mao, X. Wang, Z. Han, T. Ma, *Mater. Sci. Eng. A* **587**, 110 (2013)
- [25] Y. Zhang, H. Sun, A.A. Volinsky, B. Tian, K. Song, Z. Chai, P. Liu, Y. Liu, *Dynamic Recrystallization Behavior and Processing Map of the Cu–Cr–Zr–Nd Alloy*, vol. 5 (SpringerPlus, Berlin, 2016)
- [26] Y.X. Liu, Y.C. Lin, Y. Zhou, *Mater. Sci. Eng. A* **691**, 88 (2017)
- [27] D.G. He, Y.C. Lin, M.S. Chen, J. Chen, D.X. Wen, X.M. Chen, *J. Alloys Compd.* **649**, 1075 (2015)
- [28] D.X. Wen, Y.C. Lin, Y. Zhou, *Vacuum* **141**, 316 (2017)
- [29] H. Wei, G. Liu, X. Xiao, H. Zhao, H. Ding, R. Kang, *Mater. Sci. Eng. A* **564**, 140 (2013)
- [30] S. Serajzadeh, A. Karimi Taheri, *Mater. Des.* **23**(3), 271 (2002)
- [31] B. Zhang, G. Zhao, L. Jiao, G. Xu, H. Qin, D. Feng, *Acta Metall. Sin.* **41**(4), 351 (2005). (in Chinese)
- [32] C. Zener, J.H. Hollomon, *J. Appl. Phys.* **15**(1), 22 (1944)
- [33] S. Mandal, V. Rakesh, P.V. Sivaprasad, S. Venugopal, K.V. Kasiviswanathan, *Mater. Sci. Eng. A* **500**(1–2), 114 (2009)
- [34] C. Sun, J. Liu, R. Li, Q. Zhang, J. Dong, *Rare Met.* **30**(1), 81 (2011)

- [35] Y. Lin, M. Chen, J. Zhong, *Comput. Mater. Sci.* **42**(3), 470 (2008)
- [36] L. Tang, J.D. Baeder, *SIAM J. Sci. Comput.* **20**(3), 1115 (1998)
- [37] M.P. Phaniraj, A.K. Lahiri, *J. Mater. Process. Technol.* **141**(2), 219 (2003)
- [38] H. Ziegler, *Some Extremum Principles in Irreversible Thermodynamics, with Application to Continuum Mechanics*. 1962: Swiss Federal Institute of Technology
- [39] A.K. Kumar, Criteria for predicting metallurgical instabilities in processing, in M. Sc.(Eng.) Thesis, (Indian Institute of Science, Bangalore, India, 1987)
- [40] M. Karami, R. Mahmudi, *Mater. Lett.* **81**, 235 (2012)
- [41] R. Bhattacharya, B.P. Wynne, W.M. Rainforth, *Scr. Mater.* **67**(3), 277 (2012)
- [42] Y.V.R.K. Prasad, K.P. Rao, *Mater. Sci. Eng. A* **391**(1–2), 141 (2005)
- [43] Y. Sun, W.D. Zeng, Y.Q. Zhao, X.M. Zhang, Y. Shu, Y.G. Zhou, *Mater. Sci. Eng. A* **528**(3), 1205 (2011)
- [44] S. Anbu Selvan, S. Ramanathan, *Trans. Nonferrous Met. Soc. China* **21**(2), 257 (2011)
- [45] G. Anan, S. Nakajima, M. Miyahara, S. Nanba, M. Umemoto, A. Hiramatsu, A. Moriya, T. Watanabe, *ISIJ Int.* **32**(3), 261 (1992)