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Deformation Behavior and Processing Map during Isothermal Hot Compression of 49MnVS3 Non-Quenched and Tempered Steel

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Abstract: The deformation behavior of 49MnVS3 non-quenched and tempered steel was studied using isothermal compression tests at the temperature range of 750–1000 °C and the strain rate range of 0.1–50 s⁻¹ on a Gleeble-3500 thermal mechanical simulator. The results indicated that the flow stress increases significantly with decreasing temperature and increasing strain rate. Under the whole deformation conditions, the dependence of flow stress on deformation temperature and strain rate was analyzed by hyperbolic sine equation. Besides, the hot deformation activation energy and stress exponent were calculated to be 323.56 kJ/mol and 6.99, respectively. In addition, the processing map based on dynamic material model was established, and the optimum processing condition of the 49MnVS3 non-quenched and tempered steel can be determined to be at the temperature range of 809–850 °C and strain rate of 36.6 s⁻¹–50 s⁻¹.

Keywords: 49MnVS3 non-quenched and tempered steel, hot deformation behavior, constitutive equation, processing map, microstructure

Introduction

As an energy-efficient steel, non-quenched and tempered steel combines forging and heat treatment into one process, which is not required for quenching and heat treatment [1]. Till now, non-quenched and tempered steel has been extensively employed in numerous fields such as automobile, agricultural machinery, machine tool and engineering machinery forging, shafts and bolts [2–4]. Non-quenched steel was first introduced in the 1970s and the German Thyssen steel company successfully

developed the first generation of 49MnVS3 non-quenched and tempered steel [5]. So far, the steel has replaced the 50Mn, 40Cr and a series of quenched and tempered steel, which has been used in the manufacture of automobile crankshaft forging [6]. However, compared with quenched and tempered steel, non-quenched and tempered steel has relatively poor toughness, especially for medium carbon steel. Although higher carbon content makes the strength better, it remains more challenging to obtain better toughness than low carbon steel when the carbon content is high. Studies have already shown that the intragranular ferrite can effectively refine the ferrite grains, which is an efficient way to improve the toughness of non-quenched and tempered steel [7–9]. In the forging process, the deformation resistance and microstructure of the material would occur some complicated changes under different deformation conditions, such as strain, strain rate, deformation temperature and deformation degree. To control the forming process, it is essential to study the thermal deformation behavior and processing of non-quenched and tempered steel under different hot working conditions so as to determine the optimum process parameters.

Constitutive equation can be used to describe the relationship between flow stress, strain, strain rate and deformation temperature during material deformation. Thus, the constitutive equation is the basic data that can predict the flow behaviors of material [10]. So far, the processing maps based on the dynamic materials model (DMM) have already been successfully used to study the deformation behavior of various metals and their alloys, such as aluminum alloy, copper alloy, magnesium alloy, titanium alloy, stainless steel and nickel base alloy [11–16]. Processing maps can evaluate the processability of the material and optimize the process parameters of material. The safe and hazardous areas of the plastic working of the material can be obtained by processing map, thereby obtaining the optimized thermal parameters of the material. Therefore, the proposed study aimed to study the high temperature deformation behavior of 49MnVS3 non-quenched and tempered steel. As a result, isothermal hot compression tests were performed

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in the temperature range of 750–1000 °C and the strain rate of 0.1–50 s⁻¹. Then, the constitutive equation and processing map of 49MnVS3 non-quenched and tempered steel were developed to optimize the process parameters.

Experimental procedure

The material used in this investigation was hot rolling 49MnVS3 non-quenched and tempered steel with a diameter of 39.24 mm. Table 1 presents the chemical composition (mass %) and the initial microstructure is shown in Figure 1, from which, it can be seen that the microstructure of the 49MnVS3 non-quenched and tempered steel consists of distributed uniformly reticular and massive ferrite and pearlite. Cylindrical compression specimens of 49MnVS3 non-quenched and tempered steel were 10 mm in diameter and 15 mm in height. Isothermal compression experiments at a constant strain were carried out on Gleeble-3500 thermal mechanical simulator. To ensure a homogenous temperature distribution throughout the specimen, each specimen was heated up to 1150 °C with a heating rate of 5 °C/s and held for 5 min prior to deformation. Then, the specimens were cooled down to 750, 800, 850, 900, 950, and 1000 °C at a cooling rate of 5 °C/s and held for 30 s to eliminate the temperature gradient. Afterwards, the specimens were compressed to 0.9 strain with the strain rate of 0.1, 1, 10, and 50 s⁻¹, followed by quenching in water so as to preserve the deformed microstructure. The deformed specimens were axially sectioned and prepared by standard metallographic techniques to observe the microstructure.

Results and discussion

Flow stress behavior

Figure 2 shows the typical true stress-strain curves of 49MnVS3 non-quenched and tempered steel obtained at

Figure 1: The original microstructure of the 49MnVS3 non-quenched and tempered steel.

different deformation temperatures and strain rates, presenting that the flow stress decreases significantly with the increasing temperature and the strain rate decreases, which indicates that the flow stress is sensitive with the deformation temperature and strain rate. The reason is that the average kinetic energy of the atoms increases significantly and the high temperature critical slip shear stress decreases rapidly, causing that the obstruction of dislocation movement decreases with increasing temperature [17]. In addition, at higher temperatures and lower strain rates, the reduction of the dislocation density counteracts the hardening caused by deformation, consequently reducing the flow stress of the 49MnVS3 non-quenched and tempered steel.

According to Figure 2 (a), at the lower temperature (800 °C), the stress-strain curves at each strain rate have no obvious peak stress. Similarly, at the higher strain rate (10 s⁻¹), the stress-strain curves at each deformation temperature have the same characteristics, as shown in Figure 2 (d). The flow stress increases monotonously with the increase of strain under certain deformation conditions, and the rate of increase of flow stress decreases with the increase of strain, showing that the experimental steel softens to the main dynamic response

Table 1: Chemical composition of 49MnVS3 non-quenched and tempered steel.

Type of steel	C	Si	Mn	P	S	Cr	Ni	Mo	Cu	Al
49MnVS3	0.50	0.26	0.83	0.010	0.047	0.08	0.04	0.02	0.06	0.024

Figure 2: True stress-strain curves of 49MnVS3 non-quenched and tempered steel obtained under different deformation conditions: (a) 800 °C; (b) 1000 °C; (c) 0.1 s⁻¹; (d) 10 s⁻¹.

at higher strain rates. The flow stress rapidly reaches the peak stress at the beginning of the stress - strain curves due to the work hardening. The work hardening in the initial stage relates to the formation of poorly developed subgrain boundaries and the increase of dislocation density [18, 19]. In the latter stage of compression, the driving force caused by the material deformation is sufficient to complete the thermal deformation such as dynamic recovery and dynamic recrystallization. Thus, the material softens. When the softening behavior is stronger than the work hardening, the stress curve decreases. When the softening and work hardening tend to balance, the curve shows a gentle state.

In addition, oscillatory flow curves are observed particularly at strain rate of 50 s⁻¹. Besides, similar phenomenon has also been reported in the GH3535 superalloy [20], IC396LZR alloy [21] and Ni3Al-based superalloy [22]. The main reason may be that under the high strain rate, the heat generated during the metal deformation can not

be quickly released, resulting in a large amount of deformation heat accumulating, which creates favorable conditions for the flow softening. While the flow softening and deformation hardening process alternated and the response to the stress-strain curve showed an oscillating shape [21].

The effect of deformation parameters on the peak flow stress for 49MnVS3 non-quenched and tempered steel are shown in Figure 3, showing that the peak stress decreases with the increase of the deformation temperature and the decrease of strain rate. The overall decrease at lower strain rates has no obvious difference with that at high strain rates.

The softening of flow stress can be described by the difference between the peak stress and the steady-state stress [23]:

$$\Delta\sigma = \sigma_p - \sigma_s \quad (1)$$

Figure 3: Peak stress in the isothermal compression of 49MnVS3 non-quenched and tempered steel at different deformation conditions.

Where σ_p is the peak stress and σ_s is the steady-state stress (use $\sigma_{0.9}$ as σ_s).

According to eq. (1), the flow softening curves can be obtained, as shown in Figure 4, which presents that when the strain rate is above 10 s^{-1} , $\Delta\sigma$ of the single-phase region (below A_{C3}) is higher than that of the two-phase region. Strain rate exerts a great influence on the softening. $\Delta\sigma$ at the 50 s^{-1} strain rate is significantly higher than that the other strain rates in the single-phase region (below T_a). While in the single-phase region (above T_a), $\Delta\sigma$ at 0.1 s^{-1} strain rate is slightly higher than that at other strain rates. In general, the adiabatic effect and microstructure evolution are the

Figure 4: Flow softening of 49MnVS3 non-quenched and tempered steel with increasing temperature for various strain rates.

main reasons of flow softening [24]. As shown in Figure 4, when the temperature is higher than the austenite single-phase region (T_a), the flow softening stress at a specific strain rate with increasing temperature remained stable. Additionally, the effect of temperature on $\Delta\sigma$ also reflects changes in the microstructure of the alloy, such as grain size, volume and composition of the phase, and the number of dislocations [17].

Constitutive equation

The high temperature deformation behavior of the material is represented by the Zener-Hollomon parameter (Z), which correlates the strain rate, the deformation temperature and the activation energy by the following expression [25]:

$$Z = \dot{\epsilon} \exp\left(\frac{Q}{RT}\right) \quad (2)$$

where $\dot{\epsilon}$ represents strain rate (s^{-1}), R is the gas constant ($8.314 \text{ J}\cdot\text{mol}^{-1}\cdot\text{K}^{-1}$), T is the absolute deformation temperature (K) and Q is the activation energy of hot deformation which reflects the ease or difficulty of material thermal deformation ($\text{J}\cdot\text{mol}^{-1}$).

The Arrhenius equation is extensively used to describe the relationship between flow stress, strain rate and temperature [26]

$$\dot{\epsilon} = AF(\sigma) \exp\left(-\frac{Q}{RT}\right) \quad (3)$$

in which,

$$F(\sigma) = \sinh(\alpha\sigma)^n \quad (4)$$

Where σ is the instantaneous stress at different strain, A and α are material constants, n is a constant related to the strain rate. Through combining eq. (3) and (4), the following relationship can be acquired:

$$\ln[\sinh(\alpha\sigma)] = \frac{\ln \dot{\epsilon}}{n} + \frac{Q}{nRT} - \frac{\ln A}{n} \quad (5)$$

According to eq. (5), the following equation can be obtained:

$$Q = Rn \left(\frac{d\{\ln[\sinh(\alpha\sigma)]\}}{d(1/T)} \right) \quad (6)$$

The value of α can be determined using $\alpha = \beta/n_1$, and β can be acquired using $\beta = \frac{\partial \ln \dot{\epsilon}}{\partial \ln \sigma}$, $n_1 = \frac{\partial \ln \dot{\epsilon}}{\partial \ln \sigma}$ and n can be

represented by $\frac{1}{n} = \frac{d[\ln(\sinh(\alpha\sigma))]}{d(\ln \dot{\epsilon})}$. Through applying linear regression of $\ln \sigma - \ln \dot{\epsilon}$ and $\sigma - \ln \dot{\epsilon}$, as shown in Figure 5 (a) and (b), the mean values of the slopes are taken as the values of n_1 and β which can be calculated to be 9.56 and 0.0448, respectively. Subsequently, the value of α can be drawn as 0.004684. Therefore, the value of n is 6.99, and Q can be obtained from the slope of line from $\ln[\sinh(\alpha\sigma)] - 1/T$, as presented in Figure 6. The average value of Q obtained from different strain rate can be calculated to be 323.56 kJ/mol, which is slightly lower than the result of Chen (350.98 kJ/mol) [27]. As a result, the hyperbolic sine function for the studied 49MnVS3 non-quenched and tempered steel can be expressed as:

$$\dot{\epsilon} = 6.05 \times 10^{14} [\sinh(0.004684\sigma)]^{6.99} \exp\left(-\frac{323560}{RT}\right) \quad (7)$$

Figure 7 illustrates the relationship between the peak stress and the Zener-Hollomon parameter for 49MnVS3 non-quenched and tempered steel. Z parameter increases consistently with the increase of peak stress. Figure 7 shows that the correlation coefficient for the linear regression is 0.991, which indicates the high accuracy of eq. (7) for describing the hot deformation behavior of 49MnVS3 non-quenched and tempered steel.

Processing map

Processing map based on DMM is widely used for optimizing the hot working processing parameters, which is obtained by superposition of a power dissipation map and an instability map. The stable and unstable regions during plastic deformation can be easily obtained using the processing map.

Figure 5: The value evaluation of (a) n_1 by plotting $\ln \sigma$ versus $\ln \dot{\epsilon}$, and (b) β by plotting σ versus $\ln \dot{\epsilon}$.

Figure 6: The linear relationships of (a) $\ln[\sinh(\alpha\sigma)]$ and $\ln \dot{\epsilon}$ (b) $\ln[\sinh(\alpha\sigma)]$ and $1000/T$.

Figure 7: Relationship between $\ln Z$ and $\ln[\sinh(\alpha\sigma)]$.

The DMM regards the hot working piece in the plastic deformation process as an energy dissipation system, and the total power P absorbed in the system consists two complementary parts [28, 29]:

$$P = \sigma \dot{\epsilon} = G + J = \int_0^{\dot{\epsilon}} \sigma d\dot{\epsilon} + \int_0^{\sigma} \dot{\epsilon} d\sigma \quad (8)$$

where G represents the power dissipated by plastic work, most of which is converted to viscoplastic heat, and a small amount of the remaining power is stored in a lattice defect [30]. J denotes the energy consumed by the dynamic microstructure evolution. The stress and strain rate of the hot working workpiece have the following power function at the plastic deformation:

$$\sigma = K \dot{\epsilon}^m \quad (9)$$

where K is a constant and m is the strain rate sensitivity factor, which is calculated by:

$$m = \frac{\partial J}{\partial G} = \frac{\dot{\epsilon} \partial \sigma}{\sigma \partial \dot{\epsilon}} = \frac{\partial \ln \sigma}{\partial \ln \dot{\epsilon}} \quad (10)$$

The power ratio for microstructural evolution during hot deformation can be expressed by the efficiency of power dissipation (η), which is defined as [28, 29]:

$$\eta = \frac{J}{J_{\max}} = \frac{2m}{m+1} \quad (11)$$

The variation of η with the deformation temperature and strain rate constitutes the power dissipation map. In the

process of plastic deformation, all kinds of damage or metallurgical changes dissipate power. Therefore, the power dissipation map can be used to analyze the deformation mechanism in different regions by metallographic observation.

The interior processing performance of the material depends on the power consumed through processing the unstable energy. Prasad [28, 31] and Krishna [32] applied the principle of maximum entropy generation to the theory of large plastic rheology to obtain the criterion of rheological instability zone. The criteria for instability are as following:

$$\xi(\dot{\epsilon}) = \frac{\partial \ln(\frac{m}{m+1})}{\partial \ln \dot{\epsilon}} + m < 0 \quad (12)$$

where $\xi(\dot{\epsilon})$ is a dimensionless instability parameter. The variation of $\xi(\dot{\epsilon})$ with the deformation temperature and strain rate constitutes the instability map, flow instability is predicted where the $\xi(\dot{\epsilon})$ becomes negative.

Then, the processing map of 49MnVS3 non-quenched and tempered steel is obtained by superposing the instability map on the power dissipation map. Figure 8 presents the processing map for 49MnVS3 non-quenched and tempered steel at strain of 0.9. The contour numbers in the figure indicate percent efficiency of power dissipation (η), and the gray areas represent the flow instability region during deformation. The higher value of η does not necessarily indicate that the processing conditions are more adapted to processing [33]. Figure 8 shows that this non-quenched and tempered steel exhibits two different instability regions: (I) the first region is at the

Figure 8: Processing map of 49MnVS3 non-quenched and tempered steel at strain of 0.9.

range from 0.1 s^{-1} to 1.9 s^{-1} lower strain rates and within the range from $750 \text{ }^{\circ}\text{C}$ to $887 \text{ }^{\circ}\text{C}$ lower deformation temperatures; (II) the second region locates in the deformation temperatures from $909 \text{ }^{\circ}\text{C}$ to $984 \text{ }^{\circ}\text{C}$ and the strain rates from 0.13 s^{-1} to 2.2 s^{-1} . Accordingly, it is necessary to stay away from these regions.

Microstructural observations

To investigate the mechanism of microscopic deformation and verify the reliability of sample deformation under specific processing parameters, microstructures of typical deformation are characterized and analyzed.

Figure 9 shows the microstructure of 49MnVS3 non-quenched and tempered steel at the $750 \text{ }^{\circ}\text{C}$, 0.1 s^{-1} and $850 \text{ }^{\circ}\text{C}$, 0.1 s^{-1} , corresponding to the instability zone in Figure 8. Figure 9 (a) shows that the ferrite is distributed at the pearlite boundary. The pearlite is coarse, and ferrite is massive and precipitates along grain boundaries, which is mainly because at lower deformation temperature, the deformation is insufficient to cause the austenite undergoing dynamic recrystallization (DRX) before the phase transition and thus the grain cannot be refined. Moreover, the small deformation leads to less distortion in the grains, which cannot provide the location of the ferrite nucleation in the process of phase transformation. As a result, the ferrite is mainly concentrated in the distortion of the grain boundary. From Figure 9 (b), it can be observed that part of the ferrite is coarse and unevenly distributed. This is the fact that some of the ferrite in the deformed structure is not completely austenitized, which is detrimental to controlling the ferrite content. These features indicate that 49MnVS3

non-quenched and tempered steel has a poor workability at the lower temperature and lower strain rate.

Generally, dynamic recrystallization occurs in the region of $\eta > 0.3$ [34, 35]. Based on Figure 8, at a high strain rate, the maximum of η is 0.28. Therefore, it is considered that no dynamic recrystallization occurs in this region. In most cases, the non-quenched and tempered steel after rolling is the most widely used with large amount of application. Rolling of the non-recrystallized zone is usually performed below the termination temperature of recrystallization [36]. In the austenite non-recrystallization zone rolling, the process of austenite into ferrite makes ferrite and pearlite grain size refined. Grain refinement can improve the strength and toughness of the metal material [37].

Figure 10 shows the microstructure of 49MnVS3 non-quenched and tempered steel at temperature of $850 \text{ }^{\circ}\text{C}$

Figure 10: Microstructures of 49MnVS3 non-quenched and tempered steel at temperature of $850 \text{ }^{\circ}\text{C}$ and strain rate of 50 s^{-1} .

Figure 9: Microstructures of 49MnVS3 non-quenched and tempered steel deformed at the strain rate of 0.1 s^{-1} and temperature of (a) $750 \text{ }^{\circ}\text{C}$ and (b) $850 \text{ }^{\circ}\text{C}$.

with the strain rate of 50 s^{-1} , presenting that the ferrite is refined and uniformly distributed. The steel is in the austenite single phase region, which also locates in the non-austenite dynamic recrystallization zone. The austenite grain will not be excessively small and thus the grain boundary will not increase. Therefore, to obtain homogeneous and fine microstructures, the optimum process region from the processing map is the deformation temperature range of 809°C to 850°C and strain rate range of 36.6 s^{-1} to 50 s^{-1} .

Conclusions

To conclude, hot deformation behavior of 49MnVS3 non-quenched and tempered steel was analyzed by constitutive equations and processing map in this study. The main results could be summarized as follows:

- (1) The flow stress increases significantly with the decrease of temperature and increase of strain rate. Meanwhile, when the strain rate is above 10 s^{-1} , the flow stress curves reveal higher flow softening in the single-phase region (below A_{c3}) than that in the two-phase region.
- (2) The relationship between flow stress and Zener Hall parameters was successfully analyzed by the hyperbolic sine function over the entire deformation condition. Thus, the apparent activation energy (Q), and the material constants of A and n were determined to be 323.56 kJ/mol , 6.05×10^{14} and 6.99 , respectively. The relationship between strain rate, stress and deformation temperature can be expressed as: $\dot{\epsilon} = 6.05 \times 10^{14} [\sinh(0.004684\sigma)]^{6.99} \exp\left(-\frac{323560}{RT}\right)$. The results reflected the good prediction ability of the constitutive equation of 49MnVS3 non-quenched and tempered steel.
- (3) The processing map was established for the 49MnVS3 non-quenched and tempered steel, and the distribution of power dissipation values η were verified by microstructural observation. To obtain fine and homogeneous microstructures, the optimal process region from the processing map was temperature ranging from 809 to 850°C and strain rate ranging from 36.6 s^{-1} to 50 s^{-1} .

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References

- [1] X. Li, X.X. Li, M.Y. Li and X.C. Wu, *Adv. Mater. Res.*, 634 (2013) 2843–2848.
- [2] M.L. Li, F.M. Wang, C.R. Li, Z.B. Yang, Q.Y. Meng and S.F. Tao, *Int. J. Miner. Metall. Mater.*, 22 (2015) 589–597.
- [3] Y.J. Xia, F.M. Wang, J.L. Wang and G.Z. Li, *Int. J. Miner. Metall. Mater.*, 18 (2011) 562–569.
- [4] Z.B. Yang, F.M. Wang, S. Wang and B. Song, *Steel Res. Int.*, 79 (2008) 390–395.
- [5] D.I.S. Engineer, D.I.B. Huchtemann and D.I.V. Schüler, *Steel Res. Int.*, 58 (1987) 369–376.
- [6] B. Zhou, Y. Shen, L. Tan, H.X. Yang, W.Q. Cao and Y.Z. Bao, *Phys. Procedia.*, 50 (2013) 25–31.
- [7] G. Dini, V.M. Monir and A. Shafyei, *Isij Int.*, 46 (2006) 89–92.
- [8] M.T. Ma, H.Y. Yang, H.S. Mei and H.Z. Lu, *Adv. Mater. Res.*, 509 (2012) 75–81.
- [9] T.H. Man, P. Gao, Y.G. He, Z.L. Zhan, L. Tan and Y.Z. Bao, *Adv. Mater. Res.*, 936 (2014) 1179–1183.
- [10] H. Shi, A.J. McLaren, C.M. Sellars, R. Shahani and R. Bolingbroke, *Mater. Sci. Technol.*, 13 (1997) 210–216.
- [11] Y.J. Guo, L. Deng, X.Y. Wang, J.S. Jin and W.W. Zhou, *Adv. Mater. Res.*, 815 (2013) 37–42.
- [12] J. Cai, K.S. Wang and W. Wang, *Rare Metal Mat. Eng.*, 45 (2016) 2549–2554.
- [13] H.M. Liao, M. Zeng, S.Y. Long and H.X. Cao, *Mater. Sci. Forum*, 686 (2011) 168–175.
- [14] Y.H. Yang, D. Liu, G.J. Gao and J.G. Wang, *Adv. Mater. Res.*, 904 (2014) 262–268.
- [15] W. He, J. Liu, S. Sun, H. Chen and H. Guo, *Mater. Res. Innov.*, 15 (2015) 245–248.
- [16] Y. Kong, P. Chang, Q. Li, L. Xie and S. Zhu, *J. Alloys Compd.*, 622 (2015) 738–744.
- [17] W. Peng, W. Zeng, Q. Wang, Q. Zhao and H. Yu, *Mater. Sci. Eng. A*, 593 (2014) 16–23.
- [18] D. Cai, L. Xiong, W. Liu, G. Sun and M. Yao, *Mater. Charact.*, 58 (2007) 941–946.
- [19] Z.X. Shi, X.F. Yan, C.H. Duan, J.G. Song, M.H. Zhao and J. Wang, *J. Iron Steel Res. Int.*, 24 (2017) 625–633.
- [20] Y.Q. Ji, S.D. Qu and W.X. Han, *Trans. Nonferrous Met. Soc. China*, 25 (2015) 88–94.
- [21] Y.V.R.K. Prasad, S. Sasidhara and V.K. Sikka, *Intermet.*, 8 (2000) 987–995.

- [22] Y.T. Wu, Y.C. Liu, C. Li, X. Xia and Y.P. Huang, *J. Alloys Compd.*, 712 (2017) 687–695.
- [23] B.H. Lee, N.S. Reddy, J.T. Yeom and C.S. Lee, *J. Mater. Process. Technol.*, 187 (2007) 766–769.
- [24] M. Jackson, N.G. Jones, D. Dye and R.J. Dashwood, *Mater. Sci. Eng. A*, 501 (2009) 248–254.
- [25] K. Wu, G. Liu, B. Hu, F. Li, Y. Zhang, Y. Tao and J. Liu, *Mater. Charact.*, 61 (2010) 330–340.
- [26] K. Wu, G.Q. Liu, B.F. Hu, C.Y. Wang, Y.W. Zhang, Y. Tao and J.T. Liu, *Mater. Sci. Eng. A*, 528 (2011) 4620–4629.
- [27] Y.F. Chen, X.D. Peng, H.B. Xu, H.D. Jiang and G.H. Guan, *Strength Mater.*, 46 (2014) 198–207.
- [28] Y.V.R.K. Prasad and K.P. Rao, *Mater. Sci. Eng. A*, 487 (2008) 316–327.
- [29] C. Zhang, L. Zhang, W. Shen, M. Li and S. Gu, *J. Mater. Eng. Perform.*, 24 (2015) 149–157.
- [30] S. Huang, L. Wang, X.T. Lian, B.J. Zhang and G.P. Zhao, *Acta Metal. Sin.*, 27 (2014) 198–204.
- [31] Y.V.R.K. Prasad and T. Seshacharyulu, *Mater. Sci. Eng. A*, 243 (1998) 82–88.
- [32] V.G. Krishna, Y.V.R.K. Prasad, N.C. Birla and G.S. Rao, *J. Mater. Process. Technol.*, 71 (1997) 377–383.
- [33] Y.C. Lin, L.T. Li, Y.C. Xia and Y.Q. Jiang, *J. Alloys Compd.*, 550 (2013) 438–445.
- [34] J.L. Zhang, H.Z. Guo and H.Q. Liang, *Rare Met.*, 35 (2016) 118–126.
- [35] Z.X. Shi, X.F. Yan and C.H. Duan, *J. Alloys Compd.*, 652 (2011) 30–38.
- [36] A.H. Nobari and S. Serajzadeh, *Appl. Therm. Eng.*, 31 (2011) 487–492.
- [37] H.L. Yi, Z.Y. Liu, G.D. Wang and D. Wu, *J. Iron Steel Res. Int.*, 17 (2010) 54–58.