Exploring the influence of Al content on the hot deformation behavior of Fe-Mn-Al-C steels through 3D Processing Map

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Abstract

The effect of Al content on the hot deformation behavior of high-Mn low density Fe-Mn-Al-C steels was investigated by the 3D processing map at the temperatures of 850-1050 °C and the strain rates of 0.001-10s⁻¹. The high-Al steel showed a higher flow stress and a greater activation energy (443 kJ/mol) in contrast to the low-Al steel (394 kJ/mol). The microstructures of 8Al steel and 10Al steel depends on the deformation parameters. The initial hot rolling microstructure has been displaced by the recrystallized structure and substructures for both steels, while the unstable zone has deformation bands and flow localization. As the Z content increases with increasing Al content, this leads to a significant inhibition of the DRX process, resulting in an unstable domain with deformation zones and flow localization. For high-Al steel, the morphology and distribution of ferrite transform from the continuous band ferrite to discontinuous granular ferrite, moreover, the flow localization features cannot be observed with the decrease of strain rate to 0.001s⁻¹. Furthermore, the proportion of instability region for each of the strains increases with the increasing of the Al content, the high Al content weakens the workability of the Fe-Mn-Al-C steels.

Key words: Deformation behavior; Low density steel; 3D Processing map; Fe-Mn-Al-C

1. Introduction

The development of advanced high strength steels with low density has been a hot topic in the

automotive industry. Recently, High Mn Fe-Mn-Al-C steels have received substantial attention as a type of advanced high strength steel due to their wonderful ductility and superb strength, for instance twinning induced plasticity (TWIP) steel[1]. Despite the recent great progress reviewed in TWIP steels, another type of high Mn and high Al steels has been intensively investigated owing to exceptional advantages such as the specific weight reduction. In previous work, the mechanical properties of Fe-(20-28)Mn-(6-12)Al-(0.6-1.2)C (wt.%) alloys were described to be same as TWIP steels [2-5]. The optimization of the chemical constituents [6, 7] and the heat treatment parameters in this high Mn low density Fe-Mn-Al-C steels have been intensively investigated [8-11]. In most studies of high Mn low density Fe-Mn-Al-C steels, hot deformation behavior has been emphasized with attention being given to forming operations such as forging, extrusion or rolling. It is necessary to explore the hot deformation behavior of these steels because the thermomechanical processing is an important part of steel production [12]. Recently, several excellent reviews have described the hot deformation behavior of high Mn (>12, wt.%) Fe-Mn-Al-C steels. For example, Mohamadizadeh et al. have illustrated that the effects of deformation parameters on the activation energy evolution of Fe-18Mn-8Al-0.8C steel. Hamada et al. investigated hot deformation behavior of Fe-20Mn-1.5Al-0.8C steel by the physically based modeling[13-15]. As we know, Al is the major alloying elements of the present alloy, and Al plays a significant role on the ferrite stabilizer. However, it should be noted that very little work has been carried out on the influence of Al content on the hot deformation behavior of these steels. Therefore, it is essential to examine the effect of Al content on the flow behavior and microstructure evolution of low density Fe-Mn-Al-C steels.

It is commonly known that the processing map is a useful technique to investigate the flow stress and optimize the hot processing parameters[16-21]. Although the conventional 2D plots can exhibit

instability zones and efficiency contours at different hot working parameters, unable to reveal the influence of strain in one visual representation[15, 22]. Therefore, the 3D processing map is established to reflect the effect of strain at various thermomechanical conditions[15], which can reveal the relevance between microstructural evolution and processing map.

In this paper, the effect of Al on the high-temperature flow behavior of Fe-Mn-Al-C steels was revealed. The relevance between the microstructural evolution and flow behavior was established. The optimum processing parameters of Fe-Mn-Al-C steel are determined in a 3D processing map.

2. Materials and Methods

Fe-26Mn-8/10Al-1C (wt.%) steels, are hereafter appointed 8Al and 10Al steels, respectively. These steels were melted in an argon gas induction furnace. First of all, these steels were homogenized at 1200 °C for 2h, then the ingot was hot rolled into 20mm thickness, and the final rolling temperature was also controlled to 950 °C. The phase diagram for 8Al and 10Al steels is shown in Figure 1. As seen from this figure, the 8Al steel presents single austenite phase in temperature range of 800-1300 °C, which may be result from the high content of C and Mn. The ferrite starts to form at about 800°C with increasing Al content in temperature range of 800-1300 °C. The microstructure in the hot-rolled condition is given in Figure 2, the hot-rolled microstructure of 8Al is composed of a single austenite, while the 10Al steel which consists of a small amount of strip-like ferrite in austenitic matrix. This indicates a good agreement between the microstructure observed and calculated phase composition. The hot compression tests were carried out on a Gleeble-3800 thermomechanical simulator at temperatures of 850-1150°C and strain rates of 0.001-10s⁻¹. The specimens for compression tests were prepared with a 10mm diameter and 15mm height. In order to obtain uniform samples, the specimens were heated to 1000 °C and held for 1800s followed by cooling at a rate of

10 °C/s to the deformation temperature. All specimens were deformed to a strain of 0.6 and quenched immediately in water. The hot compressed specimens were cut parallel to the compression axis. All specimens were polished and then etched in 4% nital solution for microstructure obtained using Leica microscope.

Figure.1 Equilibrium phase diagram: (a) 8Al, (b) 10Al.

Figure.2 The hot-rolled microstructures: (a) 8Al, (b) 10Al.

3. Results and Discussion

3.1. The effect of Al on the flow behavior for Fe-Mn-Al-C steels

The flow curves of 8Al and 10Al steels deformed at temperatures of 850-1150 °C and strain rates of 0.001-10 s⁻¹ are presented in Figures 3 and 4, respectively. When deformation temperature is constant, flow stress increases with increasing strain rate. In contrast, the flow stress decreases with the increase of deformation temperature at a certain strain rate, which confirms that the present steel is impressible to the hot deformation parameters. This phenomenon is due to the increase of atomic

activation energy and the reduction of critical shear stress with increasing the temperature. In addition, the obstacles of dislocation slip are reduced at high temperature[23]. At deformation temperature of 850°C for high strain rate of 8Al (Figure 3a), a slight flow softening is observed. This tendency of flow softening is also exhibited at 850 °C for 10Al steel as shown in Figure 4a. Moreover, by increasing the deformation temperature, most of the flow stress curves have the same trend for 8Al and 10Al steels, as shown in Figure 4. The flow stress curves show a board stress peak, which may be caused by the dynamic recrystallization (DRX). The flow stress curves in Figures 3 and 4 show that the 10Al steel exhibits a higher strength than 8Al steel. This result is consistent with the observations of the study that the Al element is beneficial to the strengthening of austenitic steel [24]. This enhancement may be due to solute atoms suppressing dislocation motion and improving the deformation barrier [23,24].

Figure.3 The actual (solid lines) and predicted (dots) hot compression curves of 8Al steel under different deformation parameters: (a) 850 °C, (b) 950 °C, (c) 1050 °C, and (d) 1150 °C.

Figure.4 The actual (solid lines) and predicted (dots) hot compression curves of 10Al steel under different deformation parameters: (a) 850 °C, (b) 950 °C, (c) 1050 °C, and (d) 1150 °C.

In order to understand how the deformation conditions respond to the flow stress, the constitutive equations were developed. The power law equation is used to establish the relationships among flow stress, strain rate and deformation temperature at the high temperature deformation conditions [25]:

$$\dot{\epsilon} = A \sigma^n \exp[-Q/RT] \quad (1)$$

where $\dot{\epsilon}$ is the strain rate (s^{-1}), A , n , α and β are the materials constants, $\alpha = \beta/n$, R character denotes the gas constant ($8.314 J \cdot mol^{-1} K^{-1}$), Q as the activation energy for deformation (KJ/mol), T as the absolute temperature (K). Taking the logarithm of Equation 1 gives [26]:

$$\ln[\sinh(\alpha\sigma)] = -\frac{1}{n} \ln A + \frac{1}{n} \ln \dot{\epsilon} + \frac{1}{n} \frac{Q}{RT} \quad (2)$$

From Equation 2, the stress exponent n can be calculated as[27]:

$$n = \frac{\partial \ln \dot{\epsilon}}{\partial \ln \sigma} \quad (3)$$

Figure 5 shows the variation of $\ln \dot{\epsilon}$ with $\ln \sigma$ for different deformation temperatures of 8Al and 10Al, it can be seen that the peak stress decreases with the decreasing of strain rate and the

increasing of deformation temperature. This is because the lower strain rate allows enough time for energy accumulation and dislocation annihilation. However, the higher deformation temperature favors the nucleation and growth of DRX grains. [28, 29]. The n value is given by the slope of $\ln \dot{\epsilon}$ versus $\ln \sigma$.

$$Q = nR \frac{\partial \ln \sinh(\alpha \sigma)}{\partial (1/T)} \quad (4)$$

By using Equation 4 [30], the activation energies for the hot deformation Q are 394 and 443 kJ/mol for the 8Al and 10Al steels, respectively. In general, Q value depends on the chemical constitution and increases with alloying element content. Hence, the observed effect of Al content is quite consistent with this trend. Taking natural logarithm on Equation 1 gives:

$$\ln Z = \ln A + n[\sinh(\alpha \sigma)] \quad (5)$$

The relationship of $\ln Z - \ln[\sinh(\alpha \sigma)]$ of 8Al and 10Al steels is illustrated in Figures 6(a) and (b), respectively. On the basis of these figures, it is found that the correlation coefficients for the linear regression of $\ln Z$ and $\ln[\sinh(\alpha \sigma)]$ of 8Al and 10Al steels are 0.932 and 0.950, respectively, indicating that the plot presents a superior linear relevance between the flow stress and Z value.

Figure.5 Plots of $\ln \sigma$ versus $\ln \dot{\epsilon}$ for 8Al (a) and 10Al (b) steels.

Figure.6 Relation between Z parameter and peak stress: (a), 8Al; (b), 10Al.

Based on the calculation above, the constitutive equation for both steels can be expressed as:

8al:

$$\dot{\epsilon} = 8.396447 \times 10^{14} [\sinh(0.00604\sigma)]^{4.938272} \exp\left(\frac{-394145}{RT}\right)$$

10al:

$$\dot{\epsilon} = 3.574441 \times 10^{14} [\sinh(0.012545\sigma)]^{1.904762} \exp\left(\frac{-443413}{RT}\right)$$

In order to better predict the flow stress through the constitutive equations under different hot processing conditions, the material constants (i.e. β , n , Q , α and A) are calculated under various strains within the range of 0.05-0.6 at an interval of 0.05, which can be presented by a 5th-order polynomial relationships with true strains, as shown in Equation 6. On the whole, the predicted hot compression curves have a superior agreement with the measured hot compression curves (Figures 3 and 4).

$$\begin{aligned}\beta &= O_0 + O_1\epsilon + O_2\epsilon^2 + O_3\epsilon^3 + O_4\epsilon^4 + O_5\epsilon^5 \\ n &= P_0 + P_1\epsilon + P_2\epsilon^2 + P_3\epsilon^3 + P_4\epsilon^4 + P_5\epsilon^5 \\ \alpha &= X_0 + X_1\epsilon + X_2\epsilon^2 + X_3\epsilon^3 + X_4\epsilon^4 + X_5\epsilon^5 \\ Q &= Y_0 + Y_1\epsilon + Y_2\epsilon^2 + Y_3\epsilon^3 + Y_4\epsilon^4 + Y_5\epsilon^5 \\ \ln A &= Z_0 + Z_1\epsilon + Z_2\epsilon^2 + Z_3\epsilon^3 + Z_4\epsilon^4 + Z_5\epsilon^5\end{aligned}\tag{6}$$

The material constants (i.e. β , n , Q , α and A) of 8Al and 10Al are given as follows:

8Al:

$$\begin{aligned}\beta &= 0.08087 - 0.51467\varepsilon + 2.69855\varepsilon^2 - 7.011664\varepsilon^3 + 8.95072\varepsilon^4 - 4.43764\varepsilon^5 \\ n &= 7.44451 - 31.6679\varepsilon + 151.296\varepsilon^2 - 373.785\varepsilon^3 + 466.416\varepsilon^4 - 229.131\varepsilon^5 \\ \alpha &= 0.00967 - 0.04280\varepsilon + 0.22176\varepsilon^2 - 0.54050\varepsilon^3 + 0.62956\varepsilon^4 - 0.28187\varepsilon^5 \\ Q &= 426.378 - 328.173\varepsilon + 2564.99\varepsilon^2 - 9472.83\varepsilon^3 + 15689.6\varepsilon^4 - 9564.48\varepsilon^5 \\ \ln A &= 37.9376 - 54.6522\varepsilon + 352.23\varepsilon^2 - 1094.58\varepsilon^3 + 1624.34\varepsilon^4 - 913.598\varepsilon^5\end{aligned}$$

10Al:

$$\begin{aligned}\beta &= 0.0633 - 0.30796\varepsilon + 1.33216\varepsilon^2 - 2.70255\varepsilon^3 + 2.46741\varepsilon^4 - 0.71168\varepsilon^5 \\ n &= 4.99916 - 0.73567\varepsilon - 30.7180\varepsilon^2 + 150.105\varepsilon^3 - 277.249\varepsilon^4 + 182.793\varepsilon^5 \\ \alpha &= 0.00915 - 0.04234\varepsilon + 0.21274\varepsilon^2 - 0.50684\varepsilon^3 + 0.59129\varepsilon^4 - 0.26853\varepsilon^5 \\ Q &= 449.042 - 408.329\varepsilon + 1935.30\varepsilon^2 - 4237.79\varepsilon^3 + 2798.51\varepsilon^4 - 545.584\varepsilon^5 \\ \ln A &= 38.6196 - 29.4536\varepsilon + 140.390\varepsilon^2 - 299.497\varepsilon^3 + 145.218\varepsilon^4 - 107.056\varepsilon^5\end{aligned}$$

3.2. 3D Processing Map for Fe-Mn-Al-C steels

The processing map(PM) is improved based on the Dynamic Materials Model (DMM) by Prasad et al. [31]. In this model, the η character denotes the efficiency of power dissipation, which exhibits how efficiently the energy dissipates through microstructure change during hot deformation [32, 33].

η is defined as:

$$\eta = \frac{J}{J_{max}} = \frac{2m}{m+1} \quad (7)$$

As we all know, the field of superior workability is consistent with the higher $\xi(\dot{\varepsilon})$ value[31]. The m character denotes strain rate sensitivity parameter. When the temperature and strain rate are determined, the following formula can be used to calculate the m value:

$$m = \frac{\ln \sigma}{\ln \dot{\varepsilon}} = \frac{\partial J}{\partial G} = \frac{\dot{\varepsilon} \partial \sigma}{\sigma \partial \dot{\varepsilon}} \quad (8)$$

In order to reveal whether flow instabilities happen in the hot deformation, an instability parameter $\xi(\dot{\epsilon})$ is used to characterize the instabilities regions. $\xi(\dot{\epsilon})$ is calculated by Equation 9. Flow instabilities are forecasted to happen when $\xi(\dot{\epsilon})$ is negative. The instability map is drawn by the variation of values of instability parameter with temperature and strain rate.

$$\xi(\dot{\epsilon}) = \frac{\partial \ln(\frac{m}{m+1})}{\partial \ln \dot{\epsilon}} + m < 0 \quad (9)$$

Figure.7 Processing map of the 8Al (a) and 10Al (b) steel at true strains of 0.5. (Contour numbers reflect efficiency of power dissipation. Shaded region reflects flow instability).

Figure.8 The 3D power efficiency map of the 8Al (a) and 10Al (b) steel.

Figure.9 The 3D instability map of the 8Al (a) and 10Al (b) steel.

The 2D processing maps of extruded 8Al and 10Al steels at true strains of 0.5 are shown in Figures 7a and b, respectively, where the contour numbers reflect the power efficiency and the gray zone represents “instability”. The 3D power dissipation maps of 8Al and 10Al steels are shown in Figures 8a and b, respectively. The color of the rainbow describes the efficiency of power dissipation in the 3D power dissipation map. Just as one expected, the power efficiency varies with the variation of deformation temperature and strain rate. When the true strain is 0.5, the maximum efficiency of power dissipation for 8Al steel is about 46%, which is lower than that of 8Al steel (about 54%), and higher values of efficiency of power dissipation mainly exist in the areas of 950-1150°C/0.001-0.1s⁻¹, while higher values of 10Al mainly exist in the areas of 850-1050°C/0.001-0.01s⁻¹.

Figures 9a and b are the 3D instability map of the 8Al and 10Al steels, respectively. The green zone reflects “instability”, while the violet area is “stable”. For 8Al steel, as shown in Figures 7a and 9a, the proportion of instability zone is far less than that of stable region, it can be observed that the number of instability region shifts from two instability regions to one instability region. At stain of 0.5, where only one instability region can be recognized: the instability region exhibits at the temperatures of 850-950°C and at the strain rates of 0.001-0.1s⁻¹. For 10Al steel, as shown in Figures 7b and 9b, the proportion of unstable zone for all the strains is larger than 8Al steel. It also can be

noticed from Figures 9a and b that the number of instability region extends from one to two with increasing Al content, as shown in Figures 7 and 9. The first unstable zone exhibits at the temperatures of 850-950°C and at the strain rates of 0.001-10 s⁻¹; the second unstable zone exhibits at 1050°C and at the strain rates of 1-10s⁻¹. This result confirms that the high Al content reduces the hot workability of these steels.

3.3. Correlation between microstructure examination with processing map

Figure.10 The microstructure for the steels after deformed at 950 °C/0.01 s⁻¹: (a)8Al, (b) high magnification of (a); (c) 10Al, (d) high magnification of (c).

Figures 10a and b show the optical microstructure of 8Al steel acquired at 950 °C/0.01s⁻¹, consistent with the stability zone in the PM (Figures 7a and 9a). The initial hot rolling microstructure has been displaced by the recrystallized grain and substructures, as shown in Figure 10. A large fraction of fine DRX grains form along these grain boundaries, as shown in Figures 10a and b. Moreover, the wavy or serrated grain boundaries can be observed, confirming that grain boundary

expansion is the main mechanism of DRX nucleation. Such nucleation mechanism through strain-induced grain boundary migration can be usually found in the coarser grained structure [19, 34].

Figures 10c and d show the optical microstructure of 10Al steel acquired at 950 °C /0.01 s⁻¹, consistent with the instability zone in the PM (Figures 7b and 9b). It is found that the intense flow localization has occurred at the austenite grain boundary (Figure 10c), the detail observation on the flow localization is shown in Figure 10d, which indicates that they are reconstituted by the continuous band ferrite(CBF). The flow localization can easily lead to crack formation and accelerate crack propagation. The similar reports were made in 22Cr–19Ni–3Mo steel by[35, 36]. Moreover, the initial austenite grain boundaries produce distortion deformation, and fine nucleation of austenite DRX grain occurs at the initial austenite grain boundaries as shown in Figure 10d.

By comparison, it can be observed that the proportion of austenite DRX grains reduces with increasing Al content, and the grain size of austenite DRX grain becomes smaller in 10Al steel. There is no denying that DRX is strongly affected by the hot deformation parameters. The effect of deformation temperature and strain rate can be uniform to Z parameter[23]. The study also shows that the Z value increases with increasing Al content (as shown in Equation 11, the Z value is positively linearly related to the hot deformation Q value[30], Q are 394 and 443 kJ/mol for the 8al and 10al steels, respectively). The increasing Z value means increasing strain rate or decreasing deformation temperature, which leads to inhibition the proceeding of DRX. The behavior of the correlation coefficient implies that the nucleation and growth of DRX grains could be clearly suppressed with increasing the Al content.

$$Z = \dot{\epsilon} \exp\left(\frac{Q}{RT}\right) \quad (10)$$

Figure.11 The microstructure of 10Al steel after deformed at 850 °C with different strain rates:

(a) 10 s^{-1} , (b) high magnification of (a); (c) 0.001 s^{-1} , (d) high magnification of (c).

In order to illustrate the flow localization features and the related instability mechanisms, the microstructures of 10Al steel are observed under the different deformation parameters. The microstructures of 10Al steel deformed at $850^{\circ}\text{C}/10\text{s}^{-1}$ are shown in Figures 11a and b, which is consistent with the instability zone in the PM (Figures 7b and 9b). As is seen, a host of the deformation bands or deformation twins take place at austenite grains (Figure 11a), even appears cross deformation bands in the interior of the austenite grains (Figure 11b), these deformation structures are able to trigger easily flow localization[37]. When the strain rate drops to 0.001s^{-1} , consistent with the stability zone in the PM (Figures 7b and 9b), a great mount of new fine ferrite grains displayed in the initial austenite grain boundaries owing to the lower strain rate is favorable for the DRX proceeding [38] (Figures 11c and d).

The microstructures of 10Al steel deformed at $1150^{\circ}\text{C}/10\text{s}^{-1}$ are shown in Figures 12a and b,

consistent with the instability zone in the PM (Figures 7b and 9b). The fine austenite DRX grains are distributed in the initial austenite grain boundaries, which can be confirmed in Figure 12a. This reflects that the hot deformation of 10Al steel at $1150^{\circ}\text{C}/10\text{s}^{-1}$ is insufficient for complete DRX. It also can be seen that the intense flow localization features and the CBF extend along the deformation direction (Figure 12b), moreover, with increasing of deformation temperature, the initial austenite grain boundaries have a change from serrated grains to flattened grains as illustrated in Figures 10c and 12a. When the strain rate is lowered to 0.001s^{-1} , consistent with the stability zone in the PM (Figures 7b and 9b), it can be seen that the deformation microstructure has finished the DRX processing, and the ferrite and austenite DRX grain have both grown up, especially the austenite DRX grain size coarsened remarkably (Figure 12c). It also can be noticed from Figures 12b and d that the morphology and distribution of ferrite transform from CBF to discontinuous granular ferrite(DGF), moreover, the flow localization features cannot be observed with decreasing the strain rate.

To achieve a better understanding of microstructure evolution in the two steels during the hot deformation at different parameters, the schematic diagram of microstructure evolution is established as shown in Figure 13. For 8Al steel, it is seen that the initial hot rolling microstructure has been displaced by the recrystallized structure and substructures, as shown in Figure 13. For 10Al steel, the morphology and distribution of ferrite transform from the continuous band ferrite to discontinuous granular ferrite, and the diagram also indicates that austenite grain growth is more pronounced than ferrite at all temperatures. Moreover, the initial deformation microstructure cannot be observed with the increase of the temperature and decrease of the strain rate.

Figure.12 The microstructure of 10Al steel after deformed at 1150 °C with different strain rates: (a) 10 s^{-1} , (b) high magnification of (a); (c) 0.001 s^{-1} , (d) high magnification of (c).

Figure.13 Schematic diagram of microstructural evolution during hot rolling and hot deformation at different parameters.

4. Conclusions

We investigated the effect of Al content on the hot deformation behavior by the 3D processing map at the temperatures of 850-1050 °C and the strain rates of $0.001\text{-}10\text{ s}^{-1}$. The following conclusions

are presented on the basis of our findings:

(1). The high-Al steel has a higher flow stress and a higher activation energy of deformation (443 kJ/mol) than those of the low-Al steel (394 kJ/mol). This phenomenon results from the solute hardening and a decreasing the DRX grain size with increasing Al content. The constitutive equation of both steels was established:

8Al:

$$\dot{\varepsilon}=8.396447 \times 10^{14}[\sinh(0.00604\sigma)]^{4.938272}\exp(\frac{-394145}{RT})$$

10Al:

$$\dot{\varepsilon}=3.574441 \times 10^{14}[\sinh(0.012545\sigma)]^{1.904762}\exp(\frac{-443413}{RT})$$

(2). The unstable zone of 8Al exhibits at the temperatures of 850–950 °C and the strain rates of 0.001–0.1 s⁻¹ at true strain of 0.5. The number of unstable zone extends from one to two with increasing Al content, the first unstable zone exhibits at the temperatures of 850–950 °C and the strain rates of 0.001–10 s⁻¹, the second unstable zone occurs at the temperatures of 1050 °C and the strain rates of 1–10s⁻¹. The proportion of instability region for each of the strains increases with the increasing of the Al content. This phenomenon demonstrates that the high Al steel weakens the workability of the alloy.

(3). The microstructures of 8Al steel and 10Al steel depends on the deformation parameters. The initial hot rolling microstructure has been displaced by the recrystallized structure and substructures for both steels, while the unstable zone has deformation bands and flow localization.

(4). Increasing of Z value with increasing the Al content, the proceeding of DRX could be clearly suppressed. For 10Al steel, the morphology and distribution of ferrite transform from the continuous band ferrite to discontinuous granular ferrite, moreover, the flow localization features cannot be

observed with decreasing the strain rate.

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References

- [1] G. Frommeyer, U. Brux, P. Neumann, Supra-ductile and high-strength manganese-TRIP/TWIP steels for high energy absorption purposes, *Isij International* 43(3) (2003) 438-446.
- [2] J.D. Yoo, K.-T. Park, Microband-induced plasticity in a high Mn–Al–C light steel, *Materials Science and Engineering: A* 496(1-2) (2008) 417-424.
- [3] J.D. Yoo, S.W. Hwang, K.T. Park, Origin of Extended Tensile Ductility of a Fe–28Mn–10Al–1C Steel, *Metallurgical and Materials Transactions A* 40A(7) (2009) 1520-1523.
- [4] J.D. Yoo, S.W. Hwang, K.-T. Park, Factors influencing the tensile behavior of a Fe–28Mn–9Al–0.8C steel, *Materials Science and Engineering: A* 508(1–2) (2009) 234-240.
- [5] G. Frommeyer, U. Brux, Microstructures and mechanical properties of high-strength Fe–Mn–Al–C light-weight TRIPLEX steels, *Steel Research International* 77(9-10) (2006) 627-633.
- [6] D. Raabe, H. Springer, I. Gutierrez-Urrutia, F. Roters, M. Bausch, J.B. Seol, M. Koyama, P.P. Choi, K. Tsuzaki, Alloy Design, Combinatorial Synthesis, and Microstructure–Property Relations for Low-Density Fe–Mn–Al–C Austenitic Steels, *JOM* 66(9) (2014) 1845-1856.
- [7] Z.Q. Wu, H. Ding, H.Y. Li, M.L. Huang, F.R. Cao, Microstructural evolution and strain hardening behavior during plastic deformation of Fe–12Mn–8Al–0.8C steel, *Materials Science and Engineering: A* 584 (2013) 150-155.
- [8] Z.Q. Wu, H. Ding, X.H. An, D. Han, X.Z. Liao, Influence of Al content on the strain-hardening behavior of aged low density Fe–Mn–Al–C steels with high Al content, *Materials Science and Engineering: A* 639 (2015) 187-191.
- [9] S.S. Sohn, S. Lee, B.J. Lee, J.H. Kwak, Microstructural Developments and Tensile Properties of Lean Fe–Mn–Al–C Lightweight Steels, *JOM* 66(9) (2014) 1857-1867.
- [10] W.K. Choo, J.H. Kim, J.C. Yoon, Microstructural change in austenitic Fe–30.0wt%Mn–7.8wt%Al–1.3wt%C initiated by spinodal decomposition and its influence on mechanical properties, *Acta Materialia* 45(12) (1997) 4877-4885.
- [11] I. Gutierrez-Urrutia, D. Raabe, Influence of Al content and precipitation state on the mechanical behavior of austenitic high-Mn low-density steels, *Scripta Materialia* 68(6) (2013) 343-347.
- [12] Y. Han, D. Zou, Z. Chen, G. Fan, W. Zhang, Investigation on hot deformation behavior of 00Cr23Ni4N duplex stainless steel under medium–high strain rates, *Materials Characterization* 62(2) (2011) 198-203.
- [13] A. Mohamadizadeh, A. Zarei-Hanzaki, H.R. Abedi, Modified constitutive analysis and activation energy evolution of a low-density steel considering the effects of deformation parameters, *Mechanics of Materials* 95 (2016) 60-70.

- [14] A. Mohamadizadeh, A. Zarei-Hanzaki, A. Kisko, D. Porter, Ultra-fine grained structure formation through deformation-induced ferrite formation in duplex low-density steel, *Materials & Design* 92 (2016) 322-329.
- [15] A. Mohamadizadeh, A. Zarei-Hanzaki, H.R. Abedi, S. Mehtonen, D. Porter, Hot deformation characterization of duplex low-density steel through 3D processing map development, *Materials Characterization* 107 (2015) 293-301.
- [16] P. Zhou, Q. Ma, J. Luo, Hot Deformation Behavior of As-Cast 30Cr2Ni4MoV Steel Using Processing Maps, *Metals* 7(2) (2017) 50.
- [17] C. Sun, X. Zuo, Y. Xiang, J. Yang, Investigation on Hot Deformation Behavior and Hot Processing Map of BSTMUF601 Super-Alloy, *Metals* 6(3) (2016) 70.
- [18] Y. Han, S. Yan, B. Yin, H. Li, X. Ran, Effects of temperature and strain rate on the dynamic recrystallization of a medium-high-carbon high-silicon bainitic steel during hot deformation, *Vacuum* 148 (2018) 78-87.
- [19] G. Liu, Y. Han, Z. Shi, J. Sun, D. Zou, G. Qiao, Hot deformation and optimization of process parameters of an as-cast 6Mo superaustenitic stainless steel: A study with processing map, *Materials & Design* 53 (2014) 662-672.
- [20] X. Dai, B. Yang, Study on hot deformation behavior and processing maps of SA508-IV steel for novel nuclear reactor pressure vessels, *Vacuum* 155 (2018) 637-644.
- [21] B. Gong, X.W. Duan, J.S. Liu, J.J. Liu, A physically based constitutive model of As-forged 34CrNiMo6 steel and processing maps for hot working, *Vacuum* 155 (2018) 345-357.
- [22] Y. Sun, Z. Cao, Z. Wan, L. Hu, W. Ye, N. Li, C. Fan, 3D processing map and hot deformation behavior of 6A02 aluminum alloy, *Journal of Alloys and Compounds* 742 (2018) 356-368.
- [23] D. Li, Y. Feng, Z. Yin, F. Shangguan, K. Wang, Q. Liu, F. Hu, Prediction of hot deformation behaviour of Fe–25Mn–3Si–3Al TWIP steel, *Materials Science and Engineering: A* 528(28) (2011) 8084-8089.
- [24] H.J. McQueen, S. Yue, N.D. Ryan, E. Fry, Hot working characteristics of steels in austenitic state, *Journal of Materials Processing Technology* 53(1) (1995) 293-310.
- [25] A. Galiyev, R. Kaibyshev, G. Gottstein, Correlation of plastic deformation and dynamic recrystallization in magnesium alloy ZK60, *Acta Materialia* 49(7) (2001) 1199-1207.
- [26] M.R. Rokni, A. Zarei-Hanzaki, A.A. Roostaei, A. Abolhasani, Constitutive base analysis of a 7075 aluminum alloy during hot compression testing, *Materials & Design* 32(10) (2011) 4955-4960.
- [27] Y.-C. Lin, M.-S. Chen, J. Zhang, Modeling of flow stress of 42CrMo steel under hot compression, *Materials Science and Engineering: A* 499(1–2) (2009) 88-92.
- [28] C.M. Sellars, W.J. McTegart, On the mechanism of hot deformation, *Acta Metallurgica* 14(9) (1966) 1136-1138.
- [29] Y. Liu, C. Geng, Q. Lin, Y. Xiao, J. Xu, W. Kang, Study on hot deformation behavior and intrinsic workability of 6063 aluminum alloys using 3D processing map, *Journal of Alloys and Compounds* 713 (2017) 212-221.
- [30] M. Srinivasan, C. Loganathan, R. Narayanasamy, V. Senthilkumar, Q.B. Nguyen, M. Gupta, Study on hot deformation behavior and microstructure evolution of cast-extruded AZ31B magnesium alloy and nanocomposite using processing map, *Materials & Design* 47 (2013) 449-455.
- [31] Y.V.R.K. Prasad, H.L. Gegel, S.M. Doraivelu, J.C. Malas, J.T. Morgan, K.A. Lark, D.R. Barker, Modeling of dynamic material behavior in hot deformation: Forging of Ti-6242, *Metallurgical Transactions A* 15(10) (1984) 1883-1892.

- [32] K.S. Prasad, A.K. Gupta, A Constitutive Description to Predict High-temperature Flow Stress in Austenitic Stainless Steel 316, *Procedia Materials Science* 6 (2014) 347-353.
- [33] J. Zhang, H. Di, X. Wang, Y. Cao, J. Zhang, T. Ma, Constitutive analysis of the hot deformation behavior of Fe–23Mn–2Al–0.2C twinning induced plasticity steel in consideration of strain, *Materials & Design* 44 (2013) 354-364.
- [34] S.L. Semiatin, D.S. Weaver, R.C. Kramb, P.N. Fagin, M.G. Glavicic, R.L. Goetz, N.D. Frey, M.M. Antony, Deformation and recrystallization behavior during hot working of a coarse-grain, nickel-base superalloy ingot material, *Metallurgical and Materials Transactions A* 35(2) (2004) 679-693.
- [35] P. Cizek, Characteristics of shear bands formed in an austenitic stainless steel during hot deformation, *Materials Science and Engineering: A* 324(1) (2002) 214-218.
- [36] Z. Wang, W. Fu, B. Wang, W. Zhang, Z. Lv, P. Jiang, Study on hot deformation characteristics of 12%Cr ultra-super-critical rotor steel using processing maps and Zener–Hollomon parameter, *Materials Characterization* 61(1) (2010) 25-30.
- [37] H. Sun, Y. Sun, R. Zhang, M. Wang, R. Tang, Z. Zhou, Hot deformation behavior and microstructural evolution of a modified 310 austenitic steel, *Materials & Design* 64 (2014) 374-380.
- [38] A.S. Hamada, A. Kisko, A. Khosravifard, M.A. Hassan, L.P. Karjalainen, D. Porter, Ductility and formability of three high-Mn TWIP steels in quasi-static and high-speed tensile and Erichsen tests, *Materials Science and Engineering: A* 712 (2018) 255-265.