Processing maps for hot working of HSLA pipeline steel

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Abstract
The hot deformation features of HSLA pipeline steel are researched in the temperature from 800 °C–1100 °C and strain rate from 0.001–10 s⁻¹ using isothermal compression tests. According to the variation of power dissipation efficiency with the temperature and strain rate, the hot working process map of HSLA pipeline steel was obtained. The power dissipation map of the HSLA pipeline steel exhibits three domains. The first area exists in the temperature from 885 °C–1100 °C and strain rate from 0.01–1 s⁻¹. The second domain occurs in the temperature from 800 °C–860 °C and strain rate from 0.012–1 s⁻¹. The third area occurs in the temperature from 800 °C–825 °C and strain rate from 0.001–0.008 s⁻¹. When the specimen was deformed at higher temperature and lower strain rate, the efficiency of the power dissipation is more than 0.3, which is considered suitable for DRX to occur and may be considered a feasible condition for hot working.

1. Introduction

HSLA pipeline steel was developed in the early 1970s when there was an economic incentive to develop the methods and materials for corrosion resistance in steel [1, 2]. Large pipeline steel is often made from HSLA steel plates using thermomechanical controlled processing (TMCP) by controlling the hot deformation processes. TMCP provides a final fine-grained microstructure because of an austenite transformation to a non-equilibrium ferrite phase [3]. However, it is difficult to deform and strain HSLA pipeline steel a large amount in a single pass. The severe plastic deformation will produce heat which shall cause grain growth of the ferrite phase. Therefore, it is needed to consider the relationship between the hot working parameters and microstructural mechanisms to determine optimal hot processing parameters. A processing map is considered an effective tool to determine the optimal hot working parameters of alloy. Prasad et al [4] presented the processing map based on the dynamic material model (DMM) was often used to determine the hot workability of materials under various deformation parameters.

According to DMM theory, the dissipated power are associated with interior entropy production’s rate because of the metallurgical processes that are involved and the partitioning of the full power between that owning to the temperature increase and microstructural evolution is listed below:

\[ \eta = \frac{2m}{m + 1} \]  

(1)

Here m is the flow stress’ strain rate sensitivity. More explanation on constructing processing maps can be found elsewhere.

The instability represents the regions of the flow instability and was grown based on the irreversible thermodynamics which used to the big plastic strain flow as follows:

\[ \xi = \frac{\partial \ln (m/m + 1)}{\partial \ln \dot{\varepsilon}} < 0 \]  

(2)

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which \( \xi \) represents the instability parameter, which can be evaluated as a function of strain rate and temperature to get instability maps, where metallurgical instability during plastic flow occurs in regimes where \( \xi (\dot{\varepsilon}) \) is negative. Therefore, the deformation parameters have to select the ones where the \( \xi \) is larger than zero during the plastic deformation.

At present, many scholars have established processing maps to optimize the hot working parameters of alloys [5–7]. Aneta et al [8] investigated the hot deformation behavior of 4340 steel by isothermal compression tests, and optimized processing parameters of the investigated steel by the established processing maps. Gong et al [9] established a physically based constitutive model of the as-forged 34CrNiMo6 steel, and the optimum hot working parameters were determined based on the processing maps. Wu et al [10] studied the hot deformation behavior of Fe-Mn-Al-C steels by the 3D processing map at the temperatures of 850 °C–1050 °C and the strain rates of 0.001–10 s\(^{-1}\), and the study showed that the high Al steel weakens the workability of the alloy. Rastegari et al [11] investigated the flow behavior of vanadium microalloyed eutectoid steel and established a processing map for the steel. A safe processing window for the defect-free warm deformation of the steel was determined based on the processing maps. These reports have indicated that a processing map is an effective tool to optimize the hot working parameters and analyze the evolution mechanism of microstructure during hot deformation.

Previous reports on HSLA pipeline steel focused on the mechanical properties, welding properties and corrosion. However, few scholars focused on the establishment of a constitutive equation and optimization of hot working parameters for HSLA pipeline steel. In particular, research on the evolution mechanism of microstructures for HSLA pipeline steel during hot working is even rarer. Therefore, the aim of this paper is to study the hot workability and microstructure evolution mechanism of HSLA pipeline steel.

2. Experimental procedure

HSLA pipeline steel was prepared with the composition of 0.003 wt% C, 0.151 wt% Si, 0.76 wt% Mn, 0.517 wt% Ni, 0.014 wt% V, 0.0408 wt% Nb, 0.002 wt% Ti. The cylindrical specimens with \( \Phi 8 \times 15 \) mm were prepared. The isothermal compression experiment was carried out on a MMS-200 testing machine with a temperature scope of 800 °C–1100 °C and a strain rate scope of 0.001–10 s\(^{-1}\). The experimental procedure is shown in figure 1. The microstructure of the deformed specimens were observed and analyzed by JEM-1200 transmission electron microscope (TEM) and Electron back scattering diffraction (EBSD) examination was performed on a SEM (Hitachi S-3400N, Hitachi, Japan).

3. Results and discussion

3.1. Physical and thermo-physical properties of as-cast HSLA steels

The work of various previously published has demonstrated the ability of thermodynamic software to calculate the thermophysical properties of metals. The phase fractions of HSLA steel calculated by JMatPro software were illustrated in figure 2. According to the distribution of phase at temperatures from 100 to 1600 °C, it can be seen that the liquid phase begins to solidify into \( \alpha \)-ferrite at 1500 °C. After complete solidification, austenite was
formed at 1380 °C and gradually decreases owing to transformation from austenite to $\alpha$-ferrite when the temperature drops below 800 °C. It is noted that the HSLA steel which has equal phase proportions of ferrite and austenite at a temperature of approximately 820 °C in HSLA steel, and Martensite does not form in this case.

The crystalline phase has been investigated by XRD. Figure 3 reveals XRD pattern of HSLA specimens: three main peaks visible and correspond to the three orientations $\langle 110 \rangle$, $\langle 200 \rangle$ and $\langle 211 \rangle$ with detected as a single $\alpha$-ferrite phase. Therefore, XRD analysis shows that austenite and martensite transformation do not exist in the microstructure of steel.

3.2. Hot workability of HSLA steel

Figure 4 shows the true stress- strain curve of HSLA pipeline steel in the temperature from 800 °C–1100 °C and strain rate from 0.001–10 s$^{-1}$. As shown in figure 4, the isothermal compression process can be divided into three stages. In the early stage during isothermal compression (stage I), the flow stress increases rapidly with increasing strain. During the initial stage ($\varepsilon < 0.05$), dislocations multiply and accumulate rapidly with increasing strain, resulting in significant work hardening. Although dynamic recovery occurs at this stage, the dynamic recovery is too weak to counteract the work hardening effect \cite{12, 13}. Therefore, work hardening at this stage is the main reason for an increase in the flow stress with an increase in the strain. With further deformation (stage II), the flow stress increases slowly. In stage II, DRX occurs when the strain reaches a critical strain, and the
dynamic softening effect is gradually enhanced with increasing strain [14, 15]. Therefore, the work hardening effect is partially offset by the dynamic softening effect. When the strain increases further (stage III), the flow curves can be divided into two types: a dynamic recovery-type stress-strain curve, where the flow stress remains constant or increases slowly as the strain increases; and a dynamic recrystallization-type curve, the flow stress gradually decreases with additional strain until a steady state stress is reached. Dynamic softening plays a major role in this stage because dynamic recrystallization grain nucleation and growth can consume dislocations, resulting in a significant reduction in dislocation density. When the dynamic softening and work hardening reach equilibrium again, the flow stress remains stable.

The relationships between work hardening rate ($\theta$) and true stress at different strain rates are shown in figure 5. Figure 5 shows that the work hardening rate decreases rapidly with increasing stress in the early stages of deformation (stage I). With further deformation (stage II), the work hardening rate decreases slowly and then reaches zero at the peak stress ($\sigma_p$). When the work hardening rate reaches zero again (stage III), the corresponding stress value is called steady state stress ($\sigma_{ss}$). In addition, figure 5 shows that the peak stress decreases with increasing temperature or decreasing strain rate. The decrease of flow stress is related to DRX.
behavior. DRX can be carried out more fully at higher deformation temperature and lower strain rate. Thus, the dynamic softening effect is improved and the peak flow stress is reduced [16–19].

3.3. Development of strain dependent constitutive analysis

Constitutive laws for describing the inelastic deformation behavior of metals have been under development for the past four decades. Several researchers [20–22] have proposed phenomenological approaches to the construction of constitutive equations based upon micromechanics of plastic flow. An objective of these efforts aims to establish a concise relationship between flow behavior and microstructure. The practical feasibility of these sophisticated constitutive models with respect to accuracy and computational efficiency has not yet been established. Hence, empirical relationships based on the power law or logarithmic expressions are usually used to describe the high-temperature deformation behavior of materials.

Constitutive model is widely application for the modeling of hot deformation of metals at high temperatures and strain rates. The relationship between temperature, flow stress and strain rate can be expressed by the Arrhenius type constitutive equation. The expression of the equation is as follows:

$$\dot{\varepsilon} = Af(\sigma)\exp\left(-\frac{Q}{RT}\right)$$ \hspace{1cm} (3)

where $\sigma$ is the flow stress (MPa), $\dot{\varepsilon}$ is the strain rate ($s^{-1}$), $Q$ is the apparent activation energy (KJ.mol$^{-1}$), $T$ is the absolute temperature (K), $A$ is a material constant and $f(\sigma)$ is a stress function with the following expression:

$$f(\sigma) = \begin{cases} \sigma^{n_1} & \alpha\sigma < 0.8 \\ \exp(\beta\sigma) & \alpha\sigma > 1.2 \\ [\sinh(\alpha\sigma)]^n & \text{all } \sigma \end{cases}$$ \hspace{1cm} (4)

Where $\beta, n_1, \alpha$ and $n$ are material constants. The relationship between temperature and strain rate during hot deformation can be expressed by the Zener–Hollomon parameter [23, 24] as follows:

$$Z = \dot{\varepsilon} \exp\left(-\frac{Q}{RT}\right)$$ \hspace{1cm} (5)

Here $Z$ is the Zener–Hollomon parameter and $\alpha = \beta/n_1$ which is stress multiplier. At low ($\alpha\sigma < 0.8$), high ($\alpha\sigma < 1.2$), and all stress levels, substituting equations (4) into (3) with the appropriate function, and then taking the natural logarithms on both sides of the equation. Equations (6)–(8) can be obtained as follows:

$$\ln\dot{\varepsilon} = n_1 \ln\sigma + \ln A - Q/RT$$ \hspace{1cm} (6)

$$\ln\dot{\varepsilon} = \beta\sigma + \ln A - Q/RT$$ \hspace{1cm} (7)

$$\ln\dot{\varepsilon} = \ln A + n \ln[\sinh(\alpha\sigma)] - Q/RT$$ \hspace{1cm} (8)

For all stress levels, equation (31) can be rewritten as the hyperbolic sine law:

$$\dot{\varepsilon} = A[\sinh(\alpha\sigma)]^n \exp\left(-\frac{Q}{RT}\right)$$ \hspace{1cm} (9)

According to equations (6) and (7), the values of $n_1$ and $\beta$ can be calculated from the slopes of $\ln\dot{\varepsilon} - \sigma$ and $\ln\dot{\varepsilon} - \ln\sigma$, as illustrated in figures 6(a) and (b). The constant values obtained by using linear regression and mean values are as follows: $n_1 = 9.25$, $\beta = 0.089$ and $\alpha = \beta/n_1 = 0.0096$. For all stresses in equation (9), the slopes of the linear regression fits for the $\ln\dot{\varepsilon} - \ln[\sinh(\alpha\sigma)]$ plots can be calculated, as illustrated in figure 6(c). The average value of the stress exponent $n$ was calculated to be 6.79.

The activation energy ($Q$) can be obtained by using the slopes of the linear regression fits for the $\ln[\sinh(\alpha\sigma)]$ and 1000/T plots in figure 6(d). Therefore, the value of $Q$ can be calculated by equation (10). The value of $Q$ obtained from the experiment for HSLA steel is 278.435 kJ mol$^{-1}$.

$$Q = Rn\left[\frac{\partial \ln[\sinh(\alpha\sigma)]}{\partial(1/T)}\right]_{\dot{\varepsilon}}$$ \hspace{1cm} (10)

Combining equation (5) with equation (9) as results in:

$$Z = \dot{\varepsilon} \exp\left(-\frac{Q}{RT}\right) = A[\sinh(\alpha\sigma)]^n$$ \hspace{1cm} (11)

Take the natural logarithm of the two sides of the equation (11), equation (11) can be rewritten as follows:

$$\ln Z = \ln A + n[\ln[\sinh(\alpha\sigma)]]$$ \hspace{1cm} (12)

Figure 7 shown the linear-correlation between $\ln Z$ and $\ln[\sinh(\alpha\sigma)]$. The value of $\ln A$ is the intercept of the plot of $\ln Z$ versus $\ln[\sinh(\alpha\sigma)]$. Finally, the constant $A$ can be found averaging all the intercepts on the plot and the value is $A = 1.94 \times 10^{10}$ for temperatures above the solvus temperature.
In such a way, the material constants values \( (n, \beta, \alpha, \eta \text{ and } \ln \Lambda) \) and Q can be calculated for whole condition of HSLA steel. Table 1 displays the values of constitutive constants and Q of the steel. Then, substituting \( A, n, \alpha \text{ and } Q \) into equation (9), the constitutive equation for HSLA steel can be expressed:
The materials constants $\beta$, $\alpha$, $n$ and $\ln A$ are strongly affected by the strain. However, it is often believed that strain has a large effect on the flow stress at elevated temperatures. Therefore, strain compensation may have an important impact on the accuracy of flow stress prediction and be considered in an appropriate constitutive model. In order to establish a strain-compensated constitutive equation, the values of the material constants were measured at intervals of 0.05 in the true strain range of 0.05 to 0.55. The variation of material constants with strain is shown in figure 8. It can be seen that the relationship between the strain and the material constants can be well described by using a fifth-order polynomial fitting, as shown in equation (14). The polynomial fitting results for material constant coefficients of HSLA steel are listed in table 2.

$$
\dot{\varepsilon} = 1.94 \times 10^{10} \left[ \sinh(0.0096\sigma) \right]^{0.79} \exp \left( -\frac{278435}{RT} \right) \tag{13}\n$$

The materials constants $\beta$, $\alpha$, $n$ and $\ln A$ are strongly affected by the strain. However, it is often believed that strain has a large effect on the flow stress at elevated temperatures. Therefore, strain compensation may have an important impact on the accuracy of flow stress prediction and be considered in an appropriate constitutive model. In order to establish a strain-compensated constitutive equation, the values of the material constants were measured at intervals of 0.05 in the true strain range of 0.05 to 0.55. The variation of material constants with strain is shown in figure 8. It can be seen that the relationship between the strain and the material constants can be well described by using a fifth-order polynomial fitting, as shown in equation (14). The polynomial fitting results for material constant coefficients of HSLA steel are listed in table 2.

$$
\alpha = \alpha_0 + \alpha_1 \varepsilon + \alpha_2 \varepsilon^2 + \alpha_3 \varepsilon^3 + \alpha_4 \varepsilon^4 + \alpha_5 \varepsilon^5 \\
n = n_0 + n_1 \varepsilon + n_2 \varepsilon^2 + n_3 \varepsilon^3 + n_4 \varepsilon^4 + n_5 \varepsilon^5 \\
Q = Q_1 \varepsilon + Q_2 \varepsilon^2 + Q_3 \varepsilon^3 + Q_4 \varepsilon^4 + Q_5 \varepsilon^5 \\
\ln A = A_1 \varepsilon + A_2 \varepsilon^2 + A_3 \varepsilon^3 + A_4 \varepsilon^4 + A_5 \varepsilon^5 \tag{14}\n$$

From the hyperbolic law definition, flow stress shall be described as a function of the Zener-Hollomon parameter. In combination with the equation (3), the constitutive equation of high temperature deformation behavior to predict the flow stress shall be expressed by equation (15).
In order to verify the predictive ability of constitutive equation of HSLA steel, a comparative study of test and predicted values was conducted to verify the accuracy of the developed constitutive model in predicting the flow stress of HSLA steel during the hot deformation (figures 9(a)–(e)). It can be seen that the predicted flow stress values have the same trend and good consistency with the experimental values [25]. The predictive ability of the constitutive model can also be quantified by using standard statistical parameters, such as correlation coefficient (R) and average absolute relative error (AARE).

The correlation between experimental values and predicted values.

\[
\sigma = \frac{1}{\alpha} \left\{ \left( \frac{Z}{A} \right)^{1/n} + \left[ \left( \frac{Z}{A} \right)^{Q/n} + 1 \right]^{1/2} \right\} \tag{15}
\]

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The predictive ability of the constitutive model can also be quantified by using standard statistical parameters, such as correlation coefficient (R) and average absolute relative error (AARE).

\[
R = \frac{\sum_{i=1}^{N} (E_i - \bar{E})(P_i - \bar{P})}{\sqrt{\sum_{i=1}^{N} (E_i - \bar{E})^2(P_i - \bar{P})^2}} \tag{16}
\]

\[
AARE = \frac{1}{N} \sum_{i=1}^{n} \left| \frac{E_i - P_i}{E_i} \right| \times 100 \tag{17}
\]

Among them, \(E_i\) is the experimental value and \(P_i\) is the predicted value from the constitutive equation. \(\bar{E}\) and \(\bar{P}\) are the average values of \(E_i\) and \(P_i\), respectively. \(N\) is the total number of data used in this work.

The plot of tested values and predicted values obtained using the developed constitutive models are shown in figure 9(f). Obviously, most of the data points are very close to the line, and \(R\) and \(AARE\) for the constitutive models are 0.941 and 7.28%, respectively. The results also confirm the accuracy of the constitutive models with strain compensation of HSLA steel and the developed constitutive models can effectively predict the flow behaviour of steel that is used to simulate the hot forming process using Arrhenius constitutive modelling. Therefore, the analysis of the high temperature deformation process of this material is appropriate and reliable.

| α   | Value | n  | Value | Q   | Value | In A | Value |
|-----|-------|----|-------|-----|-------|------|-------|
| α₀  | 0.0161| 1  | 16.389| Q₀  | 472.447| A₀   | 39.216|
| α₁  | −0.0735| n₁ | −117.910| Q₁  | −3032.743| A₁   | −205.756|
| α₂  | 0.3797| n₂ | 645.922| Q₂  | 20 936.590| A₂   | 1436.707|
| α₃  | −1.0354| n₃ | −1844.799| Q₃  | 71669.361| A₃   | −5318.279|
| α₄  | 1.4319| n₄ | 2589.393| Q₄  | 120694.584| A₄   | 9802.797|
| α₅  | −0.7839| n₅ | −1409.23 | Q₅  | −79513.685| A₅   | −6964.102|

Table 2. Coefficient of fifth-order polynomial fitting curve for material constants.

Figure 9. Comparison of predicted and measured flow stress curves of HSLA steel at strain rate of (a) 0.001 s⁻¹, (b) 0.01 s⁻¹, (c) 0.1 s⁻¹, (d) 1 s⁻¹ and (e) 10 s⁻¹. (f) The correlation between experimental values and predicted values.
3.4. Microstructure and deformation mechanism

As shown in figure 10, the dislocation structure was observed after hot compression at a strain rate of 0.1–10 s$^{-1}$ with a high temperature of 1100 °C. The dislocation density increased, and a complex dislocation structure formed, as shown in the dark regions. The dislocation lines are entangled and some dislocation walls can be formed with high-density dislocation tangle [26] included subgrains (figure 10(a)). The morphology and phase structure of the deformed specimens were determined by TEM. Figures 10(b) and (c) shows a selected area electron diffraction (SAED) patterns of HSLA pipeline steel after compression at a deformation temperature of 1100 °C and strain rates of 1 and 10 s$^{-1}$, respectively. The selected area electron diffraction spots from the α-Fe matrix and dislocation area for the HSLA pipeline steel are marked accordingly.

3.5. Processing map

The processing map that is based on DMM is comprised of an instability map and a power dissipation map. The contour map represents power dissipation’s constant efficiency, and the grey and white regions represent the instability and stability domains, respectively. Figure 11 shows the processing maps of the investigated HSLA pipeline steel obtained in the temperature range from 800 °C–1100 °C with strain rates of 0.001–10 s$^{-1}$. The power dissipation map for HSLA pipeline steel exhibits three areas. The first area occurred in the temperature range from 885 °C–1100 °C with strain rates of 0.01–1 s$^{-1}$. The second area occurred in the temperature range from 800 °C–860 °C with strain rates of 0.012–1 s$^{-1}$. The third area occurred in the temperature range from 800 °C–825 °C with strain rates of 0.001–0.008 s$^{-1}$.

When the specimen was deformed at different temperatures at a strain rate of 0.1 s$^{-1}$, the microstructures appeared different. In the instability domains, dynamic recovery was the primary mechanism at moderate temperatures and moderate strain rates, as illustrated in figure 11(a). The related dissipation efficiency was 0.14, although the steady flow at a temperature of 800 °C still provided a small grain size. Figure 11(b) shows a heterogeneous microstructure at deformation temperatures of 900 °C. As the deformation temperature increases, the microstructure was more uniform and the number of regions that did not recrystallize decreased [27]. Although the recrystallized nuclei have been formed on the interfaces, the growth of these grains was slow while the specimen was deformed at 1000 °C, as demonstrated in figure 11(c).

It is generally known that DRX is often considered to take place for an efficiency value in the range of 30%–50% [28, 29]. From these results, the processing maps of the HSLA pipeline steel presented efficiency values between 30%–38% at a strain rate of 0.01–1 s$^{-1}$ and a temperature range of 975 °C–1100 °C, which indicated that dynamic recrystallization can occur in this condition and is very effective for hot forming processes. Therefore, high temperatures and low strain rates are considered suitable for DRX to occur [30, 31], consistent with the results in figures 11(d)–(f).

To provide more information for the HSLA pipeline steel processing maps and about the DRX mechanisms of the HSLA pipeline steel during hot deformation, an EBSD study was also conducted. Figure 11 shows the EBSD images of HSLA steel deformed at a strain rate of 0.1 at 1100 °C. Most of the grains are horizontally expanded in the direction of vertical compression. The inverse pole figure (IPF) map (in figure 11(d)) shows the change in crystal orientation within a grain. This owning to the grain misorientation from the plastic deformation, where certain grain orientations enabled crystalline slip. Moreover, at lower strain rates and higher temperatures, grain growth was predominant, and the size of grain was driven [20] by the mobility of the interfaces. Figure 11(e) shows the fractions of low-angle grain boundaries (LAGBs) along with high-angle grain boundaries (HAGBs) during the hot deformation of the experimental steels. The fractions of LAGBs long with HAGBs during the hot deformation of the experimental steels were found. In this case, the fraction of LAGBs was 38%, while the fraction of HAGBs was 62%, this finding indicates that the LAGBs are the typical
characteristic of substructure formation during hot deformation, whereas HAGBs are usually caused by DRX where grains are formed by the merging of other grains. The crystallographic texture can be represented as a stereographic projection by using pole figures are demonstrated in figure 10(f). The colors in the figure show the different orientations such as red, green and blue were indicated by the {001}, {110}, and {111} planes, respectively which exhibit random and homogeneous structure with no texture in the specimen. Thus, it is
noted that the pole figures are oriented such that the longitudinal direction of the bar is normal to the plane of the figure. This orientation was selected due to the convenience in describing the textured character.

4. Conclusions

The peak flow stress decreases with increasing temperature or decreasing strain rate. The strain compensated constitutive model of the investigated steel was established, and the predicted flow stress values have the same trend and good consistency with the experimental values. The processing map of HSLA pipeline steel with a strain of 0.5 was established, and the processing maps revealed that the optimal hot working parameters for this steel are at a deformation temperature of 975 °C–1100 °C and higher and a strain rate range of 0.01–1 s⁻¹.

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