1. Introduction

Hot rolling of steels is usually performed at a temperature as low as practically possible in the austenite phase range to ensure that the work rolls and mill stands are not overloaded, and to contribute to increased productivity and reduced energy consumption. For high temperature working of steels, the occurrence of recrystallization during deformation is referred to as dynamic recrystallization. Specifically, in low stacking fault energy materials, dynamic recrystallization occurs during hot working due to the formation of localized regions of high dislocation density that lead to the nucleation of strain-free grains which subsequently grow to eliminate dislocation density differentials. During deformation of austenite at an isothermal temperature, the characteristic form of the stress/strain curve at a constant strain rate illustrates a peak stress at a certain (peak) strain; this is an indication of dynamic recrystallization. The strain at the start of recrystallization (the 'critical' strain) is actually slightly lower than the peak strain because the peak strain is due to the work hardening being exactly offset by dynamic recrystallization. In other words, the strength decreases only when the level of dynamic recrystallization is sufficient to counter the work hardening. The critical strain ranges in most cases, between 0.65 to 0.8 $\varepsilon_p$ (where is the $\varepsilon_p$ strain to the peak stress), which has also been established from direct microstructural observation.\(^1\) An increase in strain past $\varepsilon_p$ results in decrease in the stress until steady state is reached ($\sigma_{ss}$). At this point, there is a balance between the growth of new strain-free grains and the work hardening of dynamically recrystallized grains.\(^1\)

2. Experimental Procedure

2.1. Materials

Four hypereutectoid steels, with chemical compositions as outlined in Table 1, were examined. The steels were prepared and cast by vacuum induction heating and soaking treatment for homogenization at 1 250°C during 48 h. These steels were hot rolled to wire of 13 mm of diameter and machined into compression specimens of height 11.4 mm and diameter 7.6 mm, leading to the height to diameter ratio of 1.5 that promotes homogeneous deformation during compression testing. (All materials were made and processed at POSCO.)

2.2. Experimental Equipment

All the mechanical compression tests were carried out on a computerized MTS machine (Model 840) adapted for high temperatures. Basically, this equipment consisted of a load frame rated for a maximum load of 100 kN capacity, a hydraulic power supply, and closed loop servohydraulic and computerized outer loop systems. A hydraulic actuator controlled by a servovalve generated the force and linear displacement. The force exerted on the specimens was measured by a load cell, while the linear displacement was monitored by a LVDT (linear variable differential transformer).

Table 1. Chemical compositions of specimen used in experimental.

|   | C  | Mn | Si |
|---|----|----|----|
| A | 0.81 | 0.3 | 0.19 |
| B | 0.9 | 0.29 | 0.19 |
| C | 1.01 | 0.3 | 0.21 |
| D | 1.08 | 0.3 | 0.20 |
The temperatures tested were attained with a Research Incorporated radiant furnace, interfaced with a computer control system, which was used to generate commands, record data and perform real-time decision making during tests.

During compression testing, thin sheets of mica (50–80 \( \mu \text{m} \) thick), separated by a layer of boron nitride powder, were placed between the face of the specimens and the anvils in order to maintain uniform deformation and avoid sticking problems during quenching. The specimen and the anvils were enclosed within a quartz tube, in which argon gas was passed to prevent oxidation of the specimen.

2.3. Single Hit Compression

These tests were conducted to determine the effect of carbon on the dynamic recrystallization characteristics. Most of the tests involved reheating at a constant rate of 1.5\(^{\circ}\text{C} \cdot \text{s}^{-1}\) to 1050\(^{\circ}\text{C}\), holding at 5 min, then cooling the test temperature (1050–800\(^{\circ}\text{C}\)) and then held for 3 min to homogenize the temperature within the specimen. The specimen was then deformed at strain rates of 0.01, 0.1 and 1\( \text{s}^{-1}\), followed by quenching.

3. Results and Discussion

3.1. Single Hit Flow Curves

The flow curves obtained for the steel D deformed to a strain of 0.7 were plotted for the deformation temperatures (800–1050\(^{\circ}\text{C}\)) at constant strain rate of 1\( \text{s}^{-1}\), as shown in Fig. 1(a), and the various strain rates at the deformation temperature 1050\(^{\circ}\text{C}\) for steel D, as shown in Fig. 1(b). All flow curves displayed a peak in the flow stress followed by work softening to a steady state region. This shape indicates that dynamic recrystallization is occurring. Either a rise in temperature or a decrease in strain rate (i.e. a decrease in the Zener–Hollomon parameter), with other parameters held constant, lowers the peak and steady state stresses. This decrease can be attributed to i) faster kinetics, for the higher temperature ii) more time spent in attaining the peak strain, in the case of decreasing the strain rate.\(^{4,5}\)

In Fig. 2, it can be seen that there is a slight but consistent difference between the flow behaviors of the four steels. For example, steel A shows a slightly higher steady state flow stress, and a slightly higher peak strain as compared with the other steels.

3.2. Hot Deformation Characteristics

Various empirical equations have been proposed to quantify the resistance to hot deformation. The one used most frequently is that due to Sellars\(^{6}\):

\[
\sigma_{\text{ss}} = A \times \dot{\varepsilon} \exp \left( \frac{Q_{\text{def}}}{RT} \right)
\]

Where \(\sigma_{\text{ss}}\) is the steady state stress, \(A\) is a function of the initial grain size and composition, \(\dot{\varepsilon}\) is the strain rate (\(\text{s}^{-1}\)), \(Q_{\text{def}}\) is the activation energy of deformation (\(\text{J} \cdot \text{mol}^{-1}\)), \(T\) is the absolute temperature (K), \(R\) is the gas constant (\(\text{J} \cdot \text{mol}^{-1} \cdot \text{K}^{-1}\)) and \(q\) is the power law exponent. As demonstrated in Fig. 3, the data follow this description.

In the present work, the stress exponent, \(q\), was deter-

![Fig. 1. Stress–strain curves for the steel D (a) different deformation temperatures at strain rate of 1\( \text{s}^{-1}\) and (b) different strain rates at deformation temperature 1050\(^{\circ}\text{C}\).](image)

![Fig. 2. Stress–strain curves for all steels at deformation temperature 850\(^{\circ}\text{C}\) and strain rate of 0.01\( \text{s}^{-1}\).](image)

![Fig. 3. Effect of temperature on the steady state stress at strain rates of 0.01 and 0.1\( \text{s}^{-1}\).](image)
and Birchenall.8) Mead and Birchenal8) have also found that the activation energy for iron self-diffusion decreases with increasing carbon content. However, some investigators have determined that the activation energy for hot deformation increases with increasing carbon content,11) but the difference in the activation energy for hot deformation increases with increasing carbon content.11) The present observations are in good agreement with those reported by Sellars and Whiteman2) and Elwazri et al. Also, the activation energy of deformation, $Q_{\text{def}}$, was then determined from the slope of the plot of the steady state stresses and the inverse temperature, as shown in Fig. 3. The activation energy for the four steels, is presented in Table 2. The results suggest that the carbon content in this carbon range has no effect on the activation energy. The present observations are in good agreement with those reported by Sakai and Ohasi and Mead and Birchenhall.5)

The values for the activation energy for four steels tested here and those obtained by other investigators are shown in Fig. 4. Sakai et al.7) and Collinson et al. have reported that the activation energy decreased with increasing carbon content. Additionally, Gatwick and Rosenqvist found that the activation energy for a low carbon steel (0.1% C) was higher than that of a high carbon steel (1.06% C). This was attributed to the lower diffusivity of the iron atoms at temperatures within the austenite phase region for the high carbon steel as compared to the low carbon steel. Moreover, Mead and Birchenall5) have also found that the activation energy for iron self-diffusion decreases with increasing carbon content. However, some investigators have determined that the activation energy for hot deformation increases with increasing carbon content,11) but the difference in the activation energy behavior was not related to hot deformation characteristics. As carbon is an interstitial in Fe and thus has a high diffusivity, it is expected that carbon would have minor effect on the high temperature dislocation behavior. However, since increasing the carbon content increases the diffusivity of carbon in iron, it is possible that increasing carbon could decrease any dislocation drag effect. On the other hand, increasing carbon content should then increase any carbon drag effect. This is a possible explanation for the divided opinion on the effect of carbon on the activation energy for deformation.

3.3. Critical Stress and Strain

It is possible to determine this critical strain from the flow curves. The method developed by Ryan and McQueen12) defines the critical strain as the strain at which the experimental flow curve deviates from the idealized dynamic recovery curve, based upon differences in the strain-hardening behaviors associated with dynamic recrystallization and dynamic recovery. The conventional strain hardening rate is obtained from the derivative of the flow stress and is given as

$$\theta = \left( \frac{\partial \sigma}{\partial \varepsilon} \right)$$

Extrapolation of the $\theta–\sigma$ curves to $\theta=0$ produces “idealized” curves corresponding to softening by dynamic recovery alone.

For the flow stress curves given in Figs. 1 and 2, the stress–strain data in the prior vicinity of the peak stress were used to calculate the values of the strain-hardening rate, $\theta$, according to Eq. (3), as shown in Fig. 5. The raw stress–strain data were smoothed using a 9th order polynomial regression; the smoothed curves were then differentiated with respect to strain to determine the strain-hardening rate. The variation in the derivative of the true stress can be several orders of magnitude larger than the mean value. In order to solve this problem, previously it was demonstrated that a smoothed stress–strain curve can be obtained by modeling the region of interest (the region of plastic deformation that encompasses the stress peak) as a 9th order polynomial. The critical stress and strain for the initiation of dynamic recrystallization cannot be determined directly from the flow stress–strain data. To define the critical strain precisely, it was necessary to determine the minimum points in the $–\delta\theta/\delta\sigma–\sigma$ curves. In this way the critical stress and strain were obtained from the flow stress–strain data.

The dependence of the peak stress and strain are plotted as a function of the inverse temperature in Fig. 6. The peak stress and strain depend linearly on $1/T$. It can be seen that the peak strain of all four steels increase with decreasing deformation temperature. The slopes of peak strain and...
peak stress–1/T plots for these steels are approximately equal. The dependence of the peak stress and strain on temperature and strain rate confirm that the deformation is controlled by a thermally activated process.17,18)

The dependence of the peak strain and stress on the carbon content is presented in Fig. 7. The peak strain and stress appear to be little affected by changes in composition. It can be seen that the peak stress of steel D (high carbon) is lower than that of steel A. This suggests that there is no problem to rolling the hypereutectoid steels.

A comparison of the critical stress for the start of dynamic recrystallization and steady state stress is made in Fig. 8. It can be seen that the critical stress for the start of dynamic recrystallization is almost the same as the steady state stress. Nominally, the critical stress for the start of dynamic recrystallization and the steady state stress both indicate the upper limit of the amount of energy which can be stored in the material above which the energy is released by dynamic recrystallization, as shown in Fig. 9. However, this will vary with grain size since the amount of strain required to trigger dynamic recrystallization depends on the nucleation rate of dynamic recrystallization. The grain size prior to recrystallization will be presumably larger than the dynamically recrystallized grain size, thus suggesting a critical stress that would be different from the steady state stress. However, this is not the case. It may be that the grain size prior to dynamic recrystallization and at steady state is the same, or that match between critical and steady state stress is coincidental. More work is required to examine
4. Conclusions

(1) The steel A with the lowest C content shows a slightly higher steady state flow stress as compared with the other steels contain higher C content.

(2) The results suggest that the carbon content has no effect on the activation energy of deformation.

(3) The peak stress of steel D (high carbon) is lower than of steel A. This suggests that there is no problem to rolling the hypereutectoid steels.

(4) Comparison of the critical stress for the start of dynamic recrystallization and steady state stress shows that the critical stress for the start of dynamic recrystallization is almost same as the steady state stress. This was explained by i) internal energy and ii) grain size.

Acknowledgments

The authors are thankful to Professor J.J. Jonas for his beneficial discussions.

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