



On the opposition of dynamic recrystallization and solute dragging in steels



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ARTICLE INFO

Article history:

Received 25 April 2014

Received in revised form 22 September 2014

Accepted 25 September 2014

Available online 20 October 2014

Keywords:

Metals and alloys

Mechanical properties

Grain boundaries

Solute drag

ABSTRACT

The influence of chemical composition on the hot deformation behavior and dynamic recrystallization was studied by conducting hot compression tests on two low alloy steels, AISI 4135 and VCN200, over a temperatures range of 1000–1150 °C and at strain rates of 0.001 s⁻¹ to 1 s⁻¹. The activation energy of dynamic recrystallization in 4135 and VCN200 was determined as 374 kJ/mol and 435.3 kJ/mol, respectively. The different apparent activation energies (about 17%) were attributed to about 30% difference between the carbon equivalents of the steels. The results confirmed that the higher the alloying elements, the higher the peak stress and strain of DRX flow curves. The results were associated with the dragging force of solute atoms on the grain boundaries. A new formula was proposed as a chemical factor to quantify the influence of solute dragging on the dynamic recrystallization behavior. Simple power equations described how the peak stress and strain depends on the Zener–Hollomon parameter. The material constants of the developed equations were related to the proposed composition factor. Unlike to dynamic recrystallization, dynamic recovery did not delayed by solute dragging.

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1. Introduction

Dynamic recrystallization (DRX) is the major microstructural phenomenon during hot deformation of different alloys [1–5]. In materials prone to DRX, dynamic recovery (DRV) is sluggish due to low stacking fault energy [6,7] and therefore the buildup of dislocation density during hot deformation gradually reaches a critical value for the nucleation of new grains [8,9]. DRX is not only very interesting from scientific point of view, but is very essential to the proper design of industrial hot working operations as well. In general, the role of softening mechanisms (DRX or DRV) during or after hot deformation is to reduce work hardening through the annihilation of deformation energy and thereby decreasing the required load for shaping and preventing the flow instability and cracking of the work piece. For case of plain Carbon or low alloy steels, different variants of hot rolling such as conventional controlled rolling (CCR), recrystallization controlled rolling (RCR) and dynamic recrystallization controlled rolling (DRCR) have been developed based on the concept of restoration processes [10–13]. The static recrystallization (SRX) plays an important role in the CCR and RCR processes and DRX is the controlling mechanism during DRCR process and some other operations such as hot forging

and hot extrusion. DRX not only decreases the deformation loads on the forming apparatus, but also impedes the initiation and propagation of cracks and premature fracture before reaching the final shape [14,15].

The extent of DRX in different alloys depends strongly on their alloying composition. The alloying elements that are usually added to improve strength or corrosion resistance may degrade the softening kinetics during or after hot deformation. This is associated with the segregation of solute atoms to the high angle grain boundaries (HAGBs), during hot deformation or through heating before deformation. The affinity between solute atoms and the boundaries results to an interaction termed as “solute drag”. The topic of solute dragging has drawn much attention so far [16–20]. Even though the solute atoms cannot effectively pin the boundaries, they can decrease the mobility of HAGBs [16]. This, in turn, degrades the potential of DRX that strongly depends on the mobility of HAGBs [21]. Gall and Jonas investigated the flow behavior of a Ni–S alloy and found out that the steady state stress increases with increasing the concentration of solute atoms [22]. In another research, Cram and his co-workers [20] confirmed the same results of Gall and Jonas in Cu–Sn alloys and developed a physical model to describe the influence of solute dragging. Although many attempts have been devoted to study or model the underlying phenomena of solute dragging, very few attempts have heretofore dealt with the influence of chemical composition on the kinetics of

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restoration processes, especially DRX [23–25]. In a series of publications, some authors tried to consider the influence of solute dragging on the kinetics of SRX in a wide variety of low alloy and plain carbon steels [26–28]. Despite very effective works carried out to understand the effect of solute dragging on softening processes, the practical data about the interaction with DRX in are very few or unpublished. Hence, the current research was conducted to provide information about the activation energy and kinetics of DRX at the presence of solute dragging in low alloy steels.

2. Materials and methods

The materials used in this investigation were VCN200 and AISI4135, medium carbon low alloy steels with chemical compositions given in Table 1. In order to study the influence of chromium, nickel and copper which are common alloying elements in low alloy steels the above materials were selected with nearly the same weight percents of carbon, manganese and Silicon.

Cylindrical compression samples of 10 mm in diameter and 15 mm in height were prepared according to the ASTM E209 standard. Some concentric grooves of 0.5 mm dept were machined on the flat surfaces to retain Graphite as the lubricant material between the anvils and specimens. An Instron 8502 testing machine, equipped with a fully computerized furnace, was used to perform the hot compression tests. Before testing, all the specimens were reheated at 1200 °C for 30 min to ensure that all alloying elements were in solution. Following reheating, the samples were cooled down rapidly to each deformation temperature and maintained for 10 min to acquire a uniform temperature before straining. Continuous hot compression tests were carried out to true strain of 0.7 in a temperature range of 1000–1150 °C with the intervals of 50 °C and at strain rates of 10^{-3} s⁻¹ to 1 s⁻¹.

3. Results and discussion

3.1. Flow analysis and activation energy

The representative flow curves of the studied steels at different deformation conditions are shown in Figs. 1 and 2. The flow curves of both steels are characterized by evident peaks that indicate the occurrence of DRX. The flow curves of Fig. 1 illustrates that the flow stress level of both steels increases as strain rate rises. In both steels, multi-peaks appear on the flow curves as strain rate decreases from 0.01 s⁻¹ to 0.001 s⁻¹. This is attributed to the change in the mechanism of DRX from cyclic to continuous nucleation with increasing strain rate [29]. At low strain rates and high temperatures (low Zener–Hollomon parameter, Z), the nucleation of DRX becomes completed before the start of the next cycle. This is due to the high tendency for recrystallization at low-Z conditions. Such a deformation condition gives rise to the cyclic flow behavior due to the cyclic nature of recrystallization. In contrast, when the materials are deformed at high strain rates and low temperatures (high Z) the nucleation cannot be completed before the start of the next cycle. This would lead to a partial nucleation at each cycle and a single-peak flow behavior. Similarly, Fig. 2 represents the influence of deformation temperature on the flow curves of the steels at constant strain rate of 0.01 s⁻¹. As expected, the flow stress level of both steels decreases by increasing temperature from 1100 °C to 1150 °C. The results of Figs. 1 and 2 are summarized as the delay in start of DRX with decreasing temperature and increasing temperature. This is implied because the position of the peak (peak stress, σ_p and peak strain, ε_p), that is considered as the starting point of DRX, moves to higher stress and strain as

temperature declines or strain rate rises. The results are reasonable because high strain rates or low temperatures increase the activation energy required for the movement of HAGBs and therefore the start of DRX needs a higher driving force that is provided at a peak with higher stress and strain. Despite both steels exhibit similar flow behavior under the hot deformation regimes, the higher strength of VCN200 can be attributed to its more alloying elements with respect to 4135.

It is well-known that solute atoms generally segregate to the grain boundaries at hot deformation temperatures and decrease their mobility [16]. Hence, the higher the solute atoms in VCN200, the lower, the tendency for grain growth during reheating. This may be the first reason for higher strength of VCN200 with respect to AISI 4135. In addition, the interaction of solute atoms with dislocations, especially those of interstitial atoms that is often termed as “Cottrell atmosphere” [30], should be also considered. However, the solid solution hardening does not seem to play a great role in this case, because the carbon contents of the steels are nearly the same.

Another difference in the flow curves of the steels is that DRX in VCN200 starts with a delay with respect to that in 4135 and the flow curves are characterized by wider peaks comparing to those of 4135. These differences can be also accounted for by accepting the solute dragging of Cr, Ni, Cu and Ti to HAGBs in VCN200.

Previous investigations on the mechanism of DRX have provided strong evidence that the local grain boundary migration is an indispensable step for the nucleation of new grains [31,32]. Hence, any factor that influence the mobility of HAGBs like solute dragging, affects the kinetics of DRX. The results obtained related to the flow curves can be verified quantitatively by calculating the required activation energies for the initiation of DRX in the steels. The apparent activation energy for the initiation of DRX is the required energy to reach the peak point of flow curves illustrated in Figs. 1 and 2. Therefore, it can be determined by describing the peak flow stress in terms of temperature and strain rate. Under hot working conditions, the dependence of peak stress on strain rate and temperature is often described by the hyperbolic sine function. It has been proved that this equation can be used over a wide range of low and high flow stresses. The equation can be written as follows [33]:

$$Z = \dot{\varepsilon} \exp\left(\frac{Q}{RT}\right) = A[\sinh(\alpha \cdot \sigma_p)]^n \quad (1)$$

where Z is the Zener–Hollomon parameter, Q stands for the apparent activation energy and α , A and n are the material constants. By doing some algebraic operations, the value of Q can be defined as follows:

$$Q = R \left[\frac{\partial \ln \dot{\varepsilon}}{\partial \ln \sinh(\alpha \sigma_p)} \right] \cdot \left[\frac{\partial \ln \sinh(\alpha \sigma_p)}{\partial (1/T)} \right] \quad (2)$$

Figs. 3 and 4 indicate the variations of peak stress with strain rate and the reciprocal of temperature, respectively. The average slopes of the plots in Figs. 3 and 4 suggest that, for both steels, peak stress depends more on temperature than on strain rate. This observation may be associated with the solute dragging phenomenon, because both the mobility of HAGBs and the diffusivity of solute atoms are strongly sensitive to temperature.

The value of n ($=1/S_{avg}$) in Fig. 3, was determined as 4.76 and 5, respectively for VCN200 and 4135. The adjustable parameter α was obtained as 0.014 to bring the curves in the most parallel and linear case. By using the values of n and α and the average slopes of the curves in Fig. 4 (that is equal to Q/nR) Q was determined as 374 kJ/mol and 435.3 kJ/mol for 4135 and VCN200, respectively. As expected, the values of Q confirm that DRX in VCN200 needs more energy to initiate than in 4135 and any reasoning should

Table 1
Chemical composition of low alloy steels used in this investigation.

Steel	Wt.%								
	C	Si	Mn	Cr	Ni	Mo	Cu	Ti	Fe
4135	0.3	0.4	0.6	1.2	–	0.3	–	–	Rem.
VCN200	0.29	0.38	0.55	1.97	2.2	0.34	0.34	0.06	Rem.

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