Grain Refinement and Superplastic Behavior in a Commercial Bearing Steel*

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Synopsis

An ultra-fine-grained SUJ-2 steel was obtained by applying one of thermomechanical processing routes developed in UHC-steel. This material showed the high elongation-to-failure values of 400~840 % at warm temperature between 650°C and 730 °C over a range of initial strain rates from 1×10^{-4} to $8 \times 10^{-4} \text{ sec}^{-1}$. The value of $m \simeq 0.33$ and the activation energy of 195 kJ/mol was obtained for the deformation process in this range.

From these results, it could be concluded that grain boundary sliding is not totally dominating the deformation process of the present material although superplastic flow plays an important role in the range of the present testing conditions.

I. Introduction

Since 1975 Sherby and his co-workers^{1,2)} have developed a variety of thermomechanical processing routes which makes ultrahigh carbon (UHC) steels containing 1~2.1 wt% carbon to have a very fine structure of spheroidized cementite particles in ferrite. The materials processed by these methods are of much interest as new steels since they are superplastic at warm temperature, and strong and ductile³⁾ at room temperature. Furthermore, it has been already demonstrated that this thermomechanical processing can be applied to provide superplasticity for some commercial high carbon steels such as 52 100 bearing steel.⁴⁾

In the present work, similar thermomechanical processing was applied to a SUJ-2 commercial bearing steel which is most widely used in Japan. The purpose of this paper is to show the behavior and the mechanism of superplastic flow of the processed SUJ-2 steel at warm temperature between 650 °C and 730 °C (in the ferrite plus cementite temperature range).

II. The Material and the Thermomechanical Processing

The SUJ-2 steel used in this work was provided in spheroidized-annealed condition in the form of 80 mm diameter bars. The chemical composition of the steel is shown in Table 1. The microstructure of this steel in the as-received condition is shown in

Table 1. Chemical composition of the steel. (wt%)

Element	C	Si	Mn	P	S	Cr	Mo	O
SUJ-2	0.75	0.29	0.32	0.015	0.014	1.29	0.02	0.0012

Photo. 1, which represents ferrite matrix of grain size 10~20 μm with spheroidized cementite particles of 0.5~2 μm . These bars were forged to billets having dimensions 55 mm $\times$ 50 mm $\times$ 1 000 mm at temperature around 1 100 °C. The blocks of about 50 mm length, cut from the billet were thermomechanically processed by the schedule shown in Fig. 1(a). The blocks were air cooled to 650 °C after heating at 1 150 °C for 1 hr, and then, isothermally rolled at 650 °C. This was carried out by rolling from the thickness of 55 mm to the final thickness of 4 mm (ϵ , true strain, of -2.62) by repeated roll passes (16 passes) of around 15 % reduction per pass with furnace reheats between passes. The rolled plates were then thermal cycled with the schedule shown in Fig. 1(b).

The scanning electron micrographs of the specimens in the as-rolled and the thermal cycled conditions are shown in Photo. 2. In the as-rolled condition, the structure mostly consists of fine-grained ferrite with fine cementite particles, but a small amount of pearlite are still observed. This is not a suitable microstructure to expect superplasticity. However, such a pearlitic structure is fully removed by the thermal cycling step and the structure becomes almost homogeneous. As shown in Photo. 2(b), the structure in the cycled condition consists of fine-grained ferrite (average grain diameter $\simeq$ 0.8 μm) containing fine cementite particles (average particle diameter $\simeq$ 0.3 μm).

In addition to such homogenizing the thermal cycling is believed to convert low angle boundaries

Photo. 1. Scanning electron micrograph of the steel in the as-received condition.

* Originally published in *Tetsu-to-Hagané*, **67** (1981), 2710, in Japanese. English version received February 24, 1981.
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Fig. 1. Schematic diagrams for the thermomechanical processing and the thermal cycling.

to high angle boundaries and thereby to enhance the superplastic properties.⁵⁾ In this work, the material in the cycled condition was used for the following tests.

III. Experimental Methods

Two types of tests were performed using an Instron-type universal testing machine accompanied by an infrared focussed radiation furnace. The tests were carried out in a flowing gas mixture (90 % N₂ + 10 % H₂) in order to prevent the oxidation of the specimens during testing. Figure 2 shows the dimensions of the tensile specimens used in the following tests.

1. Tensile Ductility Tests

To measure the tensile ductility, constant crosshead speed tests to fracture were carried out at several temperatures between 650 °C and 730 °C over a range of initial strain rates from 1 to 8 × 10⁻⁴ sec⁻¹.

2. Change-in-strain-rate Tests⁶⁾

The specimen was pulled isothermally at a low crosshead speed until a steady state flow stress was approached. Then the crosshead speed was changed repeatedly with only a small increment of strain at each change. In such a manner the flow stress-strain rate relation was established at each deformation temperature (650, 665, 680, 695, 710 and 730 °C).

IV. Results and Discussion

All of the specimens deformed to fracture are shown in Photo. 3. Elongation-to-failure obtained from these results are also shown in Fig. 3 as a function of strain rate. At each deformation temperature, elongation-to-failure increases with decreasing strain rate except for the case at 730 °C. At a given strain rate, elongation-to-failure increases with increasing deformation temperature. A maximum elongation-to-failure value of 840 % is found at 730 °C at initial strain rate of 2.8 × 10⁻⁴ sec⁻¹. At lower strain rate of 1.4 × 10⁻⁴ sec⁻¹, the value of elongation-to-failure decreases to 800 %. This may be attributed to the occurrence of grain growth due to the prolonged testing time at such a high temperature. In fact, as

Photo. 2. Scanning electron micrograph of the specimens.

Fig. 2. Dimension of tensile specimens.

seen in Photo. 4, the specimen tested to failure at this testing condition showed remarkable grain growth in gauge length.

The flow stress-strain rate curves obtained from the results of the change-in-strain-rate tests at various temperatures were shown in Fig. 4. These results show that the values of strain rate sensitivity exponent, *m*, is close to 0.33 (*n* ≈ 3, where *n* is stress exponent) in almost the whole range of the experiments except for the small region extending to high strain rates at low temperature where the value of *m* is close to 0.20 (*n* ≈ 5). The value of *m* ≈ 0.33 is comparatively lower than that obtained from the superplastic flow region of UHC-steel (typically *m* ≈ 0.5).⁶⁾ However, the large elongation-to-failure values (400 ~ 840 %) obtained in this study implies that superplastic flow plays an important role in the deformation process of the material.

It is well known that the superplastic flow rate of fine-grained materials, when grain boundary diffusion is rate controlling,^{7,10)} is given by the phenomenological equation:

$$\dot{\epsilon}_{sp,f} = A \frac{b D_{gb}}{d^3} \left(\frac{\sigma}{E} \right)^2 \dots\dots\dots (1)$$

where *A* ≈ 10⁸, *d* is the grain size, *b* is Burgers vector, *D_{gb}* is the grain boundary diffusion coefficient in the matrix phase of the superplastic material, *σ* is the creep stress and *E* is the unrelaxed dynamic Young's modulus. However, it is difficult to fit the present flow stress-strain rate data to this equation since the value of the stress exponent is close to 3. Hence, the activation energy for the plastic deformation process was tentatively calculated from these data assuming

Photo. 3.
Specimens of the SUJ-2 steel
after testing to failure.

Fig. 3. Elongation to failure as a function of strain rate.

the relation⁵⁾:

$$\dot{\epsilon} = K \left(\frac{\sigma}{E} \right)^n \exp \left(-\frac{Q}{RT} \right) \dots \dots \dots (2)$$

where K is constant, n is the stress exponent for creep, Q is the activation energy for plastic flow, R is the gas constant and T is the absolute temperature. From Eq. (2), the activation energy for plastic flow is defined as

$$Q_{|\sigma=\text{const}} = -R \frac{d \ln (\dot{\epsilon} E^n)}{d \left(\frac{1}{T} \right)} \dots \dots \dots (3)$$

or

$$Q_{|\dot{\epsilon}=\text{const}} = R \frac{d \ln \left(\frac{\sigma}{E} \right)^n}{d \left(\frac{1}{T} \right)} \dots \dots \dots (4)$$

The plots for determination of the activation energy corresponding to Eqs. (3) and (4), using $n=3$, are given in Figs. 5 and 6 respectively. In these plots, the temperature dependence of modulus, E , for the present steel is assumed to be the same as that for α -iron. The value of E for α -iron as a function of

Photo. 4. Scanning electron micrograph obtained from the gauge length of the specimen tested to failure at 730°C at initial strain rate of 1.4×10^{-4} sec⁻¹.

Fig. 4. Flow stress-strain rate relation for fine grained SUJ-2 steel.

temperature was given by Köster.¹¹⁾ As can be seen in these plots, an average activation energy for the deformation process is 195 kJ/mol. This is somewhat higher than that for grain boundary diffusion of iron in α -iron ($Q_{gb} \approx 170$ kJ/mol)⁶⁾ but lower than that for the lattice diffusion of iron in α -iron ($Q_L \approx 250$ kJ/mol).⁶⁾ The result is seen to be consistent with the fact that the obtained value of m (≈ 0.33) is smaller than the expected value of m (≈ 0.5) for ideal fine-structure superplasticity. As seen in Table 1, the carbon content (0.75 wt%) of the present steel is