The High Temperature Torsional Deformation of a 0.06% C Mild Steel*

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**Synopsis**
A low carbon steel was deformed by torsion under the conditions of true strain rates from about $10^{-3}$ to 30 sec$^{-1}$ and over the range 830°C to 1 200°C.

The temperature and strain rate dependences of the flow stress and the ductility of this steel were evaluated in this experimental range, and from the determination of its flow stress behavior and optical microstructural observations, the rate controlling mechanisms of the dynamic restoration process responsible for the occurrence of the steady state deformation were investigated.

The relation between the ductility (true strain to failure) and test temperature showed a well marked maximum in the case of higher true strain rates. At the strain rates below 5.77 sec$^{-1}$, the ductility was increased with increasing strain rate, but above this strain rate the opposite was the case. The dependence of the flow stress and ductility on experimental variables were discussed in connection with thermal softening mechanisms operating during high temperature deformation.

When the flow stress values of both the maximum and steady states for this material were plotted against the Zener-Hollomon parameter, Z, calculated by using the activation energy for self diffusion below a critical stress, which is approximately 10 kg/mm$^2$ in this case, a power law relationship was observed with a stress exponent of 6.53, and above this critical stress a steeper stress dependence occurred.

The dynamically recrystallized grain size of the specimens which were deformed at temperatures of both $\alpha\rightarrow\gamma$ and $\gamma$ ranges were measured by using the line intercept technique, from which the average recrystallized grain size $d$ was determined. When the reciprocal grain size $1/d$ for both $\alpha\rightarrow\gamma$ and $\gamma$ structures were plotted against the Zener-Hollomon parameter, a straight line was obtained for each case. Therefore, it was confirmed that the operative restoration process during high temperature deformation of this material was the dynamic recrystallization under the condition that the power law relationship was preserved, since there was a relationship between the recrystallized grain size and deformation parameters.

I. Introduction

Working of metals at high temperatures is adopted extensively in industrial processes, because a higher reduction can be given at low stresses without intermediate annealing. While a higher rate of working is more advantageous, it also increases the stress needed. The knowledge of the interrelation of flow stress, strain rate, and temperature is required to evaluate the optimum parameters for a given deformation process. A considerable amount of data on the high temperature, high strain rate deformation of metals is also necessary. However, at high temperatures of about 1 000°C and under a wide range of strain rates, relatively small numbers of authors investigated these interrelations in the ferrous materials on the scale of laboratory work.

The present investigation has following two major subjects.

One of them is to give fundamental information available in industrial hot working processes, by estimating the effects of experimental variables on two factors which determine the forming properties of a material; they are its strength and ductility which determine the maximum allowable deformation without risk of fracture. A hot torsion test of a 0.06% C steel was conducted under the conditions of the temperature at the vicinity of $\alpha\rightarrow\gamma$ transformation point and strain rates of about $10^{-3}$ to 30 sec$^{-1}$.

Secondly, the characteristics of high temperature deformation is an occurrence of the steady state deformation in which the work-hardening rate equilibrates with the dynamic restoration rate. In view of the various rate controlling mechanisms found in different temperature ranges during the high temperature steady state deformation of the ferrous materials, i.e., recovery at $\alpha$ and $\alpha+\gamma$ regions and recrystallization at $\gamma$ region, the dynamic restoration process responsible for the occurrence of the steady state deformation behavior of the 0.06% C steel was examined from various approaches. However, there are some objections about the validity of the occurrence of dynamic recrystallization as a restoration process for high temperature deformation. Recently, although the author reported the data to support the dynamic recrystallization in copper and $\gamma$-Fe, it is intended in the present work to re-examine the deformation behavior of the 0.06% C steel over a wider range of deformation conditions than have previously been employed to allow a more critical assessment of the explanation.

II. Material and Experimental Procedure

The test material was a 0.06% C mild steel and its billet was hot forged to solid bars of 30 mm$^3$, from which hollow torsion test pieces of inner diameter, $D_2$, 9.5 mm, outer diameter, $D_1$, 16 mm, and gauge length 10 mm, as shown in Fig. 1, were machined. The result of chemical analysis of the test material is given in Table 1. Since it was required to obtain a precise stress-strain relation, the test piece was machined in the hollow type in which plastic strain was homogenized. Table 2 gives the dimensions of each

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section of the test piece, the ratio of outer diameter to inner diameter, and true strain per an unit revolution.

Two testing machines giving slow speeds of 0.05 to 24 rpm and fast ones of 100 to 1500 rpm were used. Details of testing machines have already been described elsewhere. Heating of a test piece has been carried out by using a high frequency induction heating apparatus which could heat the specimen rapidly and up to higher temperatures than a conventional electric furnace. Test pieces were heated to the testing temperature at 200°C/min and held for at least 5 min before testing to reduce the temperature gradient of the gauge section. During this time they were free to move axially, but during testing axial movements were restrained. All tests were conducted under the positive pressure of nitrogen gas which greatly reduced, but did not completely eliminate the surface oxidation.

Shear stress, \( \tau \), and shear strain, \( \gamma \), were calculated by using the following relationships:

\[
\tau = \frac{12M}{\pi(D_1^2 - D_2^2)} \quad \text{(1)}
\]

\[
\gamma = \frac{D_1 + D_2}{4L} \phi \quad \text{(2)}
\]

where \( M \) is torque, and \( L \) is the gauge length of the test piece which was taken short to avoid buckling of the test piece at large strain, and \( \phi \) is twisting angle. It is necessary for comparison with tension and compression data, to convert the shear stress and shear strain to true stress and true strain. The true stress–true strain curves in this paper were calculated from the shear stress-shear strain data by using von Mises criterion (\( \sigma = \sqrt{3} \tau \) and \( \varepsilon = \gamma / \sqrt{3} \), where \( \sigma \) is the true stress and \( \varepsilon \) is true strain).

Experimental variables were the test temperature and strain rate. Since one of the aims of the present investigation is to examine the deformation behaviors at temperatures before and after the \( \alpha-\gamma \) transformation point which is shown in Fig. 2, four test temperatures were chosen from \( \alpha+\gamma \) range to \( \gamma \) range, which is shown in Fig. 2 and Table 3. The interval of test temperature was taken about 100°C from 830° to 1200°C. The range of available strain rate was limited to between \( 1.96 \times 10^{-3} \) to \( 28.87 \sec^{-1} \) in true strain rate (\( \dot{\varepsilon} \)), according to the lower and upper limits of the slow and high speed machines, respectively. Between these limits four speeds were added, and in total six speeds were used. The lower four speeds were attained in the low speed machine, and the other in the high speed machine. The ratio between any neighboring strain rates was aimed to be nearly constant. Table 4 shows the used true strain rate and test speed.

### III. Experimental Results

#### 1. A Series of Flow Stress Data

Typical examples of true stress–true strain curves obtained in hot torsion of 0.06% C steel are given in Fig. 3(a), as a function of strain rate at a constant temperature of 1200°C and in Fig. 3(b) as a function of temperature at a constant strain rate of \( 9.12 \times 10^{-1} \sec^{-1} \). General features of the type of stress-strain

![Fig. 1. Shape and dimension of test piece](image)

![Fig. 2. Constitutional diagram of Fe-C system, showing test temperatures and the belonging phase in 0.06% C steel](image)

![Fig. 3. Typical true stress-true strain curves in hot torsion of 0.06% C steel](image)

<p>| Table 1. Chemical composition of low carbon steel in wt% |
|------------------|--------|--------|--------|--------|--------|</p>
<table>
<thead>
<tr>
<th>C</th>
<th>Si</th>
<th>Mn</th>
<th>P</th>
<th>S</th>
<th>Al</th>
<th>N</th>
<th>Cu</th>
</tr>
</thead>
<tbody>
<tr>
<td>0.06</td>
<td>0.32</td>
<td>0.011</td>
<td>0.023</td>
<td>0.065</td>
<td>0.027</td>
<td>0.04</td>
<td></td>
</tr>
</tbody>
</table>

<p>| Table 2. Dimensions of hollow torsion test piece |
|------------------|--------|--------|--------|--------|</p>
<table>
<thead>
<tr>
<th>Outer dia. ( D_1 )</th>
<th>Inner dia. ( D_2 )</th>
<th>Gauge length ( L )</th>
<th>( D_2/D_1 )</th>
<th>True strain per rev.</th>
</tr>
</thead>
<tbody>
<tr>
<td>16.0</td>
<td>9.5</td>
<td>10.0</td>
<td>0.599</td>
<td>2.35</td>
</tr>
</tbody>
</table>

| Table 3. Test temperature (Tm: 1530°C) |
|------------------|--------|--------|--------|--------|
| T (°C) | 830    | 940    | 1060   | 1200   |
| T/Tm   | 0.61   | 0.67   | 0.74   | 0.82   |

| Table 4. Selected strain rate |
|------------------|--------|--------|--------|--------|
| Twisting speed (rpm) | \( 5.06 \times 10^{-2} \) | \( 3.49 \times 10^{-1} \) | \( 2.53 \) | \( 23.76 \) | \( 150 \) |
| True strain rate (s⁻¹) | \( 1.96 \times 10^{-3} \) | \( 1.34 \times 10^{-2} \) | \( 9.76 \times 10^{-2} \) | \( 9.12 \times 10^{-1} \) | \( 5.77 \) | \( 28.87 \) |
curves are to exhibit a marked peak in flow stress after initial work hardening region, and after that they show steady state deformation in which the work-hardening rate equilibrates with the dynamic restoration rate.

Figure 4 shows the relationships between maximum flow stress and strain rate in log-log plot, for each temperature, and the equation: \( \sigma_{\text{max}} = K \dot{\varepsilon}^m \), held for range of the experiment. The slope of the curve in Fig. 4 gives the strain rate sensitivity exponent, "\( m \)" value in the above equation. Values of the constant \( K \) and \( m \) in that equation are shown in Table 5 for each deformation temperature, and from this the resistance to deformation can be predicted by informing the strain rate in practical hot working processes.

The relation between the flow stress and homologous temperature is given in Fig. 5 in semi-log plot, for each test strain rate. It is observed roughly from this graph that the higher the strain rate, the smaller becomes the dependence of flow stress on temperature, although the slope of these curves shows only a little change and some irregularity with strain rate.

2. Ductility-Temperature Relation

The relation between the ductility (true strain to failure) and temperature in hot torsion of 0.06% C steel is shown in Fig. 6, for six different strain rates. In the range of strain rates from \( 1.96 \times 10^{-3} \) to \( 5.77 \) sec\(^{-1} \), the value of the ductility was increased with increasing strain rate, but the ductility at the highest strain rate, 28.87 sec\(^{-1} \), was smaller than that at the strain rate of \( 9.12 \times 10^{-1} \) sec\(^{-1} \). Although the ductility was increased with increasing temperature at low strain rate range, in the case of higher two strain rates, 5.77 and 28.87 sec\(^{-1} \), the ductility–temperature curves which have a peak point at 1 060°C (0.74 Tm) and 940°C (0.67 Tm), respectively, were obtained. From the above observation, it can be seen that the strain rate of 5.77 sec\(^{-1} \) exhibits the best ductility at each temperature except the highest one.

Fig. 3. Effect of strain rate (a) and temperature (b) on torsional stress-strain curve of 0.06% C steel, where shear stress and shear strain are converted to true stress and true strain by von Mises criterion

Fig. 4. The strain rate dependence of maximum flow stress

Table 5. Values of \( K \) and \( m \) for hot torsion of 0.06% C steel

<table>
<thead>
<tr>
<th>T (°C)</th>
<th>830</th>
<th>940</th>
<th>1060</th>
<th>1200</th>
</tr>
</thead>
<tbody>
<tr>
<td>( K ) (kg/mm(^2 ))</td>
<td>13.1</td>
<td>9.4</td>
<td>7.2</td>
<td>5.2</td>
</tr>
<tr>
<td>( m )</td>
<td>0.136</td>
<td>0.139</td>
<td>0.154</td>
<td>0.163</td>
</tr>
</tbody>
</table>

Fig. 5. Temperature dependence of maximum flow stress

Fig. 6. Temperature-ductility (true strain to failure) relation in hot torsion of 0.06% C steel
3. Interdependence of the Flow Stress, Strain Rate and Test Temperature

The several mathematical relationships which represent the correlation between the flow stress (\(\sigma\)), strain rate (\(\dot{\varepsilon}\)), and temperature (\(T\)) in the high temperature deformation have been proposed, but none of these have a strict theoretical foundation. In the present analysis, the data would be arranged by using the following power law relationship which is used for the low stress range such as creep deformation:

\[
\dot{\varepsilon} = A\sigma^n \exp\left(-\frac{Q}{RT}\right) \quad \quad \quad \quad (3)
\]

where, \(A\) and \(n\) are constants determined from the test data, \(R\) is its usual meaning, and \(Q\) is an activation energy for self diffusion. If the data fit to Eq. (3) reasonably, it can be deduced that the mechanism involves the diffusion of vacancies to climbing dislocations.

The rearrangement of Eq. (3) to the form

\[
Z = \dot{\varepsilon} \exp\left(\frac{Q}{RT}\right) = A\sigma^n \quad \quad \quad \quad (4)
\]

permits a correlation of the data for different temperatures on a single straight line, where \(Z\) is the Zener-Hollomon parameter. This type of plot provides a reliable method for interpolating data to obtain values of flow stress at any temperature or strain rate within the ranges studied.

The flow stress data obtained in the present work are plotted in Fig. 7, against the Zener-Hollomon parameter (\(Z\)) calculated using the activation energy for self diffusion shown in the figure.\(^9\) The form of the curves is similar for both the peak stress (\(\sigma_p\)) and steady state flow stress (\(\sigma_s\)) and the curves are similar to those found in a number of materials\(^8,10\) studied over a wide range of strain rate, namely, below a critical stress, in this case approximately 10 kg/mm², a power law is observed with a stress exponent of 6.53, and above the critical stress a steeper stress dependence occurs. The critical experimental conditions for changing the stress exponent are summarized in Table 6. Below these limits the stress exponent of 6.53 is obtained.

4. Metallographical Observations

In order to obtain the sample for metallographic examination, water jets were poured onto the gauge section of a test piece by opening a solenoid valve at the instance of interruption of deformation. And the gauge section of the test piece was machined by a wheel cutter to the sample for metallographic examination. Optical microstructures were observed on samples prepared as follows. Quenched and machined specimens of hot deformed 0.06% C steel were electrolytically polished on both transverse and longitudinal sections with a 90% glacial acetic acid+10% perchloric acid solution,\(^11\) and the inspection of the microstructures quenched from each temperature range was carried out by using the following methods. The microstructure after quenching from \(\sigma+\gamma\) range could be inspected by electrolytically etching the surface of the specimen to the state of the approximately mirror like, while in the observation of that quenched from the \(\gamma\) range, it was aimed to inspect the original austenite grain boundary as the ferrite structure at room temperature, in order to know the state of grain structure of austenite during deformation, and such the structure could be inspected by etching in nital (3% nitric acid+95% ethyle alchol) after the electrolytic polish.

Photograph 1 shows the microstructures after deformation up to the steady state region and then quenched from \(\sigma+\gamma\) range (830°C), for three strain rates. The calculation of the quantitative relation of the structural components which are obtained at room temperature in a 0.06%C steel quenched from 830°C show that the structure consists of approximately 95% ferrite and 5% martensite transformed from austenite. As observed in the photographs, the major part of the structure is ferrite and the minor part is martensite which is dark. The grain size became large with decreasing the strain rate or flow stress as usually, and according to the observation of these ferrite grains, it can be estimated that these structures are recrystallized ones, because these grains are equiaxed without any directionality although they are the deformed structure, and a small recrystallized grain is seen at grain

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Table 6. Critical condition of strain rate and equivalent flow stress for changing the stress exponent in Fig. 7.

<table>
<thead>
<tr>
<th>Limiting experimental conditions</th>
<th>True stress (kg/mm²)</th>
</tr>
</thead>
<tbody>
<tr>
<td>Limiting experimental conditions</td>
<td>1000</td>
</tr>
<tr>
<td>Temperature (°C)</td>
<td>830</td>
</tr>
<tr>
<td>Strain rate (sec⁻¹)</td>
<td>9.12 x 10⁻¹</td>
</tr>
<tr>
<td>Flow stress (kg/mm²)</td>
<td>12.1</td>
</tr>
</tbody>
</table>

\[
Z = \dot{\varepsilon} \exp\left(\frac{Q}{RT}\right) (sec^{-1}) \quad 5.85 \times 10^{12} \quad 2.49 \times 10^{12} \quad 1.12 \times 10^{12}
\]
boundary of greater grains for each strain rate.

Secondly, the microstructure of the specimen quenched from the $\gamma$ range after deforming up to the steady state region is shown in Photo. 2 in the relation of test temperature ($940^\circ - 1200^\circ$ C). In these structures, the austenite grain is decorated by ferrite grains and dark center part is the martensite structure. According to the estimation of these structures, it is considered that recrystallization occurs as a dynamic restoration process, because of the absence of directionality in these austenite grains in spite of the observations carried out on a parallel plane with the axis of the test piece.

Thus, according to the results of the observations of the deformed structure in $\alpha + \gamma$ and $\gamma$ ranges, respectively, it is concluded that the steady state deformation at the both temperature ranges gave the occurrence of recrystallized grains at the both test conditions that satisfy the power relationship in Fig. 7.

IV. Discussion

1. Flow Stress

The stress-strain relation gives a fundamental information on the plastic behavior of metals and alloys. The authors have already reported that the shapes of flow curves for steady state deformation in a hot torsion test could be classified into two types; that of aluminium and that of copper. It is considered to be due to the operation of dynamic restoration processes associated with the proceeding

<table>
<thead>
<tr>
<th>Strain rate</th>
<th>Temperature</th>
<th>Interrupted strain</th>
<th>Magnification</th>
</tr>
</thead>
<tbody>
<tr>
<td>(a) 9.76 x 10^{-2}</td>
<td>830°C (0.61 Tm)</td>
<td>0.439</td>
<td>300</td>
</tr>
<tr>
<td>(b) 1.34 x 10^{-2}</td>
<td>830°C (0.61 Tm)</td>
<td>0.418</td>
<td>300</td>
</tr>
<tr>
<td>(c) 1.96 x 10^{-3}</td>
<td>830°C (0.61 Tm)</td>
<td>0.181</td>
<td>300</td>
</tr>
</tbody>
</table>
of the deformation. With aluminium in which the dislocation climb is rapid, the initial work-hardened structure is already modified into an imperfect substructure by the time when the maximum stress is reached and a further deformation leads to sharpening of the subboundaries, followed by the migration when the boundary angle is large enough. Since the restoration is rapid, only a small peak, if any, is observed. This is an indication of dynamic recovery as the operative restoration process. Ferritic alloys belong to this type. In contrast, in copper and austenitic alloys, the dislocation climb is slower, the initial work-hardening cannot be removed sufficiently rapidly by polygonization, and enough strain energy is available to initiate recrystallization. Therefore, in these materials, the stress-strain curves have a marked peak due to the occurrence of softening during work hardening. According to the roundups of the above considerations, it is understood that the stress-strain curves for the aluminium type have not a peak whereas those of the copper type show a marked peak.

The true stress-true strain curves up to large strains in hot torsion test of the present 0.06% C steel have a marked peak as previously shown in Fig. 3. The behavior resembles the copper type as mentioned above. Therefore, it can be estimated that recrystallization is occurring as an operative dynamic restoration process in the 0.06% C steel for this temperature range from the observations of the shape of the stress-strain curve.

As mentioned previously, the characteristics of high temperature deformation is an occurrence of the steady state deformation. This steady state deformation is often treated as follows,10 by considering the exponent of the stress dependence of strain rate, \( n \) value. That is, the flow stress (\( \sigma \)) can be expressed as \( \sigma = a \cdot t \cdot D \), by assuming that the flow stress is generally a function of strain (\( \varepsilon \)) and time (\( t \)). Thus, at a constant temperature the exact derivative of flow stress (\( d\sigma \)) could be given as follows:

\[
\frac{d\sigma}{d\varepsilon} = \frac{\partial\sigma}{\partial\varepsilon} \frac{d\varepsilon}{dt} + \frac{\partial\sigma}{\partial t} \frac{dt}{d\varepsilon} = \frac{H}{R} \frac{dt}{d\varepsilon} = \frac{H}{R} \frac{d\varepsilon}{dt} \quad (5)
\]

At a steady state, since \( d\sigma = 0 \), the steady state strain rate (\( d\varepsilon/dt \)) can be expressed by following relationship:

\[
\frac{d\varepsilon}{dt} = -\frac{\partial\sigma}{\partial t} \frac{d\varepsilon}{\partial\varepsilon} = \frac{R}{H} = k \quad (6)
\]

where, \( R \) is \( -\frac{\partial\sigma}{\partial t} \), and represents the rate of dynamic restoration, and \( H \) is \( \frac{\partial\sigma}{\partial\varepsilon} \), and represents the rate of work hardening. Thus, the steady state strain rate is determined by the ratio of \( R \) and \( H \). If the work hardening and the restoration rate are expressed as \( H \propto \sigma^k \) and \( R \propto \sigma^r \cdot D \) \((D: \text{diffusion coefficient})\), and the strain rate can be given as \( \varepsilon \propto \sigma^{-k} \cdot D \). Therefore, the stress exponent of the power law relation as shown in Eqs. (3) and (4) can be expressed by \( n = r - k \).

In the previous work,12 this stress exponent was estimated in each metals and alloys in connection with the dynamic restoration mechanisms due to observations of the deformed structures, and the results are summarized in Table 7. From this, it is concluded that recovery occurred when \( n \) was smaller than 5, recrystallization occurred when \( n \) was larger than 6, and recovery and recrystallization occurred concurrently when \( n \) was between 5 and 6.

Applying this classification to the results of the present investigation, the stress exponent of 6.53 shown in Fig. 7 belongs to the range where the dynamic recrystallization is an operative restoration process. Thus, this is a part of the evidence that the present 0.06% C steel represents the occurrence of recrystallization as a dynamic restoration process within the range of the present experiments.

2. Ductility

Recently, many investigators have used the hot torsion test to assess the hot workability of ferrous materials. Tests in a range of temperature up to the melting point give a ductility-temperature curve exhibiting a peak point which corresponds closely to the optimum temperature for hot piercing and is slightly higher than the optimum temperature for hot forging and extrusion. Further, under the certain conditions, the true strain to failure can be used to assess the behavior of the materials in various working operations.

Reynolds and Tegart14 have concluded that the ductility data cannot be used as a fundamental parameter because it is dependent on numerous factors, e.g., the geometry of the specimen, rate of deformation, and atmosphere of test. However, they discussed that, for a given set of test conditions, the ductility data can be used to compare the influence of number of variables on mechanical properties, and furthermore, changes in ductility can be related to structural changes during deformation.

The ductility-temperature curves of the 0.06% C steel shown in Fig. 6 pass through a peak point at a particular temperature related to the test strain rate. The occurrence of this peak seems to depend on the temperature rise in the specimen due to an adiabatic heating in the higher strain rate deformation. Hughes15 has also observed a similar behavior to this in the ductility-temperature relation in his 0.10% C steel and he showed that in the specimens of higher carbon contents, the peak in ductility-temperature curve yields at about the same temperature for each test speeds used. The ductility in hot torsion test increases generally with increasing the strain rate and the temperature in the usual b.c.c. and f.c.c. materials, without considering the effect of the temperature rise due to the adiabatic heating. The curves for the present steel, Fig. 6, show that in the strain rate lower

<table>
<thead>
<tr>
<th>Material</th>
<th>Values of ( n )</th>
<th>Restoration process</th>
</tr>
</thead>
<tbody>
<tr>
<td>Al</td>
<td>4.0-4.3</td>
<td>dynamic recovery</td>
</tr>
<tr>
<td>( \alpha )-Fe</td>
<td>5.7</td>
<td>dynamic recovery and recrystallization</td>
</tr>
<tr>
<td>SUS 28</td>
<td>5.2</td>
<td></td>
</tr>
<tr>
<td>Cu</td>
<td>6.5</td>
<td>dynamic recrystallization</td>
</tr>
</tbody>
</table>
than 5.77 sec⁻¹ and below 1 060°C, increasing test speed causes increasing ductility. However, above this strain rate, the opposite is the case. Although the structural observation of the broken torsion test piece has not carried out in the present work, the internal cracking due to the occurrence of the red shortness is probably the main factor causing the downward displacement of the curves of Fig. 6 with increasing test speed. Once this cracking started, the prevailing stress conditions would probably determine the duration of deformation before fracture. In other words, the higher the test speed the sooner, in terms of deformation, the failure occurs.

As described previously, the variations of ductility can be related to the microstructural changes during high temperature deformation. In other words, the ductility is also controlled by the operation of the dynamic restoration processes. Thus, at lower temperatures in polycrystalline materials, the specimen fails at the maximum flow stress and the grains are heavily deformed with no sign of recrystallization and recovery. At high temperatures, however, dynamic restoration occurs and strained grains are replaced by new grains or subgrains, which can in turn be deformed further to produce an equilibrium between work-hardening and thermal softening processes. This onset of the dynamic restoration process leads to a marked increase in ductility so that the ductility-temperature curve shows a change in slope around 0.6 Tm.

The present results may be considered as followings in relation to the operative restoration process in each test condition. In the α+γ range, the low ductility is associated with the duplex structure which yields a massive oxide phase at α-γ interface during transformation, and with a stable substructure which prevents boundary migration; they promote intercrystalline fracture. At high temperatures in the α+γ range and γ range, the high ductility is associated with a recrystallized structure shown in Photos. 1 and 2, which developed by a process of oriented growth of subgrains formed during deformation.

3. Rate Controlling Mechanisms of Dynamic Restoration Process

Through the examinations of the mechanical properties and the microstructural observations, the evidence for the occurrence of dynamic recrystallization as a restoration process for high temperature torsional deformation of the 0.06%C steel has been submitted in the previous sections. Thus, in this section, the further evidence for dynamic recrystallization is provided by finding a relation between the mechanical properties which were described previously and the metallographical characteristics.

This was done by examining the correlation between the structure of the rapidly quenched specimen deformed up to the steady state and the deforming conditions which are temperature and strain rate. For this analysis, the dynamically recrystallized grain diameter (d) was measured by using the line intercept technique in the photographs of deformed structure obtained in each test condition. The reciprocal grain diameter (1/d) is plotted against Zener–Hollomon parameter (Z), according to the method by Sellars et al., in Fig. 8(a) and (b) for α+γ and γ regions, respectively. Here, the Z value of the material for α+γ region was calculated by using the activation energy for self-diffusion of α-Fe, because the ferrite structure occupied the major part in this material as mentioned previously in Section III. 4. The recrystallized γ grain size was measured by the original austenite grain boundary which is seen white in Photo. 2. From Fig. 8(a) and (b), it can be seen that there are a linear relationship between log Z and 1/d, in spite of some scatter for each structure; the recrystallized grain size is only determined by the temperature and strain rate of deformation uniquely. Thus, it can be seen that the recrystallized grain size becomes smaller at lowering temperatures and higher strain rates, and the dependence of the recrystallized grain size of the γ structure on the deforming conditions was smaller than that of the α+γ structure.

In the region where the power law relationship as described previously in Fig. 7 is shown with the stress exponent of 6.53, a linear relation between log Z and 1/d is not theoretically self-explanatory. However, it is confirmed that the dynamic restoration process of steady state deformation in the present 0.06%C steel is rate-controlled by the recrystallization process for the range of test conditions of the present investigation because there is a linear relationship between the recrystallized grain size which is observed by rapid quenching of the structures deformed up to the steady state region and the steady state flow stress, i.e., Z. According to the above analysis and the result of other authors, it can be seen that the recrystallized grain size can be predicted by Zener–Hollomon parameter (Z), which represents the deforming conditions.

4. Critical Strain for Dynamic Recrystallization

The strain to attain the maximum flow stress (εm)
and the strain to establish the steady state ($\epsilon_s$) in the stress–strain curves in the hot torsion test of the 0.06% C steel systematically change for the strain rate and temperature, respectively. Figure 9 shows the relation between the maximum stress and the strain to attain this stress, and it is seen that the strain on such a state increases with increasing the stress. The above fact shows that the initial work-hardening rate slightly decreases with increasing the test temperature and decreasing the strain rate.

Investigations of recrystallization during creep show that recrystallization is arised at a critical strain which is a function of the stress, temperature and purity. A similar critical strain should exist under constant strain rate conditions in hot torsion, although, since the stress has been continuously increasing in this case, the absolute values can not be comparable. Metallographic observations in the previous and other work after hot torsion indicate that at the maximum flow stress new grains appear in the region of the grain boundaries. Thus, the strain to the maximum flow stress should be closely related to the critical strain for dynamic recrystallization. This is also substantiated by the similarity in the form of the curves shown in Fig. 9 and those for the critical strain to recrystallization in creep. Furthermore, it is also simply thought that recrystallization is initiated at the strain to attain maximum flow stress because the flow stress begins to decrease at this strain.

According to the above discussion, when dynamic recrystallization occurs, the strain to attain maximum stress is closely coincident with the critical strain for the onset of recrystallization ($\epsilon_s$), and the rate of decrease in the flow stress after the maximum is determined by the rate of recrystallization. Thus, it is apparently significant to examine the dependence of these strains, $\epsilon_s$ and $\epsilon_m$ on the deforming conditions. The strain to the maximum stress or to the onset of steady state in stress–strain curves having no peak is plotted as a function of $Z$ in Fig. 10. The data fall into relatively wide bands within which there is some temperature dependence, for $\alpha+\gamma$ and $\gamma$ regions, respectively. The mean value of $\epsilon_m$ should lie roughly at the bottom of the band for $\epsilon_m$, Fig. 10, at low $Z$ values, but rapidly approach the mean value of $\epsilon_m$ at $Z \approx 10^{11}$ sec$^{-1}$ when $\epsilon_m \approx \epsilon_r$, where $\epsilon_r$ is the strain to the onset of steady state when only recovery occurs. This $Z$ value corresponds to the stress level for the change in stress–strain behavior. The relationship between $\epsilon_s$ and $\epsilon_r$ deduced from the data for the present steel, Fig. 10, is shown in the inset in that figure. This illustrates clearly that $\epsilon_s$ becomes much smaller than $\epsilon_r$ at low $Z$ values and there is a transition in the dynamic restoration process from recrystallization to recovery at $Z \approx 10^{11}$ sec$^{-1}$.

The strain occurring during the decrease in the stress from the maximum stress, $\epsilon_s$, measured as indicated in the inset in Fig. 11 is plotted as a function of $Z$ in Fig. 11 for the conditions exhibiting a distinct maximum point in the stress–strain curve. The data fall into bands for $\alpha+\gamma$ and $\gamma$ regions, respectively, with significantly steeper slopes than those for $\epsilon_m$ and within the bands there is no evidence of a systematic temperature dependence. The mean values of $\epsilon_s$ for $\gamma$ region are about twice those for the $\alpha+\gamma$ region over the whole range of $Z$. This indicates that the rate of dynamic recrystallization is higher in $\alpha+\gamma$ region.

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![Figure 9](image-url)

Fig. 9. Stress dependence of the strain to attain maximum flow stress for hot torsion of 0.06% C steel

![Figure 10](image-url)

Fig. 10. Dependence of strain ($\epsilon_m$) to attain the maximum flow stress or to the onset of steady state when no initial peak is observed on $Z$ for hot torsion of 0.06% C steel. Insert shows a schematic interpretation of the data in terms of the strain ($\epsilon_m$) to the onset of steady state if recovery were the only restoration process and of the critical strain ($\epsilon_r$) for the onset of dynamic recrystallization. Used activation energy $Q=57.2$ for $\alpha+\gamma$ region and 64.6 kcal/mol for $\gamma$ region

![Figure 11](image-url)

Fig. 11. Dependence of strain ($\epsilon_m$), measured as indicated in the inset, on $Z$ for hot torsion of 0.06% C steel
than in $\gamma$ region. The absence of temperature dependence of $\varepsilon$ when plotted against $Z$ thus indicates that the activation energy for recrystallization is close to the value for self diffusion used here ($a+\gamma$: 57.2, $\gamma$: 64.6 kcal/mole).

In Section IV, I, it was considered that the rate controlling mechanisms of dynamic restoration process could be classified into two types of aluminum and of copper. Thus, the change in relationship of $\varepsilon$ with deformation conditions observed here, Fig. 10, suggests that in the present experimental condition, the used 0.06% C steel may belong to the group of the copper type behavior described above. In aluminum type materials a relative increase in $\varepsilon$ with respect to $\varepsilon_0$ would lead to recovery as the only softening process under hot working conditions. Conversely, in copper type materials which have slower recovery rates, a relative increase in $\varepsilon$, with respect to $\varepsilon_0$ would be expected, leading to the occurrence of dynamic recrystallization during hot working to large strains. In these materials a transition in behavior similar to that observed here in Fig. 10, might also occur at sufficiently high values of $Z$.

V. Summary

Temperature and strain rate dependences of flow stress up to large strains and ductility which was estimated by true strain to failure were evaluated by hot torsion test under the conditions of homologous temperatures ranging from 0.61 Tm (830°C) to 0.82 Tm (1 200°C) and true strain rates from about $10^{-3}$ to 30 sec$^{-1}$ with hollow specimens of a 0.06% C steel. Then, the examinations of high temperature deformed structures were carried out after interrupting the test and quickly quenching the samples. The results obtained can be summarized as follows.

1) The type of stress-strain curves was the one of exhibiting a marked peak after an initial work hardening region, and after that the specimen showed a steady state deformation in which the stress was constant with increasing the strain in the whole test temperature ranges including the $a+\gamma$ range except for the case of the highest strain rate. It is a general feature of the materials in which the dynamic recrystallization as a restoration process occurs in the high temperature deformation.

2) Ductility-temperature curves exhibited a peak point at 1 060°C and 940°C for higher two strain rates of 5.77 and 28.87 sec$^{-1}$, respectively. From this, the temperature at which a peak appeared could be thought the optimum hot working temperature of this material in these strain rates.

3) There was a linear relationship between the dynamically recrystallized grain size and Zener-Hollomon parameter which is a function of the deforming conditions, for $a+\gamma$ and $\gamma$ ranges, respectively.

According to the above observations, the rate controlling mechanism of the restoration process could be thought as dynamic recrystallization.

4) In the temperature range of 830°C to 1 200°C the flow stress of 0.06% C steel was controlled by dynamic recrystallization which occurred once a critical strain was exceeded, when Zener-Hollomon parameter ($Z$) was smaller than a critical value. Above the value, dynamic recovery occurred at all strains. This transition in the restoration process occurred at the values of $Z$ less than 10$^{11}$ sec$^{-1}$, and this value corresponded to the critical values in which the stress-strain behavior changed.

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