Ductile Fracture of Carbon Steel
under Cold Metal Forming Conditions

(2nd Report, Effect of Metallurgical Structure)*

By Kozo OSAKADA**, Akihiro WATADANI***
and Hideo SEKIGUCHI****

Torsion tests under pressures of up to 4 kbars are carried out to investigate the mechanism of ductile fracture of 0.033 ~ 0.54% carbon steels of various microstructures. It is found that the strain at which fracture-nuclei begin to grow is affected by the carbon content, shape of cementite and hydrostatic pressure, whereas the strain from the beginning of nuclei-growth to fracture is not affected by these factors. A criterion for ductile fracture which accommodates the metallurgical factors is proposed. Microscopic observation of the fracture process of 0.25% carbon steel is carried out under a scanning electron microscope. The observed result is related with the initiation strain of nuclei-growth.

1. Introduction

The effects of microstructure on ductility of metals have been investigated by many researchers. It is well-known that the ductile-fracture strain of carbon steels decreases as the volume fraction of dispersed cementite or manganese-sulfide increases and that ductility increases when the cementite particles are spheroidized. Other metallurgical factors such as the dimension of pearlite colonies or of spheroidized cementite particles also give effects on ductility. Although some ductile fracture criteria have been proposed, the above facts are not included quantitatively.

In the previous paper, a method was proposed for determining the strain at which fracture-nuclei begin to grow; the environmental pressure is changed during deformation in tension and torsion tests. Moreover, the growth rate of fracture-nuclei was obtained and an experimental fracture criterion of 0.25% carbon steel was derived.

In this paper, the effects of micro-structure (volume and shapes of cementites) on ductility are studied by the use of the same method described in the previous paper. From the metallographic observations under a scanning electron microscope, macroscopical fracture behaviour, which is examined in torsion tests under pressures, is clarified from the viewpoint of microscopical fracture process.

2. Experimental procedures

Torsion tests under high pressures were conducted using the set-up described previously. Hollow specimens shown in Fig. 1(a) were used to obtain the fracture strain, and solid specimens, which were machined to have a long gauge-length, in Fig. 1(b), were used for microscopic observations. The surface of each specimen was polished carefully with emery paper No. 400, since the surface roughness may have serious effects on the fracture strain in torsion tests. The shear strain and shear stress in the deformation zone of the hollow torsional specimen are approximately uniform in the radial direction, and the specimens are used for measuring the stress-strain curves.

The materials used in this investigation are a commercial pure iron (CH 1) and five plain carbon steels of S15C, S20C, S25C, S50C and S55C. The chemical compositions and the volume fractions of cementite of these materials are shown in Table 1. Normal pearlitic structures are obtained by heat treatments as shown in Table 2; the microstructures after annealing are shown in Fig. 2.

* The values of volume fraction are calculated by assuming that the density of cementite is 7.74 g/cm³ and that the carbon content is fully precipitated as cementite.

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** Assistant Professor, Faculty of Engineering, Kobe University.
*** Postgraduate Student, Kobe University.
**** Assistant Professor, Nara Technical College.
3. Experimental results

3.1 Fracture in pearlitic structures

Fig. 3 shows the equivalent stress - equivalent strain curves for the annealed materials in torsion tests under a pressure of 4 kbars. The stress and the strain have been defined in the previous paper). It has already been confirmed that the hydrostatic pressure has almost no effect on the stress-strain curves of the materials used, except at the strains near fracture.

Fig. 4 shows the relationships between the fracture strain and the environmental pressure in torsion-test of the annealed materials. The fracture strain increases with increasing pressure, and the pressure dependence coefficients of the fracture strain, i.e. the slopes of the lines, are more or less the same.

Table 1 Chemical compositions and volume fractions of cementite of the materials used

<table>
<thead>
<tr>
<th>Material</th>
<th>Chemical Composition (wt. %)</th>
<th>Volume Fraction of Cementite γ</th>
</tr>
</thead>
<tbody>
<tr>
<td>CH 1</td>
<td>0.003 C 0.20 Si 0.18 Mn 0.063 P 0.010 S</td>
<td>0.005</td>
</tr>
<tr>
<td>S15C</td>
<td>0.18 C 0.27 Si 0.42 Mn 0.014 P 0.010 S</td>
<td>0.018</td>
</tr>
<tr>
<td>S20C</td>
<td>0.20 C 0.22 Si 0.49 Mn 0.023 P 0.011 S</td>
<td>0.027</td>
</tr>
<tr>
<td>S25C</td>
<td>0.25 C 0.23 Si 0.39 Mn 0.011 P 0.018 S</td>
<td>0.030</td>
</tr>
<tr>
<td>S50C</td>
<td>0.49 C 0.26 Si 0.77 Mn 0.020 P 0.014 S</td>
<td>0.074</td>
</tr>
<tr>
<td>S55C</td>
<td>0.54 C 0.23 Si 0.80 Mn 0.028 P 0.023 S</td>
<td>0.087</td>
</tr>
</tbody>
</table>

Table 2 Annealing conditions for pearlitic structures

<table>
<thead>
<tr>
<th>Material</th>
<th>Anneal. Temp. (°C)</th>
<th>Anneal. Time (min)</th>
</tr>
</thead>
<tbody>
<tr>
<td>CH 1</td>
<td>950</td>
<td>60</td>
</tr>
<tr>
<td>S15C</td>
<td>880</td>
<td>90</td>
</tr>
<tr>
<td>S20C</td>
<td>850</td>
<td>90</td>
</tr>
<tr>
<td>S25C</td>
<td>850</td>
<td>90</td>
</tr>
<tr>
<td>S50C</td>
<td>830</td>
<td>90</td>
</tr>
<tr>
<td>S55C</td>
<td>810</td>
<td>90</td>
</tr>
</tbody>
</table>

The strain at which fracture-nuclei begin to grow (the initiation strain of nucleigrowth), which has been defined in the previous paper, was measured by the following procedures: a specimen was deformed to a certain amount of pre-strain under atmospheric pressure, and then the pressure was changed to 4 kbars, under which the specimen was deformed until it fracture. Fig. 5 shows the total fracture strain as a function of pre-strain. Each experimental curve may be approximated by a horizontal line and a line having a downward slope. The value of pre-strain at the intersecting point of these two lines coincides with the initiation strain of nucleigrowth εi.

The initiation strain of nucleigrowth and the fracture strain under atmospheric pressure are plotted against the volume fraction of cementite, γ, in Fig. 6. Both the initiation strain and the fracture strain decrease as the volume fraction of cementite increases. The decreasing rates of these strains become smaller when the volume fraction exceeds about 0.04, which corresponds to a carbon content of 0.25%.

Fig. 3 Stress-strain curves for annealed specimens in torsion tests under 4 kbars

(a) CH 1  (b) S15C

(c) S20C  (d) S25C  (e) S50C  (f) S55C

Fig. 2 Microstructures of annealed steels

| 20μ |
The strain from the initiation strain of nuclei-growth (ei) to the fracture strain (ef) is defined as the growth strain for fracture-nuclei (eg): eg = ef - ei. The strain eg seems to be independent of the volume fraction of cementite (see Fig. 6).

3.2 Effect of shape of cementite on ductility

Spheroidized cementite is known to bring a larger ductility than lamellar pearlite; heat treatment for spheroidization of carbide particles is often adopted to increase the ductility of high carbon steels in cold forging. Therefore, the effect of shape of cementite on the process of ductile fracture was studied.

The particles of spheroidized cementite having various shapes and diameters were obtained by the heat treatments shown in Table 3. Quenching and tempering were combined to give the structures ①, ②, and the structure ③ was prepared by repeated tempering (see Fig. 7). Fig. 8 shows the micro-structures of these specimens.

The mean diameter of the cementite particles was defined by a parameter \( \alpha = \sqrt[3]{\text{shortest diameter} \times \text{longest diameter}} \), and the aspect ratio of the cementite particle was defined by \( \beta = \text{shortest diameter/longest diameter} \). Table 4 shows the values of \( \alpha \) and \( \beta \), which were obtained by measuring the diameters of 20 or more particles on micro-photographs for each specimen. The aspect ratio \( \beta \) is from 0.07 to 0.1 for pearlitic structures, and from 0.5 to 0.6 for well-spheroidized particles.

Fig. 9 shows the equivalent stress - equivalent strain curves for the spheroidized and the annealed materials under atmospheric pressure. In the case of S25C, the flow stress drops a little by spheroidization, but the flow stress is not significantly affected by the diameter of the cementite particles. In the higher carbon steel, S50C, the spheroidization of the cementite particles reduces the flow stress considerably.

Table 3: Heat treatments for spheroidization of cementite particles

<table>
<thead>
<tr>
<th>Material</th>
<th>Specimen</th>
<th>Quenching</th>
<th>Tempering</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td></td>
<td>Condition</td>
<td>Temp. (°C)</td>
</tr>
<tr>
<td>S25C</td>
<td>structure</td>
<td>oil a.</td>
<td>900</td>
</tr>
<tr>
<td>S25C</td>
<td>structure</td>
<td>oil a.</td>
<td>900</td>
</tr>
<tr>
<td>S25C</td>
<td>structure</td>
<td>water a.</td>
<td>900</td>
</tr>
<tr>
<td>S50C</td>
<td>structure</td>
<td>oil a.</td>
<td>900</td>
</tr>
<tr>
<td>S50C</td>
<td>structure</td>
<td>water a.</td>
<td>1000</td>
</tr>
</tbody>
</table>
Fig. 10 shows the relationships between the fracture strain and the environmental pressure for S25C and S50C. The fracture strain is increased by spheroidizing treatment; the pressure dependence coefficients are almost the same as those of the annealed materials.

To find the initiation strain of nuclei-growth under atmospheric pressure, torsion tests of the specimens of the structure ④ and ⑧ of S25C were carried out by the pressure changing method. Fig. 11 shows the relationships between the fracture strain and the pre-strain when the pressure is changed from 0 to 4 kbars. The result for the annealed S25C is also presented in the figure. The initiation strain of nuclei-growth varies with the degree of spheroidization, but the strain from the initiation to final fracture appears to vary very little. This shows that a larger fracture strain of a well-spheroidized specimen may be resulted from a larger value of the initiation strain of nuclei-growth.

Table 4 Mean diameters and aspect ratios of cementite particles

<table>
<thead>
<tr>
<th>Material</th>
<th>Specimen</th>
<th>Mean dia. (μm)</th>
<th>Aspect ratio β</th>
</tr>
</thead>
<tbody>
<tr>
<td>S25C</td>
<td>annealed</td>
<td>1.57</td>
<td>0.08</td>
</tr>
<tr>
<td></td>
<td>structure④</td>
<td>1.34</td>
<td>0.30</td>
</tr>
<tr>
<td></td>
<td>structure⑧</td>
<td>3.10</td>
<td>0.54</td>
</tr>
<tr>
<td></td>
<td>structure⑥</td>
<td>1.07</td>
<td>0.45</td>
</tr>
<tr>
<td>S50C</td>
<td>annealed</td>
<td>0.80</td>
<td>0.09</td>
</tr>
<tr>
<td></td>
<td>structure④</td>
<td>1.38</td>
<td>0.34</td>
</tr>
<tr>
<td></td>
<td>structure⑧</td>
<td>0.90</td>
<td>0.54</td>
</tr>
</tbody>
</table>

Fig. 9 Stress-strain curves for specimens of spheroidized cementite particles under atmospheric pressure.

Fig. 10 Relationships between fracture strain and environmental pressure (specimens of spheroidized cementite particles).
Fig. 12 shows the relationships between the fracture strain and the aspect ratio of cementite, \( \beta \), for S25C and S50C. The fracture strain increases linearly with the value of \( \beta \) for each material. Although the carbon content of S50C is almost twice as much as that of S25C, the difference in the fracture strain is not so large between these two materials and their pressure dependence coefficients are almost the same.

The fracture strains of the above data were plotted against the mean diameters of the cementite particles, but there was no typical relationship. This means that the fracture strain does not depend on the diameter of the cementite particle as long as the size of the particle is not too small.\(^2\)

![Fig. 11 Relationships between fracture strain and pre-strain when pressure is changed from 0 to 4 kbars (spheroidized and annealed structures of S25C)](image)

![Fig. 12 Relationships between fracture strain and aspect ratio of cementite in S25C and S50C](image)

4. Experimental fracture criterion

In the previous paper, a fracture criterion has been proposed as follows:

\[
f_{\text{ef}}^{*} < c - aP - b > d \varepsilon = c
\]

where \( a \), \( b \), and \( c \) are constant, and the function \( <\varepsilon> \) is defined by

\[
<\varepsilon> = \begin{cases} x & x \geq 0 \\ 0 & x < 0 \end{cases}
\]

The initiation strain of nuclei-growth, \( \varepsilon_{io} \), under a constant pressure \( P_0 \) (the hydrostatic component in kg/mm\(^2\)) is given by

\[
\varepsilon_{io} = aP_0 + b
\]

When a specimen is deformed under the constant pressure \( P_0 \) until it fractures, Eq. (1) can be rewritten as follows:

\[
f_{\text{ef}}^{*} < e - aP_0 - b > d \varepsilon = \frac{1}{2} (\varepsilon_{io} - \varepsilon_{go})^2 = \frac{1}{2} \varepsilon_{io}^2 = c
\]

where \( \varepsilon_{io} \) and \( \varepsilon_{go} \) are respectively the fracture strain and the strain for growth of nuclei under the pressure \( P_0 \). Equation (4) means that the strain for growth of nuclei is constant for a material regardless of the pressure, provided that it is deformed up to fracture under a constant pressure. Equation (4) is rewritten as

\[
\varepsilon_{io} = \varepsilon_{io} + \varepsilon_{go} = aP_0 + b + \sqrt{2c}
\]

Therefore, the pressure dependence coefficient, \( a \), of fracture strain is the same as that of the initiation strain of nuclei-growth.

Table 5 gives all the experimental values of the pressure dependence coefficients of the initiation strain and that of the fracture strain, the initiation strain of nuclei-growth, the strain for growth of fracture-nuclei, and the fracture strain, under the pressures of 0 and 2 kbars.

<table>
<thead>
<tr>
<th>Material</th>
<th>Pressure Dependence Coefficient ( a ) ((\varepsilon_{io}))</th>
<th>Pressure ( P_0 )(kbar)</th>
<th>Initiation Strain ( \varepsilon_{io} )</th>
<th>Growth Strain ( \varepsilon_{go} )</th>
<th>Fracture Strain ( \varepsilon_{ef} )</th>
</tr>
</thead>
<tbody>
<tr>
<td>CH 1 (A)</td>
<td>0.016</td>
<td>0</td>
<td>2.5</td>
<td>0.37</td>
<td>2.87</td>
</tr>
<tr>
<td>S15C (A)</td>
<td>0.016</td>
<td>1.5</td>
<td>1.0</td>
<td>0.33</td>
<td>1.83</td>
</tr>
<tr>
<td>S20C (A)</td>
<td>0.027</td>
<td>0.7</td>
<td>1.0</td>
<td>0.42</td>
<td>1.12</td>
</tr>
<tr>
<td>S25C (A)</td>
<td>0.02</td>
<td>0.6</td>
<td>2</td>
<td>0.10</td>
<td>1.0</td>
</tr>
<tr>
<td>S25C (S)</td>
<td>0.02</td>
<td>1.6</td>
<td>1.0</td>
<td>0.48</td>
<td>2.08</td>
</tr>
<tr>
<td>S25C (S)</td>
<td>0.02</td>
<td>2.1</td>
<td>1.5</td>
<td>0.44</td>
<td>1.94</td>
</tr>
<tr>
<td>S50C (A)</td>
<td>0.02</td>
<td>0.6</td>
<td>0.6</td>
<td>0.48</td>
<td>1.08</td>
</tr>
<tr>
<td>S50C (S)</td>
<td>0.02</td>
<td>1.0</td>
<td>0.6</td>
<td>0.48</td>
<td>1.15</td>
</tr>
<tr>
<td>S55C (A)</td>
<td>0.02</td>
<td>0.4</td>
<td>0.4</td>
<td>0.63</td>
<td>1.03</td>
</tr>
</tbody>
</table>

(4): Annealed (5): Spheroidized
The values of the pressure dependence coefficient of $\epsilon_p$ of pearlitic steels are 0.16 except for S20C, and are slightly larger than those of spheroidized steels. The mean value of the pressure dependence coefficients of $\epsilon_p$ is 0.020 and it agrees with the coefficient of $\epsilon_p$ for S25C. Since the pressure dependence coefficient of $\epsilon_p$ is not affected very much by carbon content and microstructure, we may assume the value to be a constant:

$$a = 0.02$$  \hspace{1cm} (6)

Almost all of the experimental values of $\epsilon_p$ drop in the range from 0.4 to 0.6, and the mean value is 0.50. The value of $c$ may also be assumed to be independent of the structures. Thus,

$$c = \frac{1}{2} (\epsilon_p)^2 = 0.12$$  \hspace{1cm} (7)

Since $a$ and $c$ in Eqn. (1) are the same for different materials, only the value of $b$ is affected by the structural factors such as the volume fraction of cementite, $\gamma$, and the aspect ratio of cementite, $\beta$. From the results shown in Figs. 6 and 12, we can derive the following experimental relation:

$$b = k \left( \frac{1 - \gamma}{\gamma} \right) + n \beta$$  \hspace{1cm} (8)

where $k = 0.1$, $m = 0.58$ and $n = 1.5$.

Thus, a criterion of ductile fracture is obtained as follows in which the effects of strain history, stress history and micro-structure are considered:

$$f_{\sigma}^e = \epsilon - 0.020 + 0.1 \left( \frac{1 - \gamma}{\gamma} \right)^{0.58} + 1.5\beta$$  \hspace{1cm} (9)

Fig. 13 shows the comparison between the experimental fracture strain and the calculated fracture strain from Eq. (9). The experimental results include all the data for pearlitic structures and spheroidized structures under various pressure histories. It can be said that the calculated values agree well with the experimental ones in spite of the approximations made to derive Eq. (9).

5. Observation of fracture process

Many metallographic observations have been conducted for various materials\(^8\)~\(^{11}\), but few attempts have been made to correlate the observed results with theoretical fracture criteria\(^2\). In this study, the physical meaning of the initiation strain of nuclei-growth is investigated to clarify the ductile fracture process.

Since voids and cracks were often observed at the outer surface or in the inside of materials\(^13\), the specimens of annealed S25C with pearlitic lamellar cementite were subjected to observation under a scanning electron microscope after torsion tests.

5.1 Inner cracks

Solid specimens, Fig. 1 (b), were metallographically examined at sections cut parallel to the longitudinal axis after torsion tests. Pearlitic structures were etched with 5% Nital. Fig. 14 shows typical inner cracks. Small cracks were formed along the boundaries or in the inside of pearlite colonies; the length of these cracks is more or less the same as the mean diameter of the colonies. Most of the observed cracks seem to be formed by fracturing of pearlite colonies, and others are caused by decohe-

![Figure 13](image1.png)

Fig. 13 Comparison between experimental fracture strain and theoretical result for various microstructures in torsion tests under different pressure histories.

![Figure 14](image2.png)

Fig. 14 Inner cracks in torsion-test at a strain of 1.78 under 4 kbars.
strain. When the strain exceeds about 0.5 under atmospheric pressure and about 1.2 under 4 kbars, the longest crack begins to grow quickly. These values of strain coincide well with the values of the initiation strain of nuclei-growth which can be calculated by Eqs. (3), (6) and (8); these are 0.6 under atmospheric pressure and 1.4 under 4 kbars.

The mean diameter of the pearlite colonies is about 6 μm in this specimen, and the large colonies have diameters of about 10 ~ 15 μm. Therefore, it can be said that the cracks stay in the pearlite colonies up to a strain at which they begin to grow rapidly into the ferrite matrix, and that the strain coincides with the initiation strain of nuclei-growth.

5.2 Surface cracks

The surfaces of solid specimens were polished with emery paper and etched electrolytically. They were then deformed to various strains under atmospheric pressure and 4 kbars. Fig. 17 shows a photograph of surface cracks observed by a scanning electron microscope. Most of the cracks at the surface are extended in the circumferential direction. When the specimen was deformed to a large strain, it was difficult to distinguish the cracks from the wrinkles of the surface.

Fig. 18 shows the relationships between the lengths of the longest cracks and strain. The surface cracks appear at a strain much smaller than the initiation strain of nuclei-growth. Although the growth rate of the surface cracks is larger than that of the inner cracks, the processes of initiation and growth of surface and inner cracks are almost the same.

Fig. 17 Surface cracks in torsion-test at a strain of 0.77 under 4 kbars

Fig. 18 Relationships between length of the longest surface cracks and strain under atmospheric pressure and 4 kbars
6. Discussion

Based on the above experimental results, the mechanism of ductile fracture of pearlitic steels is discussed here.

As shown in the section 5.1, the inner cracks in torsion-test are initiated at a very small strain. One of the authors has already reported that cementite of platelet shape in pearlite is fractured at a plastic strain not exceeding a few percent. Micro-cracks in pearlite colonies induce stress concentration, which causes successive fracture of the lamellar cementite platelet. After, the cracks grow rapidly. However, after passing through a pearlite colony, the crack stops growing when it comes across the ferrite matrix. Meanwhile, cracks grow and the number of cracks increases in other pearlite colonies. Therefore the length of the largest crack does not exceed the dimension of the pearlite colony. This can be presumed from the fact that the maximum length of cracks at small strains is almost in the same dimension as the diameter of pearlite colonies, as shown in Fig. 16.

Now, the conditions of crack-propagation into the ferrite matrix is considered. For the propagation of the crack, existence of a large local tensile stress at a crack-tip is necessary. There are many factors which may affect the state of stress at a crack-tip: (a) the tensile stress by piled-up dislocations, (b) macro-scopical external stresses, and (c) the stress concentration which depends on the volume, shape and dimension of cracks and inclusions. Since the number of dislocations is nearly proportional to the strain, the factor (a) corresponds to the first term of the left side of Eq. (1) (the term of strain). While, the factor (b) corresponds to the second term of Eq. (1) (the term of pressure), and the factor (c) is associated with the third term of the equation. Although the factors (a) and (c) can be explained by the existing dislocation theories, it is rather difficult to estimate the local stress quantitatively which depends on micro-structure such as the volume and the shape of cementite particles.

When many cracks in pearlite colonies begin to propagate into the ferrite matrix, the apparent flow stress is reduced by the existence of cracks, as was observed in gray cast iron. In the present experiment, the flow stress under atmospheric pressure was lower than that under a high pressure at a large strain just before fracture. When the slope of the nominal stress-strain curve in shearing deformation is changed to negative, the deformation will be concentrated into a narrow band. The macro-scopical fracture of a torsion specimen may be caused by the concentration of deformation after the growth of the cracks.

7. Conclusions

Torsion tests under high pressures were carried out for a commercial pure iron and five plain carbon steels. The results are clarified as follows.

(1) The fracture strain varies with the volume fraction and the shape of the cementite particles. The variation of fracture strain is caused by the difference in the initiation strain of fracture-nuclei growth.

(2) The fracture strain increases linearly with the aspect ratio of the cementite particles.

(3) A fracture criterion which includes the terms of volume and shape of cementite is derived. The calculated values of fracture strain agree well with the experimental values for various combinations of microstructures and pressure histories.

(4) Surface cracks and inner cracks are observed by a scanning electron microscope after torsional straining. The micro-cracks stay in pearlite colonies until the strain reaches a certain amount, and then they begin to propagate rapidly into the ferrite matrix. The strain at which cracks begin to grow rapidly is found to correspond to the initiation strain of nuclei-growth.

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