Effect of Temperature and Deformation Rate on Fracture Strength of Sintered Uranium Dioxide

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The fracture strength of two kinds of UO₂ specimens possessing pores of different maximum sizes (60 and 140 μm) was measured in the range of room temperature ~ 1,300°C by means of diametral compression testing. The fracture strength thus obtained proved to be smaller than any of the values reported by previous authors who mainly used bending tests. Finite element analysis showed that the method used in the present study should logically yield results close to the true tensile fracture strength. The discrepancies noted with the results reported from the other studies were attributable to the differences in the methods used for the measurements.

The fracture strength was found to remain almost constant in the relatively low temperature region (R.T.~800°C) beyond which the value increased with temperature (intermediate temperature region of 1,000~1,300°C). Electron-microscopic observations of the fractured surface indicated that the brittle-to-ductile transition temperature (Tc) was situated between 800 and 1,000°C when the strain was applied slowly. Raising the strain rate proved to affect both fracture strength and Tc. These dependences of temperature and strain rate on the fracture strength are explained from the relation between dislocation velocity and deformation rate. Griffith’s theory is cited to describe the relation between the largest pore size and fracture strength.

KEYWORDS: fracture strength, very high temperature, temperature dependence, diametral compression method, finite element analysis, brittle-to-ductile transition temperature, strain rate, Griffith’s theory, dislocation velocity

I. INTRODUCTION

The high thermal stresses occasioned in oxide fuels by the creation of steep temperature gradients during irradiation are liable to bring about fuel body fracture, and consequent generation of cracks. Such cracks have long been known to seriously affect fuel performance, with such symptoms as restructuring, pellet-cladding mechanical interaction and fission gas-release. These circumstances add importance to accurate determination of the fracture strength of oxide fuels.

A number of methods have been reported for measuring the fracture strength of oxide fuels. The earlier methods mostly incorporated bending tests, while one worker has experimented with loading UO₂ specimens shaped into helical form. These studies revealed that the fracture strength was influenced considerably by the presence of impurities and by the degree of porosity of the specimen (or the possession of large flaws). It was also indicated that three distinct patterns could be discerned in the fracture behavior of UO₂ according to the range of testing temperature. But

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so far no definite theory has been established to explain this change in temperature-dependence pattern observed on the fracture strength of UO₂. Strain rate-dependence on the fracture strength, on the other hand, has been pointed out by Canon et al.⁵ (for UC₂) and Roberts et al.⁶ (for (U, Pu)O₂) who showed that the brittle-to-ductile transition temperature shifted toward higher temperature with increasing strain rate. Agreement between these two reports, however, is not seen in respect of the relationship holding between the fracture strength and testing temperature.

More recently, Ainscough & Messer⁷⁻⁸ measured the fracture strength of UO₂ using an advanced testing technique, which is the ring test, and these authors indicated that the bending method was likely to give unrealistically high values of the fracture strength. This study revealed that the minimum measured fracture strength at room temperature was associated with the maximum flaw size.

The present work was designed to examine the effects of differences in maximum pore size, test temperature and deformation rate on the fracture strength of UO₂ in the range of room temperature ~ 1,300°C (corresponding to the lower and intermediate among the three temperature regions referred to earlier⁴⁻⁵). A diametral compression method⁹⁻¹⁰ was used to measure the fracture strength (reason for its adoption given in Chap. IV).

II. EXPERIMENTAL PROCEDURE

1. Test Specimens

Two kinds of UO₂ pellet designated types A and B were prepared by press-forming fine UO₂ powder and subsequent sintering at 1,500°C for 4 h in a reducing atmosphere. A spectrographic analysis of the raw material is given in Table 1.

Table 1 Spectrographic analysis of UO₂ powder

<table>
<thead>
<tr>
<th>Elements</th>
<th>Al</th>
<th>B</th>
<th>C</th>
<th>Ca</th>
<th>Cd</th>
<th>Cl</th>
<th>Cr</th>
<th>F</th>
<th>Fe</th>
<th>Mg</th>
<th>N</th>
<th>Ni</th>
<th>V</th>
</tr>
</thead>
<tbody>
<tr>
<td>Content (ppm of U content)</td>
<td>&lt;14</td>
<td>0.2</td>
<td>&lt;10</td>
<td>&lt;0.3</td>
<td>&lt;10</td>
<td>&lt;8</td>
<td>&lt;10</td>
<td>70</td>
<td>6</td>
<td>21</td>
<td>10</td>
<td>&lt;3</td>
<td></td>
</tr>
</tbody>
</table>

The pellets were cut into disk form with a diamond wheel slicing machine; the surfaces were then polished with emery paper (#600) in readiness for the fracture tests. Photograph 1 shows examples of the polished surfaces thus obtained. Closed pores are seen to be randomly distributed, the largest pore sizes being 60 μm in A type and 140 μm in B type specimens. All pellets were weighed and measured to determine their densities.

(a) A type specimen
(b) B type specimen

Photo. 1 Microphotographs of sliced surface of specimens
As it may be seen from Table 2, the specimens of both types A and B resemble each other quite closely in their characteristics, except for their bodily sizes and the largest pore diameters.

2. Testing Procedure

The fracture strength was measured by means of the diametral compression method. The disk specimen was sandwiched between tungsten plates, inserted under the crosshead of an Instron type compressing machine, maintained for 20 min at the prescribed temperature, and then compressed diametrically. The movement of the push rod and the applied load were recorded directly on an X-Y recorder. This testing machine assembly was contained in a glove box as described in a previous paper(16).

The tests were performed in vacuum (~10⁻⁵ Torr). The temperature was measured by Pt-Pt/Rh thermocouples attached to the specimen, and maintained by P. I. D. controller within 0.5% of the preset temperature. The normal compression rate, applied in the entire test temperature range of room temperature ~ 1,300°C was 1.0 mm/min. Additional measurements were made at 20, 100 and 500 mm/min, to examine the effect of differences in strain rate. These additional tests were performed on the B type specimens in the range between 800 and 1,300°C.

After the compression tests, the surface parallel to the direction of compression was observed under optical microscope to examine the crack patterns. Fractographs also were taken of the specimens after the tests by two-stage replica electron-microscopy.

III. RESULTS

Two types of UO₂ disk specimens with different pore sizes were diametrically compressed and fractured in the temperature range of room temperature ~ 1,300°C. All the tests were performed on specimens of almost identical grain and pore sizes as well as O/U ratio. Figure 1 plots the values of O/U ratio determined on various specimens both before and after the fracture tests.

The mode of failure observed, as typified in Photo. 2, indicates that the initial breakages usually occur along the centerline representing the axis of the applied

<table>
<thead>
<tr>
<th>Type</th>
<th>A</th>
<th>B</th>
</tr>
</thead>
<tbody>
<tr>
<td>Diameter (mmφ)</td>
<td>5.5</td>
<td>13.6</td>
</tr>
<tr>
<td>Thickness (average) (mm)</td>
<td>3.1</td>
<td>7.7</td>
</tr>
<tr>
<td>Pellet density (%T.D.)</td>
<td>94±1</td>
<td>94±1</td>
</tr>
<tr>
<td>Average grain size (μm)</td>
<td>~17</td>
<td>~17</td>
</tr>
<tr>
<td>Largest pore diameter (μm)</td>
<td>60</td>
<td>140</td>
</tr>
<tr>
<td>O/U ratio</td>
<td>2.01±0.01</td>
<td>2.01±0.01</td>
</tr>
</tbody>
</table>

Fig. 1 Variation of O/U ratio with testing temperature, determined by oxidation reduction method

Photo. 2 Pattern of cracks observed on UO₂ after fracture test
load. Typical load-displacement (change in diameter) curves for B type specimens are shown in Fig. 2. It is seen that the load and displacement proceed in linear relation to each other right up to fracture of the specimen when the test temperature does not exceed 1,000°C whereas at 1,200°C and beyond the curve comes to deviate from linearity prior to fracture, revealing that the region of ductile fracture is attained between 1,000 and 1,200°C.

If a disk of elastically isotropic material is compressed diametrically by applying a pair of knife edges at diametrically opposite ends, the maximum principal tensile stress occurs on the plane containing the lines of contact of the knife edges. This maximum stress is in the direction normal to the plane and tends to pull the disk asunder. The stress intensity is constant over a certain range on both sides of the axis, and is theoretically given by

\[ \sigma_1 = \frac{2P}{\pi DT}, \]

where \( P \) is the applied load, \( D \) the diameter and \( T \) the thickness of the disk. This formula was used in the present instance to determine the fracture strength, based on the diametral compression load at which the disk fractured, in conformity with the general practice followed in the diametral compression method. The resulting data proved to scatter less than those obtained by more classical methods.

Figure 3 presents, as function of testing temperature, the fracture strengths obtained at the lowest compression rate of 1.0 mm/min on specimens of both types A and B. The higher fracture strengths obtained on the A type specimens clearly evidence the favorable effect brought upon mechanical property by smaller pore size. It is also revealed that the fracture strength is independent of temperature in the range below 800°C, and then tends to increase with rising temperature.

The fracture strength of B type speci-
mens is presented in Fig. 4 as function of compression rate, at constant temperatures of 800, 1,000, 1,200 and 1,300°C. The strain rate $\dot{\varepsilon}$ represented on the scale appearing in the upper margin of this figure has been derived using the formula

$$\dot{\varepsilon} = \frac{\varepsilon}{Dt} = \frac{\sigma_f}{E} \frac{1}{Dt}, \quad (2)$$

where $\varepsilon$ and $\sigma_f$ are the tensile fracture strain and stress, respectively, $E$ is Young's modulus and $Dt$ the time to fracture. Upon insertion into Eq. (2) of the values of

![Graph showing the effect of compression rate on fracture strength of UO₂ at different temperatures.](image)

**Fig. 4** Effect brought on fracture strength of UO₂ by differences in compression rate (strain rate) at constant temperatures of 800, 1,000, 1,200 and 1,300°C.

Surfaces fractured at room temperature (a), at 1,000°C (b) and at 1,200°C (c); slip lines are seen around the pore in the specimen fractured at 1,000°C (indicated by arrow in (d)).

**Photo. 3** Electron-micrographs of B type specimens fractured at 1.0 mm/min compression rate.
18,000 kg/mm² (18) for $E$, 1.8±0.8 kg/mm² for $\sigma_f$ (Fig. 3) and $5 \times 10^{-3}$ h for $d_f$ (determined from the crosshead motion), it proves that the compression rate of 1.0 mm/min corresponds to a strain rate of 0.02±0.009 h⁻¹. Figure 4 indicates that, at 800°C, the fracture strength is little affected by differences in the compression rate (strain rate), whereas such differences come to influence the strength quite significantly at higher temperatures beyond 1,000°C.

The fracture behavior was studied by electron-microscopy. Typical aspects shown by the fractured surfaces of B type specimens compressed at 1.0 mm/min are presented in Photo. 3, which reveals that the fractures occur in both transgranular and intergranular modes, with microcracks appearing around the pores. (Photo. (a), (b) and (c)). In the specimens tested above 1,000°C, many slip lines were observed around these pores, indicating the occurrence of plastic flow in these locations (cf. Photo. 3(d)).

IV. DISCUSSION

1. Effect on Fracture Strength due to Differences in Testing Method

Table 3 represents a comparison between the studies by various authors including the present. The fracture strengths obtained in the present study are seen to be appreciably smaller than given by previous authors.

<table>
<thead>
<tr>
<th>References</th>
<th>Conditions of specimen preparation</th>
<th>Testing procedure</th>
<th>Fracture strength (kg/mm²)</th>
<th>Remarks</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td>Sintering temp. (°C)</td>
<td>O/U</td>
<td>Density</td>
<td>Grain size (μm)</td>
</tr>
<tr>
<td>Evans et al. (4)</td>
<td>1,650°C</td>
<td>2.000</td>
<td>97</td>
<td>8</td>
</tr>
<tr>
<td></td>
<td>1,750°C</td>
<td>~2.001</td>
<td>25</td>
<td></td>
</tr>
<tr>
<td>Canon et al. (5)</td>
<td>1,750°C</td>
<td>2.00</td>
<td>98</td>
<td>8</td>
</tr>
<tr>
<td></td>
<td>×4 h</td>
<td>~2.01</td>
<td>31</td>
<td></td>
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<tr>
<td></td>
<td></td>
<td>84</td>
<td>12</td>
<td></td>
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<td></td>
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<td>12</td>
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<td></td>
<td></td>
<td>94</td>
<td>15</td>
<td></td>
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<tr>
<td></td>
<td></td>
<td>95</td>
<td>17</td>
<td></td>
</tr>
<tr>
<td>Roberts et al. (7)</td>
<td>—</td>
<td>2.00</td>
<td>91</td>
<td>12</td>
</tr>
<tr>
<td></td>
<td></td>
<td>94</td>
<td>15</td>
<td></td>
</tr>
<tr>
<td></td>
<td></td>
<td>95</td>
<td>17</td>
<td></td>
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<tr>
<td>Solomon (9)</td>
<td>1,800°C</td>
<td>~30 min</td>
<td>92</td>
<td>15</td>
</tr>
<tr>
<td></td>
<td></td>
<td>95~96</td>
<td>15</td>
<td></td>
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<tr>
<td>Ainscough et al. (8)</td>
<td>1,650°C</td>
<td>—</td>
<td>98.6</td>
<td>12</td>
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<tr>
<td></td>
<td></td>
<td></td>
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<td></td>
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<tr>
<td>Kennedy et al. (14)</td>
<td>—</td>
<td>—</td>
<td>91</td>
<td>2</td>
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<tr>
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<td></td>
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<td>95</td>
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<tr>
<td></td>
<td></td>
<td></td>
<td>96</td>
<td>11</td>
</tr>
<tr>
<td>Ainscough et al. (15)</td>
<td>1,650°C</td>
<td>—</td>
<td>98.5</td>
<td>11</td>
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<tr>
<td></td>
<td></td>
<td></td>
<td></td>
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<tr>
<td>Bandyopadhyay et al. (15)</td>
<td>1,650°C</td>
<td>2.00</td>
<td>96</td>
<td>11</td>
</tr>
<tr>
<td>Present work</td>
<td>1,500°C</td>
<td>×4 h</td>
<td>94</td>
<td>17</td>
</tr>
<tr>
<td></td>
<td></td>
<td>~2.01</td>
<td></td>
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It will be recalled that, in the present work, the fracture strength was determined from the measured compression fracture load using Eq. (1). Strictly speaking, this
equation gives the true value only when the UO₂ disk is constrained by a load-imparting blade through a line of contact of negligibly small width. This width \( W \) can be calculated from the equation \( W = 1.6 \sqrt{PD(1-\nu^2)/TE} \),

\[ (3) \]

where \( P, D, T \) and \( E \) have already been defined, and \( \nu \) is Poisson's ratio of the disk material. Insertion into Eq. (3) of published values \(^{(17)}^{(18)} \) for \( \nu \) and \( E \), and the actual data for \( P, D \) and \( T \) results in a value of \( W \) smaller than 1.0 mm at fracture. The validity of this estimation was ascertained by measurements made of the actual values of \( W \) after the fracture tests.

Further, in order to examine the influence on stress imparted by increasing \( W \), some elastic analyses were performed to estimate the distribution of the principal tensile stresses generated in a diametrically compressed specimen, using the finite element method \(^{(19)}^{(20)} \) with \( W \), the width of contact area varied from 0 to 1.0 mm. Examples of the results, representing the cases of \( W=0 \) and 1.0 mm, are reproduced in Fig. 5. It is seen that, even between these two extreme cases, there is little difference in the intensity of maximum principal tensile stress which dominantly influences the fracture; the two results differ from each other mainly in respect of the pattern of stress distribution.

The influence upon the maximum principal stress brought by the finiteness of \( W \) is illustrated quantitatively in Fig. 6, where this influence is represented by a “decrease factor” \( \alpha \), which is defined as the ratio of maximum principal tensile stresses between that for \( W=0 \) and that for \( W > 0.0 \) mm. It is revealed that, if \( W < 1.0 \) mm, Eq. (1) will assure accuracy exceeding 95% in converting the compression fracture load into the fracture strength. This substantiates the validity of the diametral compression test in providing reasonably accurate values of tensile fracture strength for UO₂; the discrepancies seen in the measured fracture strengths given in the case by various authors could possibly be attributed, at least in part, to differences in their testing methods.

Berenbaum & Brodie \(^{(13)} \) have examined three different methods for measuring the tensile fracture strength of brittle materials, which they compared with the conventional tensile test. They showed that the fracture strength of plaster measured by bending
was about twice the value measured by two other methods—indentation and diametral compression. The latter two methods gave results in good agreement with those obtained by the conventional tensile test, but the bending method was found to be sensitively influenced by the surface conditions of the specimens. This circumstance might also apply to the bending tests performed on UO₂ specimens by past authors, to account for the particularly high values of fracture strengths they have reported (cf. Table 3).

The fracture strengths obtained in the present study are still smaller than those obtained by the previous workers from ring tests(10)(11) and diametral compression tests(14)(15). This might possibly be due to difference in the microstructure of the specimens used. In this connection, several workers(4)~(6) have pointed out that the fracture of UO₂ at temperature up to about 1,500°C are attributable to extension of large pores and/or flaws. Such a mechanism was also indicated in the present instance from photomicrographs (cf. Photo. 4). The authors have interpreted the cracks seen around the pores to have resulted from stress concentrations around the pores, which had led to fracture. It was also reasoned that the larger pores would tend to generate around them correspondingly larger stress concentrations, meaning that the fracture strength would be dominantly influenced by the size of the largest pore.

![Photo. 4](image-url)

Surfaces fractured at 800°C (a) and 1,300°C (b)

The smaller strengths registered in the present study are explained by means of Griffith’s theory(21) and a modification(22) thereof; Griffith’s theory stipulates that, for brittle materials containing penny-shaped flaws of radius \( R_\ast \), the fracture strength

\[
\sigma_\ast = \sqrt{2E\gamma_e/\pi R_\ast},
\]

and for similar materials having spherical pores of radius \( R_{**} \),

\[
\sigma_{**} = \sqrt{\pi E\gamma_e/2(1-\nu^2)R_{**}},
\]

where \( \sigma_\ast \) and \( \sigma_{**} \) are the theoretical fracture strengths, \( E \) Young’s modulus, \( \gamma_e \) the effective surface energy, and \( \nu \) Poisson’s ratio(17). The present results cannot easily be compared with those reported by Ainscough & Messer(10)(11) who measured tensile fracture strengths on highly densified UO₂ (98.5% T.D.), in consideration of the observation by Roberts & Ueda(7) that the fracture strength is strongly influenced by the density.
The data given by Kennedy and Bandyopadhyay, on the other hand, are devoid of values indicating the size of the pores contained in the specimens used in their diametral compression test, but their micrograph of a fractured surface reveals the presence of flaws of about 20 μm diameter resulting from heat treatment (annealing at 1,920°C for 18 h after quenching at 700°C). The fracture strengths of these specimens were about 2.4 kg/mm² at room temperature. Now, the effective surface energy $\gamma_e$ is difficult to estimate, and we will here use Eqs. (4) and (5) to derive the ratio of fracture strength values between that given in the reports cited above and that obtained in the present work. Inserting into Eqs. (4) and (5) the values of 10 μm for $R_*$, 50 μm (average between present specimens A and B) for $R_{**}$, and 0.30 for $\nu$ and assuming that the values of $E$, $\gamma_e$ and $\nu$ are common to the two studies, we obtain 1.5 as the ratio of fracture strengths $\sigma_*/\sigma_{**}$. This value practically agrees with the ratio of 2.4/1.5=1.6 obtained from the actually reported values. This justifies the smaller strengths obtained in the present study.

2. Influence of Largest Pore Size on Fracture Strength

The A and B types specimens used in the present study were prepared from a common source powder, so that the difference found between them in respect of fracture strength should confirm with Griffith's theory, relating the strength of brittle materials to flaws. The largest pores present in the A type specimen have a radius of 30 μm, and the corresponding pores of B type are 70 μm. Applying these pore sizes to Eq. (5), we derive the ratio to be expected between the strengths of the A and B types specimens:

$$\frac{\sigma_A}{\sigma_B} = \sqrt{\frac{70}{30}} = 1.5.$$ 

The corresponding value of the ratio, determined from the actual strengths measured on the two types of specimen (Fig. 3) is (in the range of room temperature ~ 800°C) 2.0, which is close enough to the calculated value of 1.5 given above. Hence, in the lower temperature range of brittle fracture, Eq. (5)—base on Griffith's theory—may be considered to provide a practical indication of the effect on fracture strength brought by differences in the size of the largest pores.

At high temperatures, with the onset of plastic flow in the strain field, Griffith's theory must be modified in accordance with Orowan's suggestion. This provides for the addition in Eq. (5) of a plastic work function $F$ to the surface energy $\gamma_e$, which latter, besides, is in many cases small compared with $F$ as to be negligible. The Orowan modification permits Eq. (5) to be rewritten in the form

$$\sigma = \sqrt{\pi E F/2(1-\nu^2)R_{**}}.$$  

From Orowan's definition of $F$, it is commonly accepted that the value of $F$ in Eq. (6) corresponds to the amount of plastic deformation. Now it can be logically expected that plastic deformation would take place more extensively in specimens possessing larger radius pores and at higher temperatures; this would imply that the slopes presented by the plots in Fig. 3 at the higher temperatures beyond 1,000°C should be somewhat steeper for $\sigma_B$ than for $\sigma_A$. And actually, the ratio of measured fracture strengths between A and B types is seen to narrow down from 1.9 at 1,000°C to 1.4 at 1,300°C.

3. Brittle-to-ductile Transition Temperature ($T_c$)

It was seen in Fig. 2 that, for the tests conducted at the compression rate of 1.0
mm/min, specimen deformation (change in diameter of disk specimens) was roughly proportional to the applied compression load, up to 1,000°C, above which the curve came to deviate from linearity prior to rupture. This change in the stress-strain relationship suggested the existence—in macroscopic sense—of a brittle-to-ductile transition temperature $T_c$ \cite{15}, situated between 1,000 and 1,200°C. Electron-microscopy, on the other hand, indicated 800-1,000°C to be the range that included $T_c$; on the surface fractured at 1,000°C there were observed a number of slip lines (Photo. 3(d)), which were absent from the specimen tested at 800°C.

At the highest compression rate of 500 mm/min ($\varepsilon=10.0$ h$^{-1}$), $T_c$ shifted upward beyond 1,300°C (Fig. 2(f)). Canon et al.\cite{5} measured the UO$_2$ fracture strength from room temperature to 1,800°C to determine its temperature dependence at three different strain rates of 0.092, 0.92 and 9.2 h$^{-1}$, making use of bending method; he reported that $T_c$ rose from 1,000 to 1,450°C with the increase of strain rate. The values of $T_c$ obtained in the present work agree well with those given above by Canon et al., and a similar tendency has also been observed in respect of strain rate-dependence on $T_c$.

A mechanism for explaining the influence of strain rate on $T_c$ has been proposed by Roberts & Wrona\cite{6}. They found from a series of bending tests on (U, Pu)O$_2$ specimens that the relation between $T_c$ and strain rate (0.15 to 0.73 h$^{-1}$) lent itself to Arrhenius-type expression. Nadeau\cite{24}, on the other hand, showed that deformation of UO$_2$ under high temperature at rate of strain ranging of 0.02~7.2 h$^{-1}$ was primarily controlled by dislocation motions and not by pure diffusion. Moreover, Yust & McHargue\cite{25} found that, with single UO$_2$ crystals, the dislocation velocity was controlled mainly by the temperature. Thus, in the low temperature region, the deformation should have to be applied at a correspondingly low rate in order to permit the plastic flow generated by dislocation motion to keep up with the applied strain and let the specimen yield without fracture, while a higher strain rate could be sustained without fracture when the temperature is higher; this explains the rise with strain rate shown by $T_c$, the temperature of transition to ductile property.

4. Effect of Differences in Temperature on Fracture Strength

Evans et al.\cite{4} and Canon et al.\cite{5} have examined the temperature dependence of fracture strength in the temperature ranges up to 1,500 and 1,800°C, respectively. The first-cited study presented the relation to temperature in the form of a convex curve in the range of room temperature $\sim$ ca. 1,400°C, while for the same lower and intermediate temperature regions, the latter work indicated a gradual increase of the fracture strength with rising temperature. Thus the two authors give mutually differing patterns of temperature dependence, neither of which, moreover, agrees with the present results, presented in Fig. 3, where, for both A and B specimens, the plots maintain a constant level in the range up to 800°C, and then abruptly rise beyond 1,000°C. Evans et al. explained the temperature dependence of fracture strength in terms of an increase of the effective surface energy and linking of microcracks which might be caused by plastic deformation. The mechanism suggested by Canon et al. for the gradual increase of fracture strength with temperature was a strengthening provide by the plastic deformation\cite{5}.

The model we propose for describing the temperature dependence observed in our study brings into account the extension of pores and dislocation motions. It is inspired by Mordike’s recent experimental work\cite{27} on yield and flow in UO$_2$ single crystals.
which showed that slip—*i.e.* dislocation motion in UO$_2$—became pronounced above 800°C, and that below this temperature it remained negligibly small, with plastic deformation contributing little to the fracture strength of UO$_2$. This would imply that fracture below 800°C occurs primarily by extension of the pores where stresses concentrate. This aptly explains the almost complete absence of temperature dependence observed in the present work in the lower temperature range below 800°C.

In the intermediate temperature region (1,000~1,300°C), the movement of dislocations pointed out by Mordike would contribute to dissipating the stress concentrations at the pores. The occurrence of such relaxation is substantiated in the present instance by the slip lines observed in Photo. 3(d) around pores in the UO$_2$ fractured at 1,000°C. The foregoing observations indicate stress dissipation to be the cause of the increase in fracture strength seen with rising temperature in the intermediate temperature region.

5. **Effect on Fracture Strength Brought by Differences in Strain Rate**

The relation between fracture strength and strain rate was examined by Canon *et al.*(5) for UO$_2$ and by Roberts *et al.*(6) for (U, Pu)O$_2$. These two studies have yielded mutually inconsistent results, probably due to the difference in specimen material. The data from the present study, on the other hand, agree in substance with the latter of the two studies cited above, for the intermediate temperature region—*i.e.* decreasing fracture strength with increasing strain rate at temperatures above 1,000°C (Fig. 4). To explain this behavior, Roberts *et al.* proposed a mechanism that took into consideration differences in strain rate: This mechanism stipulated that, in this intermediate temperature region, $T_c$ would shift toward higher temperature with increasing strain rate, induced by a thermally-activated *i.e.* diffusion-controlled-mechanism. This explanation, however, should validly apply only to cases of low strain rate (*e.g.* high temperature creep deformation), since at the higher strain rates, plastic deformation might be governed, more likely, by a mechanism other than diffusion$^{(16)(24)}$.

The mechanism proposed here takes account of the value of dislocation velocity in its relation to the strain rate, as described in Sec. IV-3. This mechanism implies that, the fracture strength is independent of the strain rate in the low temperature region below 800°C, where the fracture is not affected by plastic deformation.

In the temperature range above 800°C, dislocation motion becomes pronounced, to influence the fracture strength quite significantly. For a given temperature in this region, the amount of plastic deformation accepted without fracture decreases with increasing strain rate (deformation rate), with less time allowed for the movement of dislocations to let the specimen yield without fracture. Thus even in this temperature range favorable to plastic flow and consequent dissipation of the stresses concentrated around pores, increasing the strain rate will tend to undermine the ductility acquired at the higher temperature. This explanation should be considered more convincing than that given by Roberts *et al.*, who called into play a thermally activated process for this intermediate temperature region.

**V. CONCLUSIONS**

The following conclusions can be derived from the results of diametral compression tests applied to UO$_2$ specimens in the range of room temperature ~ 1,300°C:

1. The measured fracture strength in UO$_2$ varies according to the method used for testing; diametral compression method used in this study should provide tensile
fracture strengths close to the true values.

(2) The fracture strength of sintered UO₂ is sensitively influenced by the conditions of specimen preparation; but Griffith's theory appears to be applicable, by which the strength is inversely proportional to the square root of the largest pore size.

(3) The brittle-to-ductile transition temperature is affected by the strain rate. This is explained in terms of the relation between dislocation velocity and strain rate.

(4) The pores present in the UO₂ structure extend under stress, leading to fracture, and up to about 800°C the fracture strength is independent of temperature. In the intermediate temperature region (1,000~1,300°C), the fracture strength increases with temperature, due to onset of plastic flow, which serves to dissipate the stresses concentrated at the pores.

(5) The temperature dependence of the fracture strength weakens with increasing strain rate, which allows less time for plastic deformation to dissipate the rising stress concentrations around the pores.

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