The Effect of Co Content on Fatigue Crack Growth Characteristics of WC-Co Cemented Carbides

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Abstract

In order to study the influence of Co content on crack propagation behavior of WC-Co cemented carbides, cyclic fatigue crack growth tests were carried out over a wide range of fatigue crack growth rates. The precrack was introduced by the EC method. Fatigue crack growth tests were conducted at a constant stress ratio \( R (=0.1, 0.5) \) and constant frequency \( f (=10\text{Hz}) \). The Paris law was confirmed between \( da/dN \) and \( \Delta K \) on these materials. The fractography of the fracture surface shows different fracture patterns depending on \( R \). It was found that phase transformation occurred in the Co phase on the fracture surface by X-ray observation.

Introduction

Recently, it was clarified that the Co phase transformed from \( \gamma \)-Co to \( \delta' \)-Co with an increase in external energy and the failure of the Co phase in WC-Co cemented carbides was generated by structure changes. Thus, it is thought that the phase transformation of Co exerts a strong influence on the fatigue crack progress behavior, but there are few papers which have investigated in detail the relation between the two[1][2].

Therefore, in this study a fatigue crack growth test was done using WC-Co cemented carbides into which a precrack was introduced by the EC(Edge Compression) method[3] and an investigation of the influence of the Co content and the stress ratio \( R (=P_{\text{min}}/P_{\text{max}}) \) on the crack growth characteristics was performed. Moreover, fractography was performed on the fractures obtained and the mechanism of fatigue crack progress was examined in detail.
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In addition, the phase transformation from $\gamma$-Co (fcc structure) to $\varepsilon'$-Co (hcp structure) was examined by X-ray diffraction as was the influence of the phase transformation on fatigue crack growth.

**Experimental Procedure**

**Materials and specimens**

The materials supplied for the experiment were WC with 16 and 25wt%Co cemented carbides which were sintered in vacuum. Average WC grain size after sintering was 3 $\mu$m. Each material was recorded as follows: 16Co3 and 25Co3. Typical mechanical properties of the materials are shown in Table 1. Figure 1 shows the shape and size of the test specimen used for this study. The test piece contained a 1mm notch produced by mechanical processing and a precrack of about 1mm was introduced into each of the specimens by the EC method (Fig.2). The specimen surfaces were polished to a mirror finish with diamond powder.

### Table 1. Mechanical properties.

<table>
<thead>
<tr>
<th>Materials</th>
<th>$E$(GPa)</th>
<th>$\sigma_f$(MPa)</th>
<th>$K_{IC}$(MPa$\sqrt{m}$)</th>
<th>$HV$</th>
</tr>
</thead>
<tbody>
<tr>
<td>WC-16%Co</td>
<td>516</td>
<td>2925</td>
<td>14.4</td>
<td>969</td>
</tr>
<tr>
<td>WC-25%Co</td>
<td>456</td>
<td>2666</td>
<td>16.3</td>
<td>812</td>
</tr>
</tbody>
</table>

$E$: Young’s modulus  
$\sigma_f$: Bending strength  
$K_{IC}$: Fracture toughness value  
$HV$: Vickers hardness

![Fig.1. Shape and dimensions of specimen](image1)

![Fig.2. Schematic illustration for the EC Method](image2)

**Fatigue crack growth test**

Load-controlled fatigue tests in 3-point bending were carried out in Shimazu servohydraulic testing machine at room temperature using a sinusoidal wave form. The frequency of stress cycling was 10Hz and the test environment was air of 40% r.h. at 20°C. Stress ratios, $R (=K_{min}/K_{max})$ of 0.1
and 0.5 were used. In order to minimize any effect of the precracking procedure on the test results, the precrack was extended under cyclic loading about 1mm before taking test data. The crack length, \( a \), was determined at a magnification of 200 times with an optical microscope. The stress intensity factors, \( \Delta K \) were calculated from the following equation,

\[
\Delta K = \frac{3S \cdot \Delta P}{2B \cdot W^{3/2}} \cdot F(a/W)
\]

where \( S \) is the length of the span, \( \Delta P \) is the load range, \( B \) is the thickness, \( W \) is the breadth, and \( F(a/w) \) is a modifying factor.

**Observation of crack closure behavior**

The crack closure behavior was observed by the unloading elastic compliance method. The output of a clip gauge installed on the specimen was taken into an analyzing recorder and the crack closure load was obtained. The effective stress intensity factor range, \( \Delta K_{\text{eff}} \) was obtained from equation (2).

\[
\Delta K_{\text{eff}} = K_{\text{max}} - K_{\text{op}} \quad (K_{\text{op}} > K_{\text{min}})
\]

\[
= K_{\text{max}} - K_{\text{min}} = \Delta K \quad (K_{\text{op}} \leq K_{\text{min}})
\]

where \( K_{\text{max}} \) is the maximum stress intensity factor, \( K_{\text{min}} \) is a minimum stress intensity factor, and \( K_{\text{op}} \) is the crack closure stress intensity factor obtained from \( P_{\text{op}} \), the crack closure load.

**Experimental Results and Discussion**

**Fatigue crack growth behavior**

Figure 3 shows the relation between crack propagation rate, \( da/dN \), and stress intensity factor range \( \Delta K \) of the 16Co3 and 25Co3 materials. The arrows in Fig.3 show the threshold stress intensity factors, \( \Delta K_{\text{th}} \). When the relation \( da/dN-\Delta K \) of both materials was obtained, it was possible to approximate from the low crack growth rate region to the high crack growth rate region respectively using one curve. There is a general trend for \( da/dN \) to increase with increasing stress ratio, \( R \) in 16Co3 when the stress intensity factor range is the same. On the other hand, some differences are observed between the data for \( R=0.1 \) and \( R=0.5 \) in the high crack growth rate region, thought to be due to the influence of static fatigue in 25Co3. However, the data settled comparatively well in the low crack growth rate region and the influence of the stress ratio is not observed.

In general, for experimentally obtained data from metallic materials, it is possible to represent the crack growth rate, \( da/dN \) within the range of \( \Delta K \) in a specific region by a function of the type
shown in equation (3) known well as the Paris law. The $da/dN$-$\Delta K$ diagram of WC-Co cemented carbides obtained in this study followed the straight line relation:

$$da/dN = C \cdot (\Delta K)^m$$  \hspace{1cm} (3)

Table 2. Values of constants $C$ and $m$.

<table>
<thead>
<tr>
<th>Materials</th>
<th>$R$</th>
<th>$C$</th>
<th>$m$</th>
</tr>
</thead>
<tbody>
<tr>
<td>16Co3</td>
<td>0.1</td>
<td>$2.23 \times 10^{-15}$</td>
<td>10.36</td>
</tr>
<tr>
<td></td>
<td>0.5</td>
<td>$2.77 \times 10^{-14}$</td>
<td>11.14</td>
</tr>
<tr>
<td>25Co3</td>
<td>0.1</td>
<td>$1.15 \times 10^{-11}$</td>
<td>6.92</td>
</tr>
<tr>
<td></td>
<td>0.5</td>
<td>$2.32 \times 10^{-13}$</td>
<td>9.33</td>
</tr>
</tbody>
</table>

Crack closure was not able to be observed with 25Co3 though it was observed in 16Co3 since the unloading elastic compliance method was used. Figure 4 shows the fatigue crack growth data for an effective stress intensity factor range, $\Delta K_{eff}$, worked out from the measurement of the crack opening level ($K_{op}$) using equation (4).
\[
\frac{da}{dN} = C' (\Delta K_{\text{eff}})^{m'}
\]  

(4)

where \(C', m'\) are material constants. 25Co3 behavior does not differ from that in Fig.3 because crack closure behavior is not observed even if the stress ratio changes, by defining \(\Delta K_{\text{eff}} (K_{\text{op}}=K_{\text{min}})\). On the other hand, the fatigue crack growth characteristics of 16Co3 cannot be completely explained with \(\Delta K_{\text{eff}}\) alone. This is based on fractography and X-ray diffraction measurement results on the fracture surface.

**Fractography**

Figure 5 shows an SEM photograph of the fracture surface of 25Co3. If the fracture surface is observed macroscopically, the fracture path is as follows: fracture follows the interface between carbide crystals, then the interface of the carbide-cobalt, and finally the rupture of the cobalt phase. Next, fracture surfaces were compared for the brittle fracture present in the Co phase at \(R=0.1\) and the ductile fracture present in the Co phase at \(R=0.5\). Observation of the WC part of the fracture at \(R=0.5\) showed river patterns which confirmed cleavage fracture. It was clear that the fracture form was different depending on the stress ratio for 25Co3, moreover, the fracture surface pattern of 16Co3 was basically similar to 25Co3.

![SEM photograph](image)

(a) \(R = 0.1\)  \(\Delta K_{\text{eff}} = 6.0 \text{ MPa} \sqrt{\text{m}}\)  
(b) \(R = 0.5\)  \(\Delta K_{\text{eff}} = 6.1 \text{ MPa} \sqrt{\text{m}}\)

Crack growth direction is from left to right \(1 \mu\text{m}\)

Fig. 5. Scanning electron micrographs of fatigue fracture surfaces of 25Co3.
Phase transformation of Co phase

To examine the role which the phase transformation of Co played in the fracture mechanism of the WC-Co cemented-carbides, the extent of the phase transformation of Co on the fatigue fracture surface was examined by the X-ray diffraction method using an automatic X-ray diffraction device. \(\varepsilon'\) - Co 101 was mainly observed at 40kV and 100mA with CrK\(\alpha\). The X-ray irradiation was done on specimens in the range of \(K_{\text{max}} = 10\text{~MPa} \sqrt{\text{m}}\). Figure 6 shows the result of X-ray diffraction measurement on the fracture surface of 25Co3. Next, table 3 shows the ratio of the integrated intensity \(I_{N0}\) of \(\gamma\) - Co 111 and the integrated intensity \(I_N\) of \(\varepsilon'\) - Co 101 obtained from the fracture surface. The phase transformation of Co is generated on each fracture surface and the largest integrated intensity was found at \(R = 0.1\) in 25Co3. Moreover, the Co phase transformation was promoted by a low stress ratio and by a high content of the Co phase.

![X-ray diffraction patterns](image)

**Fig. 6.** X-ray diffraction patterns of 25Co3.
\((K_{\text{max}} = 10\text{~MPa} \sqrt{\text{m}})\)

<table>
<thead>
<tr>
<th>Materials</th>
<th>(R)</th>
<th>(I_N / I_{N0})</th>
</tr>
</thead>
<tbody>
<tr>
<td>16Co3</td>
<td>0.1</td>
<td>0.297</td>
</tr>
<tr>
<td></td>
<td>0.5</td>
<td>0.110</td>
</tr>
<tr>
<td>25Co3</td>
<td>0.1</td>
<td>0.701</td>
</tr>
<tr>
<td></td>
<td>0.5</td>
<td>0.269</td>
</tr>
</tbody>
</table>

It is thought that the reason for the amount of phase transformation in 25Co3 being larger than in 16Co3 for the same stress ratio, is as follows: The Co phase cannot undergo enough plastic...
deformation and fracture because the mean free path of the Co (thickness) becomes small when the Co content is small. However, the contact ratio of the WC particle is smaller in 16Co3 than in 25Co3 because the content of Co in the alloy is smaller. Moreover, the amount of the plastic deformation increases with Co content because the mean free path of Co becomes thick. As a result, it is thought that the amount of phase transformation of Co in 25Co3 was greater than 16Co3. It is thought that when the content of the transformation dislocations increases, the amount of transformation of the Co increases, too.

Next, it is thought that a large amount of phase transformation is observed at $R=0.1$ as follows. The cyclic deformation experienced by the Co is sure to increase because the change of $K$ which acts at the crack tip is large for $R=0.1$. Moreover, because the stacking fault content in Co increases when the amount of the cyclic deformation increases, it is thought that a large amount of phase transformation is caused at $R=0.1$. Vasel[2] reported that this phase transformation is remarkably promoted by cyclic load compared with the case of a monotonic load. Moreover, the slip planes are limited in the surrounding untransformed phase and the $\varepsilon'$ phase is thought to cause an adverse effect on the fracture. The fatigue embrittlement mechanism produced by the effect of the phase transformation is schematically shown in Fig.7. When the crack progresses through the binder phase, the binder phase located at the crack tip causes the phase transformation by cyclic deformation. On transforming a part of the $\gamma$-Co phase into the other crystal structure ($\varepsilon'$-Co), the transformed phase produced in this way receives an extremely strong restraint from the surrounding $\gamma$-Co phase which therefore, causes a high stress in $\varepsilon'$ phase. To ease this stress, slip or twinning is generated internally in the $\varepsilon'$-Co phase. The deformation to follow cannot take place because the number of slip systems is limited for $\varepsilon'$-Co compared with $\gamma$-Co and a crack is formed in the Co phase. It is thought that the difference in the fracture pattern of the Co was caused by the difference in the phase transformation behavior of such Co as shown in Fig.7 (a) and (b).

![Crack growth direction](image)

(a) WC-16%Co  (b) WC-25%Co

Fig. 7. Schematic of phase transformation at the crack tip.
Conclusions

The main results obtained in the present study are summarized as follows:

(1) The Paris law was confirmed between $\frac{da}{dN}$ and $\Delta K$ and it was shown that $\frac{da}{dN} = C(\Delta K)^n$. The influence of the stress ratio was not evident in 25Co3 though it was observed in 16Co3. The relation between $\frac{da}{dN}$ and $\Delta K$ for the WC-Co cemented carbides did not follow that of other metals, but showed a characteristic between that of metal and of alumina ceramics.

(2) The fractography of the fracture surface shows that brittle fracture occurs in the Co binder phase at a stress ratio $R=0.1$ and ductile fracture occurs in the Co binder phase at $R=0.5$.

(3) The Co phase in the neighborhood of the crack tip undergoes a phase transformation by a fatigue process and Co becomes brittle. It has been understood that the effects of Co content and Co phase transformation are closely related to the fatigue crack growth characteristic of WC-Co cemented carbides.

References