Effect of Hydrogen on Fatigue Crack Propagation in Vanadium

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The influence of hydrogen on fatigue crack propagation in unalloyed vanadium and several hydrogen-charged vanadium alloys has been investigated. The Paris-Erdogan equation, \( da/dN = C(\Delta K)^m \), was approximately obeyed for all alloys. Crack growth rates were lowest in vanadium and dilute vanadium-hydrogen alloys, and were not very sensitive to volume fraction of hydrides in more concentrated alloys. The crack growth exponent, \( m \), is inversely proportional to the cyclic strain hardening rate, \( n' \), and the rate constant \( C \) is inversely proportional to the square of the ultimate tensile stress, \( \sigma_{UTS} \). Metallographic examination showed hydride reorientation and growth in the originally hydrided alloys. No stress-induced hydrided were observed in V-H solid-solution alloys. Fractures in hydrided materials exhibited cleavage-like features, while striations were noted in unalloyed vanadium and dilute solid-solution alloys.

The effects of hydrogen on monotonic mechanical behavior of the group Va refractory metals (Ta, V and Nb) have been extensively investigated. Strengthening by hydrogen in solution and hydride precipitates in these metals occur within certain temperature ranges, but reduced ductility (hydrogen embrittlement) at low temperature also has been observed. Little attention, however, has been directed to the influence of hydrogen on cyclic deformation and fatigue crack propagation in the refractory metals.

Wilcox showed that hydrogen in solution and in the form of hydrides significantly reduced the fatigue life of tantalum and concluded that hydrogen embrittlement was associated with increased ease of both crack initiation and propagation. In contrast, Lee and Stoloff recently reported significant improvements in high cycle tension-compression (T-C) fatigue life when hydrogen was introduced into vanadium as hydrides. Furthermore, fatigue lives of hydrided vanadium were several times greater than for an alloy with 132 ppm hydrogen in solution when tested at a similar stress level.

The present investigation extends the work of Lee and Stoloff to take into account the separate effects of hydrogen content on the time for crack initiation and the rate of crack propagation in vanadium. The study utilizes six different hydrogen contents in polycrystalline vanadium; unalloyed, two alloys with hydrogen in solution and three hydrided alloys. Stress ranges and mean stresses have been investigated in stress controlled tests at room temperature. Fatigue crack growth characteristics are compared with tensile properties and fracture modes were observed by means of metallography and scanning electron microscopy.

EXPERIMENTAL PROCEDURE

Vanadium was received from Wah Chang Corp. in the form of annealed sheet with a grain size ~50 \( \mu \)m. Fatigue specimens of the single-edge notch type: 0.3 cm thick \( \times 2.54 \) cm gage length \( \times 1.43 \) cm wide, (notch depth = 0.127 cm) were machined from the original material. Following mechanical polishing to 6 \( \mu \)m diamond compound, electrolytic polishing was conducted in a \( \text{H}_2\text{SO}_4-\text{H}_2\text{O} \) solution, applying 30 volts at 0°C. All samples were then annealed for two hours at 760°C in a vacuum of 10\(^{-5} \) torr to obtain the final grain size of \( \sim 80 \) \( \mu \)m. Predetermined amounts of hydrogen were produced by the thermal decomposition of \( \text{ZrH}_2 \) in a vacuum furnace for two hours at 660°C. All materials were furnace cooled. The chemical analyses of vanadium and the vanadium alloys, as well as the size of hydrides, are listed in Table I. The microstructures of the alloys were similar to those reported in Ref. 5. The increased oxygen content associated with high hydrogen levels has previously been shown to have little or no effect on fatigue behavior. The solubility limit of hydrogen in vanadium at room temperature is reported to be \( \sim 500 \text{ ppm} \) (\( \sim 2.5 \text{ at. pct} \)).

Fatigue tests were conducted in tension-tension (T-T) loading in a closed-loop electrohydraulic machine at room temperature in air using a sinusoidal loading cycle at a cyclic frequency of 20 Hz. During the fatigue tests measurements of crack length were made by use of thin plastic replicating tape placed at the fracture surface; accuracy was estimated to be \( 2 \times 10^{-4} \) cm. The tests were terminated when the crack had reached a length equal to 60 pct of specimen width. The stress intensity range, \( \Delta K \), was determined from a standard \( K \) calibration.

Strain control T-C tests were run on additional specimens of all alloys to establish cyclic strain hardening data, and tensile tests were performed to determine base-line mechanical properties. Yield stress, \( \sigma_y \), ultimate tensile strength, \( \sigma_{UTS} \), and total strain, \( \epsilon \), are listed in Table I.

RESULTS

Fatigue Crack Propagation

Crack growth rates, \( da/dN \), are presented in Figs. 1 to 4 in the form of log-log plots of \( da/dN \) vs \( \Delta K \):

\[ da/dN = C(\Delta K)^m \]

as described by Paris and Erdogan.\(^{11,12} \)
Figure 1 shows data plotted for annealed vanadium and several V-H alloys at a stress ratio, \( R(\sigma_{\text{min}}/\sigma_{\text{max}}) = 0.4. \) (Every second point has been omitted from the 1500 ppm alloy curve for clarity. The full curve is replotted in Fig. 4.) It can be seen that the crack growth rates for a given \( \Delta K \) are similar for annealed vanadium and the 200 ppm H alloy although the yield stress of the latter is significantly higher than that of annealed vanadium (see Table I). The crack growth rates of the high hydrogen alloys deviate significantly from linearity at high and low \( \Delta K \), as is commonly observed for engineering materials; however, we will continue to use Eq. [1] as a basis for discussion, and straight lines are drawn through the data for intermediate \( \Delta K \) levels in Fig. 1. Increasing hydrogen content in solution to 420 ppm is seen to significantly increase both the slope of \( da/dN \) vs \( \Delta K \) and the rate of crack growth for a given \( \Delta K \). In the 1500 ppm H alloy (hydrides) the slope is increased; however, crack growth is slower than in the 420 ppm H alloy. Figure 2 shows data for the high hydrogen alloys, with \( R = 0.05 \). The crack growth rate of the 830 ppm H alloy was the lowest among the three hydrided alloys, and was very similar to that of the solid solution 420 ppm H alloy.

The influence of stress ratio, \( R \), on \( da/dN \) for V-420 ppm H is shown in Fig. 3. (The curve for \( R = 0.4 \) is the same as in Fig. 1 except for the two points between \( 10^{-5} \) and \( 10^{-4} \) m/cycle which were left off Fig. 1.) An increase in \( R \) from 0.05 to 0.4 produces a significant increase in growth rate at all values of \( \Delta K \); however, no further increase in \( da/dN \) is noted for \( R = 0.516 \). In the 1500 ppm H alloy on the other hand, the crack growth rate, as shown in Fig. 4, is hardly influenced by \( R \) increasing from 0.05 to 0.40, but there is a distinct increase in \( da/dN \) as \( R \) is increased further to 0.516. An increase in growth rate with increasing mean stress (higher \( R \)) is characteristic of brittle steels.

Table II is a summary of all crack growth data, showing that \( m \) increases with hydrogen content and perhaps with increasing mean stress. Also shown in Table II are values of the cyclic strain hardening exponent, \( n' \), which were determined for each alloy; \( n' \) decreased sharply with increased hydrogen content, in agreement with previous observations. Information on fatigue crack nucleation (actually initial extension of 10 pct in the length of the initial

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**Table I. Chemical Analyses and Tensile Properties of V-H Alloys**

<table>
<thead>
<tr>
<th>Specimens</th>
<th>H, ppm</th>
<th>O, ppm</th>
<th>N, ppm</th>
<th>( \sigma_y ) (MN/m²)</th>
<th>( \sigma_{UTS} ) (MN/m²)</th>
<th>Total ( e, )%</th>
<th>Microstructure</th>
</tr>
</thead>
<tbody>
<tr>
<td>V-Annealed</td>
<td>6</td>
<td>65</td>
<td>27</td>
<td>136.5</td>
<td>220.6</td>
<td>75</td>
<td>Single phase</td>
</tr>
<tr>
<td>V-200 H</td>
<td>200</td>
<td></td>
<td></td>
<td>146.1</td>
<td>232.5</td>
<td>45</td>
<td>Single phase</td>
</tr>
<tr>
<td>V-420 H</td>
<td>420</td>
<td>170</td>
<td>39</td>
<td>173.3</td>
<td>245.4</td>
<td>40</td>
<td>Single phase</td>
</tr>
<tr>
<td>V-530 H</td>
<td>520</td>
<td>130</td>
<td>12</td>
<td>181.7</td>
<td>282.6</td>
<td>34</td>
<td>Hydride (20 ( \mu )m)</td>
</tr>
<tr>
<td>V-830 H</td>
<td>830</td>
<td>190</td>
<td>43</td>
<td>193.3</td>
<td>293.2</td>
<td>22.5</td>
<td>Hydride (40 ( \mu )m)</td>
</tr>
<tr>
<td>V-1500 H</td>
<td>1500</td>
<td>480</td>
<td>57</td>
<td>203.7</td>
<td>293.2</td>
<td>22.5</td>
<td>Hydride (70 ( \mu )m)</td>
</tr>
</tbody>
</table>