Effect of microstructure on fatigue properties of WC-Co hard metals

Thomas Klünsner, Stefan Marsoner, Reinhold Ebner, Reinhard Pippan, Johannes Glätzle, Arndt Püschel

Materials Center Leoben Forschung GmbH (MCL) Roseggerstrasse 12, 8700 Leoben, Austria
Erich Schmid Institut, Jahnstrasse 12, 8700 Leoben, Austria
Ceratizit Austria, Metallwerk-Plansee-Straße 71, 6600 Reutte, Austria

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Abstract

Due to the inherent hardness and brittleness of WC-Co hard metals the determination of mechanical properties such as yield strength, ultimate strength and fatigue strength is a difficult task. A testing setup to determine stress-strain curves of hard metals under static and cyclic uniaxial compression and tension loading was developed. The investigated hard metal grades varied in WC grain size and Co-binder content ranging from 0.2 to 1.3 μm and 6 to 12 wt. %, respectively. The static stress-strain behaviour was analysed in uniaxial compression and tensile tests and specific material data like Young’s modulus and yield strength were determined. The cyclic stress-strain response along with the ratchetting behaviour as well as subcritical fatigue crack growth were investigated under uniaxial loading conditions. Fractographic studies of the fracture surfaces were conducted using scanning electron microscopy to identify the nature of crack origins. Based on Murakami’s concept the threshold of the stress intensity factor range at which fatigue crack growth starts from inhomogeneities was determined to be about 4.3, 6.2 and 9 MPa√m at stress ratios of 0.1, -1 and -3 respectively.

Keywords: Hard metal; uniaxial tension and compression test; Young’s modulus; yield strength; cyclic behaviour; threshold of stress intensity factor range

1. Introduction

Hard metals are so far mainly utilized for wear applications and metal machining. Today there is an increasing demand in using these materials for highly fatigue loaded tools or components. In the last years the property spectrum was significantly widened and today new hard metal grades are available with improved hardness, bending strength and/or fracture toughness. The main achievements are based on WC grain refinement leading to fine and ultra fine grained hard metal grades. A prerequisite for the exploration of the potential of hard metals for highly loaded tools or components is a sound knowledge of their static and cyclic behaviour. In order to design fatigue-proof tools or structural components made of hard metals, mechanical and design engineers require material data like yield strength or fatigue strength.
So far the mechanical behaviour of hard metals has mainly been characterised by bending strength [1-6], hardness and fracture toughness [1, 6-12]. Material data like yield strength and ultimate tensile or compressive strength are not as well established in the literature mainly due to the high hardness and the low ductility which makes the determination of reliable material data a challenging task. In general values of ultimate tensile strength and yield strength are higher in bending than in uniaxial testing conditions due to the differences in tested effective volume [13] and due to the non-linear stress distribution in the bent specimens which are partially deformed plastically before fracture [14].

The cyclic behaviour of hard metals was investigated by several authors [15-24]. The results from literature on microstructure and properties can be summarised as follows: For highly loaded hard metals the binder content ranges between 3 and 30 wt. %, the WC grain size between 0.1 and 20 μm. Hardness values vary from 1000 to 2200 HV30, fracture toughness varies from 7 to 20 MPa m\(^{1/2}\) and bending strength varies from 2000 to 4600 MPa. For ultrafine hard metals with a WC grain size of about 200 to 300 nm the maximum value of bending strength was found at about 20 wt. % Co [21]. Fracture toughness rises with an increase in mean free path in the Co binder (\(\lambda_{\text{Co}}\)) and there is also a dependence on the crystallographic structure (hcp/fcc ratio) of the binder [10]. In general WC grain size and hardness as well as hardness and fracture toughness are inversely related, although there are indications for an enhanced toughness to hardness ratio at extremely small WC grain sizes [11]. No complete understanding of the behaviour of toughness to hardness ratio in WC-Co hard metals with nanometer sized WC grains is available today [12].

Fatigue crack initiation at material inhomogeneities and subsequent subcritical fatigue crack growth (FCG) until the maximum stress intensity \(K_{\text{max}}\) at the tip of the growing crack approaches the fracture toughness \(K_{\text{IC}}\) can be regarded as the main factor affecting fatigue failure [15]. Under static loading cracks propagate through both, the carbides and the ductile binder phase [16], whereas during subcritical crack growth under cyclic loading conditions the crack mainly propagates in the ductile binder phase [17] that is possibly also subjected to a mechanically induced phase transformation from fcc to hcp [18]. According to Kursawe the localization of fatigue damage in shear bands and areas of transformed and therefore embrittled binder phase are reasons why the fatigue sensitivity is not a simple function of binder content in hard metals with an average WC grain size of 1 μm [19]. On the other hand Llanes states that the fatigue sensitivity, defined as the ratio between applied stress intensity factors corresponding to FCG threshold and fracture toughness, increases if the stress ratio \(R = \sigma_{\text{min}}/\sigma_{\text{max}}\) is decreased from 0.7 to 0.1 and the effective ductility of the constrained Co binder rises [20]. The materials become more fatigue sensitive as \(\lambda_{\text{Co}}\) increases from 0.15 μm to 0.75 μm and a transition from ceramic-like to metal-like fatigue crack growth behaviour is observed, i.e. the prevalence of \(K_{\text{max}}\) over \(\Delta K\) as the fatigue controlling parameter becomes less pronounced. Sailer’s investigations on ultrafine hard metals support Llanes’ findings with regard to fatigue sensitivity [21]. Fatigue limit rises with a decrease in \(\lambda_{\text{Co}}\) at room temperature [21] and also at elevated testing temperatures [22]. Roebuck finds significant higher fatigue life in hard metals at 300°C testing temperature compared to room temperature tests [23]. Varying testing temperatures up to 700°C [24] encounters lifetimes significantly below the values determined at room temperature. The behaviour of hard metals under cyclic loads cannot be deduced from their behaviour under static loading conditions [24].

This paper concentrates on static and cyclic material properties of high strength hard metals in uniaxial tension and compression loading and their dependence on the microstructure.

2. Experimental techniques and materials

The materials investigated in this work characterized by cobalt binder content, hardness, bending strength, average WC grain size and fracture toughness as given by the material manufacturer Ceratizit Austria are presented in Table 1. For all these hard metal grades a special kind of monotonous uniaxial test in compression and tension described in earlier work [25] was performed. The specimen geometry used in these tests is hourglass shaped i.e. of non-constant diameter with a minimum value of 6 mm [25]. At least three specimens of each grade were tested to determine Young’s modulus in compression and tension, fracture stress in tension as well as yield strength \(R_p\) and ultimate strength \(R_m\) in compression which does not provide a statistical significant data set but gives a first impression about the material’s properties. To investigate the role of stress ratio \(R\) on the value of the threshold of stress intensity range \(\Delta K_{\text{th}}\) at which FCG starts from micro-defects, load controlled cyclic tests at a test frequency of 10 Hz and a stress ratio \(R\) equal to 0.1, -1 and -3 were performed for the hard metal grade X8 with a binder content of 12 wt. % and a WC grain size of 0.2 to 0.5 μm. The cyclic stress-strain response including ratchetting behaviour was studied.
performing load controlled uniaxial cyclic tests at a test frequency of 0.25 Hz and a stress ratio $R_{\text{eff}}$ for the grades referred to as X2, X5 and X8. The grades vary in average WC grain size from fine (0.8 to 1.3 μm) over submicron (0.5 to 0.8 μm) down to ultrafine (0.2 to 0.5 μm). They exhibit a comparable binder content of 11.8 wt. % for the grade X2 and 12 wt. % for the grades X5 and X8.

Table 1. Materials tested under uniaxial loading conditions and corresponding typical parameters to characterize hard metals

<table>
<thead>
<tr>
<th>Grade</th>
<th>Co [wt. %]</th>
<th>HV 30</th>
<th>$\sigma_{\text{testing}}$ [MPa]</th>
<th>WC grain size [μm]</th>
<th>$K_{IC}$ [MPa$\cdot$\text{m}^{1/2}]</th>
</tr>
</thead>
<tbody>
<tr>
<td>X1</td>
<td>12</td>
<td>1340</td>
<td>3000</td>
<td>0.8 to 1.3 – fine</td>
<td>12</td>
</tr>
<tr>
<td>X2</td>
<td>11.8</td>
<td>1380</td>
<td>3000</td>
<td>0.8 to 1.3 – fine</td>
<td>12</td>
</tr>
<tr>
<td>X3</td>
<td>9</td>
<td>1470</td>
<td>2800</td>
<td>0.8 to 1.3 – fine</td>
<td>10.9</td>
</tr>
<tr>
<td>X4</td>
<td>6</td>
<td>1610</td>
<td>2300</td>
<td>0.8 to 1.3 – fine</td>
<td>9.9</td>
</tr>
<tr>
<td>X5</td>
<td>12</td>
<td>1460</td>
<td>4000</td>
<td>0.5 to 0.8 – submicron</td>
<td>9.9</td>
</tr>
<tr>
<td>X6</td>
<td>10</td>
<td>1590</td>
<td>3700</td>
<td>0.5 to 0.8 – submicron</td>
<td>9.4</td>
</tr>
<tr>
<td>X7</td>
<td>9</td>
<td>1590</td>
<td>3600</td>
<td>0.5 to 0.8 – submicron</td>
<td>10.4</td>
</tr>
<tr>
<td>X8</td>
<td>12</td>
<td>1730</td>
<td>4600</td>
<td>0.2 to 0.5 – ultrafine</td>
<td>7.5</td>
</tr>
<tr>
<td>X9</td>
<td>8.2</td>
<td>1930</td>
<td>4400</td>
<td>0.2 to 0.5 – ultrafine</td>
<td>7.5</td>
</tr>
<tr>
<td>X10</td>
<td>6.5</td>
<td>2080</td>
<td>4300</td>
<td>0.2 to 0.5 – ultrafine</td>
<td>7.2</td>
</tr>
</tbody>
</table>

3. Results and Discussion

3.1. Static uniaxial tests

The results of the uniaxial compression and tension tests of the investigated hard metal grades are displayed in Fig. 1 and listed in Table 2. The fracture strength in tension ranges from 800 MPa to 3300 MPa. As any other materials hard metals are not entirely free of defects which causes the scatter of the results. The Young’s modulus is not significantly different in tension and compression and varies between 480 and 600 GPa for the investigated grades as displayed in Table 2 along with the materials’ properties $R_{p0.05}$, $R_{p0.1}$, $R_{p0.2}$ and $R_{m}$ under compression loading. In case of tension loading all hard metal grades but the grades X1, X2, X5 and X8 containing the highest binder content exhibit a linear stress-strain behaviour before failure. The mentioned grades which contain 12 wt. % Co binder show some small amount of ductility also in tension loading. In case of compression loading even the hardest hard metal grades show a ductile behaviour. Plastic flow starts at yield strength levels ($R_{p0.05}$) between about 2400 and 5700 MPa. Please note that under compressive loading most samples did not fracture either after having reached the maximum load applicable by the testing equipment or the maximum strain value given for the chosen laser extensometer setup. Therefore values of compressive strength $R_{m\text{compression}}$ indicated by “>” in Table 2 are higher than the given values but their real values were not determined. The highest strength levels are achieved in the grades with the lowest binder content and the smallest WC grain size.

In the used uniaxial test [25] strain values need to be corrected via simulation due to the non constant sample diameter. Please note that stress-strain curves in compression were determined up to strain values of 3 % strain but due to reasons of simulation accuracy in strain correction [25] they are displayed only up to smaller strain values in Fig. 1.
Fig. 1. Stress-strain curves of hard metals in compression and tension

Table 2. Young’s modulus, yield strength $R_p$ in compression and ultimate strength of tested materials

<table>
<thead>
<tr>
<th>Grade</th>
<th>$E_{compression}$ [GPa]</th>
<th>$R_{p0.05}$ [MPa]</th>
<th>$R_{p0.1}$ [MPa]</th>
<th>$R_{p0.2}$ [MPa]</th>
<th>$R_{m,compression}$ [MPa]</th>
<th>$E_{tension}$ [GPa]</th>
<th>$R_{m,tension}$ [MPa]</th>
</tr>
</thead>
<tbody>
<tr>
<td>X1</td>
<td>509 ± 8</td>
<td>2560 ± 270</td>
<td>3000 ± 200</td>
<td>3560 ± 170</td>
<td>&gt;5100</td>
<td>497 ± 8</td>
<td>2391 ± 68</td>
</tr>
<tr>
<td>X2</td>
<td>497 ± 15</td>
<td>2610 ± 100</td>
<td>3030 ± 70</td>
<td>3570 ± 50</td>
<td>&gt;5420</td>
<td>513 ± 30</td>
<td>2663 ± 148</td>
</tr>
<tr>
<td>X3</td>
<td>540 ± 5</td>
<td>2870 ± 170</td>
<td>3440 ± 100</td>
<td>4080 ± 70</td>
<td>5610</td>
<td>528 ± 8</td>
<td>1426 ± 178</td>
</tr>
<tr>
<td>X4</td>
<td>566 ± 4</td>
<td>3690 ± 160</td>
<td>4250 ± 120</td>
<td>4850 ± 70</td>
<td>6470</td>
<td>557 ± 15</td>
<td>2110 ± 237</td>
</tr>
<tr>
<td>X5</td>
<td>521 ± 13</td>
<td>2710 ± 90</td>
<td>3210 ± 70</td>
<td>3810 ± 60</td>
<td>&gt;5770</td>
<td>496 ± 14</td>
<td>2910 ± 470</td>
</tr>
<tr>
<td>X6</td>
<td>524 ± 8</td>
<td>3470 ± 30</td>
<td>3980 ± 30</td>
<td>4580 ± 35</td>
<td>&gt;6580</td>
<td>528 ± 11</td>
<td>2761 ± 220</td>
</tr>
<tr>
<td>X7</td>
<td>533 ± 22</td>
<td>3690 ± 270</td>
<td>4150 ± 210</td>
<td>4740 ± 150</td>
<td>&gt;6330</td>
<td>551 ± 8</td>
<td>2433 ± 481</td>
</tr>
<tr>
<td>X8</td>
<td>533 ± 21</td>
<td>3570 ± 360</td>
<td>4150 ± 320</td>
<td>4890 ± 250</td>
<td>&gt;7580</td>
<td>515 ± 14</td>
<td>2539 ± 560</td>
</tr>
<tr>
<td>X9</td>
<td>540 ± 11</td>
<td>4820 ± 270</td>
<td>5370 ± 230</td>
<td>6080 ± 180</td>
<td>&gt;8500</td>
<td>535 ± 9</td>
<td>2449 ± 123</td>
</tr>
<tr>
<td>X10</td>
<td>590 ± 8</td>
<td>5710 ± 90</td>
<td>6280 ± 80</td>
<td>7000 ± 60</td>
<td>&gt;8500</td>
<td>592 ± 9</td>
<td>1500 ± 700</td>
</tr>
</tbody>
</table>

Fig. 2 shows the strong influence of the binder content and the WC grain size on yield strength $R_{p0.05}$, $R_{p0.1}$ and $R_{p0.2}$ under compressive loading. In earlier work of the authors the same kind of study was performed for a smaller number of hard metal grades [25]. It can be clearly seen that the values of the flow stress rise significantly as the mean WC grain size and binder content decrease. This can be explained by the increased resistance dislocations have to overcome when moving through the thinner binder channels: Channels between WC grains become larger.
and the constraint for the dislocation motion decreases in grades of same WC grain size but increasing binder content. The same happens if WC grain size is increased while the binder content is kept constant [25].

![Fig. 2. Effect of binder content and WC grain size on yield strengths Rp0.05, Rp0.1 and Rp0.2 in compression](image)

3.2. Cyclic uniaxial tests

Hard metal used for structural components may be subjected to a broad variety of loading conditions ranging from fluctuating tensile loads via alternating loads to fluctuating compression load. But, in many applications as tools hard metals are subjected to cyclic loads with a high compressive mean stress leading to stress ratios between $R = -1$ and $R = -5$, if residual stresses caused by local plastic deformation are taken into account. Stress ratio $R$ was thus varied from 0.1 via -1 to -3 in cyclic uniaxial tests at a test frequency of 10 Hz under load control for the hard metal grade X8 to study the influence of stress ratio on the cyclic material behaviour. Results of the fatigue tests, which were performed up to $10^6$ cycles, are summarised in the form of S-N curves in Fig. 3, Fig. 4 and Fig. 5.

The results indicate that the fatigue strength increases with decreasing stress ratio i.e. with increasing compressive mean stress. All S-N-curves show a significant scattering of the fatigue data and especially the results for $R = 0.1$ and $R = -1$ exhibit a rather low slope of the S-N-curve. Please note that the fatigue tests were performed at specimens with an hourglass shape and that in many cases fracture occurred not at the minimum sample diameter but at distances up to 10 mm away from the minimum diameter. Therefore, the S-N curves contain a set of data points that considers the nominal stress at the smallest cross section (filled squares) and a second set of data points for which the stress was determined at the position of the fracture origin (open squares).

As fracture origins pores or aggregates of WC grains that are larger than the average grain size or that have a reduced binder content were identified as crack origins. Different sizes of defects from which fatigue crack growth starts account for the large scatter of the data. Large crack origins generally cause low numbers of cycles to failures because only very little fatigue crack growth is necessary before the crack reaches its critical length which is determined by the local stress and the fracture toughness of the hard metal.
An estimate of the threshold of stress intensity factor range $\Delta K_{th}$ was determined on the basis of Murakami's $\sqrt{\alpha A}$ concept [26] using Eq. 1 and Eq. 2 for surface flaws and internal flaws respectively [26].

\[
\Delta K_s = 0.65 \Delta \sigma \sqrt{\alpha A} \quad \text{for fracture origins at the surface} \tag{1}
\]
\[
\Delta K_i = 0.5 \Delta \sigma \sqrt{\alpha A} \quad \text{for internal fracture origins} \tag{2}
\]

The symbol $A$ describes the projected area in the direction of the maximum stress, $\Delta \sigma$ represents the stress range ($\Delta \sigma$ equals two times the applied stress amplitude). The area $A$ of the origins of fracture was determined from scanning electron microscopy (SEM) images taken from fracture surfaces. For each fatigue crack origin the initial stress intensity factor range $\Delta K$ was calculated ($\Delta K_{\text{start}}$) on the basis of Eq. 1 or Eq. 2 taking the stress range at the position of the fracture origin and the projected area of the initiation flaw from the SEM micrographs. Those samples that gave the minimum value of $\Delta K_{\text{start}}$ at the respective stress ratio are indicated by grey circles in Fig. 3, Fig. 4 and Fig. 5.

Fig. 6 shows a typical example for the fatigue crack origins found in the investigated hard metals. In the immediate surroundings of the micro pore no “dimples” or “ridges” as they are typically formed during unstable crack growth [27] are visible. This is an indication of stable fatigue crack propagation in the material [27]. For clarification the approximate sizes of the fatigue crack origin and the area of stable fatigue crack propagation are indicated by white dashed lines in Fig. 6.
The lower bounds of the initial stress intensity factor range values $\Delta K_{\text{start}}$ found for the hard metal grade X8 at different stress ratios $R$ are displayed in Table 3. These values represent an estimate for the upper bound of $\Delta K_{\text{th}}$ values. The results indicate that $\Delta K_{\text{th}}$ rises with decreasing stress ratio i.e. with enhanced compressive mean stress similar to results found in the literature for tools steels [28].

Table 3. Minimum start values of stress intensity factor range at which FCG started in samples of hard metal grade X8 under various stress ratios

<table>
<thead>
<tr>
<th>$R$</th>
<th>$\Delta K_{\text{start, min}}$ [MPa(\sqrt{\text{m}})]</th>
</tr>
</thead>
<tbody>
<tr>
<td>0,1</td>
<td>4.3</td>
</tr>
<tr>
<td>-1</td>
<td>6.2</td>
</tr>
<tr>
<td>-3</td>
<td>9</td>
</tr>
</tbody>
</table>

3.3. Ratchetting behaviour

Another aspect of the cyclic material behaviour is the material response in cases in which the stresses exceeds the yield strength. Fig. 7 shows the stress-strain response for the hard metal grade X8 at a test frequency of 0.25 Hz and a stress ratio $R=-\infty$. This means that the sample is loaded between 0 MPa and a certain compressive stress value. It can be concluded from Fig. 7 that the largest plastic deformation occurs in the first loading cycle and there is a clear indication for ratchetting during subsequent load cycles. Please note that in this curve the necessary strain correction which accounts for the hourglass shape of the specimen was not performed.
The ratchetting behaviour of the three hard metal grades X2, X5 and X8 was studied in load controlled tests at a stress ratio $R = -\infty$. For each of the three grades five samples were subjected to five different stress ranges $\Delta \sigma$ that correspond to the yield strength values $R_{p0.05}$, $R_{p0.1}$, $R_{p0.2}$, $R_{p0.3}$ and $R_{p0.5}$. The development of the residual strain present in the stress free specimen as a function of the number of cycles is displayed in Fig. 8, Fig. 9 and Fig. 10. The experiments were performed up to 2000 cycles for cases in which the material response had stabilised, in other cases tests were performed also up to higher loading cycles. The results in Fig. 8, Fig. 9 and Fig. 10 indicate that ratchetting is completed in most cases within a few hundred load cycles. The results of the ratchetting experiments lead to the conclusion that the residual strain increases with increasing stress level and that finer WC grain sizes lead to less residual strain at the same stress level (see Fig. 11).
4. Conclusion

The main results can be summarised as follows:

- The stress-strain behaviour of WC-Co hard metals in uniaxial tension and compression loading up to the highest strength levels was determined.
- Static uniaxial tension and compression tests were performed at ten different hard metal grades varying in average WC grain size from “ultrafine” (0.2 to 0.5 μm) via “submicron” (0.5 to 0.8 μm) to “fine” (0.8 to 1.3 μm) and cobalt binder content ranging from 6 to 12 wt. %. The average strength levels of the investigated hard metal grades varied from about 1400 to 2900 MPa in case of tension loading and from about 5000 to 8500 MPa in case of compression loading. Onset of plastic deformation, as characterized by the Rp0.05 yield strength, takes place between about 2400 and 5700 MPa.
- S-N curves were determined for a hard metal grade with 12 wt. % cobalt binder content and an ultrafine WC grain size in stress controlled uniaxial tests at stress ratios R of 0.1, -1 and -3. Fatigue fracture typically originates from inhomogeneities such as micro pores or aggregates of WC grains. The fatigue strength of the hard metal grades is comparable to that of high strength steels, scatter of the results is mainly caused by a few larger inhomogeneities. The threshold of the stress intensity factor range at which fatigue crack growth starts from inhomogeneities was estimated to be about 4.3, 6.2 and 9 MPa√m at stress ratios of 0.1, -1 and -3 respectively.
- Cyclic stress-strain behaviour in case of loading to stresses higher than the elastic limit was studied under uniaxial loading conditions at a stress ratio R=∞. The results show that most of the plastic deformation happens in the first loading cycle followed by ratchetting during the subsequent loading cycles. Ratchetting slows down to zero after a few hundreds of load cycles. Residual strains are lowest for grades with ultrafine WC grain size at the same applied stress range.

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