Fatigue 2010

Microstructural effects on the fatigue crack nucleation in cold work tool steels

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Received 5 March 2010; revised 9 March 2010; accepted 15 March 2010

Abstract

Crack nucleation mechanisms in two types of cold work tool steels were evaluated under monotonic and cyclic loading conditions. The effect of the microstructural constituents: the primary alloy carbides and the tempered martensite matrix, and their interaction, was identified through fractographic analysis and determination of mechanical properties such as the bending strength, \( \sigma_b \), the fatigue limit, \( \Delta \sigma_{fat} \), and the fracture strength of the primary carbides under static tensile stressing, \( \sigma_{RC} \). The response under monotonic loading was found to be governed by the fracture of primary carbides. The cracks nucleated when the applied stresses were higher than the carbides fracture strength, accordingly depending on their properties, morphology and arrangement. Under cyclic loading, despite failure origins were located at the primary carbides and cracks emanating from them and propagating through the metallic matrix were evidenced, crack nucleation phenomena could not be explained as in monotonic loading since the applied stresses in fatigue were lower than the determined \( \sigma_{RC} \). Carbides fracture was then probably caused by damage observed in the metallic matrix. Primary carbides acted as stress concentrators and strain localization was more likely occurring in the matrix around them. Thus, fatigue failure was attributed to the destabilization of the tempered martensite of the matrix, induced by the strain localization processes around carbides, which produced their breakage and gave rise to fatigue propagating cracks.

Keywords: Carbides; Microstructure; Fatigue; Micromechanical; Toughness

1. Introduction

Intensive efforts on the automotive industry focus on further reduction of CO2 emissions and higher energy efficiency in all phases of vehicles life cycles. Reduction of fuel consumption and corresponding CO2 emissions can be achieved by development of an innovative sustainable structural design. Accordingly, the usage of Ultra high Strength Steels (UHSS) has been broadened in the past few years to satisfy the target conflict of weight reduction – safety related requirements with functionality – economic criteria.

Despite showing excellent strength to weight ratios, the extensive application of UHSS is being mainly hampered...
by the high loads acting on cold forming and cutting tools, which considerably decrease the tooling performance. Traditionally, failure mechanisms like wear or sinking were overcome by increasing the hardness levels of tools. However this strategy can no longer be applied for forming UHSS since high toughness levels are also required to avoid premature failure of tools. Forming tools are actually subjected to cyclic load solicitations, and thereby their fatigue performance must be well understood to increase their efficiency. Thus, the hardness – toughness relationship must be optimized through an appropriate microstructural design, taking into account not only the micromechanical properties of the metallic matrix and the embedded carbides, but also their interaction under cyclic loading conditions.

For this reason, during the last two decades some studies [1 - 9] have been published on the fatigue behavior of cold work tool steels. These works generally suggest that fractured carbides, or the debonding at the carbide – matrix interface, can initiate fatigue. Berns et al. [1, 2] performed fatigue fracture studies on conventional ingot metallurgical DIN 1.2379 (AISI D2) type cold work tool steel. The authors stated that the primary alloy carbides represent the largest microstructural discontinuity group with a high frequency of occurrence and can act as fatigue starting sites. Fukaura et al. [3] tested a JIS-SKD 11 (1.2379 type equivalent tool steel) up to $10^7$ cycles and also observed fatigue crack initiation at one or more primary carbides. They also determined, using Acoustic Emission (AE) techniques, the stress level needed to start breaking primary carbides, and this was about 1100 – 1800 MPa in JIS-SKD 11. However, after these results the authors found no explanation for the fracture of carbides in fatigue, since the fatigue limit was determined at about 700 MPa.

There is a lack of knowledge regarding the role of the primary carbide - metallic matrix interaction on fatigue behavior. Even though, some studies by Shiozawa et al. [4 - 6] on the fatigue properties of high-strength steels such as high speed steels and bearing steels, showed that at the regimes of the $10^7 - 10^9$ cycles micro-cracks tend to initiate around a defect particle by decohesion of small spherical carbides from the matrix, resulting in the formation of a short fatigue crack and forming the so called “Granular Bright Facet” (GBF). However, the work by Sohar at al. [7 – 9] also evidenced the presence of GBFs at the vicinity of carbide clusters in fatigue tests of 1.2379 up to the $10^{10}$ cycles. The better knowledge about the interaction between the metallic matrix and the primary and secondary carbides in cold work tool steels may help to explain the fatigue crack nucleation step.

The aim of this study is to better understand the effect of microstructural constituents of cold work tool steels, i.e. the primary alloy carbides and the tempered martensite matrix, on the crack nucleation during fatigue. The work also wants to clarify effects of carbide size, shape and arrangement on fatigue behavior by means of studying two different types of tool steels exhibiting different microstructures. The response of these steels under cyclic loading is compared to that under monotonic loading through fractographic analysis and determination of mechanical properties such as the bending strength, $\sigma^b$, and the fatigue limit, $\Delta\sigma^{\text{fat}}$, as well as the fracture strength of the primary carbides under static tensile stressing, $\sigma^{\text{RC}}$.

## Nomenclature

| Symbol | Description |
|--------|-------------|
| UHSS   | Ultra High Strength Steels |
| AE     | Acoustic Emission |
| PSB    | Persistent Slip Band |
| GBF    | Granular Bright Facet |
| $R$    | Fatigue load ratio |
| $\Delta\sigma^{\text{fat}}$ | Fatigue limit |
| $\sigma^b$ | Bending strength |
| $\sigma^{\text{RC}}$ | Fracture strength of carbides |
| $H$    | Hardness |
| $E$    | Elastic modulus |
| $K_C$  | Fracture toughness |
2. Experimental procedure

In this work, two wrought cold work tool steels were studied: a conventional DIN 1.2379 (AISI D2) and a special grade of tool steel developed by the company ROVALMA named as UNIVERSAL. These steels show different microstructural features in terms of arrangement, size and shape of primary alloy carbides. Their chemical composition is shown in Table 1. Heat treatments were applied in steels to get a hardness level of 60-62 HRC, as scheduled in Table 2.

The microstructure of the two studied steels is shown in Fig. 1. In 1.2379 steel the primary carbides (M₇C₃) are embedded in the matrix forming clusters and following bands along with the rolling direction. In UNIVERSAL two types of primary carbides are found (MC and M₇C₃) forming smaller clusters since the carbides are found more homogeneously distributed in the matrix. In addition, carbides in UNIVERSAL steel are of smaller size and more spherical shape. The microstructure of these two steels was investigated in a previous work by Picas et al. [10].

| Steel    | C    | Cr  | Mo   | V    | W  |
|----------|------|-----|------|------|----|
| 1.2379   | 1.5  | 1.6 | 11.0 | 0.6  | 0.8|
| UNIVERSAL| 0.9  | 1.2 | 6.8  | -    | 2.5|

Table 1. Chemical composition of the studied steels (in wt. %)

| Steel      | Austenizing (quench in oil) | Tempering   | HRC  |
|------------|-----------------------------|-------------|------|
| 1.2379     | 1050 ºC for 30 min          | 550 ºC for 2 h (x2) | 60 - 62 |
| UNIVERSAL  | 1060 ºC for 35 min          | 540 ºC for 2 h (x3) | 60 - 62 |

Table 2. Heat treatments applied to the studied steels and obtained hardness

Fig. 1. Tool steels microstructures: a) 1.2379 and b) UNIVERSAL
All specimens were machined to the desired geometry from cylindrical bars in annealed condition. Prismatic specimens of 5x4x50 mm were extracted with their axis in the rolling direction. Specimens were ground and the corners were rounded to avoid stress magnifications and machined flaws. The surfaces were polished to mirror-like finish. The bending strength, $\sigma_R$, was evaluated by three point bending in an articulated fixture to minimize torsion effects and using a universal testing machine.

Fatigue tests were conducted in a resonance machine at a load ratio $R = 0.1$ and at room temperature. The fatigue limit, $\Delta \sigma_{fat}$, was evaluated at $2 \times 10^6$ cycles using the “stair-case” method. Scanning electron microscopy was used to examine the fracture and fatigue surfaces.

The fracture strength of the primary carbides under static tensile stressing, $\sigma_{RC}$, in 1.2379 and UNIVERSAL, was determined by means of loading a sample at a stress level lower than $\sigma_R$, unloading and inspecting the surface using optical and confocal microscopy. The sample was loaded and reloaded with increasing applied loads until the first broken carbides were detected. The local stress acting on the carbide, calculated by measuring the distance from the central point of the sample (where the stress is the maximum in a three point bending test) was defined as the fracture strength of the carbide, $\sigma_{RC}$.

3. Results

The obtained $\sigma_R$ and the $\Delta \sigma_{fat}$ for 1.2379 and UNIVERSAL are summarized in Table 3. UNIVERSAL shows higher values of $\sigma_R$ and $\Delta \sigma_{fat}$. However and despite that $\sigma_R$ is clearly higher in UNIVERSAL than in 1.2379, $\Delta \sigma_{fat}$ in UNIVERSAL is only slightly above $\Delta \sigma_{fat}$ in 1.2379.

The inspection of the fracture surfaces revealed the presence of broken primary carbides, or carbides clusters, at the failure initiating sites in the two studied steels under both monotonic and cyclic loading (Fig. 2 and 3).

| Steel     | Bending strength, $\sigma_R$, MPa | Fatigue limit, $\Delta \sigma_{fat}$, MPa |
|-----------|---------------------------------|------------------------------------------|
| 1.2379    | 2874 ± 96                       | 856 ± 116                                 |
| UNIVERSAL | 3974 ± 119                      | 887 ± 48                                  |

Table 3. Bending strength, $\sigma_R$, and fatigue limit, $\Delta \sigma_{fat}$, of 1.2379 and UNIVERSAL steels

Fig. 2. Fracture surfaces at monotonic load: a) 1.2379 at $\sigma = 2760$ MPa; b) UNIVERSAL at $\sigma = 4017$ MPa
Fracture strength of carbides under static tensile stressing, $\sigma^{RC}$, is shown in Table 4 together with the crystallographic structure, chemical composition and mechanical properties of each type of carbide, determined in a
previous work by Casellas et al. [11]. Carbides M\(_7\)C\(_3\) and MC in UNIVERSAL steel show higher values of \(\sigma^{RC}\) than M\(_7\)C\(_3\) carbides in 1.2379, and that is understandable after their mechanical properties and morphology characteristics. Despite it is clear that the higher the carbide toughness, the higher their fracture strength, carbide clustering must be also taken into account. Rammerstorfer et al. [12] presented a hierarchical modeling strategy which allowed computer simulation of the thermo-mechanical behavior of high speed tool steels and showed that stresses were magnified at the vicinity of carbides, especially when those were distributed within clusters. Thus, the higher values of \(\sigma^{'RC}\) for carbides in the UNIVERSAL steel can be explained by their higher fracture toughness together with their lower size, more spherical shape and homogeneous distribution within the matrix.

Table 4. Crystallographic structure, chemical composition, mechanical properties [10] and fracture strength of the primary carbides in 1.2379 and UNIVERSAL steels

| Steel     | Carbide structure | Chemical composition (in wt. %) | H (GPa)   | E (GPa)   | Kc (MPa·m\(^{1/2}\)) | \(\sigma^{RC}\) (MPa) |
|-----------|-------------------|--------------------------------|-----------|-----------|----------------------|-----------------|
| 1.2379    | M\(_7\)C\(_3\)    | 40 - 45 Fe, 46 - 51 Cr, 6.0 - 6.5 V, 2 - 3 Mo | 18.2 ± 2.4 | 294 ± 17  | 2.3 ± 0.8            | 800 - 1200      |
| UNIVERSAL | M\(_7\)C\(_3\)    | 35 - 45 Fe, 35 - 45 Cr, 10 - 20 V, 2 - 4 W | 14.5 ± 2.5 | 270 ± 18  | 2.6 ± 0.9            | 2000 - 2800      |
| UNIVERSAL | MC                | 2 - 6 Fe, 8 - 10 Cr, 50 - 70 V, 7 - 12 W | 25.7 ± 3.6 | 315 ± 23  | 3.7 ± 0.6            | 2300 - 2800      |

4. Discussion

The fracture event during monotonic loading can be mainly explained by the crack nucleation at primary carbides and further propagation. The metallic matrix shows a more ductile behavior in comparison to the fragile hard carbides, which act as discontinuities for the homogeneous distribution of strain. The stresses generated at the carbide – matrix interface and the accumulated strain can produce the carbide fracture, as no signs of decohesion in primary carbides have been found yet neither by the authors nor by others [3, 8].

The stress is locally concentrated in the matrix at the vicinity of carbides depending on their size, shape and arrangement, as shows the work by Rammerstorfer et al. [12]. When such stress attains \(\sigma^{RC}\) the carbides break (Fig. 4 a)). The increasing load allows the matrix to attain plastic deformation at the surrounding of the nucleated crack in the carbide (Fig. 4 b)). As the load is progressively increased, the matrix locally deforms around the broken carbide until a small crack is formed at the highest deformation zones, as shown in Fig. 4 c). Further increasing load give rise to crack propagation to the final rupture.

![Fig. 4. a) M\(_7\)C\(_3\) primary carbide in UNIVERSAL steel broken between 2200 – 2400 MPa; b) the same carbide observed using confocal microscopy at 3000 MPa; c) the same carbide at 3400 MPa. Plastic deformation in the matrix is evident from confocal microscope topographies](image-url)
Fatigue failure origin in tested samples is located in or around primary carbides, as shown by fractographic observations (Fig. 3) and previous works [3]. However, when comparing the applied stress amplitude in fatigue with the fracture strength determined for carbides in monotonic loading (Fig. 5), it can be seen that the fatigue limit for the studied steels is far below the $\sigma^{RC}$ for the UNIVERSAL and close to $\sigma^{RC}$ for the 1.2379. Thus, crack nucleation under cyclic loading cannot be solely explained by the fracture of primary carbides because the applied stress is higher than $\sigma^{RC}$. It means that the crack nucleation mechanism in cyclic loading is not the same as in monotonic loading.

![Graph showing bending strength $\sigma^{b}$, fatigue limit $\Delta \sigma^{fat}$, considering that the load ratio $R = 0.1$; $\Delta \sigma^{max}$ = $\Delta \sigma^{fat}/0.9$ and fracture strength of carbides determined under monotonic loading, $\sigma^{RC}$, for 1.2379 and UNIVERSAL.](image)

Fig. 5. Bending strength, $\sigma^{b}$, fatigue limit (in terms of the maximum applied stress in fatigue, $\Delta \sigma^{max}$, considering that the load ratio $R = 0.1$; $\Delta \sigma^{max}$ = $\Delta \sigma^{fat}/0.9$) and fracture strength of carbides determined under monotonic loading, $\sigma^{RC}$, for 1.2379 and UNIVERSAL.

Aimed at understanding fatigue crack nucleation in primary carbides, the interaction between the carbide and the metallic matrix during cyclic loading should be analyzed. A first explanation may be the carbide decohesion from the metallic matrix, but there are no experimental evidences for such behaviour. Moreover, broken carbides with a crack emanating from them and propagating through the metallic matrix during cyclic loading can be clearly discerned at the polished surface of runout specimens (Fig. 6), meaning that fatigue cracks are associated with broken carbides.

It is clear that carbides do not fail by fatigue, since they are brittle monocrystals and as it is well established for brittle materials, only polycrystal ceramics undergo fatigue. However, that is not the case of the metallic matrix, which consists of tempered martensite with secondary precipitates. A detailed inspection of the fatigue runout specimens shows the presence of damage localization in the tempered martensite of the matrix, as shown in Fig. 7. Damage can be associated to the destabilization of the tempered martensite structure, induced by bands where strain gets localized. Fukaura et al. [3] already perceived a cyclic softening during fatigue under stress-controlled conditions in the overpeak-tempered JIS-SKD 11 steel, which was attributed by the authors to the formation of Persistent Slip Bands (PSBs) and vein formation.

Dougherty et al. [13] studied the fatigue crack closure in a 1070M steel, with a tempered martensite microstructure, and attributed the progressive decrease in cyclic stress response at low strain amplitudes to the formation of slip bands. They stated that in absence of local stress and strain concentrators, the intensified slip bands are the most likely sites for crack initiation. This observation by Dougherty et al. [13] is especially relevant to analyze the behavior of cold work tool steels. The fracture of primary carbides under fatigue at stress amplitudes lower than their $\sigma^{RC}$, seems then plausible by the strain localization in the martensitic matrix around carbides, since they can act as stress concentrators.

Mughrabi [14] stated that in cyclic loading, a major part of fatigue life is spent in the bulk microstructural
processes leading to cyclic strain localization in the form of PSBs and subsequent crack initiation. That observation could be one explanation for the close values of $\Delta \delta_{\text{fat}}$ found for 1.2379 and UNIVERSAL. If the major part of the fatigue life is expected to be spent by the formation of PSBs in the tempered martensite matrix, regardless of the presence of the primary carbides strain localization will take place by similar mechanisms. In both steels the metallic matrix consists in tempered martensite, with the same hardness (the hardness evaluated by nanoindentation was 9.9±0.3 GPa in 1.2379 and 9.9±0.4 GPa in UNIVERSAL. Such similar metallic matrix for both steels could explain the small differences in fatigue, despite the great differences in the fracture strength, where the behavior is mainly governed by the properties of the primary carbides.

Fig. 6. Fatigue propagating cracks starting from broken primary carbides at the surface of fatigue UNIVERSAL specimens

Fig. 7. Damage localization observed in the matrix of the 1.2379 fatigue specimens

5. Conclusions

Based on fatigue and fracture results and fractographic observations, it has been determined that crack nucleation mechanisms under cyclic loading are distinct from those under monotonic loading. The role of each microstructural constituent of tool steels, the primary carbides and the tempered martensite matrix, and their interaction in the crack nucleation processes have been identified:

- In monotonic loading cracks nucleate at the primary carbides when the applied stress is higher than the fracture strength of the carbide. Accordingly, tool steels having tougher, smaller and more homogeneously distributed primary carbides show higher fracture strengths.
- Under cyclic loading, failure origins are still located at the primary carbides. However, they do not break at the first load cycle, i.e. carbides break at lower applied stresses than their fracture strength and some load cycles are needed to nucleate cracks. Damage has been observed in the metallic matrix, probably due to strain localization of the tempered martensite. Primary carbides can act as stress concentrators and thus,
localization of strain in the matrix is more likely to take place at their vicinity. Such strain localization around carbides can induce their fracture, giving rise to fatigue propagating cracks.

- Thus, the similar metallic matrix for both 1.2379 and UNIVERSAL steels could explain the similar fatigue behavior results; meanwhile, the distinct carbide properties and morphologies might be the responsible for the clear differences observed under monotonic loading.

Acknowledgements

This work is part of the tasks held by CTM and UPC in the Forma0 research project. It is funded by CDTI (National Board for Technological and Industrial Development) within the frame of Spanish National funds.

References

[1] Berns H, Trojahn W. Einfluss der Wärmebehandlung auf das Ermüdungsverhalten ledeburitischer Kalartbeitsstähle. VDI-Z 1985;127(22):889-92.
[2] Berns H, Lueg J, Trojahn W, Wählung R, Wisell H. The fatigue behavior of conventional and powder metallurgical high speed steels. Powder Metall Int 1987;19:22-6.
[3] Fukaura K, Yokoyama Y, Yokoi D, Tsuji N, Ono K. Fatigue of cold work tool steels: effect of heat treatment and carbide morphology on fatigue crack formation, life and fracture surface observations. Met Mat Trans A 2004;35A:1289-300.
[4] Shiozawa K, Lu L, Ishihara S. S-N curve characteristics and subsurface crack initiation behavior in ultra-long life fatigue of a high carbon-chromium bearing steel. Fatigue Fract Engng Mater Struct 2002;24:781-90.
[5] Shiozawa K, Lu L. Very high-cycle fatigue behavior of shot peened high carbon-chromium bearing steel. Fatigue Fract Engng Mater Struct 2002;25:813-22.
[6] Shiozawa K, Morii Y, Nishino S, Lu L. Subsurface crack initiation and propagation mechanisms in high-strength steel in a very high cycle fatigue regime. JSME Int Journal A 2006;49(1):1-10.
[7] Sohar CR, Betzwar-Kotas A, Gierl C, Weiss B, Danningher H. Gigacycle fatigue behavior of a high chromium alloyed cold work tool steel. Int Journal Fat 2008;30:1137-1149.
[8] Sohar CR, Betzwar-Kotas A, Gierl C, Weiss B, Danningher H. Fractographic evaluation of gigacycle fatigue crack nucleation and propagation of a high Cr alloyed cold work tool steel. Int Journal Fat 2008;30:2191-2199.
[9] Sohar CR, Betzwar-Kotas A, Gierl C, Weiss B, Danningher H. Influence of residual stresses on gigacycle fatigue response of high chromium cold work tool steel. Mat.-wiss. U. Werkstofftech. 2008;39(3):248-56.
[10] Picas I, Hernández R, Casellas D, Casas B, Valls I. Mechanical performance of cold forming tools. In: Beiss P, Broeckman C, Francke S, Keyselitz B, editors. Proceedings of the 8th International Tooling conference, Aachen; 2009, p. 1037-48.
[11] Casellas D, Caro J, Molas S, Prado JM, Valls I. Fracture toughness of carbides in tool steels evaluated by nanoindentation. Acta Mater 2007;55:4277-86.
[12] Rammerstorfer FG, Plankensteiner AF, Fischer FD, Antretter T. Hierarchical models for simulating the mechanical behavior of heterogeneous materials: an approach to high speed tool steel. Mat Sci Engng 1999;A259:73-84.
[13] Dougherty JD, Srivatsan TS, Padovan J. Cyclic stress response, strain resistance and fracture behavior of modified 1070 steel. Engng Fract Mech 1996;53:589-47.
[14] Mughrabi H. On “multi-stage” fatigue life diagrams and the relevant life-controlling mechanisms in ultrahigh-cycle fatigue. Fatigue Fract Engng Mater Struct 2002;25:755-64.