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Damage evolution in sinter-hardening powder-metallurgy steels during tensile and fatigue loading

G. Straffelini *, M. Benedetti, V. Fontanari

Department of Industrial Engineering, University of Trento, via Mesiano, 77 – 30100 Trento, Italy

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ABSTRACT

The damage approach was used to compare the tensile and push-pull fatigue behaviour of two high-strength sinter-hardened powder metallurgy (PM) steels, with a density of 6.7 and 7.2 g/cm^3. In both alloys, tensile damage was found to start when the ratio of the applied stress to UTS was greater than about 0.3. The tensile damage was due to localized yielding and, in a later stage, to the formation of several micro-cracks that joined to form more than one macrocrack. Fatigue damage was followed in the finite fatigue life regime and was found to develop through three stages. During the first one, damage increased with fatigue cycling up to the attainment of a plateau (second stage) that lasted to a fraction of about 0.9 of the fatigue life. The damage recorded during the second stage was very similar to that encountered in the tensile tests at the same stress to UTS ratio. The formation of microcracks was observed in the third stage only, when the fraction of fatigue life was greater than 0.9. During this stage damage increased very sharply and this was due to the growing and joining of the microcracks to form one long crack able to lead to final fracture.

1. Introduction

High-strength steels (HSS) produced by PM are characterized by a bainitic/martensitic microstructure and are attractive candidates for replacing conventional wrought steels for the production of different mechanical parts [1,2]. In this respect, HSS production by sinter-hardening is particularly promising for different reasons. It combines sintering and heat treatment in one step, porosity remains free of oil since no oil quenching is required, the distortion of complex shaped parts is limited, the production costs are reduced [3–7].

The mechanical response of PM alloys is strongly influenced by the interplay between porosity and matrix microstructure [8–10]. As far as porosity is concerned, both tensile and fatigue properties are directly related to the fraction of load bearing section, Φ, which is mainly dictated by the total porosity content. Different models have been proposed to evaluate Φ [11,12]. Danninger et al. [13,14] showed that the simplified closed cell foam model proposed by Ashby [15] is the most effective in predicting the experimental results:

$$\Phi = \left(\frac{E}{E_0}\right)^2 \tag{1}$$

where E is the Young's modulus measured on the PM sample and E_0 is that of the matrix. Pores reduce the fraction of load-bearing section but they also may act as stress and strain concentrators during loading. This last effect is very important when matrix hardness is sufficiently high (i.e., it is greater than about 400 HV0.1 [16]) to avoid extensive plastic deformation at pores edges thus preventing local stress relaxation. This is clearly the case of PM HSS steels, for which pores may act as pre-existing cracks in the microstructure thus inducing a brittle damage during tensile or fatigue loading [5,17].

The damage evolution during tensile or fatigue loading may be successfully characterized by using the Kachanov parameter D [18], whose meaning is very close to the concept of plane porosity introduced by Slesar et al. [19]. D is defined by the following equation [20]:

$$D = 1 - \Phi \tag{2}$$

By comparing Eqs. (1) and (2) it is obtained that D can be monitored by evaluating changes in Young's modulus as a function of tensile strain or cycling loading. This approach has been used for different types of materials, including metals, polymers and ceramics [21,22]. In this respect, the residual porosity in PM alloys can be regarded as pre-existing damage in the materials microstructure.

* Corresponding author. Tel.: +39 0461 282458; fax: +39 0461 881977.
E-mail address: Giovanni.straffelini@ing.unitn.it (G. Straffelini).

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In the case of PM materials with a soft matrix, the increase in damage during tensile loading is given by the increase in the pores volume [7,9]. Two stages were recognized in the case of PM iron [23]. In the first stage (for low nominal plastic strains), plastic deformation is highly localized at the pore edges and the damage rate is very high. In the second stage, bulk plastic deformation takes place and the damage rate is lower.

So far the damage approach has not been widely used to investigate the tensile and fatigue damage in PM alloys, specially in HSS steels. In a number of papers, Chawla and coworkers [24–27] investigated the static and fatigue damage evolution in some HSS PM steels, and they found that damage initially increased very slowly with the fatigue cycles and only in proximity of the end of the fatigue life damage was found to increase sharply. The authors attributed damage evolution to an extensive plastic deformation at the necks and to microcracks formation and propagation.

In the present investigation, the role of density on the damage evolution during tensile and push-pull fatigue of sinter-hardened steels produced by Distaloy DH1 powder have been investigated. The evolution of internal damage calculated by means of Eqs. (1) and (2), was followed as a function of the applied load and the fatigue life fraction for tensile and fatigue tests respectively. The micro-mechanisms of damage were also investigated by observing selected regions on the surface of the tensile and fatigue specimens. Aim of the work was to highlight the possibility of using this approach to characterize the mechanical damage in sinter-hardened HSS, and to compare the tensile and push-pull fatigue damage evolution in this type of materials.

2. Materials and experimental procedures

The geometry of the specimens for tensile and fatigue testing is shown in Fig. 1 (ISO 3928 norm [28]). The specimens were produced in an industrial plant by mixing Distaloy DH1 powder (nominal chemical composition: 1.5%Mo prealloyed, 2%Cu diffusion alloyed, provided by Hoganaes AB, Sweden), with 0.8% lubricant and enough graphite to reach a sintered carbon level of 0.6%. Compaction was carried out to a density level of 6.8 and 7.2 g/cm^3. Delubrication was then carried out at 400 °C followed by sintering at 1120 °C for 30 min in an endogas atmosphere. Cooling was performed in two steps: slow cooling to 900 °C followed by rapid cooling at about 2.5 °C/s to room temperature, thus combining sintering and quenching. To improve the mechanical properties tempering was finally carried out at 220 °C. The specimens were tested in the as treated state, without any subsequent geometrical modification.

The microstructural characteristics of the materials were studied in detail in a previous investigation [17]. An example of the matrix microstructure and pore morphology is shown in Fig. 2. The microstructure was found to comprise martensite as predominant phase (around 80% in volume) with some amounts of upper and lower bainite, and small amounts of retained austenite. For both materials microhardness was in between 580–590 HV 0.1. Information on the pore content and morphology was achieved by optical microscopy interfaced with an image analyser. The

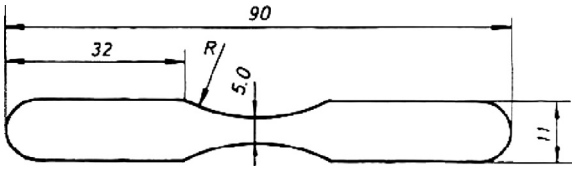

Fig. 1. Tensile and fatigue sample (the dimensions are in mm and the thickness is 5 mm).

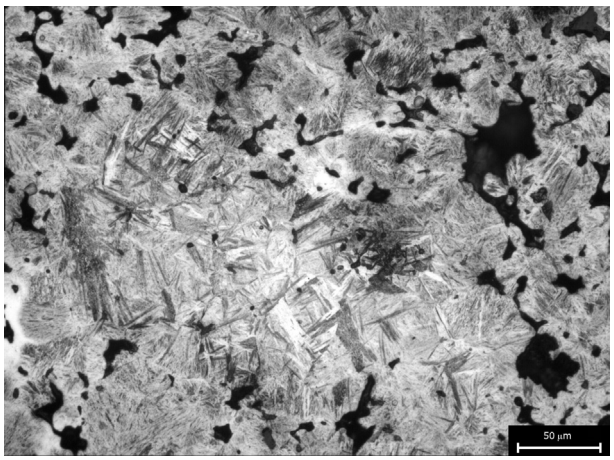

Fig. 2. Microstructure of the material with density 7.2 g/cm^3 after etching with Nital 2% to reveal the microstructure.

porosity area was found to be 14.4% and 7.9% for the materials at the two densities. The average shape factor, defined as the ratio between area and diameter of each pore, was found to be around 0.5 for both materials. As known, the shape factor ranges between 0 and 1, being equal to unity for a circular and perfectly smooth pore [29,30]. The maximum size of pores ranged between 40 and 50 µm for both materials.

Tensile tests were carried out load controlled at room temperature with an Instron Testing machine. The cross-head speed was 0.5 mm/min, and the gauge length of the extensometer was 12.5 mm. At regular load intervals the test was interrupted and a partial unloading to a stress amounting to about 10% of UTS was carried out. The elastic modulus, E, was determined as slope of the line fitting the unloading curve. Preliminary tests showed that it was not necessary to perform several unloading-reloading cycles to obtain reliable measurements of E.

Axial fatigue tests (stress ratio: $R = 0$) were carried out at room temperature and in load control, using a 20 kN Rumul resonant testing machine. The S–N curves were published in previous investigations [17,31]. In the present study further tests were carried out in the finite fatigue life regime with the aim of obtaining a fatigue life of about 10^5 cycles, in order to allow subsequent comparison with specific literature data [24,25,32]. On the basis of the S–N curves, stress amplitudes of 200 and 302 MPa for the materials with 6.8 and 7.2 g/cm^3 respectively were selected. In order to estimate the damage accumulation with fatigue cycling the following procedure was adopted. Load blocks of 5000 cycles were carried out at a frequency of 5 Hz followed by one cycle at a frequency of 0.0025 Hz to allow for a reliable Young's modulus evaluation.

To detect the micro-mechanisms of damage during tensile and fatigue testing, the specimens were mirror polished before testing and inspected during the test at specific intervals during the unloading ramps by means of a stereomicroscope mounted on the testing machine.

3. Results and discussion

3.1. Tensile behaviour

Fig. 3 shows the tensile stress–strain curves of the two materials under study. It can be observed that both materials are characterized by a continuous yielding behaviour with an early deviation from linearity. This is quite typical for HSS obtained by powder metallurgy, and it is attributed to the formation of microplastic regions at the pore corners during tensile loading [10,16,17]. The