CHARACTERIZATION OF FATIGUE CRACK FORMATION IN MECHANICALLY SURFACE TREATED AUSTENITIC STAINLESS STEEL

I. Altenberger [1], B. Scholtes [2], U. Noster [2], R. 0. Ritchie [1]

[1] Lawrence Berkeley National Laboratory, University of California, Berkeley, USA
[2] Institute of Materials Technology, University Gh Kassel, Kassel, Germany

ABSTRACT

Mechanical surface treatments such as shot peening or deep rolling are very effective tools to improve the fatigue life and endurance strength of cyclically loaded components. By introducing compressive residual stress and strain hardening in the surface layers, all stages of fatigue are significantly altered, from the first dislocation movements (cyclic hardening/softening) until the eventual propagation of micro- and macro-cracks.

Metastable austenitic stainless steels are particularly attractive for mechanical surface treatment as they exhibit very high strain hardening due to the martensitic transformation and a very complex near-surface microstructure, e.g. a thin two-phased surface layer of nanocrystallites.

In order to clarify the differences in crack formation of polished and of surface treated (shot peened or deep rolled) material states, scanning electron microscopy (SEM) studies have been carried out with conventional and high resolution instruments on samples of austenitic stainless steel AISI 304 fatigued under stress control.

The results clearly showed a distinct difference in damage mechanism depending on the surface state. Whereas polished surface states exhibited crack formation preferentially at sites of extensive multiple planar slip, cracks in mechanically surface treated states were mostly formed in a brittle manner without observable surface slip. Local loss of coverage during the shot peening process, however, led to microscopically non-peened regions which exhibited early crack initiation by formation of slip bands.

Interestingly, these preferential microcracks were rarely the crack sites from which fatal cracks originated, which suggests that the damage process is not only controlled by physical crack initiation but mainly by crack propagation conditions for microcracks. A slip line-induced microcrack formation can also be promoted in deep rolled specimens if thin surface layers of approximately 5-l 0 microns are elecrolytically removed prior to cycling, thus also removing the nanocrystalline surface regions which impede slip line formation.

KEYWORDS

shot peening, deep rolling, nanocrystallization, crack initiation

INTRODUCTION

Mechanical surface treatments can lead to very complex and depth-dependent microstructues in metastable austenitic stainless steels, such as AISI 304. A typical deformation mechanism in AISI 304 is deformation-

induced martensitic transformation [1,2,3] leading to typically 25-45% a'-martensite after deep rolling or shot peening, respectively, in near surface layers [4,5,6]. The formed martensite is lath-like and heavily twinned with high dislocation densities in the austenitic matrix. Additionally, deformation bands and nanocrystalline surface regions are formed with grains as small as 20 nms, extending into a depth of 1-2 microns (Fig. 1) [4].
Fatigue investigations on mechanically surface treated AISI 304 revealed pronounced lifetime and endurance strength improvements, especially after deep rolling [4]. Since the resultant near-surface compressive residual stress profile remained only partially stable in low cycle (higher strain) fatigue, in contrast to surface treatment-induced near surface microstructures which remained stable throughout cyclic loading, the fatigue life improvement can primarily be attributed to a higher microstructural fatigue resistance in near surface regions [4,7]. The specific role of different contributing microstructures for fatigue life-improvement of mechanically surface treated materials has not been defined in the literature. Therefore, in the present work an attempt is made to characterize the effect on fatigue of the most striking microstructural feature, -namely the nanocrystalline layer- in mechanically surface treated AISI 304 by investigating deep rolled specimens and specimens before and after the nanocrystalline layer has been removed electrolytically prior to cycling.

EXPERIMENTAL PROCEDURES

Rotation-symmetrical unnotched specimens of AISI 304, with a gage length of 10 mm and a diameter of 5 mm, were deep rolled with a "ball-point" rolling device, using a hydrostatic spherical rolling element and a rolling pressure of 150 bar (the composition and microstructure of the steel are given in ref. [4]). One set of specimens was then electrolytically polished for 20 seconds with a buthanol/perchloric acid-electrolyte in order to remove the surface treatment induced nanocrystalline surface layer. The surface roughness R_Z before and after polishing was around 0.5 microns.
Tension/compression fatigue tests were performed under stress control without mean stresses (R = -1) with a cycling frequency of 5 Hz and a stress amplitude of 365 MPa. The fatigue tests were interrupted just before fracture at the onset of macro-crack propagation; practically this was done at the point where the hysteresis loops as measured with a clip-on extensometer began to buckle. The surface crack length was then approximately 1 mm. The specimens were then taken out of the servohydraulic testing device and the surface crack topography was investigated with a Camscan S4- and a LEO 1550 Gemini (Field emission)-scanning electron microscope (SEM).

RESULTS

The fatigue damage mechanisms in the polished, un-surface treated (reference) state is characterized by extensive slip line formation at the surface (Fig. 2). Typically, in AISI 304 multiple planar slip along (11 1)-planes occurs at high stress amplitudes. The planar slip character is due to the small stacking fault energy of AISI 304 which is around 20 mJm^{-2} [8]. It has been repeatedly shown that slip line formation is a precursor to microcrack formation in electrolytically polished AISI 304 [9]. Preferential cracks emanate from the border between slipped and unslipped material due to strain incompatibilities and stress concentration, as has been shown in SEM and atomic force microscopy [10].
In mechanically surface treated (e.g. deep rolled or shot peened) AISI 304 the surface fatigue damage mechanism is quite different (Fig. 3). Here, only an extremely low amount of surface plasticity (increase of surface roughness) around crack flanks and crack tips was detected. This lack of plasticity can be attributed to the surface treatment-induced nanocrystalline layer which effectively impedes dislocation movement and hence slip line formation. According to [1 1], dislocation pile-ups in nanocrystallites with a grain size of 20 nm or smaller are not effective enough to promote dislocation movement or Frank-Read-sources in neighbouring grains, thus only allowing the formation of brittle cleavage cracks [12] to relieve stress concentrations. Other characteristic features of fatigue damage in mechanically surface treated

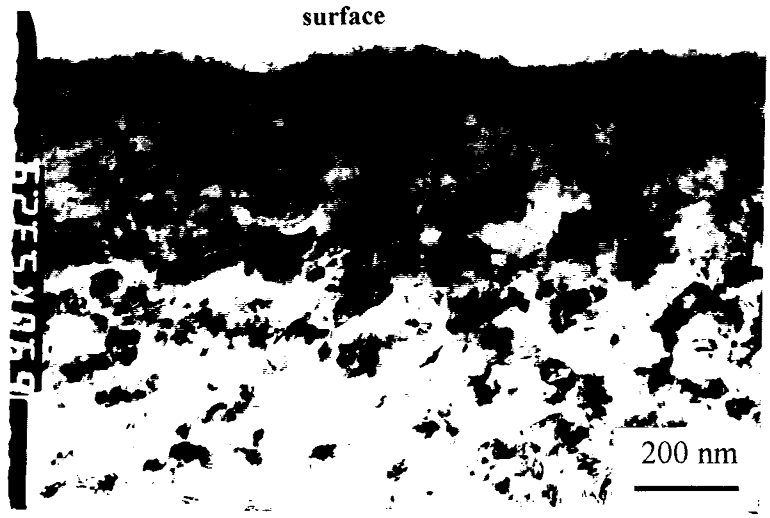

Fig. 1: Nanocrystalline layer in direct surface regions of shot peened (Almen intensity 0.175 mmA) AISI 304 (TEM-bright-field image)

Fig. 2: Typical planar (111)-slip lines in polished un-surface treated AISI 304 after high strain-fatigue, σ_a = 320 MPa, N = 1000 cycles (SEM-image, backscattering contrast)

AISI 304 compared to polished non-treated specimens are generally lower crack densities and lower crack openings as a consequence of compressive residual stresses. Crack formation without surface slip has also been reported for mechanically surface treated ferritic steels, where no nanocrystallites but extremely high dislocation densities in surface region existed [13,14].

Fig. 4 shows the microtopography of a fatigue crack in deep rolled and (prior to cyclic loading) electropolished AISI 304 for a stress amplitude of 365 MPa. The electropolishing treatment led to a layer removal of about 5-10 microns at the surface. In this case, the wake of the crack as well as the crack tip (Fig. 5) exhibit pronounced plasticity in form of slip lines that can be seen in high concentrations directly at the crack flanks but in lower density also as far away as 100 microns from the crack. Compared to Fig. 3, the surface roughening by plasticity is very pronounced here. Slip lines were found to initiate cracks but were also created by stress concentrations at the crack tip (Fig. 5). It should be noted that there was no significant difference in fatigue lifetime between this deep rolled state and the deep rolled state incorporating a

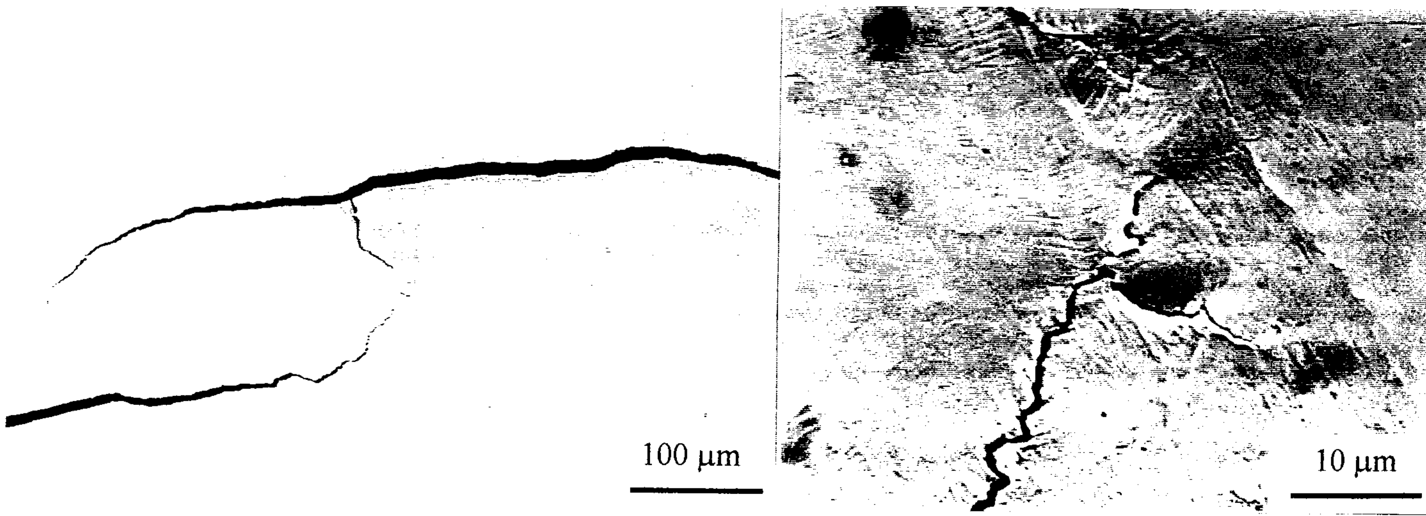

Fig. 3: Typical fatigue cracks in non-polished deep rolled AISI 304 (σ_a = 365 MPa, N = 68000 cycles)

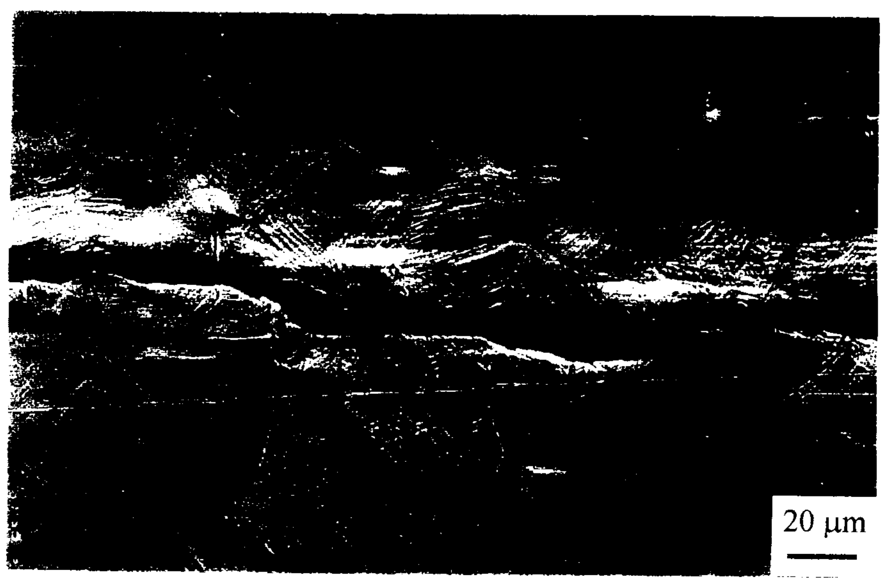

Fig. 4: SEM-Micrograph of a typical fatigue crack in deep rolled and (prior to cyclic loading) electrolytically polished AISI 304, showing extensive slip-line formation (σ_a = 365 MPa, N = 82000 cycles)

nanocrystalline surface layer. The number of cycles to fracture was N_f = 82800 for the deep rolled and prior to cycling polished state compared to N_f = 71700 for the as deep rolled unpolished state. This is within the statistical scatter of fatigue life of this material. In short, we can summarize, that the nanocrystalline surface layer does not seem to have any positive effect on the fatigue behaviour of AISI 304 under stress controlled cyclic loading. This result is consistent to similar investigations on AISI 304, where the ratio of surface-strain hardened regions to soft core regions was smaller and where a different stress amplitude was used, resulting in shorter fatigue lives of approximately 10000 cycles[14].

It is not yet clear why a thin nanocrystalline surface layer does not signifcantly influence fatigue lifetime in spite of totally different surface damage processes. One possible explanation could be that crack initiation in deep rolled samples investigated here actually takes place beneath the surface, thus rendering the surface processes insignificant. In fact, hints for a subsurface crack formation at a depth of 2-5 microns beneath the nanocrystalline surface layer have been found in [14], where cross sections of deep rolled fatigued samples

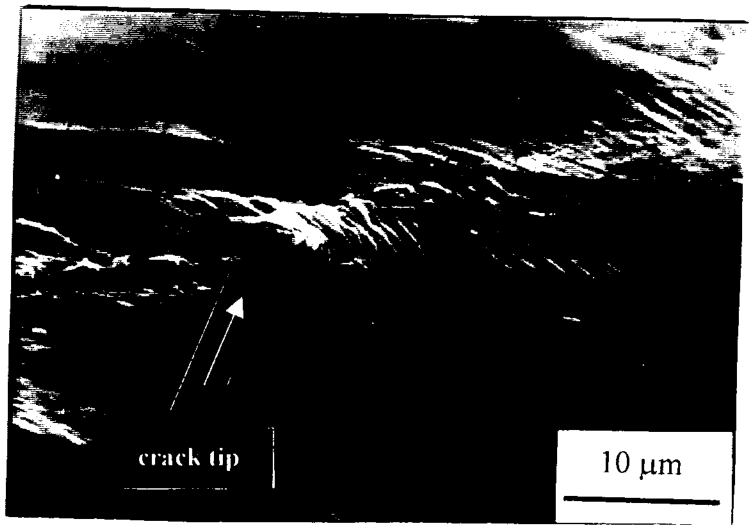

Fig. 5: Slip lines in the plastic zone of a fatigue crack tip in deep rolled and (prior to cyclic loading) electrolytically polished AISI 304 (σ_a = 365 MPa, N = 82000 cycles)

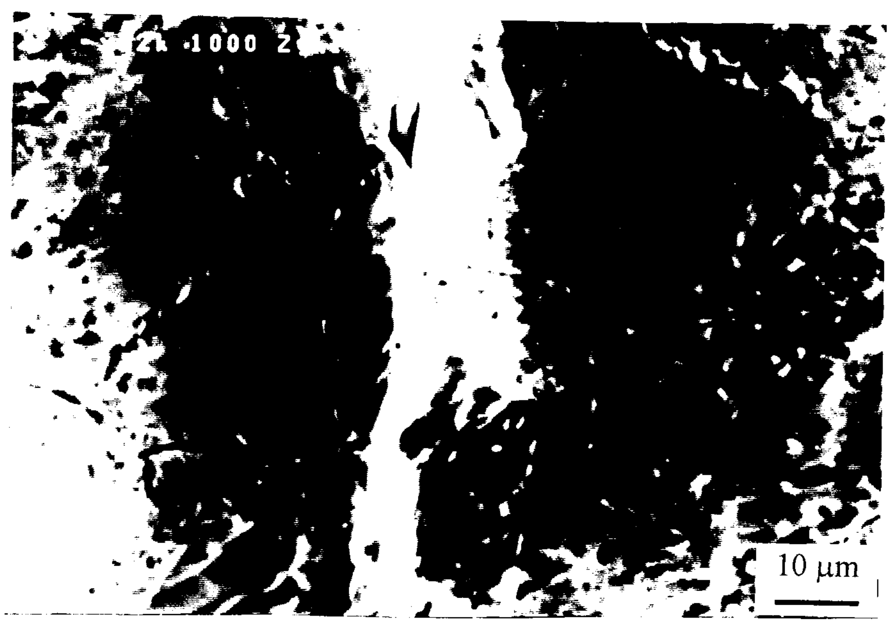

Fig. 6: Fatigue crack formation in an area of localized slip in shot peened (Almen intensity 0.175 mmA) and fatigued (σ_a = 320 MPa, N = 1000 cycles) AISI 304

revealed highest crack densities in these regions. Moreover, it has been shown in [15], that local fatigue life estimations for ductile mechanically surface treated materials are not as successful as concepts averaging over the fatigue behaviour of the whole sample like e.g. Manson-Coffin [16,17] or Smith Watson-Topper [18]. In the case investigated here, plastic strain amplitudes were identical for the polished and unpolished state, since the volume fraction of the removed layer was extremely small.

In practical applications, even in surface treated materials local plastic surface fatigue damage in form of slip lines can occur: For example, during shot peening a local loss of coverage can lead to locally non-strain hardened regions, which are prone to the same fatigue damage mechanisms as polished surfaces. Fig. 6 shows such a slip line-region in shot peened fatigued AISI 304 where a microcrack is readily formed. However, these cracks were not found to affect the fatigue life significantly and only rarely developed into 'fatal' cracks, since their propagation was retarded or stopped as soon as they encountered adjacent shot peened regions.

CONCLUSIONS

Mechanically surface treated specimens of AISI 304 exhibit brittle crack formation with little or no plasticity in direct surface regions in stress-controlled fatigue tests. This is attributed to a thin nanocrystalline surface layer. The totally different amount of surface plasticity of this layer, however, does not seem to have any significant influence on the extent of damage, since specimens without nanocrystalline surface layers exhibited quite similar fatigue lifetimes than those with nanocrystallites at the surface.
In shot peened specimens, regions not covered by the shot peening process could be found showing slip line formation similar to non-peened polished specimens. These regions were preferential sites for crack formation; however, only rarely would they develop into a fatal macro-crack, suggesting that local residual stress fields and microstructures play an important role in early crack propagation [19,20].

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