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Investigation of residual stresses in high-cycle loaded laser steel welds

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Abstract. The paper contains results of a comprehensive experimental programme aimed at an evaluation of residual stresses in and near laser welds of pressure vessels steels using X-ray diffraction technique. Generally, the manufacturing processes of a machine component introduce residual stresses that have an important and may have even essential influence on their behaviour during service life. The goal of this contribution was to find the influence of high-cycle loading on the stresses´ relaxation in and near the welds. It has been found that the surface residual stress state is not the most important factor, either positive or negative, but the fatigue life more or less depends on the specific situation and condition of the materials and welds. On the other hand, the fatigue loading test caused the reduction of microdeformation values. The analysis of microstructure changes by X-ray diffraction could be an indicator of the future cracks initialization and their propagation.

Introduction

The fatigue life of an object with a laser weld or, more generally, its service life depends on numerous structure parameters and can be significantly affected by the state of residual stresses. Even though residual stresses not yet belong to standard parameters for certification of welding processes, they can be considered as an important characteristic of the weld quality [1]. The aim of the laser welding process is not only to produce a joint but also to avoid the generation of undesirable stresses that would significantly promote crack propagation and cause the distortion of the final object. It has been found that distortion and the macroscopic residual stresses due to welding are two mutually affected phenomena and, therefore, upon controlling the residual stresses by means of laser beam energy and welding speed, one can influence the undesirable distortion as well.

Note that the macroscopic residual stresses (RS) can significantly promote fatigue resistance and decelerate life reducing fatigue initiation processes in the weld or bulk material. Moreover, the compressive RS can reduce crack growth from different defects like pores of lack of fusion if they are negative and contrarily if they are positive.

High cycle fatigue damage is a complicated process which can be divided into two categories: micro- and macrostructure changes. The microstructure changes contain two stages. Stage I. is associated with the movement, multiplication of dislocations and development of specific layout inside the material and grains. In case of high cycle fatigue loading with the stress amplitude close to endurance limit, the main effect of these movements is a reduction of microdeformations (microstresses or microscopic stresses). Cyclic softening or hardening may be associated phenomena in some materials. Stage II. corresponds to clustering of dislocations, formation of persistent slip bands, then extrusions and intrusions resulting in an occurrence of microcracks with a crystallographic character and orientation in the first stage. These microcracks result in a smooth reduction of RS and microdeformations in relation to Stage I. Due to macrocracks propagation up to final failure, Stage III. of macrostructure changes show a very high decrease of micro- and macrostresses values [2,3]. The relaxation of the mentioned structure parameters strongly depends on their original values. The main reduction of macroscopic residual stresses is connected with Stage III. It should be pointed out, however, that the reason of this reduction in the Stage III. are not fatigue processes themselves, but simply RS relaxation resulting from the material disruption. The same or similar stress relaxation would occur after any kind of cutting, which is being used, when some of the destructive methods of RS analysis are applied, like e.g. the contour method.

Experiment

The samples for high-cycle fatigue test were prepared from plates made of P355 and P460 steels (marked as "N" and "H") with a ferritic-pearlitic microstructure. The plates of thicknesses 8 mm and 10 mm were double-side laser-welded (the first-welded side is marked as "UP" and the second one as "DOWN", see Fig. 2), for welding parameters see Tab. 1.The X-ray diffraction analysis were performed only in the direction parallel to the axial axis of the samples for high-cycle fatigue test, i.e. in the direction perpendicular to the weld, see Fig. 1.

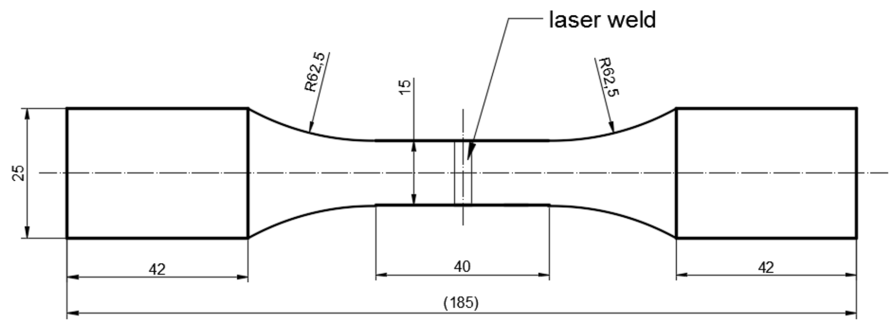

Fig. 1: Shape and dimensions of the fatigue test specimen

Table 1: Welding parameters; P – power of laser beam, v – welding speed, λ – wavelength of laser beam

P [W]	v [mm·s^{-1}]	mode	λ [nm]
3000	5.5	Continuous	900–1080

High-cycle fatigue tests were performed at stress ranges sufficiently low, bellow expected endurance limit, not to cause failures, with the aim to evaluate effects of the cyclic loading on possible relaxation of residual stresses. Load asymmetry was R = 0, frequency 33 Hz, stress range 206 and 185 MPa for the N10 and H8 specimen, respectively. The target number of cycles was 5 M and 17 M for the first and second loading test, respectively.

The *X'Pert PRO MPD* diffractometer was used to measure lattice deformations of the ferrite phase (α-Fe) using the Cr$K\alpha$ radiation. Diffraction angles $2\theta^{hkl}$ were determined from the peaks of the diffraction lines $K\alpha_1$ of the planes *{211}* of ferrite. The *Rachinger's* method was used for the separation of the diffraction lines $K\alpha_1$ and $K\alpha_2$ and diffraction lines $K\alpha_1$ were fitted by the

Pearson VII function. To determine RS, the *Winholtz & Cohen* method [4] and X-ray elastic constants $\frac{1}{2}s_2 = 5.75$ TPa^{-1}, $s_1 = -1.25$ TPa^{-1} were used.

The diffraction patterns were obtained using the *X'Pert PRO MPD* diffractometer in the Bragg-Brentano geometry with Co$K\alpha$ radiation. The Rietveld refinement performed on the *MStruct* software [5] was used for calculation of the values of microdeformations. The microstresses can be calculated using Hook's law $\sigma^{micro} = Ee$, where e and E are microdeformation and Young's modulus, respectively.

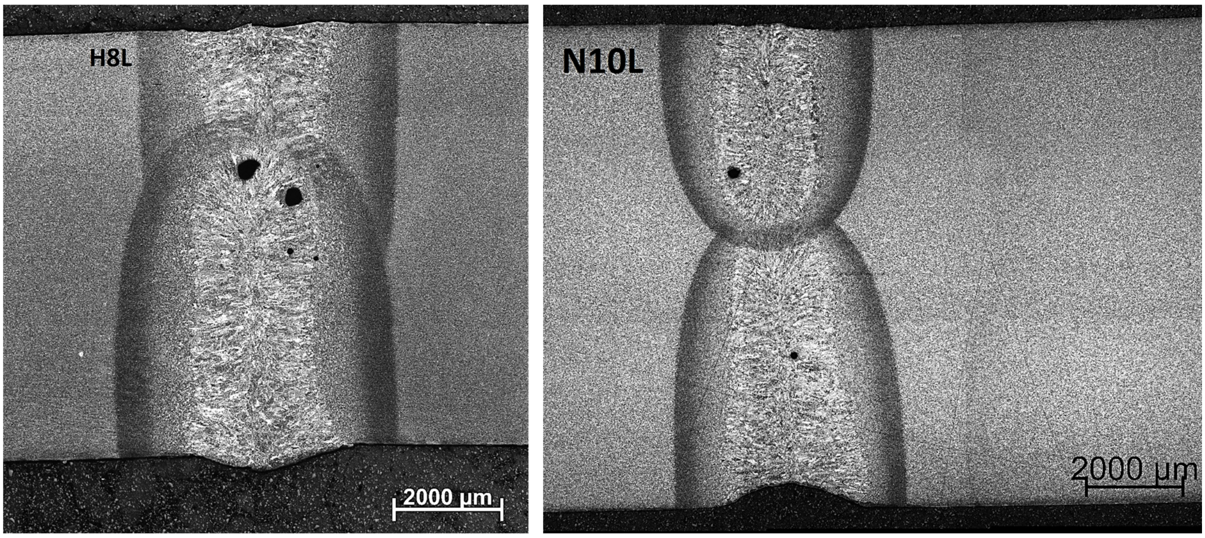

Fig. 2: Macrostructure of weld cross-section after etching

Results and discussions

Surface distributions of RS can be seen in Fig. 3. It is evident, in the case of the H8 sample, that the RS are compressive unlike the N10 sample, where they are tensile on the UP side. After the 1st (or the 2nd) fatigue cycle loading no statistically significant changes in the RS values for the H8 sample were observed. On the contrary, unexpected, premature failure of the N10 sample occurred after approx. 800000 cycles. The crack was initialized on the UP side on the range of the weld fusion and heat affected zones, where the RS were (slightly) tensile. As a result of the cracking, rather than the fatigue loading as mentioned in the Introduction, the RS relaxed to insignificant values.

More significant effect of high-cycle loading is evident from a microscopic point of view. From diffraction patterns, microdeformations and phase contents were determined using Rietveld refinement, see Figs. 4–5. All three stages, see Introduction chapter, is possible to find in Fig. 4. It was found that after the 1st fatigue cycle loading there were almost no changes (correlation with Stage I.) but after the 2nd fatigue cycle loading (correlation with the transition from Stage I. to Stage II.), the values of microdeformations of H8 sample were reduced more than two times. This effect is doubtless, interesting and important. It can be explained by the movement of dislocations, their recombination, and formation of different dislocation microstructure layout, which is a very long-time process occurring for a very high number of cycles, may last several tens of million cycles. Whether this phenomenon was connected with an occurrence of microcracks or arrested physically short cracks, it is not known. Nevertheless, further investigations in this direction would be useful and could provide a more exact explanation. The major decrease/relaxation of microdeformations of the N10 sample was evidently caused by the total specimen disruption as a result of the long main fatigue crack and final failure.

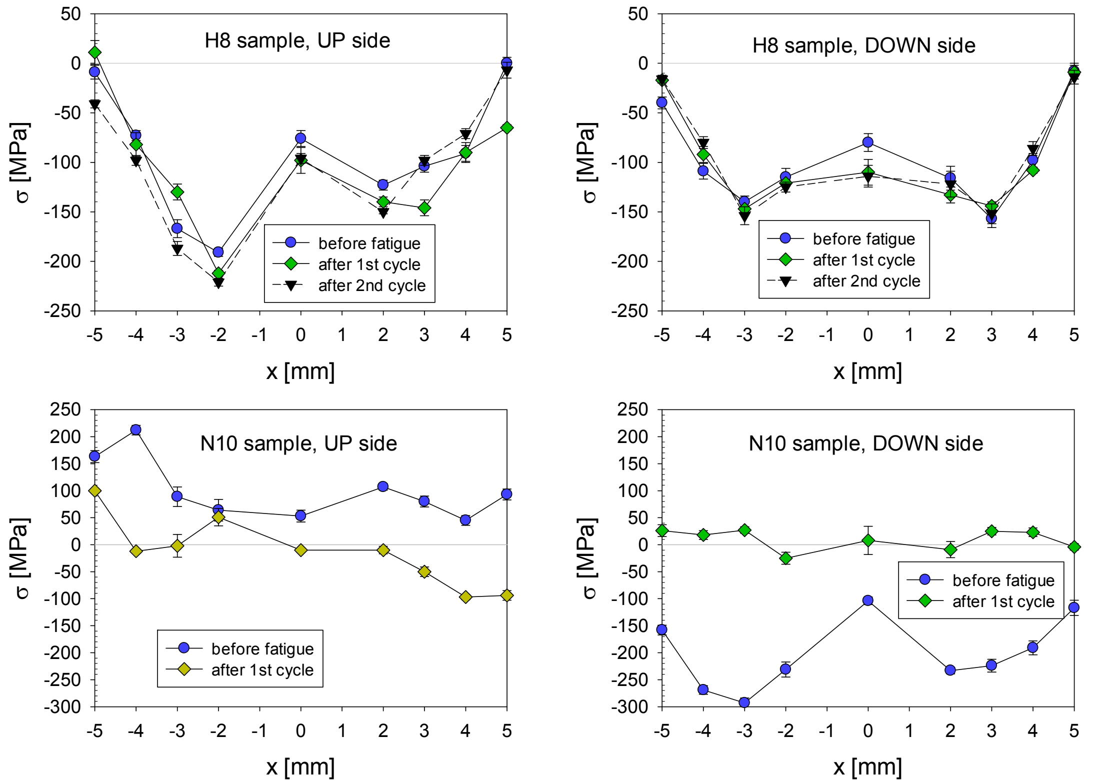

Fig. 3: Residual stresses in respect to the position from weld axis

In the process of initiation or propagation of fatigue cracks, the oxide layer can play in general an important role [6]. The premature fatigue failure of the N10 sample actually was initiated at the fusion zone on the surface with a thin oxide layer of the thickness less than 20 μm. Note that fusion zone represents a microstructural notch and is therefore typically the weakest area of the weld. The tensile RS could slightly accelerate the fatigue initiation process [7] by shifting the means stress to positive values and so increasing the load asymmetry. But such effect was likely small as the RS values were not so high in comparison with the stress amplitude.

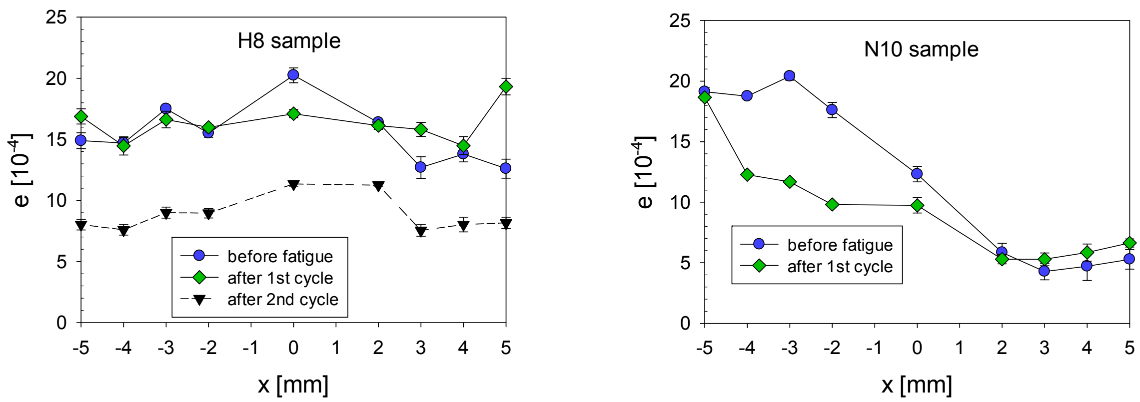

Fig. 4: Influence of high-cycle loading on the values of microdeformations on the UP side of samples, where x denote distance from the weld axis

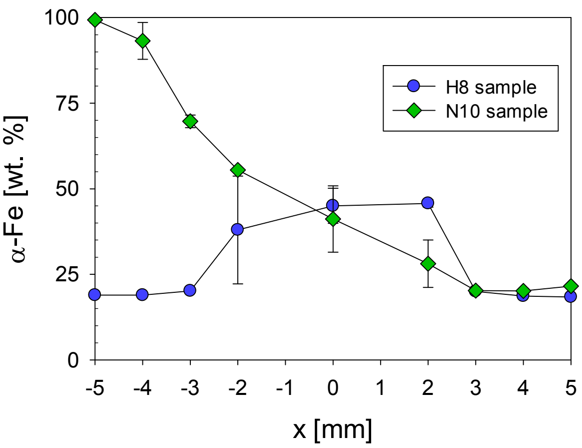

Fig. 5: Ferrite (α-Fe) content in respect to the position from weld axis in the surface layers on the UP side of samples

Conclusions

Results of the study of fatigue loading effects on the macroscopic residual stresses (RS) confirmed negligible effects of high-cycle fatigue loading with the stress amplitude close to the endurance limit on the RS. No statistically significant changes in the RS values for the H8 sample were observed. In the case of cracking of the N10 sample during fatigue loading, the RS values decreased and relaxed almost to zero due to macroscopic disruption. On the other hand, the fatigue loading test of H8 sample caused the significant changes of the microdeformation values after a high number of cycles, namely 17 million. This phenomenon can be explained by changes of dislocation layout and structure inside the material and grains as the first step of the fatigue crack initiation process. For the cracked N10 sample, the reduction of microdeformations values was observed in the area with tensile RS, high microdeformation, the thin oxide layer and within the heat affected zone. The analysis of microstructural changes by non-destructive X-ray diffraction in the surface layers indicated that this method could be a good indicator of the future cracks initialization and propagation. However, further investigations in this field would have to be performed.

Acknowledgement

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