PREDICTION OF FATIGUE CRACK GROWTH IN MARTENSITIC HIGH STRENGTH LOW ALLOY STEELS

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ABSTRACT

The fatigue crack growth behaviour of ferritic pearlitic steels can be predicted by the model of Roven and Nes assuming that the damage in the cyclic plastic zone in front of the fatigue crack is the same like in an low cycle fatigue specimen. The low cycle fatigue data and the ferrite grain size as the primary dislocation barrier will be used for the prediction of the fatigue crack growth behaviour. The model shows deviations from experimental data in case of high strength low alloy steels with martensitic structure. Therefore the model was improved. The influence of the stress ratio on the fatigue crack growth curve and the threshold value were considered by using the damage parameter of Smith Watson and Topper instead of the Manson-Coffin curve. The primary dislocation barrier in martensitic structures is the size of packets which was used in the model. There is a well correspondence between the experimental data and the predicted fatigue crack growth. The influence of crack closure effects on the threshold was not considered. This results in an underestimation of the threshold.

KEYWORDS: fatigue crack growth, HSLA-steel, microstructure, LCF-damage model

INTRODUCTION

As suggested by Roven and Nes the fatigue crack growth of ferritic pearlitic steels can be predicted with the help of low cycle fatigue data [1]. For this prediction the ferrite grain size and the striations have to be known. Extensive experimental researches have shown, that the proceedings in the cyclic plastic zone before the fatigue crack tip are the same as the proceedings in the LCF-specimen. It also turned out, that the mean values of the striations are constant in a wide area of the fatigue growth. The striations increase with increasing da/dN-values over a critical value of cyclic stress intensity factor of $\Delta K'_{eff} > 20$ MPa√m. It follows that in an area $\Delta K_{eff} < \Delta K'_{eff}$ obviously several cycles are necessary to create a striation. Up to know most crack propagation models on the base of the LCF-concept require one cycle for the creation of a striation. There is a good correspondence between the data predicted by the model of Roven and Nes and the experimental data measured at ferritic steel. The applicability of this model to martensitic high strength low alloy steels has to be examined.

THE MODEL OF ROVEN AND NES

Roven and Nes assume that the cycles N_S for the creation of one striation can be calculated by the striation width s and the macroscopic fatigue growth velocity da/dN:

$$N_S = \frac{s}{da/dN} \tag{1}$$

N_S can be calculated by the part of Manson-Coffin curve with the plastic strain range at the crack tip $\Delta\varepsilon_{ap\,tip}$:

$$\Delta\varepsilon_{ap\,tip} = 2 \cdot \varepsilon_f^{'} \cdot (2N_S)^c = 2 \cdot \varepsilon_f^{'} \cdot \left(\frac{2s}{da/dN}\right)^c \tag{2}$$

There is a relation between $\Delta\varepsilon_{ap\ tip}$ and the cyclic crack tip opening displacement $\Delta\delta$:

$$\Delta\varepsilon_{ap\ tip} = \frac{\Delta\delta}{d} \tag{3}$$

d is the primary dislocation barrier. If the linear fracture mechanics is valid, there exist an relation between $\Delta\delta$ and the cyclic stress intensity factor:

$$\Delta\delta = \frac{\Delta K_{eff}^2}{E \cdot R'_{p0,2}} \tag{4}$$

ΔK_{eff} is the effective cyclic stress intensity factor corrected with the value of crack closure. Roven and Nes only took the amplitude of the cyclic yield $R'_{p0,2}$ into consideration instead of the range $2 * R'_{p0,2}$.
From Eqn. (3) and (4) follows:

$$\Delta K_{eff} = \sqrt{2 \cdot \varepsilon'_f \cdot E \cdot R'_{p0,2} \cdot d \cdot (2N_s)^c} \tag{5}$$

So that for the fatigue crack growth is valid:

$$\frac{da}{dN} = 2s \cdot \left(2 \cdot \varepsilon'_f \cdot d \cdot E \cdot R'_{p0,2}\right)^{1/c} \cdot \Delta K_{eff}^{-2/c} \tag{6}$$

Roven and Nes showed, that the da/dN-curves of a low strength steel can be described by the model very well. Because of this the applicability of the model to martensitic steels will be examined. In table 1 the mean values for striation and packet width are given.

TABLE 1

STRIATION AND PACKET WIDTH OF HSLA S890 AND THE BUTT WELD [2]

	Striation width [µm]	Packet width [µm]	Prior austenit grain size [µm]
S890 base metal (BM)	0,32	9,4	17
S890 heat affected zone (HAZ)	0,38	27,1	55
S890 weld metal (WM)	0,28	20,9	64

Figure 1 shows the application of the model to martensitic high strength steel. There are deviations between the predicted and the experimental fatigue crack growth curve.
The limits of the model are that the threshold condition is not available automatically and the influence of the stress ratio on the fatigue crack growth curve is assumed to be caused only by crack closure effect. Crack growth measurements by Hück [3] on steel show a shifting of fatigue crack growth curve by the factor 1,3 if the stress ratio is changed from R = 0 to −1. This is an effect without crack closure because Hück has used the ΔK_{eff} - values for correction. This is in accordance with the fact, that crack closure effect arises at low stress intensity ratios.

IMPROVEMENT OF THE MODEL BY ROVEN AND NES

For improving the model following assumptions have been made:
The crack tip opening displacement has an elastic and a plastic part so the transition into the elastic region I of da/dN-curve can be described. The influence of the stress ratio on the da/dN-curve was considered by using the damage parameter of Smith, Watson and Topper [4] instead of using the Manson-Coffin curve. The yield stress $R'_{p0,2}$ in Eqn. (2) for describing the cyclic behaviour has to be doubled ($2*R'_{p0,2}$). The transition to the accelerated fatigue crack growth is considered with the term $\Delta K_{fc}/(\Delta K_{fc} - \Delta K/(1-R))$.
First we calculate the cycles N_s for a given da/dN-value and a constant striation width s. With the help of Eqn. (7) it is then possible to calculate the damage parameter P_{SWT}.

$$P_{SWT} = \sqrt{\left(\sigma_f'\right)^2 \cdot \left(2N_s\right)^{2b} + E \cdot \sigma_f' \cdot \varepsilon_f' \cdot \left(2N_s\right)^{b+c}} \tag{7}$$

σ'_f, b, ε'_f and c are constants of Manson-Coffin-Curve $\varepsilon_a = \frac{\sigma_f'}{E} \cdot (2N)^b + \varepsilon_f' \cdot (2N)^c$.

P_{SWT} is defined as:

$$P_{SWT} = \sqrt{\sigma_o \cdot \varepsilon_{a\,tip} \cdot E} \tag{8}$$

With the maximum stress $\sigma_o = \sigma_{a\,tip} * 2/(1-R)$ follows:

$$P_{SWT} = \sqrt{\sigma_{a\,tip} \cdot \varepsilon_{a\,tip} \cdot E} \cdot \sqrt{2/(1-R)} \tag{9}$$

and for the strain amplitude at the crack tip:

$$\varepsilon_{a\,tip} = \frac{(1-R) \cdot P_{SWT}^2}{2 \cdot E \cdot \sigma_{a\,tip}} \tag{10}$$

For the calculation of ΔK_{eff} a separation from $\varepsilon_{a\,tip}$ and the cyclic yield stress $R'_{p\,0,2}$ in the region of the primary dislocation barrier d is necessary (Eqn. 3 and 4).

$$\Delta\delta = 2 \cdot \varepsilon_{a\,tip} \cdot d = \frac{\Delta K_{eff}^2}{E \cdot 2 \cdot R_{p0,2}'} \tag{11}$$

That would mean a large-scale procedure by using the cyclic stress-strain-curve and the Neuber's rule. We consider, that $\sigma_{a\,tip}$ is in the order of the cyclic yield $R'_{p0,2}$ and can be used directly for the further calculations.

$$\frac{\Delta K_{eff}^2}{E \cdot 2 \cdot \sigma_{a\,tip}} = 2\varepsilon_{a\,tip} \cdot d \tag{12}$$

$$\Delta K_{eff} = \sqrt{d \cdot 2 \cdot \varepsilon_{a\,tip} \cdot E \cdot 2 \cdot \sigma_{a\,tip}} \tag{13}$$

This assumption would make the calculation easier, because $\sigma_{a\,tip}$ and $\varepsilon_{a\,tip}$ have not to be determined by the Neuber's rule. So Eqn. (13) is simplified:

$$\Delta K_{eff} = \sqrt{2 \cdot d \cdot (1-R)} \cdot P_{SWT} \tag{14}$$

P_{SWT} has to be calculated with Eqn.(1) from the cycles N_S needed for the creation of one striation s. With the help of Eqn.(14) the ΔK_{eff} -value can be calculated with a given da/dN-value. This given da/dN-value has to be corrected with the term Eqn. (15) because in the region III of the da/dN-curve an acceleration has to be taken into account.

$$\left(\frac{da}{dN}\right)_{korr} = \frac{da}{dN} \cdot \frac{\Delta K_{fc}}{\Delta K_{fc} - \frac{\Delta K}{1-R}} \tag{15}$$

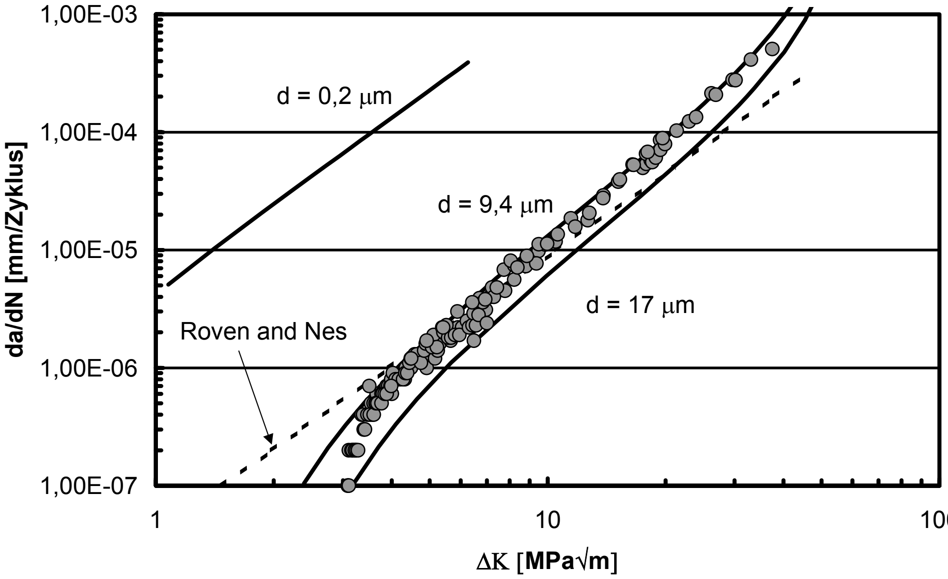

Figure 1: Influence of the primary dislocation barrier on the position of the fatigue crack growth curve (base metal S890 R=0,5)

The microstructure constant d introduced by Roven and Nes is in the case of ferritic steels similar to the grain size, because the high angle tilt boundaries have an influence as primary dislocation barriers. For martensitic and bainitic steels lattice width, packets and prior austenitic grain size are a possibility. Figure 1 illustrates, that for the base metal of S890 only the packets show an agreement between experiment and calculation. Normally lattices in the packets are not arranged exactly in the same parallel orientation. During the martensitic transformation the lattices in one packet get randomly orientated. Naylor [6] showed, that although this various orientation exists, the crack growth during a brittle fracture takes place without large angle deviations in the packets, because the several possible brittle fracture planes are located in a small angle region. Only at packet boundaries larger deviations occur. The neighbouring lattices in a packet have small angle deviations for plastic deformation, because the fracture planes are identical with the sliding planes. Lattice boundaries are a smaller barrier for plastic deformation than the packet boundaries. Because of this, packet boundaries should be defined as primary dislocation barrier. The packet boundaries correlates with the austenitic grains, but the packets are smaller then. In the case of heat affected zone, the prior austenitic grains are not verifiable anymore, so that only the using of packets is meaningful. With the values for striations s and the packets d (given in table 1) and the values of the Manson-Coffin-Curve (table 2) the fatigue crack growth curves for the high strength steel S890 and its welding joints were calculated (with Eqn.16 and 17) with different stress ratios.

TABLE 2

PARAMETER FOR THE CALCULATION [2]

	σ_f' [MPa]	b	ε_f'	c	ΔK_{fc} [MPa√m]
S890 BM	1605	-0,1	0,9924	-0,77	106
S890 HAZ	1254	-0,07	1,104	-0,8	98,5
S890 HAZ	1254	-0,07	1,104	-0,8	98,5

The figures 2 to 4 show a very good correspondence. The influence of the microstructure on the threshold region of da/dN-curve can be explained with the larger primary dislocation barrier d. When the packet width for the calculation of fatigue crack growth curve in the HAZ and in the WM is used, then the shifting of da/dN-curve to larger ΔK_{eff} -values can be understood. Also the influence of stress ratio on da/N-curve is reflected correctly. There are deviations from threshold value when the stress ratio R and the ΔK-values are small. These deviations can be explained by the crack closure effects, which were not considered.

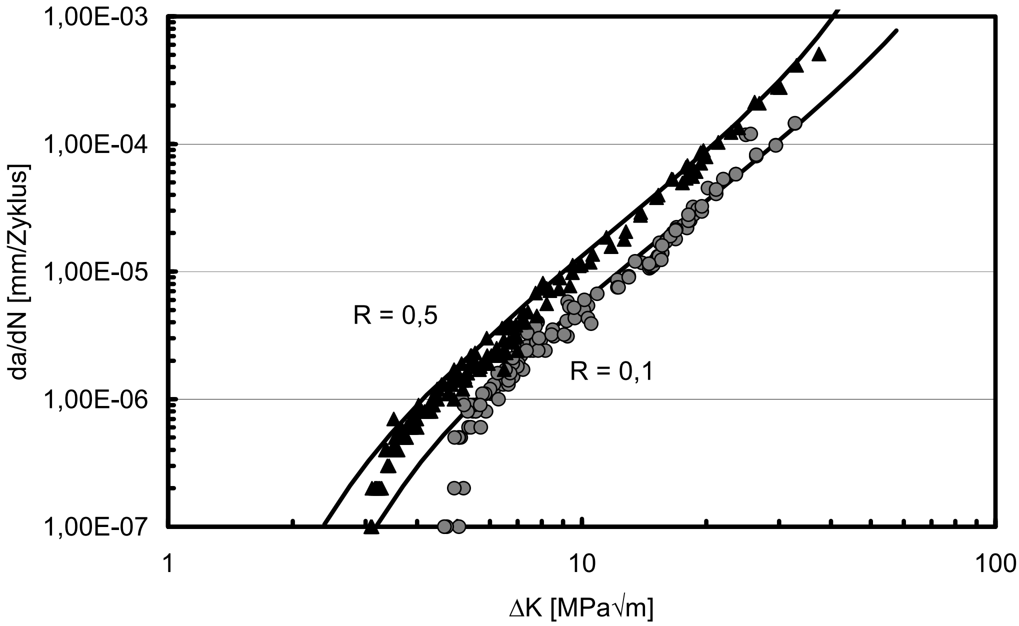

Figure 2: Experiment and calculation for S890 base metal

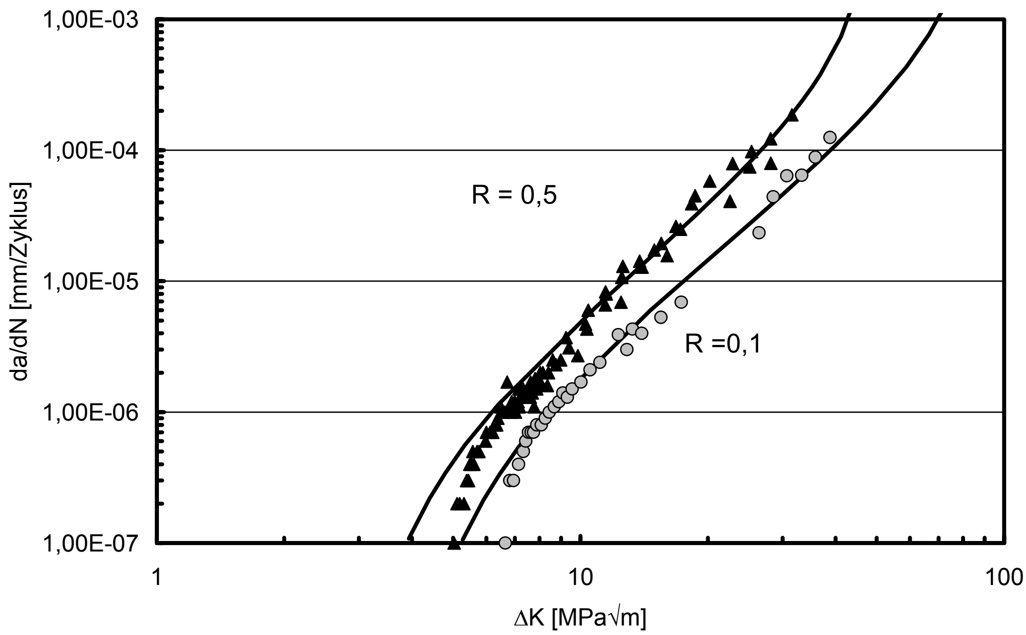

Figure 3: Experiment and calculation for S890 heat affected zone

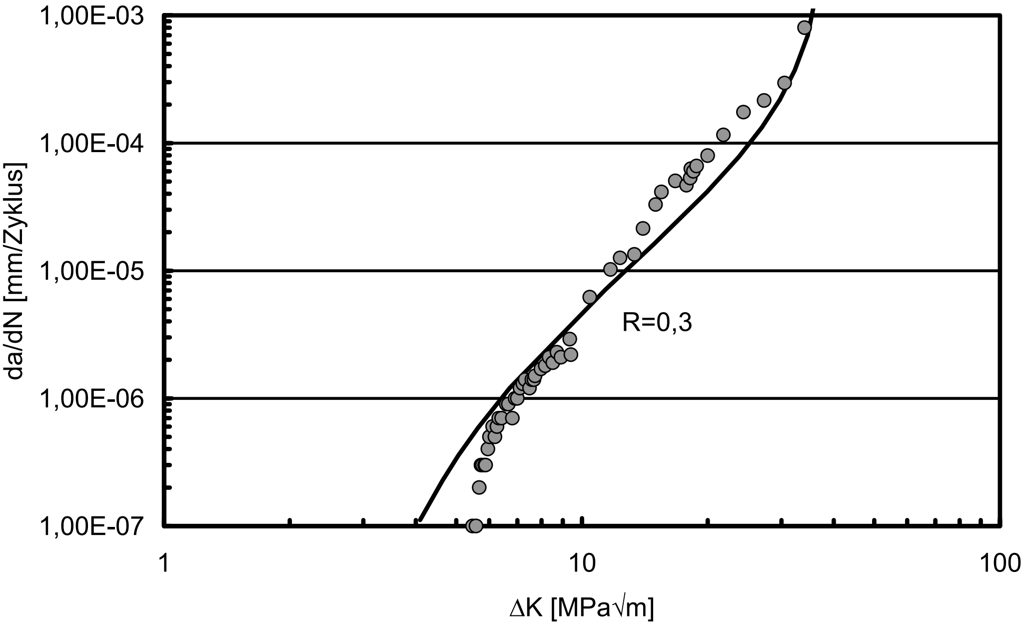

Figure 4: Experiment and calculation for S890 weld metal

REFERENCES

1. Roven, H. J. and Nes, E. (1991)
   Overwiew No. 94 Acta metall. mater. Vol. 39, No. 8, pp. 1719-1754

2. Hübner, P. (1996)
   Schwingfestigkeit der hochfesten schweißbaren Baustähle StE 885 und StE 960
   Dissertation TU Bergakademie Freiberg, Germany

3. Hück, M. (1992)
   Mikrolegierte Stähle Vorhaben Nr. 122
   Bewertung der Schwingfestigkeit der mikrolegierten Stähle 27 Mn VS 6 und 38 Mn VS 5
   Forschungshefte des Forschungskuratorium Maschinenbau Heft 163

4. Smith, K.N., Watson, P. and Topper, T.H. (1970)
   Journal of Materials, IMLSA, Vol. 5, No. 4, pp.767-778

5. Naylor, J. P. (1979) Met. Trans. A Vol. 10 A 7 pp. 861-873