FATIGUE PROPERTIES
OF RAIL STEELS
by
BERTRAND JOURNET
Ingenieur Civil des Mines De Saint-Etienne (1981)

Submitted to the Department of
Materials Science and Engineering
in Partial Fulfillment of the Requirements
for the Degree of

MASTER OF SCIENCE IN METALLURGY

at the
Massachusetts Institute of Technology

February 1983

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Signature of Author

Department of Materials Science and Engineering
January 13, 1983

Certified by

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Thesis Supervisor

Accepted by

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Chairman, Department Committee on Graduate Students

Archives

MASSACHUSETTS INSTITUTE
OF TECHNOLOGY
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ABSTRACT

Fatigue properties of pearlitic rail steels (ASTM standards) were studied. Four different areas were investigated:

- Deformation of the rail head
- Fatigue crack propagation
- Crack retardation
- Mean stress effect

Macro- and microhardness tests give a measure of the extent of cold work below the running surface of the rail. The sequence of beach marks on "detail" fractures was recorded by macrofractography and SEM analysis of replicas.

The fatigue crack growth rates in air were measured over the full range (da/dN - \( \Delta K \)). Paris' Law and Foreman's equation were verified. The R ratio effect was investigated by experiments run at four different values of R (0.05, 0.3, 0.5, 0.7). The \( \Delta K \) threshold in air decreases with increasing R. There is concomitantly an increasing susceptibility to corrosion embrittlement with increasing R. A model of \( \Delta K \)-threshold as a function of R is suggested on the basis of crack closure below R = 0.5. In this range the experimental data show that the \( \Delta K \)-threshold is a linear function of R.

Application of a tensile peak overload during fatigue crack propagation at R = 0.05 leads to delayed retardation. The validity of Wheeler's and Willenborg's models was investigated.

Relaxation of a mean tensile stress under cyclic strain control was studied. Since residual stresses are up to 2/3 the flow stress in track service, experiments were run with an initial applied mean stress of 2/3 the yield stress. A significant decrease of the mean
stress is noticed during the first ten cycles. The rate of mean stress relaxation is found to be a function of the plastic strain amplitude and the prior applied mean strain. A calculation shows that the relaxation of the mean stress corresponds to the decrease of the back stress.

Thesis Supervisor: Prof. R.M.N. Pelloux

Title: Professor of Metallurgy
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Special thanks go for Jim J. Scutti for his help and advice. He wrote an excellent Master's thesis on fatigue of rail steels.

I also thank the following persons:

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- the research group for their help and good mood,
- Lenny Sudenfield for his assistance in SEM works,
- my parents for their moral support and devotion.

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<td>$da/dN$</td>
<td>crack growth rate</td>
</tr>
<tr>
<td>$\sigma, \sigma_{\text{max}}, \sigma_{\text{min}}$</td>
<td>stress, maximum and minimum stress</td>
</tr>
<tr>
<td>$a$</td>
<td>crack length</td>
</tr>
<tr>
<td>$N$</td>
<td>number of cycles</td>
</tr>
<tr>
<td>$K, K_{\text{max}}, K_{\text{min}}$</td>
<td>stress intensity factor, maximum, minimum value</td>
</tr>
<tr>
<td>$\Delta K$</td>
<td>stress intensity factor range, $\Delta K = K_{\text{max}} - K_{\text{min}}$</td>
</tr>
<tr>
<td>$\Delta K_0$</td>
<td>stress intensity factor range necessary to produce crack growth at threshold</td>
</tr>
<tr>
<td>$\Delta K_{\text{th}}, K_{\text{max,th}}$</td>
<td>values of $\Delta K$ and $K_{\text{max}}$ at threshold</td>
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<tr>
<td>$P, P_{\text{max}}, P_{\text{min}}$</td>
<td>load, maximum, minimum value</td>
</tr>
<tr>
<td>$R$</td>
<td>load ratio $P_{\text{min}}/P_{\text{max}} = K_{\text{min}}/K_{\text{max}}$</td>
</tr>
<tr>
<td>PSB</td>
<td>Persistent Slippage Bands</td>
</tr>
<tr>
<td>$K_C$</td>
<td>critical stress intensity factor bringing about unstable crack propagation</td>
</tr>
<tr>
<td>$\Delta K_C$</td>
<td>$\Delta K_C = (1 - R)K_C$</td>
</tr>
<tr>
<td>CTOD</td>
<td>Crack Tip Opening Displacement</td>
</tr>
<tr>
<td>$E$</td>
<td>Young's modulus</td>
</tr>
<tr>
<td>$Y$</td>
<td>0.2 percent yield stress</td>
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<tr>
<td>$\delta$</td>
<td>displacement</td>
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<tr>
<td>$R_y$</td>
<td>plastic zone size ($K_{\text{max}}$)</td>
</tr>
<tr>
<td>$R_{\text{pl}}$</td>
<td>overload plastic zone size ($K_{\text{OL}}$)</td>
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<td>$m$</td>
<td>Wheeler's shaping exponent</td>
</tr>
<tr>
<td>$K_{\text{red}}$</td>
<td>residual stress intensity factor in Willenborg's model</td>
</tr>
<tr>
<td>$R_{\text{eff}}$</td>
<td>effective $R$ in Willenborg's model</td>
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<td>$K_{op}$</td>
<td>stress intensity factor for opening a crack</td>
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<td>$\Delta K_{eff}$</td>
<td>effective $\Delta K$</td>
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<td>$U$</td>
<td>ratio $\Delta K_{eff}/\Delta K$</td>
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<td>CAL</td>
<td>Constant Amplitude Loading</td>
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<td>VAL</td>
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<td>$\Delta \varepsilon_T, \Delta \varepsilon_E, \Delta \varepsilon_p$</td>
<td>total, elastic, plastic strain range</td>
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<td>$\sigma_f$</td>
<td>fatigue strength coefficient</td>
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<td>$\varepsilon_f$</td>
<td>fatigue ductility coefficient</td>
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<td>$b$</td>
<td>fatigue strength exponent</td>
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<td>$c$</td>
<td>fatigue ductility exponent</td>
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<tr>
<td>$n'$</td>
<td>cyclic strain hardening exponent</td>
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<td>$N_f$</td>
<td>number of cycles to failure</td>
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<td>$\sigma_0$</td>
<td>mean stress</td>
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<td>$\sigma_B$</td>
<td>backstress</td>
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<tr>
<td>$\sigma_{01}$</td>
<td>initial mean stress</td>
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<tr>
<td>$\sigma_{ON}$</td>
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<td>$\sigma_f, \sigma_r$</td>
<td>forward and reverse flow stress</td>
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<td>4-1</td>
<td>Values of constants in Paris' equation</td>
<td>62</td>
</tr>
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<td>4-2</td>
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<td>62</td>
</tr>
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CHAPTER 1
INTRODUCTION

After the Civil War, American railroads were conceived of as an interstate network. The Association of American Railroads took over the design of rails and cars. The low carbon steel rails were T-shaped and were required to meet light operating conditions: low train speed and minimum service requirements. Provided that the average design stress did not exceed a certain fraction of the ultimate tensile strength, the rail steel was manufactured at the lowest possible cost.

Unfortunately, rails have caused serious problems since the advent of diesel power. Their ability to withstand higher contact and shear stresses was questioned (Fig. 1-1). The controlled cooling process in the 1930's has helped in the manufacturing of a clean steel. Despite the reduction of defects such as flakes, premature failure at stress levels below the theoretical maximum strength still occurred.

A new type of carbon steel has been found to show good resistance to failure [1]. The structure is a fine pearlite, heat treated to harden the head of the rail. This gives a good trade-off between fracture resistance and wear resistance [2].

The fatigue behavior must be investigated using a scientific approach that deals with local conditions which intensify the cyclic stresses and lead to premature failure. Defects and dislocations combined with internal stresses [3,4] are the key terms.

The fatigue behavior of rail steel (Table 1-1) was investigated in
four different areas:

- in-service deformation
- fatigue crack propagation
- crack retardation
- mean stress effect.

This understanding will help to predict more accurately the fatigue life of a rail steel in order to maintain the network in good and safe conditions.
Figure 1-1. Rail head deformation; probably from a curve, right-hand side being the inner side of the track.
Table 1-1. Chemical composition and microstructure of rail steel.

<table>
<thead>
<tr>
<th>Rail</th>
<th>C</th>
<th>Mn</th>
<th>P</th>
<th>S</th>
<th>Si</th>
<th>Mo</th>
<th>Cr</th>
<th>Cu</th>
<th>Ni</th>
<th>Al</th>
<th>Co</th>
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<td>0.025</td>
<td>0.15</td>
<td>0.005</td>
<td>0.033</td>
<td>0.16</td>
<td>0.080</td>
<td>0.015</td>
<td>0.021</td>
<td>0.0038</td>
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<td>79</td>
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<td>0.034</td>
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<td>0.006</td>
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<tr>
<td>ASTM A1</td>
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<td>0.70-</td>
<td>0.04</td>
<td>0.05</td>
<td>0.10-</td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td>0.25</td>
</tr>
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<td>1.00</td>
<td>max</td>
<td>max</td>
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<td></td>
<td></td>
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<td></td>
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</tr>
</tbody>
</table>

Fine pearlitic microstructure; \( d\) (average) = 10\(\mu m\); \(d\) spacing = 0.3\(\mu m\)
(from [7])

Chemical composition & microstructure
CHAPTER 2
LITERATURE SURVEY

2.1 Failure of Rail Steels

Different research programs have been carried out to study the main defects that occur in service and lead to failure. In the case of rail steels, contact fatigue strength is dependent on the type of non-metallic inclusions that intensify the local cyclic stress [5]. The deformation of the rail head, the tensile longitudinal stresses developed by local bending of the rail [6] and a lower threshold of the stress intensity factor range for crack propagation in vacuum [7] contribute to crack initiation in the sub-surface of the rail at an inclusion stringer. This is the shelling-type defect that grows horizontally, then turns down 90° (Fig. 2-1). Shear stress reversal is induced as a train passes. It is believed that this is the dominant stress component causing fatigue crack propagation [8].

Good serviceability of rail steel, in terms of good resistance to structural failure and excessive wear during a long enough period of time, depends on strength and toughness. The mechanical properties were thought to be the relevant parameters that control the fatigue behavior of rail steel [9]. Actually the microstructure is the focus point [10,11]: a fine pearlite microstructure gives a better strength and the refining of the prior austenite grain size increases the toughness. These two parameters can be controlled independently. The interlamellae spacing and the cementite lamellae thickness characterize the deformation behavior [2]. The deformation of the
rail head (Fig. 2-2) takes place below the running surface. The shell-type crack initiates in this area.

2.2 Fatigue Crack Propagation

2.2.1 Linear Elastic Fracture Mechanics (LEFM)

The knowledge of the fatigue behavior of structures is of prime importance because that of mechanical properties is not enough to prevent catastrophic failures. This field is chiefly concerned with the propagation of crack-like imperfections. Detection service provides the user with the knowledge of the maximum size of defects present and prediction of their growth is carried out in order to remove the rail before it fails.

The crack growth rate \( da/dN \) is determined as a function of the applied stress range \( \sigma \) and the crack length \( a \) [13]: \[ da/dN = f(\sigma, a, C) \] \( (N: \) number of cycles, \( C: \) constant of the material.) A dimensional analysis prompted Frost and Dugdale [14] to establish equations such as \[ da/dN = [\sigma^n a^m /C] \]. But the small range of available data did not grant the uniqueness of the exponents.

In low stress-high cycle fatigue, the analysis of the stress field around the crack tip brought about the determination of the stress intensity factor \( K \) [15]. When the plastic zone size ahead of the crack tip is small compared to the crack length or other significant geometrical parameters, such as the thickness of the specimen, the crack growth rate is simply related to \( K \) by Paris' Law [16,17]: \[ da/dN = C(\Delta K)^n \], where \( \Delta K \) is the stress intensity factor range.
For finite plates, \( K \) is written as follows: \( K = F(a) \sigma \sqrt{a} \). \( F(a) \) is a function of the crack length that accounts for finite geometry effects and the distribution of the remote stress \( \sigma \). This is the small scale yielding condition (SSY) [18]. \( K \) is the relevant parameter that describes the stress-strain field around the crack tip under SSY conditions. This \( \Delta K \) concept contains the following philosophy [19]: "The same cause, stress intensity at the crack tip, results in the same effect, growth rate of the crack tip."

The double log ([da/dN] - \( \Delta K \)) plot of a fatigue test run under constant amplitude cyclic loading yields a typical curve (Fig. 2-3) divided into three parts:

- Slow crack growth rate with a \( \Delta K \)-threshold value below which a crack does not grow: Stage I*
- Fatigue crack propagation with a linear part (Paris' Law): Stage II**
- Fast propagation bringing about failure: Stage III***

2.2.2 Initiation and Growth of a Crack Near Threshold

Crack initiation raises the concept of a microcrack itself. Initiation is important for ground vehicles: it lasts 90 percent of 

*Large influence of microstructure, mean stress and environment.  
**Little influence of microstructure; large influence of environment and mean stress.
***Large influence of microstructure and mean stress; little influence of environment.
the life of the material and fracture takes place right after. This local phenomenon is sensitive to surface roughness, defects and bulk properties [20]. A microcrack becomes a macrocrack when the stress intensity factor has a direct relationship to its growth. The \( \Delta K \)-threshold is the value of \( \Delta K \) below which a macrocrack does not grow. Ferritic, pearlitic and martensitic steels obey the same law, i.e., \( \Delta K \)-threshold depends on \( R \) ratio and is independent of their chemical composition and their mechanical properties [21]. But the yield strength and the crack tip radius play a part in crack initiation.

Fatigue cracks are initiated by repeated plastic deformation. Cracks nucleate in persistent slip bands (PSB) due to the to-and-fro movement of dislocations. Wood's model [22] gives a schematical oscillatory movement that causes notch formation. The ridges are extrusions and the valleys intrusions. They are the features of a fracture surface. Then crevices form and grow into cracks. This reportedly occurs in local soft regions. Cottrell and Hull [23] explain this mechanism when such conditions are not met, especially at low temperatures. Frank-Read sources may contribute to the formation of extrusions when they are on two intersecting planes. Cyclic slip alters the depth and length of the two discontinuities presently formed.

Contemporary research presents models based on dislocation behavior, which accounts for PSB formation where cracks nucleate. Slip irreversibility of the dislocations in the PSB is shown to be
associated with extrusion evolution for fcc metals [24]. The basic event in the early stage of fatigue is the evolution of the microstructure into two components [25]: the matrix and the softer PSB. Therefore, the deformation becomes inhomogeneous and takes place in the PSB. The \( \Delta K \)-threshold might be therefore related to a threshold for developing PSB. PSB are observed in the bcc ferrite of low carbon steels [26] and microcrack nucleation takes place at the interface between PSB and the surrounding matrix. In pearlitic rail steels, cracks nucleate in the ferrite, close to the ferrite-cementite interface [27]. The interlamellae spacing related to the dislocation cells' size is the relevant controlling parameter.

At threshold, in addition to the microstructure, the environment also exerts a strong influence. The formation of an oxide film in moist air is responsible for the difference in \( \Delta K \)-threshold values at low and large values of \( R \) [7]. They are the same in vacuo. Surface roughness promotes crack closure [18]: near threshold, a Mode I+II displacement enhances crack closure by welding the crack open at some contact points, at low values of \( R \). A coarse grain in fully pearlitic steels is beneficial at low \( R \) [29] because it results in a higher \( \Delta K \)-threshold. This supports the model of oxide induced crack closure promoted by debris formation, which increases surface roughness at low \( R \) in air. Observation of PSB in Stage I ahead of the crack tip [30] suggests that when they are small compared to the grain size, near a grain boundary, and do not encroach upon the next grain, there is more resistance to crack extension.
2.2.3 Crack Propagation--Stage II

2.2.3.1 Semi-Empirical Laws

The tests run on compact tension specimens enable one to check different crack propagation laws. The crack propagates in a macroscopic plane perpendicular to the stress axis (Mode I opening). The double log plot of \([da/dN] - \Delta K\) yields a linear part (Stage II) which is generally best fitted by semi-empirical laws. The typical law is Paris' Law: \([da/dN] = C(\Delta K)^n = f(\Delta K, R)\). The microstructure and environment do not exert an obvious influence in rail steels [21].

The crack growth rate is strongly dependent on \(K_{\text{max}}\) and \(\Delta K\), or \(\Delta K\) and \(R\) [31]. The decrease of \(R\) results in a shift of the \([da/dN] - \Delta K\) curve to the right-hand side. This brings about different values of \(\Delta K\) threshold and \(\Delta K_c\). Some experimenters tried to find equations that described the whole \([da/dN] - \Delta K\) curve. Forman has accounted for \(R\) and \(K_c\) by saying that \(da/dN\) should be infinite (or at least very high) when \(K_{\text{max}}\) approaches \(K_c\):

\[
\frac{da}{dN} = C \frac{\Delta K^n}{(1 - R)K_c - \Delta K} \quad [32]
\]

Broek and Rice [33] thought they could find a better equation to account for the \(R\) effect in rail steels:

\[
\frac{da}{dN} = C(1 - R)^2(K_{\text{max}}^2 - K_{\text{th}}^2) \frac{K_{\text{max}}^{n-1}}{K_c - K_{\text{max}}}; \quad R > 0
\]

But it only showed good agreement for \(R = 0\).

All these empirical laws are fairly well checked out. But they do not suggest any micromechanism that accounts for crack propagation.
Fatigue crack propagation is a local phenomenon and the variability in crack growth denotes the influence of microstructure. The interlamellae spacing and the prior austenite grain size in pearlitic steels should enlighten any mechanism when the following events are considered: shear cracking of the lamellae, colony orientation, shear behavior, slip reversibility and cyclic work-hardening of the ferrite [34]. The interaction between the crack and the microstructure should explain the way the energy corresponding to the crack advance is dissipated. The key terms are slip blockage, generation of dislocations, and redirection of the crack front.

Generally speaking, five factors have to be taken into account to characterize fatigue crack propagation: geometry, load, material properties, time and environment [35].

2.2.3.2 Fracture Surface

The features of the fracture surface are fatigue striations. They have brought up controversial discussion about their formation. The plastic blunting of the crack tip during tensile loading and its resharpening upon unloading is a nice idea to explain the striations formation [36]. But this plastic blunting process is not convincing because of the lack of striations formation observed in vacuo [37]. The crack actually advances by a process of alternating shear [38]. Fatigue striations are formed in air because of irreversible shear due to the building up of an oxide film. Thus one can relate the crack growth rate to the CTOD as follows:
\[ \frac{da}{dN} = C \left( \frac{\Delta K}{E Y} \right)^2 \]

A lot of work remains to be done in order to correlate the crack extension to the dislocation cell structure that develops ahead of the crack tip [39].

2.2.3.3 Stage III or Fast Fracture

This regime of crack propagation is characterized on the \([da/dN] - \Delta K\) plot by an asymptotic vertical line giving the value of \(K_C\). Cleavage mode is the relevant fracture mechanism.

2.3 Retardation

Fatigue conditions are not actually as simple as the constant amplitude loading condition. Materials and structures are subjected to varying amplitude loading. This alters the crack growth rate; this one does not obey Paris' Law any more. A single tensile peak overload results in the retardation of the crack propagation, affecting at the same time the remaining fatigue life. In rail steels, the retardation consists of two phases (Fig. 2-4): the crack growth rate decreases to a minimum value, then increases and recovers the value that would have been effective had the peak overload not been applied (Paris' Law). When a test is run at constant \(\Delta K\), four kinds of response to an overload are generally told apart [40] (Fig. 2-5): 1) no effect, i.e., there is a threshold below which there is no retardation effect (Fig. 2-5a); 2) simple retardation (Fig. 2-5b); 3) delayed retardation
(Fig. 2-5c); and 4) lost retardation—the new curve bridges its gap with the previous one (Fig. 2-5d). The important parameters that are measured and that characterize this phenomenon are the crack length over which retardation acts and the number of cycles it takes the crack to recover its original value. A compressive overload results in much less retardation than a tensile one.

2.3.1 Mechanisms

Several concepts govern the crack tip behavior after the overload application:

- Compressive residual stresses
- Crack tip blunting
- Crack closure.

2.3.1.1 Compressive Residual Stresses (Fig. 2-6)

The overload introduces a large plastic zone and upon reversal of the load, the plastically deformed material experiences the clamping action of the elastic surrounding material that tries to recover its original size. It will thereby exert compressive stress, which will act against the propagation of the crack by closing it. This model explains the crack arrest phenomenon, retardation or the gradual return to normal growth as the crack tip paves its way through the residual compressive stress field. Delayed retardation is not accounted for because of maximum compressive stresses acting at the crack tip right after the overload application.
2.3.1.2 Crack Tip Blunting

The crack tip radius increases after an overload application. The stress intensity factor is thereby smaller and so is the crack growth rate. The crack tip blunting is not relieved immediately on unloading to the base stress level and progressive resharpening of the crack tip is thought to account for the gradual recovery of the crack growth rate to its original value. This model suggests immediate retardation. But well-established mechanisms of resharpening of the crack at each cycle question the validity of this model [41].

2.3.1.3 Crack Closure

This mechanism was introduced by Elber [42]. Crack closure occurs under a zero to tension cyclic loading, during the unloading part of the cycle [43,44]. Because of residual deformation left in the wake of the fatigue crack, the crack tip closes under a low stress [42]. The plot (load P vs displacement δ) consists of two linear segments connected by a non-linear one (Fig. 2-7). The linear segments refer to the behavior of the material when the crack is closed or fully open. Plasticity demands that the second derivative (a2P/a2δ) of the connecting curve be positive, which is not the case. Therefore there is a change of configuration of the crack. There is a stress intensity factor level K_{op} beyond which the crack opens. Therefore ΔK is not the relevant parameter for describing the advance of the crack as far as LEFM is concerned. This has to be corrected into an effective one: ΔK_{eff} = K_{max} - K_{op}. The crack propagation law
suggested by Elber is therefore: \[ \frac{da}{dN} = C(\Delta K_{\text{eff}})^n \]. The following relationship for an Al alloy was established:

\[ \Delta K_{\text{eff}} = U \Delta K; \ U = 0.5 \text{ to } 0.4 \text{ R}; \ -0.1 < R < 0.7 \]

The crack closure stress remains constant under CAL condition. Retardation is explained by the change of \( K_{\text{op}} \) after an overload application. The change of \( K_{\text{op}} \) can account for delayed retardation (Fig. 2-8).

Measurements of \( K_{\text{op}} \) are not easy. \( K_{\text{op}} \) can be correlated to the change of striation spacing [45]. The agreement is good as long as damage ahead of the crack tip is not excessive. Retardation is sensitive to plane strain and plane stress conditions. Crack closure is associated with large plastic zone and mainly occurs near free surface where the plane stress condition prevails. Bent shape crack fronts (beach marks) were observed [46]. Were \( K_{\text{op}} \) changes to explain delayed retardation, the minimum values of \( \Delta K_{\text{eff}} \) and \( \frac{da}{dN} \) should occur at the same time. Procedures such as load-displacement curves to measure \( K_{\text{op}} \) and visual measurement of the crack length on the edge surface of the specimen [47] are not consistent measurements to show that crack closure does not explain retardation. Because the second one only deals with plane stress condition. There is interaction between plane stress and plane strain conditions [45] and differences found in tests [47] are not large enough to draw positive conclusions.
2.3.2 Retardation Models

2.3.2.1 Wheeler's Model [48]

The assumption is based on the part played by the residual stresses and structural changes due to plastic strain during overload. The idea suggests that the crack length affected by retardation would be such that the current plastic zone ahead of the crack tip be contained in the one entailed by the overload (Fig. 2-9). The affected crack length is then \((a_t - a_{OL})\), \(a_t\) is such that the plastic zone \(R_y\) is tangential to plastic zone \(R_p\) (these are monotonic plastic zone sizes). Wheeler puts in a retardation coefficient \(C\) such that:

\[
C = \left(\frac{R_y}{a_p - a}\right)^m \text{ if } a + R_y < a_{OL}
\]

\[
\left(\frac{da}{dN}\right)_{OL} = C\left(\frac{da}{dN}\right)_{CAL} \quad C = 1 \text{ otherwise.}
\]

\(
\left(\frac{da}{dN}\right)_{OL} \quad : \text{crack growth rate after overload application}
\)

\(
\left(\frac{da}{dN}\right)_{CAL} \quad : \text{crack growth rate for CAL.}
\)

\(m\) is determined by fit to experimental data. The increase of \(C\) up to 1 denotes this gradual recovery of the crack growth rate.

2.3.2.2 Willenborg's Model [49]

Retardation is explained in terms of reduction of both \(K_{max}\) and \(K_{min}\) by a certain amount \(K_{red}\). \(K_{red}\) is the difference between \(K\) required \((K_{req})\) to get the plastic zone of crack length a tangential
to the overload one and $K_{\text{max}}$ applied. $K_{\text{req}}$ is related to the crack length $a_{\text{req}}$ such that: $a + a_{\text{req}} = a_p$ (Fig. 2-10). The residual compressive stresses introduced by an overload reduce the effective stress intensity factor at the crack tip:

$$K_{\text{red}} = K_{\text{req}} - K_{\text{max}}, \text{ where } K_{\text{red}} \text{ follows from:}$$

$$a_p - a = \frac{1}{2\pi} \left( \frac{K_{\text{req}}}{V} \right)^2$$

\hspace{2cm} a = \begin{cases} 
1 \text{ plane stress condition} \\
3 \text{ plane strain condition}
\end{cases}

The effective values of the stress intensity factors are:

$$K_{\text{maxeff}} = K_{\text{max}} - K_{\text{red}}$$

$$K_{\text{mineff}} = \begin{cases} 
K_{\text{min}} - K_{\text{red}} \text{ if } > 0 \\
0 \text{ otherwise}
\end{cases}$$

Hitherto:

$$\Delta K_{\text{eff}} = K_{\text{maxeff}} - K_{\text{mineff}} \text{ and } R_{\text{eff}} = [K_{\text{maxeff}}/K_{\text{mineff}}].$$

With the knowledge of Forman's equation, $da/dN$ is calculated by plugging in the effective values of $\Delta K$ and $R$.

These two models do not account for delayed retardation, complete arrest nor the difference between a single peak overload and a multiple one. On Al alloys, these models provide a fair approximation of the retardation phenomenon [50]. They predict less retardation than observed. They deserve attention because they give a fairly good approximation of the minimum value of the crack growth rate.

2.3.3 Random Loading

Random loading is not easy to deal with. For instance, in aircraft, the gust spectrum is approximated by a step function. The
load levels are rearranged into a curve of load level exceedance. The knowledge of CAL and VAL are then applied to predict the fatigue life. Whether you consider overloads occurring from time to time or you gather them in blocks of same load levels, the load sequence effects are more dramatic in the first case [51]. This method is conservative and the problem is to reduce the difference between predicted and actual data. Miner's rule is a functional relationship that is independent of load sequence. This is the reason why this rule is conservative.

Another approach for predicting fatigue life under VAL conditions deals with a statistical calculation of the stress intensity factor [20]. Discrepancies with actual data raise the issue of prior strain history of the material in terms of dK/da, which is equivalent to saying that $K_{op}$ is sensitive to VAL.

2.4 Mean Stress Effect

2.4.1 Low Cycle Fatigue (LCF)

2.4.1.1 Coffin-Manson Law

Fatigue is the result of cyclic plastic deformation. Large plastic strain is the controlling parameter in LCF. Repeated stresses of thermal origin combined with thermal expansion in pressure vessels or gas turbine were such that new considerations on fatigue behavior were taken instead of thinking only of monotonic properties. The strain life equations relative to tests run under strain control derive from Coffin's [52] and Manson's [53] works, under fully
reversed cycles conditions.

Total strain amplitude: \[ \frac{\Delta \varepsilon_T}{2} = \frac{\Delta \varepsilon_p}{2} + \frac{\Delta \varepsilon_e}{2} \]

Plastic strain amplitude: \[ \frac{\Delta \varepsilon_p}{2} = \varepsilon_f (2N_f)^c \]

Elastic strain amplitude: \[ \frac{\Delta \varepsilon_e}{2} = \frac{\sigma_f}{E} (2N_f)^b \]

(\(\varepsilon_f - \sigma_f\)) are respectively the fatigue (ductility-strength) coefficients, which are usually taken as monotonic (fracture ductility-strength) coefficients. \(2N_f\) is the number of reversals to failure. The values of the strains are derived from the hysteresis loop (Fig. 2-11a) which is representative of a test run until failure when the strains are kept constant. The double log plot of the strain-life equations (Fig. 2-11b) yields the following typical values of b and c for steels: \(-0.12 < b < -0.05; -0.7 < c < -0.5\). Ductility and strength properties of steels are fairly well approximated by the monotonic properties [54]. The b and c exponents can be determined from the knowledge of the cyclic strain hardening exponent \(n'\) [55]:

\[ b = -\frac{n'}{1 + 5n'} \quad \text{c} = \frac{-1}{1 + 5n'} \]

The CSS curve gives the value of \(n'\) and thereby fully characterizes the cyclic response of a material.

On rail steels, the strain-life data are correlated by Coffin-Manson Law [7].

Note that the plastic strain-life relationship breaks down at long
life on aluminum. The line is observed to drop off.

2.4.1.2 Cyclic Stress Strain (CSS) Curve

The cyclic response of steels is plotted as stress vs. strain in cyclic fully reversed cycles. The relative position of this curve with the monotonic stress-strain curve denotes cyclic softening, cyclic hardening, or both. On rail steels, the CSS curve denotes a transition from softening to hardening [7]. This behavior is reported to be predictable from the knowledge of the value of the monotonic strain-hardening exponent [56]. The CSS curve can be determined by different methods that are in agreement for steels over a wide range of hardness values [57]:

- multiple steps test
- incremental step test
- monotonic tension after cyclic straining
- individual hysteresis loop analysis
- tips of stable hysteresis loops from several companion specimens.

Before stabilization of the hysteresis loop, the accommodation to cyclic plastic deformation is understood from the dislocation behavior. A dislocation cells' structure is building up. This involves change of dislocation density motion and multiplication of dislocations. Saturation is reached when a dynamical equilibrium exists between bowing out and annihilation of dislocations [58]. The dislocation's stored energy has to be a minimum according to the
Second Law of Thermodynamics. The Bauschinger effect gives some information about the back stress that assists the reverse motion of dislocations. The back stress is due to pile-up of dislocations at obstacles such as grain boundaries, deformation-induced lower-Cottrell barriers or the interaction with other sets of dislocations. Textural stresses, directly connected with the grain and its orientation cannot account on their own for the Bauschiner effect [59].

2.4.2 Mean Stress Relaxation

Variable amplitude loading or straining induces a self-imposed mean stress. This phenomenon is difficult to account for in any cumulative damage analysis. Whatever method is applied for recalculating the number of cycles to failure when sequence effects are taken into account [60,61,62], a significant result is drawn. A tensile mean stress reduces the fatigue life and the knowledge of the stress-strain response under variable cyclic loading conditions remains a difficult task to achieve.

Morrow suggested that a mean stress $\sigma_0$ altered the fatigue strength coefficient $\sigma_f$ as follows:

$$\frac{\Delta \varepsilon_e}{2} = \frac{\sigma_f - \sigma_0}{E} (2N_f)^b$$

Under strain control conditions, a non-zero mean stress relaxes. The rate of decrease can be approximated by a power law [63]. The relaxation is due to the accumulation of plastic deformation [64]. On dispersion strengthened materials [65], dislocation mechanics yield a
"mean matrix stress" component that relaxes. This stress due to loops of dislocations around particles is related to the Bauschinger effect. The relaxation consists of the removal of the orowan loops and the formation of a new dislocation structure around particles.
Figure 2-1. Detail fracture. Note its origin from the shelling type crack perpendicular to it and parallel to the running surface.
Figure 2-2. Rail head deformation and concomitant appearance of subsurface cracks, from ref. [12].

Figure 2-3. Sigmoidal curve describing fatigue crack propagation in terms of log(da/dN) vs log(ΔK).
Figure 2-4. Crack growth rate evolution after application of a tensile peak overload.

Figure 2-5. Responses of crack growth to application of an overload under condition of constant applied $\Delta K$. 
Figure 2-6. Compressive residual stresses created ahead of the crack tip due to overload.

Figure 2-7. Load-displacement curve denoting change of crack configuration.
Figure 2-8. Change of crack closure stress $K_{op}$ after application of a single tensile peak overload.

$$\frac{da}{dN}_{OL} = C\left(\frac{da}{dN}\right)_{CAL}$$

$$C = \begin{cases} 
\left(\frac{R_y}{h-a}\right)^m & \text{if } a + R_y < a_p \\
1 & \text{otherwise}
\end{cases}$$

$$R_y = \frac{1}{\kappa} \left(\frac{K_{max}}{Y}\right)^2$$

$$R_p = \frac{1}{\kappa} \left(\frac{K_{OL}}{Y}\right)^2$$

Figure 2-9. Wheeler's model.
\[
\frac{a_p - a}{a} = \frac{1}{d_r} \left( \frac{K_{req}}{Y} \right)^2
\]

\[d_r = \begin{cases} 
1 & \text{plane stress} \\
3 & \text{plane strain}
\end{cases}\]

\[K_{\text{max eff}} = K_{\text{max}} - K_{\text{red}}\]

\[K_{\text{min eff}} = \begin{cases} 
K_{\text{min}} - K_{\text{red}} & i > 0 \\
0 & \text{otherwise}
\end{cases}\]

\[K_{\text{red}} = K_{\text{req}} - K_{\text{max}}\]

---

**Figure 2-10.** Willenborg's model.

---

**Figure 2-11.** Hysteresis loop in Low Cycle Fatigue.

\[\frac{\Delta \varepsilon_T}{2} = \frac{\sigma_f}{E} (2N_f)^b + \varepsilon_f^e (2N_f)^c\]

\[\frac{\Delta \varepsilon_T}{2} = \frac{\Delta \varepsilon_e}{2} + \frac{\Delta \varepsilon_p}{2}\]
CHAPTER 3

IN-SERVICE DEFORMATION

This pictorial section aims at giving an idea of the deformation on the rail head, especially on the gauge side. This hot-rolled, control-cooled, plain carbon rail steel follows the ASTM A1 standards [1].

Macro and microhardness measurements were taken in order to estimate the depth within which work-hardening takes place.

Metallography and scanning electron microscopy show the resulting pearlite lamellae configuration.

Replicas of the fatigue crack surface show beach marks.

3.1 Macrohardness

Rockwell C macrohardness tests were carried out on rail sample no. 6-18. Indentations were spaced 10 mm and taken as shown on Fig. 3-1. The specimen was ground to 600 grit and a 6 μm diamond polishing was then performed.

The measurements show that the material below the running surface is harder due to cold work during in-service.

3.2 Microhardness

Vickers hardness tests with a 200 g load were performed on a section of the gauge corner of the rail head. The length of the diagonals of indentations taken on sample no. 6-25 were about 30 μm, which was small enough to take measurements as close as 50 μm [66].
The sample was also ground to 600 grit and polished with a 6 μm paste.

Five sets of measurements such as primary readings (A, C, E) and secondary readings (B, D) were taken. An (X, Y) axis system was determined (horizontal and vertical axes of the rail). The origin 0 is the intersection between X-axis and the tangent to the gauge corner. A, B, C, D, E readings were made on vertical lines respectively marked: X = 3, 6, 9, 12, 15 mm (Fig. 3-2).

A, C, E measurements start at the edge and their Y coordinates are such that: \( Y_{i+1} = Y_i + \Delta_i \), \( i \geq 1 \), i integer.

- \( \Delta_i = 0.05 \text{ mm} \) \( 1 \leq i \leq 4 \)
- \( \Delta_i = 0.10 \text{ mm} \) \( 5 \leq i \leq 8 \)
- \( \Delta_i = 0.20 \text{ mm} \) \( 9 \leq i \leq 12 \)
- \( \Delta_i = 0.50 \text{ mm} \) \( 13 \leq i \leq 16 \)
- \( \Delta_i = 1.00 \text{ mm} \) \( 17 \leq i \leq 20 \)
- \( \Delta_i = 2.00 \text{ mm} \) \( 21 \leq i \leq 22 \)

B, D measurements start at 1 mm from the edge. Ten 1-mm spacing measurements were taken.

Fig. 3-3 shows the variations of microhardness on a magnified scale. Fig. 3-4 shows the same curves but they are all set at the same origin. The marks at the bottom indicate the Y values where the measurements were taken. The sharp decrease of curve C right after the beginning denotes the existence of a crack. The highest hardness value found is at the beginning of curve C. Large changes of hardness take place over a depth of about 6 mm.
3.3 Metallography and SEM

In order to add to the hardness measurements, optical metallography and SEM micrography were performed to give an account of the plastic deformation of the pearlite lamellae. The specimen no. 6-25 was polished with a 1 μm diamond paste. Final polishing was carried out with Nalcoag 1060. Then the specimen was etched with Nital 2 percent to attack the ferrite lamellae.

The pearlite is shown on Figs 3-5 to 3-10. The lamellae spacing is reduced and their shape is much wavier near the edge of the specimen. Each of the micrographs was taken at one of the three different depths: edge, 5 mm from the edge, and 15 mm from the edge; along a line parallel to the vertical of the rail.

3.4 Replicas

Replicas were made of the fatigue fracture surfaces of rail no. 2-11 in order to investigate their features. The replica is an acetate tape softened with acetone. SEM micrographs show the beach marks at different magnifications in Figs. 3-11 and 3-12. The micrograph in Fig. 3-11 denotes two networks of overload beach marks.
Figure 3-1. Macrophase measurements on Rockwell C scale.

Figure 3-2. Schematic of microhardness measurements pattern.

MACROHARDNESS
Rockwell C
Figure 3-3. Microhardness measurements plotted on Vickers scale vs depth of indentation from the running surface.
Figure 3-5. Metallography on rail #6-25.
-a) close to the running surface, 1000X.
-b) about 5 mm deeper, 1000X
Figure 3-6. Metallography on rail #6-25.
   -a) about 15 mm deep from the running surface, 1000X.
   -b) same spot, 200X.
Figure 3-7. SEM on rail #6-25, close to the edge.
   a) 4400X
   b) 20000X
Note the heavy wavy pearlite.
Figure 3-8. SEM on rail #6-25, at a crack close to the edge.
-a) 4400X
-b) 20000X
Note particular pearlite configuration around the crack.
Figure 3-9. SEM on rail #6-25, 5 mm from the edge.
-a) 4400X
-b) 20000X
Note pearlite is less deformed.
Figure 3-10. SEM on rail #6-25, 15 mm from the edge.
   - a) 4400X
   - b) 20000X
   Observe bulk pearlite.
Figure 3-11. SEM on a replica from detail fracture of rail #2-1-1.

- a) 11X
- b) 28X

Note the pattern of two different lines networks.
Figure 3-12. SEM on a replica from detail fracture of rail #2-1-1.

- a) 60X
- b) 1000X

No striation features.
CHAPTER 4
FATIGUE CRACK PROPAGATION

4.1 Introduction

Classical laws of fatigue crack propagation are reviewed and checked out as a function of the R ratio. Paris' and Forman's equations were especially determined with their coefficients and a linear regression was carried out to give the value of the exponent.

The discussion about the R ratio effect leads to the concept of crack closure at threshold and its calculation. This attempt was made because the experimental data fitted the theoretical equations.

4.2 Experimental Procedure

Four tests were run on compact tension specimens with a computer-controlled MTS 810 test system. The specimens were cut out of the rail (Fig. 4-1) and had a L-T orientation. The chemical composition is indicated on Table 1-1, batch no. 79.

These tests were run under load control, in air, at room temperature, at a frequency of 50 Hz. The crack length was measured on one side of the specimen with the help of a 20x travelling microscope with a resolution of 0.01 mm. The tests were started with the precracking procedure described in ASTM reference [67]. The load shedding was lower for crack growth rates below $10^{-8}$ m/cycle and the threshold was reached when the crack growth rates were in the vicinity of $10^{-10}$ m/cycle. Threshold rates were taken with increasing and decreasing $\Delta K$ and a lower limit was found. The remaining part of the
(\(da/dN - \Delta K\)) curve was determined with increasing values of \(\Delta K\).

The \(R\) ratios studied are: 0.05, 0.3, 0.5, 0.7. The specimens were respectively labeled: LT-6, LT-3, LT-5, LT-4 and their geometry is shown in Fig. 4-2.

4.3 Results

The crack front was making an angle with the normal to the surface of the specimen. The angle was 10° on specimens LT-3, LT-5, and LT-4. It was 35° on the fourth one. A correction giving the mean crack length \(a\) was performed.

The double log \((da/dN - \Delta K)\) plots on Figs. 4-3, 4-4, 4-5, and 4-6 comprise a linear part with a slope around 3.65. The coefficients of Paris' Law are indicated on Table 4-1. The classical S-shaped curve is observed on each of those plots. The asymptotic straight line down to the \(\Delta K\) scale yields in each case a value of \(\Delta K_C\) which corresponds to \(K_C = 55 \text{ MPa}\sqrt{\text{m}}\). Some of the data points indicate a pop-in fracture at high \(\Delta K\) level due to cleavage, lamellar separation and ferrite grain-boundary fracture [10].

The values of \(\Delta K\) at threshold are reported on Table 4-1. The relative crack growth rate is \(10^{-10}\) m/cycle or lower. The curve is vertical in this region. Crack growth rates below the interatomic distance are due to a nonhomogeneous crack growth. The difference in mechanical properties between ferrite and cementite and the grain's orientation should explain this behavior.

When \(R\) increases, \(da/dN\) increases but \(\Delta K\)-threshold decreases. The
plot of Fig. 4-8 depicts this trend.

An attempt in interpreting the actual data such as $\Delta K_{th}$ and $K_{max}$ at threshold as functions of $R$ (Fig. 4-9) is based on the following observations:

- if $R \leq 0.5$ $\Delta K_{th}$ is a linear function $R$ and $K_{max}$ remains constant

- if $R > 0.5$ $\Delta K_{th}$ levels off and $K_{max}$ increases.

Let us assume that crack closure occurs when $R$ is smaller than 0.5, as could happen from Elber's research [42]. Let $K_{op}$ denote the stress intensity factor required to open the crack and $\Delta K_o$ the stress intensity factor range to produce crack growth at threshold, one can assumed to be constant.

- when $R \leq 0.5$ $K_{min} < K_{op}$

$K_{max} = K_{op} + \Delta K_o$ is a constant since $K_{op}$ depends on $K_{max}$ which is constant at threshold

Therefore: $\Delta K_{th} = K_{max} - K_{min} = (1 - R)K_{max}$

$\Delta K_{th} = (K_{op} + \Delta K_o)(1 - R)$

- when $R > 0.5$ $K_{min} > K_{op}$

$\Delta K_{th} = \Delta K_o$ (crack closure is not observed)

$K_{max_{th}} = \frac{\Delta K_{th}}{1 - R}$

$K_{max_{th}} = \frac{\Delta K_o}{1 - R}$

These two equations are plotted on Fig. 4-10.

The experimental data plotted on Fig. 4-9 yields the following
relationships:

- $R \leq 0.5$

  \[ \Delta K_{th} = A(1 - BR) \]

  \[ K_{\text{max}} \text{ remains constant} = 8.71 \text{ MPa m}^{\frac{1}{2}}; \quad \{A = 8.79 \text{ MPa m}^{\frac{1}{2}} \}
  \]

- $R > 0.5$

  \[ K_{\text{max th}} \text{ seems to obey the following law:} \]

  \[ K_{\text{max th}} = \frac{\Delta K_o}{1 - R} \quad (R = 0.7, \quad K_{\text{max}} = 12.73 \text{ MPa m}^{\frac{1}{2}}) \]

We therefore draw the following set of equations:

\[ A = K_{op} + \Delta K_o = 8.71 \text{ MPa m}^{\frac{1}{2}} \]

\[ \Delta K_o = (1 - R) K_{\text{max th}} \]
\[ = (1.07) \quad 12.73 \text{ MPa m}^{\frac{1}{2}} \]

The solution is:

\[ K_{op} = 4.89 \text{ MPa m}^{\frac{1}{2}} \]
\[ \Delta K_o = 3.82 \text{ MPa m}^{\frac{1}{2}} \]

These values are typical for tests run in air and environmental effects are included. At threshold, the ratio $K_{op}/K_{\text{max}}$ is 0.56.

The use of Elber's equation enables the calculation of $K_{op}/K_{\text{max}}$ in Stage II of crack propagation. By comparing the following two equations:

Paris' \[ \frac{da}{dN} = C(R) (\Delta K)^n \]

Elbers' \[ \frac{da}{dN} = C''(\Delta K_{\text{eff}})^n, \quad \Delta K_{\text{eff}} = U\Delta K = K_{\text{max}} - K_{op} \]

\[ \Delta K_{\text{eff}} = K_{\text{max}} - K_{op} \text{ if } K_{op} > K_{\text{min}} \]
\[ \Delta K_{\text{eff}} = K_{\text{max}} - K_{\text{min}} \text{ if } K_{\text{op}} \leq K_{\text{min}} \]

\[ \Delta K_{\text{eff}} = U \Delta K \]

We draw: \( m = n \)

Therefore, assuming there is no crack

\[ C(R) = C^n U^m \]

Closure at \( R = 0.7 \):

\[
\frac{C(R = 0.05)}{U^n(R = 0.05)} = \frac{C(R = 0.3)}{U^n(R = 0.3)} = \frac{C(R = 0.5)}{U^n(R = 0.5)} = C(R = 0.7)
\]

Since \( C(R) \) is known with these four values of \( R \), one can immediately calculate \( U \) (\( n \) is taken as 3.65):

\[
\begin{array}{c|c|c|c|c}
R & 0.05 & 0.3 & 0.5 & 0.7 \\
U(R) & 0.73 & 0.87 & 0.94 & 1 \\
\end{array}
\]

A linear regression yields the following relationship:

\[ U = aR + b \]

\[ \begin{aligned}
a &= 0.73 \\
b &= 0.41 \\
r &= 0.99 \text{ correlation parameter.}
\end{aligned} \]

Since \( U \) is a function of \( R \), from an argument given in Appendix A one finds: \( K_{\text{op}} = a(R)K_{\text{max}} \). The different values of \( a \) can therefore be calculated:

\[
\begin{array}{c|c|c|c}
R & 0.05 & 0.3 & 0.5 \\
a(R) & 0.30 & 0.39 & 0.53 \\
\end{array}
\]

We observe that the calculated values of \( a \) denote less closure during Stage II than at threshold. Microstructure and environment exert a large influence on threshold.

Forman's equation gives a reliable relationship for \( R \) ratios such as 0.05, 0.3 and 0.5 (Table 4-2, Fig. 4-11). A factor of 2 tells
apart data at $R = 0.7$. Sensitivity to embrittlement at large $R$ values (0.7) may modify $K_c$ into an effective one.

Another semi-empirical law [33] was checked out (Fig. 4-12).

There is no agreement with the data.

4.4 Micrography

SEM micrographs taken on two specimens' fracture surfaces denote a mixture of fatigue striations and cleavage. At low $\Delta K$ (near threshold) striations are seen and Fig. 4-7a depicts a planar facet mingled with striations. Fatigue striations are observed at mid $\Delta K$ (mid-Stage II), the fatigue fracture surface shows regions of hills and valleys with different orientations with respect to crack propagation. This one goes from bottom to top of the figures. At high $\Delta K$ (near fast fracture), cleavage facets prevail. Since Forman's equation is more likely to account for crack propagation in Stage III, the cross-over with the power law sets an upper bound limit for Paris' Law: $\Delta K_t \text{"} \Delta K \text{ transition."

\[ C(\Delta K)^n = \frac{C(\Delta K)^n}{(1 - R)K_c - \Delta K} \Rightarrow \Delta K_t = \frac{1 - R}{2} K_c \]

<table>
<thead>
<tr>
<th>$R$</th>
<th>0.05</th>
<th>0.3</th>
<th>0.5</th>
<th>0.7</th>
</tr>
</thead>
<tbody>
<tr>
<td>$\Delta K_t$ (MPa $\sqrt{m}$)</td>
<td>26.1</td>
<td>19.2</td>
<td>13.7</td>
<td>8.2</td>
</tr>
</tbody>
</table>

The microstructural dependence of Stage III crack propagation mechanism results in a superimposition at high $\Delta K$ of a mechanism to fatigue striations (Figs. 4-13e, 4-13h).
Table 4-1: Values of constants in Paris' equation:
\[ \frac{da}{dN} = c(\Delta K)^n \]

<table>
<thead>
<tr>
<th>R</th>
<th>0.05</th>
<th>0.3</th>
<th>0.5</th>
<th>0.7</th>
</tr>
</thead>
<tbody>
<tr>
<td>n</td>
<td>3.65</td>
<td>3.64</td>
<td>3.67</td>
<td>3.66</td>
</tr>
<tr>
<td>c</td>
<td>(8.57 \times 10^{-13})</td>
<td>(1.63 \times 10^{-12})</td>
<td>(2.12 \times 10^{-12})</td>
<td>(2.67 \times 10^{-12})</td>
</tr>
<tr>
<td>(\Delta K) threshold MPa(\sqrt{m})</td>
<td>8.40</td>
<td>5.80</td>
<td>4.40</td>
<td>3.82</td>
</tr>
</tbody>
</table>

Table 4-2: Values of constants in Forman's equation:
\[ \frac{da}{dN} = C \frac{(\Delta K)^n}{(1-R)K_C - \Delta K} \]

<table>
<thead>
<tr>
<th>R</th>
<th>0.05</th>
<th>0.3</th>
<th>0.5</th>
<th>0.7</th>
</tr>
</thead>
<tbody>
<tr>
<td>n</td>
<td>3.15</td>
<td>3.14</td>
<td>3.14</td>
<td>3.14</td>
</tr>
<tr>
<td>C</td>
<td>(1.17 \times 10^{-10})</td>
<td>(1.41 \times 10^{-10})</td>
<td>(1.12 \times 10^{-10})</td>
<td>(6.96 \times 10^{-11})</td>
</tr>
</tbody>
</table>
- Orientation of CTS -

Figure 4-1. CT Specimen extraction from the rail. The chemical composition is that of batch # 79.
All dimensions are in inches

\[ \Delta K = \frac{\Delta P}{BW^{\frac{1}{2}}} \left(2 + \frac{a}{W}\right) \frac{0.886 + 4.64\left(\frac{a}{W}\right) - 13.32\left(\frac{a}{W}\right)^2 + 14.72\left(\frac{a}{W}\right)^3 - 5.6\left(\frac{a}{W}\right)^4}{\left(1 - \frac{a}{W}\right)^{\frac{3}{2}}} \]  

\( \Delta K \) and \( \frac{da}{dn} \) are calculated through the following process: a two-point linear regression between crack length \( a \) and number of cycles \( N \) is undergone. Then, the program computed on the calculator yields an average crack growth rate and a \( \Delta K \)-value corresponding to the average crack length.

Figure 4-2. CTS geometry and \( \Delta K \) calculation.
RAIL STEEL
FCP  R=0.05

\[ \frac{da}{dN} = 8.57 \times 10^{-13} (\Delta K)^{3.65} \]

Figure 4-3. FCP data for rail steel in air, at room temperature, 50 Hz, R=0.05.
RAIL STEEL
FCP  $R=0.3$

$$\frac{da}{dN} = 1.63 \times 10^{-12} (\Delta K)^{3.44}$$

Figure 4-4. FCP data for rail steel in air, at room temperature, 50 Hz, $R=0.3$. 

$\Delta K$ (MPa$\sqrt{m}$)

$da/dN$ (m/cycle)
Figure 4-5. FCP data for rail steel in air, at room temperature, 50 Hz, R=0.5.
Figure 4-6. FCP data for rail steel in air, at room temperature, 50 Hz, R=0.7.

\[ \frac{da}{dN} = 2.67 \times 10^{-9} (\Delta K)^{3.65} \]
Figure 4-7. SEM on fracture on CTS LT-3 (R=0.3).
-a) low $\Delta K$
-b) mid $\Delta K$. 
Figure 4-8. FCP data for rail steel in air, at room temperature, 50 Hz, relative to four values of R.
Barsom's relationship: [21]

\[
\Delta K_{th} = \begin{cases} 
7.02 (1 - 0.85R) & R \geq 1 \\
6.04 & R < 1 
\end{cases}
\]

curve: ——

Experimental data:

<table>
<thead>
<tr>
<th>R</th>
<th>.05</th>
<th>.3</th>
<th>.5</th>
<th>.7</th>
</tr>
</thead>
<tbody>
<tr>
<td>(\Delta K_{th} ) (MPa m)</td>
<td>8.45</td>
<td>5.85</td>
<td>4.43</td>
<td>3.92</td>
</tr>
</tbody>
</table>

curve: +—

**\( \Delta K \) Threshold**

Figure 4-9. Experimental values of \( \Delta K \)-threshold and \( K_{max} \)-threshold with respect to \( R \).

**\( K_{max} \) Threshold**

Experimental data:

<table>
<thead>
<tr>
<th>R</th>
<th>.05</th>
<th>.3</th>
<th>.5</th>
<th>.7</th>
</tr>
</thead>
<tbody>
<tr>
<td>( K_{max} ) (MPa m)</td>
<td>8.89</td>
<td>8.38</td>
<td>8.66</td>
<td>12.75</td>
</tr>
</tbody>
</table>

best linear regression:

\( R = .05, .3, .5 \)

\( \Delta K_{th} = 8.79 (1 - 0.97R) \)

correlation parameter \( r = 0.99 \)
$\Delta K_{th}$

$\Delta K_0$

$K_{min} = K_{op}$

$K_{max} = \frac{\Delta K_0}{1-R}$

$K_{min} < K_{op}$:

$K_{max} = K_{op} \times \Delta K_0$

$= \text{constant}$

$\Delta K_{th} = K_{max} - K_{min}$

$= K_{max} (1-R)$

$\Delta K_{th} = (K_{op} + \Delta K_0)(1-R)$

$K_{min} \geq K_{op}$:

$\Delta K_{th} = \Delta K_0$

$= \text{constant}$

$K_{max} = \frac{\Delta K_{th}}{1-R}$

Figure 4-10. Theoretical variation of $\Delta K$-threshold and $K_{max}$ threshold with respect to $R$. 
Figure 4-11. FCP data replotted owing to Forman's equation for the four R values.
Figure 4.12. FCP data replotted owing to equation from reference [33].

\[ \phi = \frac{K_{\text{max}} - 1}{K_{\text{max}}} \left[ 1 - \left( \frac{K_{\text{th}}}{K_{\text{max}}} \right)^2 \right] \]

- \( R = 0.7 \)
- \( R = 0.5 \)
- \( R = 0.3 \)
- \( R = 0.05 \)
Figure 4-13. SEM on fracture surface of CTS LT-4 (R=0.7).
-a) low $\Delta K$, 1000X
-b) low $\Delta K$, 4000X.
Figure 4-13. SEM on fracture surface of CTS LT-4 (R=0.7).
-c) low ΔK, 8000X
-d) low ΔK, 8000X.
Figure 4-13. SEM on fracture surface of CTS LT-4 (R=0.7).
- e) mid $\Delta K$, 1000X
- f) mid $\Delta K$, 4000X.
Figure 4-13. SEM on fracture surface of CTS LT-4(R=0.7), -g) mid $\Delta K$, 8000X.
Figure 4-13. SEM on fracture surface of CTS LT-4 ($R=0.7$).
-h) high $\Delta K$, 1000X
-i) high $\Delta K$, 4000X.
Figure 4-13. SEM on fracture surface of CTS LT-4 ($R=0.7$).
-j) high $\Delta K$, 16000X.
CHAPTER 5
CRACK RETARDATION

5.1 Introduction

The aim of this section is to watch the retardation phenomenon and to check some models such as Wheeler's and Willenborg's.

5.2 Experimental Procedure

Tests were run on compact tension specimens at room temperature. The overload ratio $K_{OL}/K_{max}$ is 1.3 and $R$ is equal to 0.05. The CTS geometry is shown on Figs. 5-1 and 4-2; batch no. 79 on Table 1-1. Test frequencies were 10 to 20 Hz. The experiments were conducted at constant applied $\Delta K$ in order to detect the end of the retardation on the $(a-N)$ plot, $a$ being the crack length and $N$ the number of cycles. The load program is depicted in Fig. 5-2.

One set of data was taken with values of $\Delta K$ in the vicinity of 15.4 MPa $\sqrt{m}$. A second set was taken for $\Delta K$ of about 16.5 MPa $\sqrt{m}$. Crack length measurements were performed with a 20x travelling microscope of 0.01 mm resolution.

5.3 Results

Right after the application of the overload, the crack growth rate experiences a sharp decrease; it reaches a minimum value and then increases slowly back to its original value (Figs. 5-3 to 5-6). This phenomenon is called delayed retardation. The crack growth rate minimum value is observed at a distance of about 0.1 mm ahead of the
overload front. Retardation ends up after the crack has gone through a distance which is roughly the length of the overload plastic zone size in plane stress. The size of the plane stress plastic zone is:

\[ R_P = \frac{1}{\pi} \left( \frac{K_{OL}}{Y} \right)^2 \]

**Wheeler's Model** [48]. Retardation is supposed to end when the current plastic zone is tangent to the overload plastic zone. This yields: \( R_Y/R_P = (K_{\text{max}}/K_{OL})^2 = 0.59 \). This model predicts a short-time effect. The determination of the shaping exponent \( m \) is explained in Fig. 5-7. This procedure yields a value of 5. The plots \((da/dN - a)\) according to Wheeler's model are drawn on Figs. 5-3 and 5-5, with \( m = 5 \). A much lower value of the crack growth rate is predicted right after overload application, about one-fourth to one-half the actual minimum value. Wheeler's model fairly well approximates the rising part of the actual \((da/dN - a)\) plot until this curves starts levelling off.

The experimental curve denotes delayed retardation this model is unable to explain.

**Willenborg's Model** [49]. The plot \((da/dN - a)\) that derives from this model is drawn in Figs. 5-3 and 5-5. There is no agreement at all with the experimental data. The crack growth rate values predicted are much higher than the actual data. The failure is explained as follows: Soon after the application of the overload, \( R_{\text{eff}} > 0.05 = R_{\text{applied}} \) and in the case where \( R \neq 0 \), \( \Delta K_{\text{eff}} = \Delta K_{\text{applied}} \). From our knowledge of \( R \) effect, the crack growth rate
value predicted by this model is greater than the one under constant amplitude loading condition ($\Delta K_{\text{applied}}$, $R_{\text{applied}}$), which is greater than the retarded crack growth rate.

\[
\frac{da}{dN}_{\text{OL Willenborg's}} > \frac{da}{dN}_{\text{CAL}} > \frac{da}{dN}_{\text{retarded}}
\]

$\Delta K_{\text{applied}} = \text{constant}$

$R_{\text{applied}} = 0.05$

**Change of Closure Stress.** From Elbers' equation:

\[
\frac{da}{dN} = C U^n \Delta K^n
\]

one can write:

\[
\frac{U}{U_0} = \left[ \frac{(da/dN)_{\text{OL}}}{(da/dN)_0} \right]^{1/n}
\]

\[
U = (K_{\text{max}} - K_{\text{op}})/\Delta K
\]

\[
U_0 = (K_{\text{max}} - K_{\text{op}})/\Delta K
\]

Subscript OL denotes during retardation; subscript 0 denotes right before overload application.

The change of $U$ will give the change of $K_{\text{op}}$. From Section 4, $U_0 = 0.73$ because $R = 0.05$ and we are in Stage II. $(da/dN)_0$ is known. From the $(da/dN)_{\text{OL}}$ data and the relationship $K_{\text{op}} = K_{\text{max}} - U\Delta K$, the plots of $K_{\text{op}}$ with the number of cycles after the overload application are drawn on Fig. 5-8.
CTS geometry is the same as the one shown on fig 4-2 and also 
ΔK calculation.
However, some differences have to be outlined:

- no chevron
- notch depth: 1.3 inch

Figure 5-1. CTS geometry for retardation tests.

\[ R = \frac{K_{\text{max}}}{K_{\text{min}}} = 0.05 \]

\[ \frac{K_{\text{eq}}}{K_{\text{max}}} = 1.3 \]

Figure 5-2. Load program for retardation tests: constant ΔK applied, and four tensile peak overload, \( K_{\text{eq}}/K_{\text{max}} = 1.3 \).
Figure 5-3. Retardation data. Test run in air, at room temperature. Crack growth rate vs distance from overload front.
Figure 5-4. Retardation data relative to fig. 5-3. Crack length vs number of cycles after overload application.
Figure 5-5. Retardation data. Test run in air, at room temperature. Crack growth rate vs distance from overload front.
Figure 5-6. Retardation data relative to fig. 5-5. Crack length vs number of cycles after overload application.
Determination of shape exponent $m$ is such that the predicting curve derived from Wheeler's model gives the same crack length as experimental data at the end of retardation.

At crack length $a_i$, $C_i = \frac{R_y}{R_{p-a}}$, and number of cycles is $N_i$. $\Delta K_i$ is the relative stress intensity factor range to $a_i$. Between two measurements, $a_i$ and $a_{i+1}$, $C_i$ is assumed to remain constant. Thus Wheeler's model prediction can be performed such as follows:

$$a_{i+1} = \left( \frac{R_y}{R_{p-a_i}} \right)^m \left( \frac{da}{dN} \right)_{c_{AL}} (N_{i+1} - N_i) + a_i$$

$$\left( \frac{da}{dN} \right)_{c_{AL}} = C' \frac{\left( \Delta K_i \right)^{n'}}{(1-R)K_e - \Delta K_i}$$

Figure 5-7. Determination of shaping exponent $m$ of Wheeler's model.
Figure 5-3. Change of crack closure stress after overload application.
- a) data relative to fig. 5-3 and 5-4
- b) data relative to fig. 5-5 and 5-6.
CHAPTER 6
MEAN STRESS EFFECT

6.1 Introduction
An interesting feature of the behavior of rail steel is the relaxation of an applied tensile mean stress when tests are run under strain control. The strain-life curves, relative to a mean stress of 2/3 the yield point, are plotted and superimposed to results under fully reversed straining [7].

6.2 Experimental Procedure
Six tests were run in order to cover a wide enough range of life. Only one test was stopped, the longer one, at 400,000 cycles. Tests were run under strain control, on hour-glass specimens (Fig. 6-1); batch no. 37 on Table 1-1, fulfilling ASTM standards requirements [68] for low cycle fatigue. The specimens were ground to 600 grit, then polished with a Wenol metal polish. The tests were run in air, at room temperature on an MTS 810 servohydraulic machine. Diametral strain was continuously recorded with stress on a Hewlett-Packard X-Y plotter using an MTS diametral model 632 serial no. 120 extensometer positioned at the smallest diameter of the specimen.

The initial mean stress of 2/3 the yield point was shot by choosing a combination of mean strain and total strain amplitude. The test was then run under this constant strain amplitude. Table 6-1 gives the strain life data of the tests.
6.3 Results

The mean stress decreases and its values with respect to the number of cycles elapsed were recorded. The decrease is dramatic during the first ten cycles (Fig. 6-2). The values of $\sigma_{\text{mean}} = \frac{1}{2}(\sigma_{\text{max}} + \sigma_{\text{min}})$ and $\sigma_{\text{ON}}/\sigma_{01}$ ($\sigma_{\text{ON}}$ is the mean stress at N cycles; $\sigma_{01}$ is the initial mean stress) are reported with the relative number of cycles on Figs. 6-3 and 6-4. At short life, the relaxation is almost complete (90 percent of the initial value). At longer life, the mean stress seems to level off.

The strain-life data (Figs. 6-5, 6-6, 6-7) are superimposed to data obtained under fully reversed straining [7]. All data points fall below the reported results. This asserts the trend that a tensile mean stress reduces the fatigue life. However there is not a large difference. A more significant effect is observed at long life.

A linear regression performed on the strain-life data yields:

<table>
<thead>
<tr>
<th>Condition</th>
<th>$b$</th>
<th>$\sigma_f$ (ksi)</th>
<th>$c$</th>
<th>$\varepsilon_f$</th>
</tr>
</thead>
<tbody>
<tr>
<td>Fully reversed [7]</td>
<td>-0.102</td>
<td>185.8</td>
<td>-0.442</td>
<td>0.163</td>
</tr>
<tr>
<td>Mean stress (2/3 Y)</td>
<td>-0.145</td>
<td>224.7</td>
<td>-0.498</td>
<td>0.185</td>
</tr>
</tbody>
</table>

The following remarks are drawn:

- Plastic strain - 2Nf: a) the slope is steeper, about 10 percent change  
  b) $\varepsilon_f$ is altered, about 10 percent change  
  c) the two lines give nearly the same plastic strain amplitude for short life.
-elastic strain - 2Nf: a) the slope is steeper, about 30 percent change
b) \( \sigma_f \) is altered, about 20 percent change
c) the two lines nearly agree for short life.

The effect of mean stress is more significant at long life. The strength resistance of rail steel is altered by the application of a mean stress of 2/3 the flow stress.

The modified (elastic-strain amplitude - 2Nf) equation suggested by Morrow:

\[
\frac{\Delta \varepsilon_e}{2} = \frac{\sigma_f - \sigma_0}{E} (2Nf)^b; \quad \sigma_0 = 2/3 Y
\]
gives a good fit to the experimental data (Fig. 6-6).

6.4 Mean Stress Relaxation

From the log-log plot of mean stress ratio versus the number of cycles, a linear regression was carried out in order to give an estimate of the average rate of decrease of the mean stress. The assumption is to approximate the curves (Fig. 6-4) with straight lines obeying the following law: \( \left( \sigma_{ON}/\sigma_{01} \right) = aN^b \). Table 6-2 gives the linear regression results.

The slope \( a \) is decreasing with increasing value of the applied mean strain and plastic strain amplitude. The plastic strain amplitude is the driving force for mean stress relaxation. The
greater the plastic strain amplitude, the greater $\beta$ is (Fig. 6-8). 

The decrease of both $\sigma_{\text{max}}$ and $\sigma_{\text{min}}$ (Fig. 5-9) suggests that there is softening in tension and hardening in compression. In terms of dislocations, there might be an obstacle building up, preventing motion on unloading and assisting it on loading.

An explanation can be found on the ground of variation of the backstress assisting dislocation motion on reverse straining. In the case of pearlitic steel, a dislocation substructure builds up due to the accumulation of plastic strain [27]. The cells of dislocations are walls perpendicular and along the ferrite-cementite interface. The pile-up of dislocations against this interface produces a backstress $\sigma_B$. Therefore one can write:

- forward flow stress $\sigma_f = \sigma_0 + \sigma_d + \sigma_B$
- reverse flow stress $\sigma_r = \sigma_0 + \sigma_d - \sigma_B$

$\sigma_0$: matrix yields stress; $\sigma_d$: hardening term.

Therefore, the mean stress is:

$$\sigma_m = \frac{\sigma_f - \sigma_r}{2} = \sigma_B$$

Thus, experimental data denote the decrease or the relaxation of $\sigma_B$. During cyclic deformation, there is mutual trapping of the dislocations on their moving back and forth. This would explain less pile-up and thereby a decrease of $\sigma_B$. 
<table>
<thead>
<tr>
<th>test #</th>
<th>$\varepsilon_m$ (%)</th>
<th>$\frac{\Delta \varepsilon}{2}$ (%)</th>
<th>$\frac{\Delta \varepsilon_{\text{p}}}{2}$ (%)</th>
<th>$\frac{\Delta \varepsilon_{\text{p}}}{2}$ (%)</th>
<th>$2N_f$</th>
<th>$\frac{\sigma_f}{\sigma_{\text{f}0}}$ (%)</th>
</tr>
</thead>
<tbody>
<tr>
<td>LCF-M2</td>
<td>6.20</td>
<td>1.36</td>
<td>1.01</td>
<td>0.35</td>
<td>324</td>
<td>8.9 (38)</td>
</tr>
<tr>
<td>LCF-M6</td>
<td>5.80</td>
<td>0.75</td>
<td>0.48</td>
<td>0.27</td>
<td>1,580</td>
<td>10.0 (37)</td>
</tr>
<tr>
<td>LCF-M4</td>
<td>4.35</td>
<td>0.50</td>
<td>0.24</td>
<td>0.26</td>
<td>6,020</td>
<td>7.4 (38)</td>
</tr>
<tr>
<td>LCF-M7</td>
<td>3.52</td>
<td>0.48</td>
<td>0.25</td>
<td>0.23</td>
<td>6,120</td>
<td>5.0 (38)</td>
</tr>
<tr>
<td>LCF-M5</td>
<td>1.73</td>
<td>0.267</td>
<td>0.097</td>
<td>0.17</td>
<td>35,420</td>
<td>31.0 (54)</td>
</tr>
<tr>
<td>LCF-M8</td>
<td>0.85</td>
<td>0.165</td>
<td>0.025</td>
<td>0.14</td>
<td></td>
<td></td>
</tr>
</tbody>
</table>

* Last measurement taken before failure. Number in brackets indicates the relative percent of life.

Table 6-1. Mean stress relaxation tests data.

![Hour glass specimen geometry](image)

Figure 6-1. Hour glass specimen geometry.
Figure 6-3. Rail steel. Mean stress relaxation data as a function of number of cycles.
Figure 6-5. Strain-life curve for rail steel. Superimposition of fully reversed data from ref.[7] to those with applied tensile mean stress.
Figure 6-6. Strain-life curve for rail steel. Superimposition of fully reversed data from ref.[7] to those with applied tensile mean stress.
Figure 6-7. Strain-life curve for rail steel. Superimposition of fully reversed data from ref.[7] to those with applied tensile mean stress.
\[ \frac{G_{om}}{G_{01}} = \alpha N^\beta \]

Linear regression results:

<table>
<thead>
<tr>
<th>Test #</th>
<th>( \beta )</th>
<th>( \alpha )</th>
<th>( r )</th>
</tr>
</thead>
<tbody>
<tr>
<td>LCF-M2</td>
<td>-0.417</td>
<td>1.082</td>
<td>0.961</td>
</tr>
<tr>
<td>LCF-M6</td>
<td>-0.343</td>
<td>1.223</td>
<td>0.985</td>
</tr>
<tr>
<td>LCF-M4</td>
<td>-0.296</td>
<td>1.469</td>
<td>0.941</td>
</tr>
<tr>
<td>LCF-M7</td>
<td>-0.306</td>
<td>1.306</td>
<td>0.944</td>
</tr>
<tr>
<td>LCF-M5</td>
<td>-0.125</td>
<td>0.978</td>
<td>0.999</td>
</tr>
<tr>
<td>LCF-M8</td>
<td>-0.030</td>
<td>0.852</td>
<td>0.899</td>
</tr>
</tbody>
</table>

Table 6-2. Calculation of the average rate of mean stress decrease through linear regression.

Figure 6-8. Slope \( \beta \) with respect to \( \Delta \varepsilon_p/2 \).
Figure 6-9. Decrease of $\sigma_{\text{max}}$ and $\sigma_{\text{min}}$ with number of cycles.
CHAPTER 7
DISCUSSION OF RESULTS

Macro and microhardness measurements on used rails indicate a value of 32 Rockwell C near the running surface of the gauge corner, and 20 RC at the center of the rail head. This is in agreement with what is usually found [69]. Vickers hardness profiles yield values of 450-500 VHN near the running surface to a lower value of 250 VHN in the bulk region which is found at a depth of 6 to 7 mm (vertical distance). The pearlite lamellae are severely cold worked in the subsurface region. This region has therefore a higher yield strength. Cracks initiate in this area. Fatigue crack initiation by slip is related to the development of PSB. The pearlite spacing is the controlling parameter. Crack initiation due to defects or inclusions is associated with yielding and plastic deformation. Reducing the extent of such a zone would decrease the probability of such an occurrence. A higher strength material is beneficial for a better wear resistance but shelling-type cracks initiate where the hardness increases [70]. The interlamellae spacing and the thickness dependence of the deformation behavior of the cementite lamellae should give the trade-off between wear resistance and fracture resistance. The transverse crack usually takes place in the soft region. Its origin is found to be close to the cold worked region. Replicas of the fatigue crack surfaces denote beach marks.

Fatigue crack propagation tests run in air show the effect of the R ratio. A large effect of R on the rate of crack growth is observed
in Regime II (R = 0.7, \(\frac{da}{dN} = 2.7 \times 10^{-12} (\Delta K)^{3.6}\); R = 0.5, \(\frac{da}{dN} = 8.6 \times 10^{-13} (\Delta K)^{3.6}\)), contrary to what is found with tests of rail steels of ref. [21]. The curves fall out of the scatter band including the results of [21]. Therefore chemical composition, mechanical properties and microstructure may exert an influence on \(\frac{da}{dN}\).

The double log plots of \(\frac{da}{dN}\) vs. \(\Delta K\) are S-shaped curves with a \(\Delta K\)-threshold an an asymptotic limit for fast fracture. The R ratio has a large effect on \(\Delta K\)-threshold. Experimental data obey a relationship different from [21]. This relationship, in air, is of the type \(\Delta K_{th} = C(1 - R)^\gamma\) where \(\gamma = 1\). The \(\Delta K\) threshold values range from 3.8 MPa \(\sqrt{m}\) (R = 0.7) to 8.5 MPa \(\sqrt{m}\) (R=0.05). \(K_{max}\) at threshold was observed to be constant for R = 0.05, 0.3 and 0.5 (8.7 MPa \(\sqrt{m}\)). It was higher for R = 0.7. This R ratio effect is explained in terms of oxide-induced crack closure at low R in air, and susceptibility to corrosion embrittlement at high R.

The fracture surfaces show different features at different \(\Delta K\) levels. Fatigue striations as well as cleavage facets are observed. In the mid \(\Delta K\) range, striations refer to pearlite lamellae [7]. They are also present near \(\Delta K\)-threshold. The micromechanisms of fatigue crack propagation change with \(\Delta K\). At low \(\Delta K\), the effects of microstructure and environment are strong. At high \(\Delta K\), a cleavage mode is superimposed to a fatigue striations mode.

**Retardation.** The application of an overload (\(K_{OL}/K_{max} = 1.3\); \(\Delta K_{applied} = 16 \text{ MPa } \sqrt{m}, \text{ R } = 0.05\)) results in a change of the crack
growth rate. Delayed retardation is observed. The crack growth rate decreases to a minimum value at a distance of 0.1 mm ahead of the overload front and it slowly increases back to its original CAL value. Wheeler's and Willenborg's models do not account for delayed retardation. Simple geometric considerations do not account for the actual mechanism that controls crack retardation. A closure stress is likely to govern da/dN during retardation because its change was reported in the literature.

On the whole, retardation takes place over a distance approximately equal to the overload plastic zone size. For rail steels, Wheeler's model gives a much lower minimum value of crack growth rate and it fits the data during the regime of increasing crack growth rate. Willenborg's model does not agree with experimental data. This is due to the fact that the calculated $R_{eff}$ rises soon enough and outnumbers the applied $R$ value when the distance of the crack tip to the overload front is barely equal to 1/4 the overload plastic zone size.

Mean stress effect. The relaxation of a mean tensile stress equal to 2/3 the flow stress is a function of the plastic strain amplitude. The relaxation is almost complete for large cyclic plastic deformations. The decrease of the mean stress is related to the decrease of the backstress. The backstress is due to dislocation pile-ups on ferrite-cementite interfaces. Dislocations trap each other by their back and forth motion.

A tensile mean stress of 2/3 the flow stress reduces the fatigue
life by a factor of 2 for the short life and 10 for long life. The Morrow procedure accounts for mean stress effects and gives a good fit to experimental data.
CHAPTER 8
CONCLUSIONS

The fatigue properties of rail steels were investigated in different areas: wear resistance, crack resistance, fatigue crack propagation, crack retardation, mean stress effect.

During service, the rail head is cold worked resulting in extensive plastic deformation of the pearlite lamellae. The macrohardness number is RC 32.5 below the running surface of the gauge corner and RC 21 in the center of the rail head. The microhardness measurements show variation of the BHN from 500 (running surface) to 250 (bulk pearlite). This change takes place over a distance of 6-7 mm.

Fatigue crack growth rate data in air are correlated by Paris' Law at different R ratios. In each case, the slope of the da/dN - \( \Delta K \) data is 3.65. Forman's equation was checked out. There is agreement with data but a factor of 2 in the coefficient \( C' \) tells apart data for \( R = 0.07 \) and the set of the others (\( R = 0.05, 0.3, 0.5 \)).

The \( \Delta K \)-threshold value is a linear function of \( R \), when \( R \) is smaller than 0.5. \( K_{\text{max}} \) at threshold is observed to remain constant within this range of \( R \) values. The \( \Delta K \)-threshold value decreases with increasing \( R \). The relationship between these two parameters was used to calculate crack closure at threshold.

Micrographs of the fracture surfaces denote a mixed modes mechanism for crack propagation. Fatigue striations as well as planar facets are observed. Cleavage mode takes over at high \( \Delta K \).
The application of a tensile peak overload during fatigue crack propagation at $R = 0.05$ results in delayed retardation on rail steel. With an applied $\Delta K$ of about 16 MPa m, the minimum crack growth rate is observed at $1/6$ the overload plastic zone size. The ratio of minimum crack growth rate to the original value is between 2 and 4. Wheeler's and Willenborg's models do not account for the delayed retardation observed on rail steel. Wheeler's model predicts a much lower minimum value of the crack growth rate. It fairly well fits the rising part of the curve $da/dN$ vs. $a$. Willenborg's model does not fit the data at all.

The relaxation of a mean tensile stress equal to $2/3$ the flow stress is a function of the cyclic plastic strain amplitude and the applied mean strain. Under strain control, the shakedown of the hysteresis loops is observed. In terms of dislocations, this corresponds to a decrease of the backstress. The general trend observed is that a mean tensile stress reduces the fatigue life when comparison is made with fully reversed straining.
CHAPTER 9

RECOMMENDATIONS FOR FURTHER RESEARCH

The following considerations are suggestions for further understanding of the fatigue behavior of rail steels:

- Micrography should be performed to investigate the 90° turndown of the shell-type cracks. Their sites of initiation should deal with stress analysis, surface defects (pits) and inclusion content as well as deformation resistance.

- Correlation between FCP, LCF, CSS data and microstructure (pearlite spacing, grain size) should be made in order to have a rationale for improving the fatigue properties.

- The study of dislocation structure should inform about the microstructure influence on threshold, on the CSS curve and on mean stress relaxation.

- More complex loading conditions should be designed to study the load sequence effects in order to predict the fatigue life under random loading conditions.

- An analysis combining thermal stresses, residual stresses, random loading and structural changes is expected to account for effective crack growth calculation.
APPENDIX A

$K_{op}$ is assumed to be a function of $R$ and $K_{max}$. The existence of its derivatives of the first order as continuous functions is assumed from the $K_{op}$ calculation procedure [42].

Therefore: $K_{op} = g(R, K_{max}), \quad g \in C^1$.

Now:

$$U = \frac{K_{max} - K_{op}}{K_{max} - K_{min}} = \frac{K_{max} - g(R, K_{max})}{(1 - R)K_{max}} = f(R). \quad (Also, f \in C^1).$$

Since $f$ only depends on $R$,

$$\frac{af}{aK_{max}} = 0$$

$$\frac{af}{aK_{max}} = \frac{-1}{(1 - R)K_{max}^2} \left[K_{max} \frac{ag}{aK_{max}} (R, K_{max}) - g(R, K_{max})\right]$$

Therefore:

$$\frac{af}{aK_{max}}(R) = 0 \iff K_{max} \frac{ag}{aK_{max}} (R, K_{max}) - g(R, K_{max}) = 0$$

$$g \in C^1$$

The solution to this differential equation is:

$$g(R, K_{max}) = \alpha(R)K_{max}$$

where $\alpha$ is a function of $R$ ratio.
REFERENCES


