# Creep and fatigue crack growth at elevated temperatures
# Kúszási és fáradási repedésterjedés növelt hőmérsékleten

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*„This paper is dedicated to the memory of Tibor Konkoly, PhD., DSc., Univ. Prof. Emeritus (1924-2000), whose support and advice will always be remembered."*

*Ezt a tudományos közleményt néhai Dr. Konkoly Tibor emeritus egyetemi tanár (1924-2000) emlékére ajánljuk, akinek értékes támogatására és tanácsaira mindig emlékezni fogunk.*


## Abstract

Engineering investigations of creep and fatigue, frequently involving fracture problems, require practical answers from modeling based on fundamental principles. It has been realized that a model suitable at atomic level may not be appropriate to solve the subsequent difficulties encountered at the macro level. However, the use of meso-fracture mechanics (FM) modeling methods, where definite assumptions and generalizations can be made, could provide an answer in macro-engineering situations. In this paper two crack growth models are described: the first, linking the micro to macro process in creep, while the second model of fatigue crack growth enables use of meso-fracture dynamics in cyclic crack growth at a range of frequencies and temperatures.

Creep crack growth can be separated into two regimes of failure, consisting of initiation (incubation) and steady cracking behavior. A creep process zone is developed at a grain size level and the application of parameter $C^*$ will enable the predicting of both creep stages. Subsequent fatigue crack growth is completed using the second model.

## Összefoglalás

A kúszás és a fáradás jelenségére vonatkozó mérnöki kutatásoknál, amelyek gyakran a törés problémáját is magukba foglalják, olyan, a gyakorlatban használható válaszokra van szükség, amelyek alapvető elvekre épülő modellekre támaszkodnak. Nyilvánvaló, hogy egy atomi méretekre alkalmazott modell nem minden esetben alkalmas arra, hogy a makroszkópikus szinten jelentkező nehézségeket ezen az alapon lehessen megoldani. Ezzel szemben a mezo-törésmechanika (FM) modellezési módszerei, amelyeknél bizonyos feltételezéseket és általánosításokat lehet tenni, alkalmasak lehetnek a makroszkópos, mérnöki kérdések megválaszolására. Az alábbi cikkben két repedés-növekedési modell szerepel: az első összekapcsolja a kúszás mikro- és makrofolyamatait, míg a másik modell a fáradásos repedés terjedésére vonatkozik, és ez lehetővé teszi a mezo-dinamikus törésmechanika alkalmazását a repedésterjedésre a frekvencia és a hőmérséklet függvényében.

A kúszási repedésterjedés a károsodás két szakaszára bontható: a kezdeti, inkubációs szakaszra és az állandósult, stabil repedésterjedésre. A repedés tövében kialakul egy *károsodó kúszási zóna (creep process zone)*, amelynek nagyságrendje a krisztallitok méretével egyezik, és a $C^*$ nemlineáris törésmechanikai paraméter alkalmazása (meghatározását lásd a Függelékben) lehetővé teszi a becslést mindkét kúszási szakaszban. Ez után vizsgáljuk a fáradási repedésterjedést a második modellel.


## Introduction

Sudden, often catastrophic failures of engineering materials, due to the application of cyclic loading are generally referred to as fatigue failures. Their importance is well recognised and the damage development in all types of materials and structures reported in the literature. Fatigue failures are usually caused at much lower stresses than under monotonic loading. Two common approaches for characterizing fatigue behavior are S – N curves (such as Woehler curves) and fatigue crack propagation (FCP) analysis used in the present paper.

Crack growth behaviour has been investigated for many materials and it was concluded that a large number of factors might substantially influence failures. Changes in temperature and cyclic frequency are of particular importance. Increasing temperature and decreasing frequency may lead to a substantial decrease of fatigue resistance. Interaction of creep at higher temperatures, as well as long hold times incorporating in structural cycling, may lead to material degradation and consequent life reduction.

The creep fracture and failures in components are usually related to initiation and growth of cracks in areas of high stress concentration, heat affected zone, and in degraded materials due to overaging and damage [1]. Localised embrittlement in a creep ductile material may also fail in a creep-brittle manner. Therefore it is possible to induce a creep crack growth in a material with a high uniaxial creep ductility if the crack tip damage is contained locally by means of geometry and constraint, or some form of material damage and degradation.

The aim of this study was to show the applicability of non-linear fracture mechanics (FM) crack growth model, at the meso-fracture mechanics level, based on $C^*$, to the initial and steady state stages of creep crack. Before creep redistribution occurs the crack tip stress profile can be described by linear fracture analysis. But with the advance of damage with time, the parameter $C^*$ will be necessary to use. Simple high temperature tests results on engineering materials are presented in order to show the usefulness of the model in a practical situation. The crack growth data are compared with the predicted calculations for cracking rates using the present transient and steady state models. The characterisation of the crack growth rates in the initial stage of damage accumulation at first loading is described in terms of initial drop in the crack growth rate followed by the gradual increase in the rate to coincide with the steady state cracking rate.

Some of the limiting properties necessary in the preliminary design are: detailed operating temperatures of the components involved in the life analysis calculated in the operation schemes, such as power plants and aerospace engineering, or those parts of the machinery specified in detailed programmes; yield stress; ultimate strength; frequency and strain rate involved in operation.

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## Creep crack growth based on a creep process zone

A new model of creep crack growth [1] is based on stress and strain fields, assuming a creep damage process zone as in *Fig. 1* and described in the following section.

Generally, when creep and time dependent mechanisms are involved, fractures are intergranular and the damage at the crack tip ranges from a local sharp crack to void ahead of the crack tip depending on the creep ductility of the material and the level of constraint. In the range of creep dominant cracking it is more appropriate to use non-linear fracture mechanics concepts and express the steady state time dependent cracking rate $\dot{a}$ as a function of the creep fracture mechanics parameter $C^*$ by assuming that the creep strain rate $\dot{\varepsilon}$ is governed by the simplified Norton's creep rate [2]

$$\dot{\varepsilon} = C\sigma^n \qquad (1)$$

or in the form

$$\frac{\dot{\varepsilon}}{\dot{\varepsilon}_o} = \left(\frac{\sigma}{\sigma_o}\right)^n \qquad (2a)$$

and earlier developed – empirical (by Garofalo)

$$\dot{\varepsilon}_s = H(\sinh\alpha\sigma)^n \qquad (2b)$$

We obtain for a cracked body

$$\dot{a} = D_0 C^{*\Phi} \qquad (3)$$

where $\Phi = n/(n+1)$, a number slightly less than unity, and $C$ are material constants, $n$ is the creep index in Norton's creep law, $C^*$ is the non-linear creep parameter evaluated at maximum load that is $C^*$ is the creep equivalent of J integral using strain rate instead of strain in the above equations. (See formula in the Appendix). $D_0$ is the proportionality factor determined chiefly by the material creep ductility and the constraint local to the crack tip. An increase in crack growth is obtained by an increase in the degree of constraint and with decrease in ductility. Therefore material cracking behaviour will be controlled by both the material creep properties, specimen geometry and test variables such as temperature and frequency. Equation (3) has been experimentally investigated and shown to describe time dependant cracking behaviour over a range of crack tip constraints. $D_0$ and $\Phi$ are dependant on temperature and stress state. Thus the crack growth rate at elevated temperatures in eq. (3) will depend on material properties, loading conditions and time during the actual crack propagation.

### Steady state behaviour

A model of creep crack growth under steady state conditions has been proposed by Nikbin, Smith and Webster [1] and is known as the NSW model. It is based on stress and strain fields characterised by $C^*$ and combined with creep damage mechanism. When the damage is accumulated in the zone, *Fig. 1*, and reaches a certain failure value, the crack is postulated to progress.

The condition for the crack growth is

$$\dot{\varepsilon}_{ij} = \int_{r=r_c}^{t} \dot{\varepsilon}_{ij}\,dt \qquad (4)$$

The model can be described as follows [1, 3, 4]: In order to predict crack growth at high temperatures a creep process zone (CPZ) was modelled at the crack tip, *Fig.1*, where cracking proceeds when an element of material experiences damage and rupture stress according to the local magnified state of stress. This non-linear stress singularity determines the rate at which the element of material accumulates damage, and failure occurs when the creep ductility $\varepsilon_f^*$ appropriate to the state of stress (or to the extent of constraint) at the crack tip is exhausted. For the case of plane stress $\varepsilon_f^* = \varepsilon_f$, where $\varepsilon_f$ is the uniaxial creep ductility, the NSW model [1] for steady state creep crack growth gives $\dot{a}_s$ in the form:

$$\dot{a}_s = \left\{(n+1)/\varepsilon_f^*\right\}\left[C^*/I_n\right]^{n/(1+n)}(Ar_c)^{1/(n+1)} \qquad (5)$$

where $I_n$ is a non-dimensional function of $n$, $r_c$ is the creep process zone, CPZ size over which each element sees the appropriate stress history and A is a material constant. This expression assumes the development of steady state damage distribution in the region of the process zone. It assumes that zero damage exists at $r = r_c$ and that progressively more damage is accumulated as the crack tip is approached. Therefore the model assumes that only little extra strain is required to break a small ligament *dr* at the crack tip since it will be almost broken before the crack will reach it. Since $D_0$ can be approximated by $3/\varepsilon_f^*$, equation (5) can be simplified for practical engineering application generally as

$$\dot{a} = 3C^{*\Phi}/\varepsilon_f^* \qquad (6)$$

known as the „approximate NSW engineering model" which describes creep crack growth from plane stress to plane strain, where $\dot{a}$ is in mm/h, $C^*$ in MJ/m²h and the creep strain $\varepsilon_f^*$ as a fraction. The value of $\varepsilon_f^*$ is taken as the material uniaxial creep ductility $\varepsilon_f$ in plane stress, and as $\varepsilon_f/50$ for the case of plane strain [4].

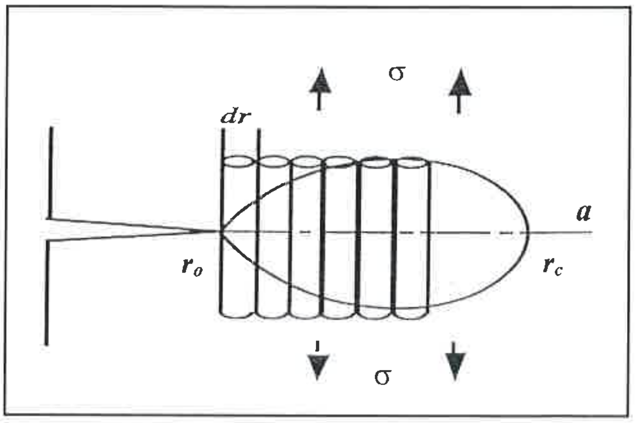


**Fig. 1** Creep cracking process zone
**1. ábra.** A kúszási repedésterjedés során keletkező károsodó zóna

### Transient analysis

It was previously reported [5] in the creep tests on low alloy steels (CrMoV and CrMo), *Fig. 2*, that the initial loading produced a „tail" at the beginning of the crack growth and discussed later. This sudden and unexpected steep crack growth can be attributed to a time dependent creep transition behaviour as shown in *Figs. 2a* and *2b* in terms of *da/dt* vs. $C^*$ [2, 6]. Similar tests on other materials confirmed that this behaviour is not singular. *Fig. 3* presents our results on Al-alloy RR58 for a range of temperatures showing the tails at the start of the crack growth in Region I on DCB specimens. An increase of specimen size and possibly the side growing may assist to the rapid cracking rate. ASTM standards were used in all experimental work with the geometry of CT (compact type), but for some tests the behaviour of DCB (double cantilever beam , contoured) was also investigated *(Fig. 4)*.

The transition time is usually described as the time taken to go from the elastic state of stress *(Fig. 5)* at the crack tip to a steady static creep stress referred to $C^*$ stress field, see also [6]. By eliminating the data points up to the transition time, the actual periods of the tails that have been measured, are still apparent in the test data. This suggests that

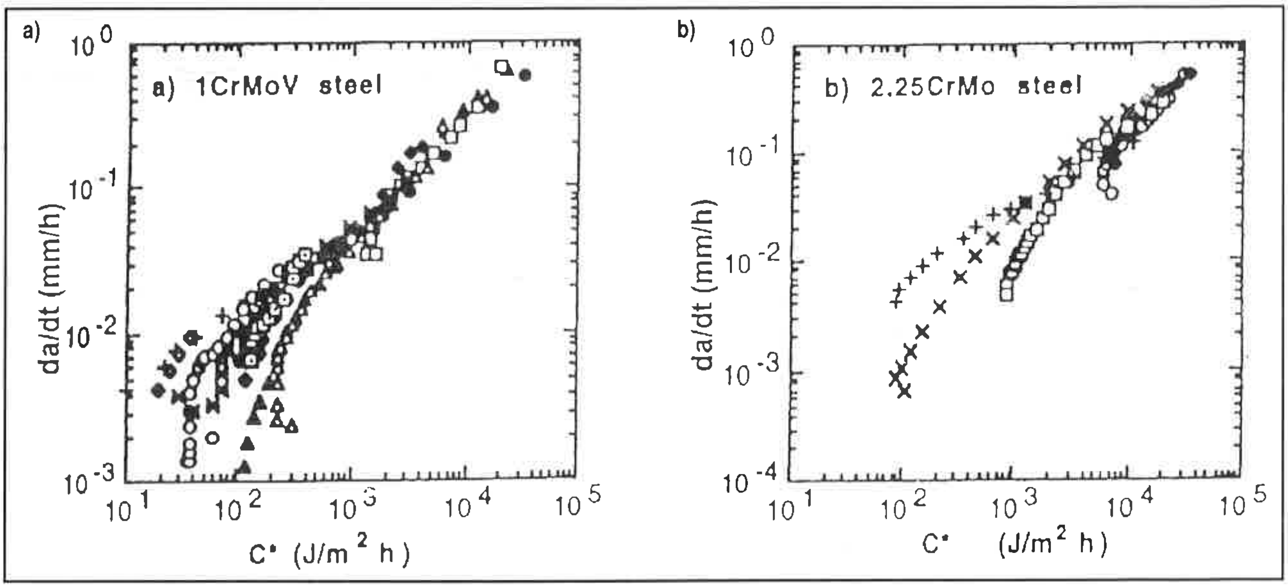


**Fig. 2** Creep crack growth rate versus C* of *a)* 1CrMoV and *b)* 2.25 CrMo steel at 550 °C in CT specimens
**2. ábra.** A kúszási repedésterjedés sebessége a C* függvényében *a)* az 1CrMoV acélra, *b)* a 2.25CrMo acélra 550 °C-on, CT próbatesteken mérve

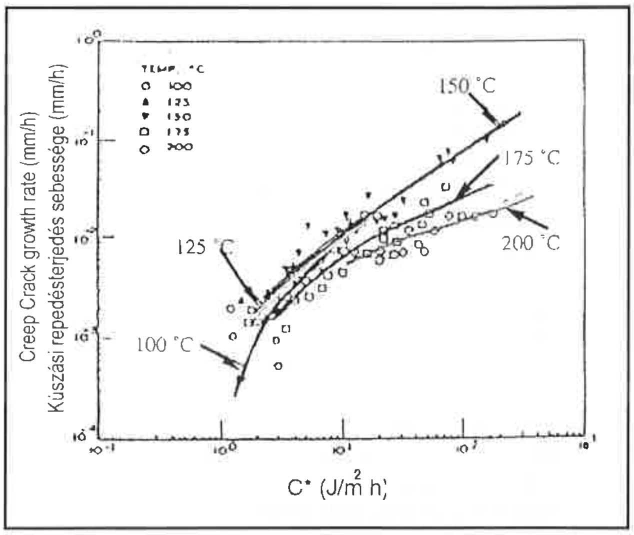


**Fig. 3** Dependence of creep crack growth rate on temperature correlated versus C*. Aluminium Alloy RR58 tested between 100–200 °C using DCB specimens
**3. ábra.** A kúszási repedésterjedési sebesség a hőmérséklet (100–200 °C) és a C* függvényében.
A vizsgált anyag: RR58 jelű alumínium ötvözet, DCB próbatestek

upon the first loading the steady state situation at the crack tip has not yet developed, *Fig. 2.* Therefore, at the first loading in Region I a stable distribution of damage will have to be built up ahead of the crack before the real steady state crack growth begins.

Earlier study by Kenyon [6b] on RR58 confirmed that the initial period $t_i$ in *Fig. 6* can be reduced with increased initial crack length. Other

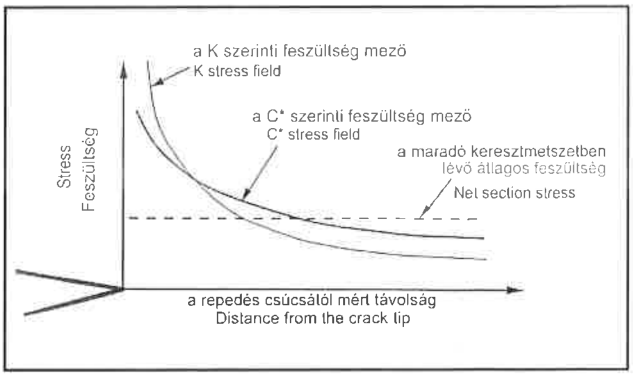


**Fig. 5** Elastic and steady state creep stress distribution ahead of a crack tip
**5. ábra.** A feszültségek eloszlása a repedés csúcsa előtt a rugalmas és az állandósult kúszás állapotában.

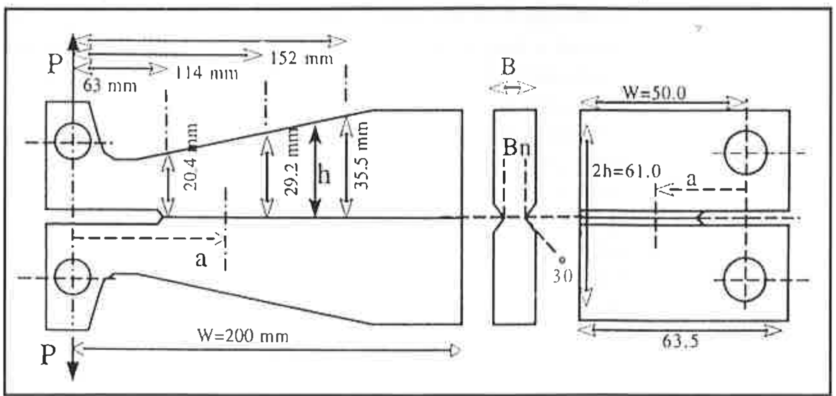


**Fig. 4** Typical dimensions of DCB-C and the CT specimen employed in the tests. The DCB-P has the same size as the DCB-C but with parallel sides
**4. ábra.** A vizsgálatoknál használt DCB-C és CT próbatestek jellemző méretei.
A DCB-P próbatestek méretei azonosak a DCB-C méreteivel, de az oldalak párhuzamosak

possibility of this time reduction was observed by soaking the specimen at the test temperature before loading, and this indicate both geometry and aging sensitivity [6b].

Reconsider now the transient creep crack growth model in *Fig. 6.* The rather small differencies between the steady and transient crack growth are due to the transient creep crack growth. This creep growth is starting from the undamaged state of material in the creep damage process zone, whereas the steady state model assumes a steady state distribution of damage in the process zone. It should be remembered that the creep damage accumulates in the micro-elements of the width *dr, Fig. 1,* ahead of the crack tip. Note also, that the steady creep crack growth can be predicted analytically.

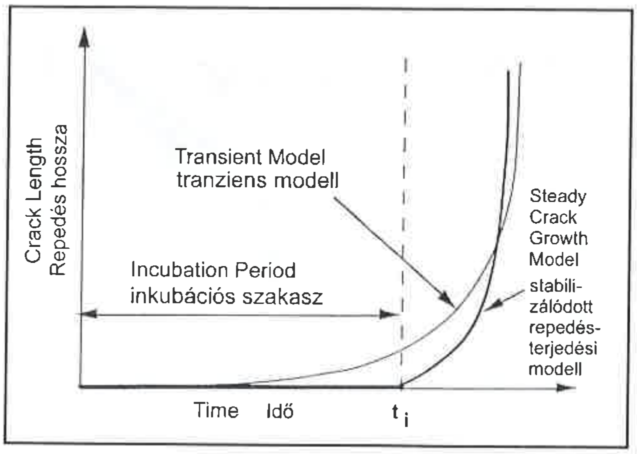


**Fig. 6** Schematic time change of crack length using the combination of the incubation time and the steady crack growth
**6. ábra.** A repedés hossza és az idő közötti összefüggés vázlatos ábrázolása, felhasználva az inkubációs idő és a stabilizálódott repedésterjedési sebesség kombinációját

The small ligament *dr* (*Fig. 1.*) will not have suffered any creep strain, and failure will not occur until a time *dt* has elapsed (cf. eq. (4)) given by

$$\varepsilon_f^* = \dot{\varepsilon}\, dt \qquad (7)$$

where $\dot{\varepsilon}$ is the creep displacement rate. This leads to an initial creep crack growth rate $\dot{a}_0$ giving

$$\dot{a}_0 = dr / dt \qquad (8)$$

and on adjustment giving

$$\dot{a}_0 = \left(1/\varepsilon_f^*\right)\left[C^*/I_n\right]^{n/(1+n)}(A\, dr)^{1/(n+1)} \qquad (9)$$

Equation (9) is very similar to that derived from the steady state damage conditions (equation (5)). It results in the relation

$$\dot{a}_0 = (1/(n+1))(dr/r_c)^{1/(n+1)}\dot{a}_s \qquad (10)$$

the ligament *dr* can be chosen to be a suitable fraction of $r_c$. However, since $dr/r_c$ is raised to a small power in equation (10) it can be reduced to

$$\dot{a} \approx (1/(n+1))\dot{a}_s \qquad (11)$$

For most engineering materials therefore the initial crack growth rate is expected to be approximately an order of magnitude less than that predicted from the steady state analysis. The cracking rate will progressively reach the steady state cracking rate $\dot{a}_s$ when the damage is accumulated. Numerical integration is required to evaluate equation (11). A computer program has been developed, using incremental crack extension, to evaluate the transition period resulting from the development and the accumulation of damage in accordance with equations (5 and 10).

The comparison of the experimental cracking rates in *Fig. 2* for the two low alloy steels and the predicted results show that the trends at the initial stages and steady state of the cracking rates, where a tail exists in the experimental data, are predicted by the model satisfactorily. The results in *Fig. 2* are for CT specimens as in ASTM E1457 and the appropriate formulae in the Appendix. The values of $D_0$ and $\Phi$ used in equation (3) determine the accuracy with which the steady state crack growth rates are predicted. There are no visible differences in trends. All show varying degrees of an initial tail and the model within a factor of two or less describe the transition to steady state. For the two alloys the calculations show that at most the transient cracking rate is approximately a factor of five less than the steady state crack growth rate. The reason for this similarity may be due to the differing constraint conditions, either due to temperature, geometry or loading, that are imposed on the materials in order that they would behave in a creep-brittle manner under plane strain conditions. The cracking rate will progressively reach steady state cracking on the completion of damage accumulation.

## Fatigue crack growth

Most engineering components which operate at elevated temperatures are subjected to non-steady loading during service. For example, electric power plant required to follow the demand for electricity, or equipment used for making chemicals may undergo a sequence of operations during the production process. The power plants have to change their operating temperature and pressure to follow the demands of electricity needed and to shut down and restart for their routine maintenance. Also aircraft experience a variety of loading conditions during take-off, flight and landing. There may, in addition, be a superimposed high frequency vibration. Similarly, equipment that is subjected to predominantly steady operating conditions may experience transients during start-up and shut down.

The gradual increase over the years in operating temperatures to achieve improved efficiency and performance of plants is causing materials to be used under increasingly arduous conditions. The first stage blades in aircraft gas-turbines can, for example, experience centrifugal stresses in the region of 150-200 MPa at gas temperatures that can exceed their incipient melting temperatures. Under these circumstances, survival is only possible if cooling is adopted to reduce average blade temperatures and coating are used to limit erosion and environmental attack. The steep temperature gradients produced by the cooling will, however, introduce thermal stresses which will be regenerated each flight cycle and which can give rise to a mode of failure called thermal fatigue. The same situation can occur during rapid start-ups and shut-downs in thick sections of other components. Thermal fatigue is the type of failure that can occur by repeated application of predominantly thermal stresses that are produced by local constraint imposed by the surrounding material.

It is apparent, depending on the material and operating conditions, that creep, fatigue and environmental processes may contribute to failure. The dominating mode of failure in particular circumstance will depend on such factors as material composition, heat-treatment, cyclic to mean load ratio, frequency, temperature and operating environment.

In the preceding sections, creep crack growth under steady loading was discussed. For the prediction of the crack growth in components at elevated temperatures, the interaction of creep and fatigue crack growth are introduced first. Then a prediction of crack growth under the combination of creep and fatigue will be outlined.

Before considering creep-fatigue crack growth, some observations concerning the characteristics of fatigue crack growth will be discussed. References [7, 8] give various description and views of the problems involved in fatigue and creep/fatigue conditions. Fatigue mechanisms usually dominate at room temperature. Procedures for measuring the fatigue crack propagation properties of material at room temperature are described in reference [10]. Fatigue crack growth is usually observed as transgranular cracking at low temperature. At elevated temperature, transgranular cracks are also observed under relatively high frequency cycles ($f > 1\,\mathrm{Hz}$) and this fatigue crack growth rate can still be characterised by elastic or elasto-plastic fracture mechanics parameters in most cases. Thus in the range of one-parameter fracture mechanics, when the effect of plasticity can be neglected, the stress field is expressed in the form

$$\sigma_{ij} = \frac{K}{\sqrt{2\pi r}} F_{ij}(\theta) \qquad (12a)$$

around a crack tip, where $F_{ij}$ is a function of the polar angle $\theta$ and $K$ is the stress intensity factor.

The crack growth per cycle $da/dN$ is often plotted against the stress intensity factor range $\Delta K$ which is defined as the difference between the maximum $K_{max}$ and minimum $K_{min}$ stress intensity factors applied each load cycle. $\Delta K$ is given [7] as

$$\Delta K = F\Delta\sigma\sqrt{a} \quad \text{(12b)}$$

where $\Delta\sigma$ is the range of the applied stress, and $F$ is the non-dimensional geometry factor. The crack growth rate $da/dN$ is correlated with $\Delta K$ using the power law relation in steady state fatigue as,

$$da / dN = C_f \Delta K^m \quad \text{(13)}$$

where $C_f$ and $m$ are material constants and $m$ is typically around 3. Typically, $da/dN$ is sensitive to the mean stress or load ratio $R$ defined by

$$R = \frac{\sigma_{min}}{\sigma_{max}} \quad \text{(14)}$$

where $\sigma_{min}$ and $\sigma_{max}$ are the minimum and the maximum stress, respectively. To include the effect of the load ratio, the effective stress intensity range $\Delta K_{eff}$ is used, several formulae to define $\Delta K_{eff}$ have been proposed. However, it should be noted, that the majority of fatigue tests were performed at sinusoidal loading, between $\sigma$ = zero and $\sigma_{max}$, as described elsewhere [9] and constant temperature.

In the range of conventional one-parameter fracture mechanics (FM) method the relation (13) is assumed to be a material curve, i.e. for a given material and a set of boundary conditions (including a particular value of $R$), the growth rates should depend only on $\Delta K$, i.e., the $da/dN$ vs. $\Delta K$ curve should be independent on the geometry of the body. However, it has been noted that with $\Delta K$ held constant, the rate of the fatigue crack propagation may depend on the geometry of a specimen and varies with the degree of constraint in front of the crack tip.

Two-parameter fracture mechanics method is based upon the assumption that the fracture behaviour of the test specimen is the same as that of the structure, if both have the same value of the stress intensity factor and the same range of constraint parameters. The elastic T-stress used in the two-parameter FM denotes a constant stress acting parallel to the crack flanks and is related to the first non-singular term in the Williams expression of the stress field [8].

In the region of two-parameter FM this stress field can be expressed in the form

$$\sigma_{ij} = \frac{K}{\sqrt{2\pi r}} F_{ij}(\theta) + T\,\delta_{li}\,\delta_{lj} \quad \text{(15)}$$

where $F$ is a function of the polar angle $\theta$.

Here the influence of constraint on fracture behaviour is described using mainly T-stress and it was shown that the fatigue crack rate is influenced substantially in cases where T-stress is negative. The effect of positive T values on the rate of propagations is much less pronounced than it is for negative T stress. The one-parameter description is rigorously correct only for T = 0, but can be accepted for positive values of T as well. Note that for three point bend specimen (and $a/w > 0.4$) and compact tension specimen the values of T-stress are positive and both approaches (one- and two-parameter) are practically identical. Further particulars on the T-stress influence are in [8].

### *Elevated temperature cyclic crack growth*

In the previous section cyclic crack growth was considered where time dependent effects were not important and where cracking was controlled mainly by fatigue mechanisms. As temperature is increased, time dependent processes become more significant. Creep and environmentally assisted crack growth can take place more readily since they are aided by diffusion and rates of diffusion increase with rise in temperature. The effects of frequency, temperature, mean stress and environment are considered elsewhere [9, 10].

Frequency effects can be investigated in continuous cycling tests, or by introducing hold periods into a cycle or by other changes. However, it is found that wave-shape is relatively unimportant compared with the temperature and mean load at which the cycling is taking place. Generally the influence of frequency on crack propagation rate is more pronounced with increase in temperature and $R$. *Figure 7* presents a schematic description of cracking rate versus $\Delta K$ for cyclic cracking at elevated temperatures showing the effects of frequency. $R$-ratio and temperature. A higher crack propagation rate is observed with increase in load range. In all cases, crack propagation rate increases slowly at first before accelerating rapidly as final fracture is approached. As a consequence, most of the time in these tests is spent in extending the crack a small distance and the number of cycles to failure is not significantly influenced by the crack size at which fracture occurs [5].

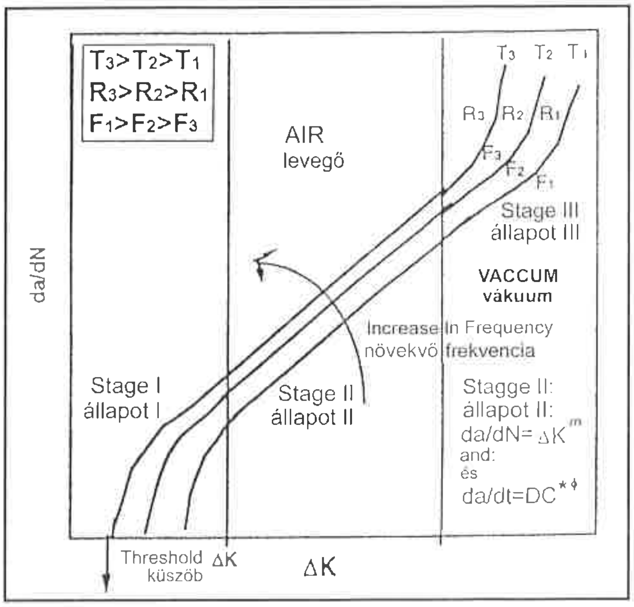


**Fig. 7** Cyclic crack growth at elevated temperatures
**7. ábra.** A fáradásos repedés növekedése növelt hőmérsékleten

### *Creep-fatigue crack growth interaction*

When alternating loads are applied to high temperature structures, the crack growth in the structures will be subject to both creep and fatigue. Interaction between creep and fatigue is expected under cyclic loading. Some of the causes of creep-fatigue interaction might be the enhancement of fatigue crack growth due to embrittlement of grain boundaries or weakening of the matrix in grains and enhancement of creep crack growth due to acceleration of precipitation or cavitation by cyclic loading. The importance of creep-fatigue interaction effects is largely dependent on the material and loading conditions. Nevertheless, the simple linear summation rule for creep-fatigue crack growth defined by the following equation has been successfully applied to predict the crack growth in several engineering metals [11]:

$$\frac{da}{dN} = \left(\frac{da}{dN}\right)_{creep} + \left(\frac{da}{dN}\right)_{fatigue} = \frac{1}{3600f}\left(\frac{da}{dt}\right)_{creep} + \left(\frac{da}{dN}\right)_{fatigue} \quad \text{(16)}$$

or

$$\frac{da}{dt} = \left(\frac{da}{dt}\right)_{creep} + \left(\frac{da}{dt}\right)_{fatigue} = \left(\frac{da}{dt}\right)_{creep} + 3600f\left(\frac{da}{dN}\right)_{fatigue} \quad \text{(17)}$$

where $da/dN$ is crack growth per cycle in mm/cycle, $da/dt$ is crack growth rate in mm/h , and $f$ is frequency in Hz.

Effects of the influence of frequency [5] on crack growth /cycle in a nickel base alloy (API), again using CT specimens, at 700 °C are shown in *Figs 8* and *9*. *Figure 8* indicates a dependence of $da/dN$ on frequency at a range of $R$. In the steady state cracking region crack growth can

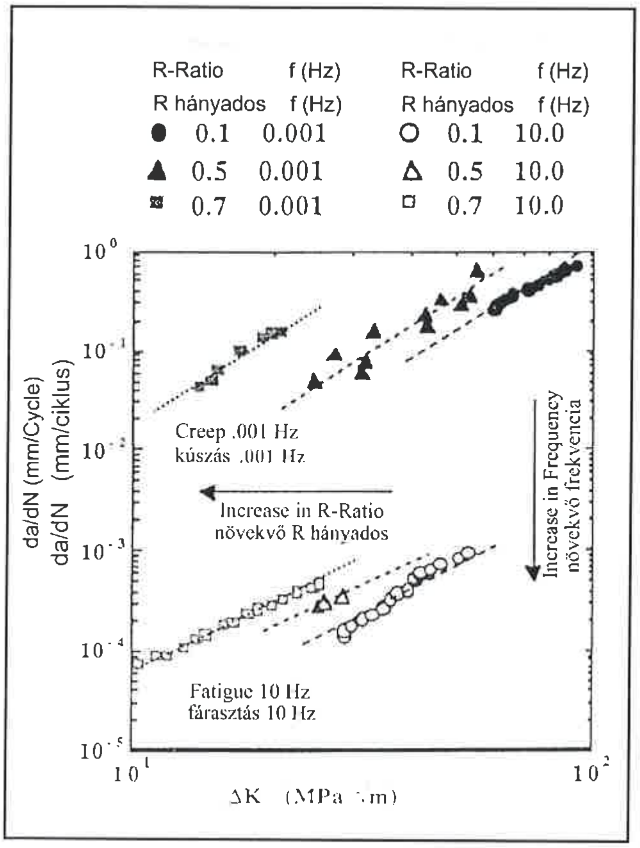


**Fig. 8** Dependence of crack growth on frequency and R-ratio for an AP1 nickel-base superalloy tested at 700 °C.

**8. ábra.** A repedésterjedés sebességének függése a frekvenciától (f) és az *R* hányadostól az AP1 nikkel alapú szuperötvözetnél 700 °C-on

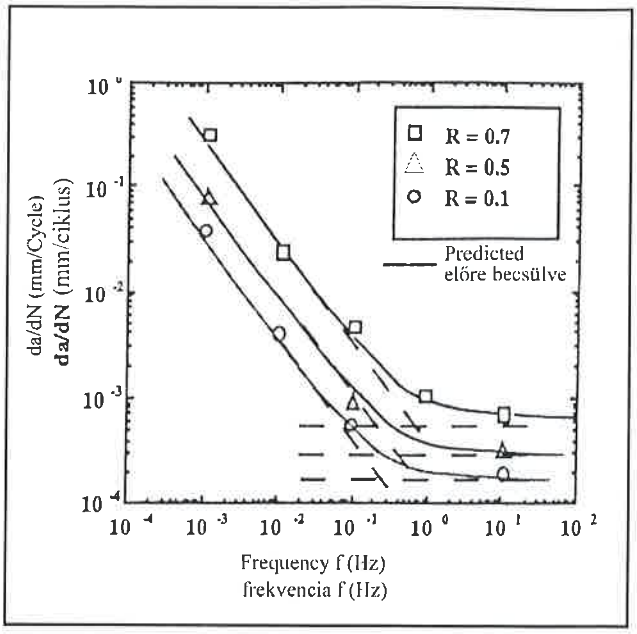


**Fig. 9** Fatigue crack growth sensitivity to frequency and R-ratio in an AP1 nickel base superalloy tested at 700 °C at constant $\Delta K = 30$ MPa.m$^{1/2}$

**9. ábra.** A fáradásos repedésterjedés (da/dN) érzékenysége a frekvenciára (f) és az R hányadosra az AP1 nikkel alapú szuperötvözetnél 700 °C-on és $\Delta K = 30$ MPa.m$^{1/2}$ állandó terhelésnél.

be described by the Paris law (see equation (13)) with $m \approx 2{,}5$. This value is within the range expected for room temperature behaviour. Correlation of results obtained at MPa$\sqrt{m}$ for specimen thickness (B = 25 mm) is depicted in *Fig. 9*. Furthermore, in Figure 9 the two dashed lines show the relative behaviour of the creep and the fatigue components of *da/dN*.

The slope of −1 indicates time dependent creep cracking and the approximate horizontal line indicates fatigue control of *da/dN*. Therefore by adding the two components it is clear that at high frequency creep is seen to have third order effect on cracking rate and conversely at low frequencies fatigue has a third order effect. From metallurgical and fractographic investigations performed on the alloy tested in the creep and creep-fatigue range similar qualitative conclusions were reached with respect to the mode of the creep-fatigue interaction.

The intermediate frequencies show a mixture of intergranular and transgranular cracking modes. These suggest that the two mechanisms work in parallel and that cumulative damage concepts proposed above can well describe the total cracking behaviour [11, 12].

## *Interpretation of the crack growth*

It was shown in the previous section that fatigue crack growth rates at a higher frequency $f = 1$ Hz can be successfully characterised by effective stress intensity factor range $\Delta K_{eff}$. However at a very low frequency, creep plays a large role in the crack behaviour [5]. To examine frequency dependence of crack growth at elevated temperatures, experiments under different loading frequencies were carried out at the laboratories of Imperial College on the 0.5CrMoV steels and new 2.25CrMo steel (to be reported later). Preliminary data are as follows: chemical composition of both low alloy steels are shown in *Table 1*. Elastic modulus E is 150 GPa at the testing temperature of 550 °C, yield stress $\sigma_0 = 100$ MPa. The ASTM formulae used for all tests are shown in the Appendix.

The relationships between frequency and crack growth per cycle for the new 2.25CrMo steel and 0.5CrMoV steel using CT specimens, B = 25 mm, are shown in *Figs. 10* and *11*. *Figure 10* is for maximum stress

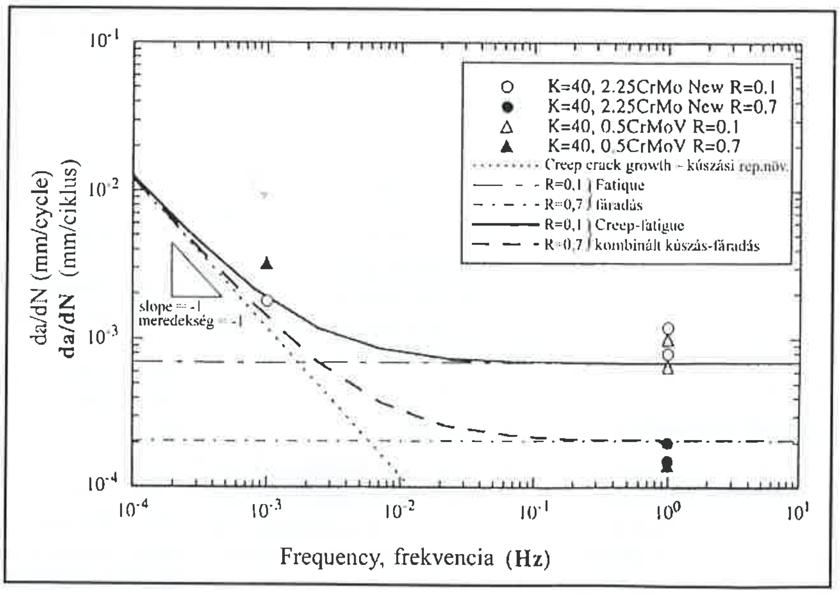


**Fig. 10** Frequency dependence of crack growth rate at $K_{max}$ = 40 MPa.m$^{1/2}$ for 0.5CrMoV steels and new 2.25CrMo steels

**10. ábra.** A repedésterjedési sebesség (da/dN) függése a frekvenciától (f) $K_{max}$ = 40 MPa.m$^{1/2}$ állandó terhelés hatására a 0.5CrMoV és az új, 2.25CrMo acéloknál

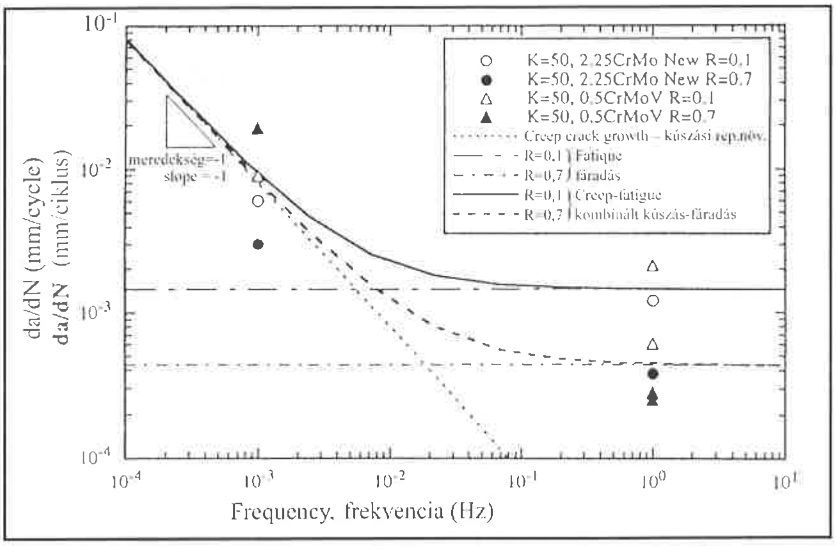


**Fig. 11** Frequency dependence of crack growth rate at $K_{max}$ = 50 MPa.m$^{1/2}$ for 0,5CrMoV steels and new 2,25CrMo steels

**11. ábra.** A repedésterjedési sebesség függése (da/dN) a frekvenciától (f) $K_{max}$ = 50 MPa.m$^{1/2}$ állandó terhelés hatására a 0,5CrMoV és az új, 2,25CrMo acéloknál

intensity factor $K_{max}$ = 40 MPa.m$^{1/2}$, and *Fig. 11* is for $K_{max}$ = 50 MPa.m$^{1/2}$. As shown earlier, *Fig. 9*, if a creep mechanism is dominant, the slope should be –1 because of constant creep crack growth in time, while a crack growth rate should be constant if fatigue crack growth dominates. Changes of crack growth per cycle between $f$ = 1 Hz and $f$ = 0.001 Hz have neither a slope of –1 or 0 in log-log plotting in *Fig. 10*. Furthermore, stress ratio dependence is observed at high frequency, but it may be not always clear at low frequency. Therefore, for the present state transition from creep dominance to fatigue dominant growth is expected to occur between these two frequencies. Predictions based on a linear accumulation rule [5] are compared with experimental results in *Figs. 10* and *11* $C^*$ values to predict creep crack growth have been determined experimentally between 0.001 Hz and 0.1 Hz.

From the above results, linear accumulation of creep and fatigue crack growth can be applied for predicting creep/fatigue crack growth for the low alloy steels, provided account is taken of threshold behaviour at low stress levels. To describe creep-fatigue interaction in creep-to-fatigue transition region, detailed experiments between 0.01 Hz and 0.1 Hz with various stress levels are in progress.

**Table 1:** Chemical Composition of Low Alloy Steels (weight%)
**1. táblázat:** Gyengén ötvözött acélok kémiai összetétele (súly%)

| Materials | 2.25 CrMo | 2.25CrMo | 2.25CrMo | 0,5CrMoV | 0,5CrMoV |
|---|---|---|---|---|---|
| Acélok | New/új | SE thin wall | SE thick wall | New/új | SE |
| C | 0,13 | 0,09 | 0,10 | 0,13 | 0,15 |
| Si | 0,29 | 0,48 | 0,35 | 0,17 | 0,29 |
| Cr | 2,31 | 2,20 | 2,13 | 0,44 | 0,42 |
| Mo | 0,96 | 1,00 | 1,04 | 0,52 | 0,53 |
| Mn | 0,54 | 0,39 | 0,39 | 0,59 | 0,48 |
| V | N/A | N/A | N/A | 0,25 | 0,30 |
| Ni | N/A | 0,12 | 0,12 | N/A | N/A |
| P | 0,026 | 0,01 | 0,01 | 0,027 | 0,02 |
| S | 0,028 | 0,01 | 0,01 | 0,013 | 0,009 |
| N | N/A | N/A | N/A | 0,003 | 0,01 |

*Notes – Megjegyzések:* SE = Service exposed material, thin/ thick wall – üzemszerűen használt anyag, vékony/vastag falú; N/A = Not available – nincs adat

## Conclusions

This paper presents an example of how to transfer fundamental concepts of creep and fatigue to engineering applications by the use of a fracture mechanics modelling technique at the intermediate stress (meso)level. Due to the number of unknowns the modelling at the atomistic level will be unable to solve the problem at the macro level. However the use of meso-fracture mechanics modelling techniques, where assumptions and certain generalisations can be made, could provide an answer to macro engineering problems. It has been shown that creep crack growth can be separated into two regimes of failure consisting of initiation/incubation and a steady cracking behaviour. A creep process zone, described at a grain size level and the modelling concept using the parameter $C^*$ are used to predict the transition and the crack growth stages. The model is used to describe crack initiation and growth behaviour in a range of engineering alloys showing, that the model does not have to perform under very strict conditions. The predicted results show that initial cracking rate $\dot{a}_0$ could be up to approximately 1/5 of the steady state crack growth rate $\dot{a}_s$. This value is consistent with the experimental data for the alloys considered. The period over which the transition takes place in the predictions has also been found to be about 40% of total life and it compares well with the first stage crack initiation and growth times found experimentally.

## Appendix

The formulae used for calculating $K$, reference stress and the $C^*$ parameter

1) Stress intensity factor for CT specimen (ASTM E 1457)

$$K = \frac{P}{\sqrt{BB_n W^{1/2}}} \frac{2 + a/W}{(1 - a/W)^{1/2}} f(a/W)$$

where:

$f(a/W)$ = 0.866+4,64$(a/W)$–13,32$(a/W)^2$+14,72$(a/W)^3$–5,6$(a/W)^4$

2) Reference stress for CT specimen

$$\sigma^{ref} = \frac{P}{mB_{eq}W}$$

where:

$$m = -\left(1 + \gamma \frac{a}{W}\right) + \sqrt{(1+\gamma)\left(\gamma (a/W)^2 + 1\right)} \qquad \gamma = \frac{2}{\sqrt{3}}$$

$$B_{eq} = B - \frac{(B - B_N)^2}{B}$$

$B$ = thickness of the specimen
$B_N$ = the net thickness

3) $C^*$ parameter (ASTM E 1457)

$$C^* \frac{P\dot{V}_c}{B_N(W-a)} \frac{n}{n+1}\left(2 + 0.522 \frac{W-a}{W}\right)$$

where: $\dot{V}_c$ = creep displacement rate
$n$ = creep exponent

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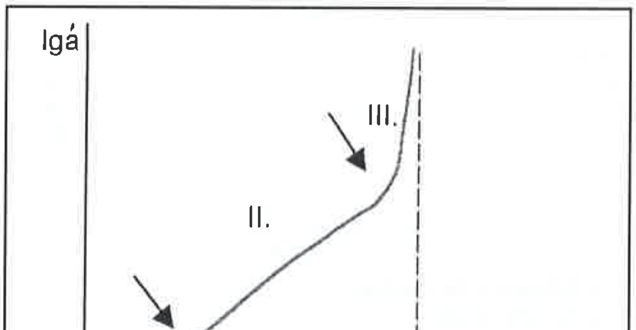
# Hozzászólás a kúszási és fáradási repedésterjedés növelt hőmérsékleten témához

**Lehofer Kornél**

A kísérleti tapasztalatok egybehangzóan azt igazolják, pl. az előző cikk, de az [1] és a [2] is, hogy a szerkezeti elemben (próbatestben) meglévő $a$ méretű repedés akkor terjed, ha a repedés csúcsánál az igénybevételt leíró $P$ törésmechanikai jellemző [vagy annak megváltozásának $(\Delta P)$] értéke nagyobb egy ún. $P_{th}$ küszöbértéknél, és akkor – fáradás estén a ciklusonkénti, kúszás esetén az egységnyi időtartam alatti – repedésnövekedés sebessége a törésmechanikai jellemző függvényében – kettős logaritmikus koordinátarendszerben – az *1. ábra* szerint változik. Az igénybevétel módjától függően a $P$ jellemző lehet, pl. a $K$ feszültségintenzitási tényező (MPa.m$^{1/2}$), vagy a $C^*$ (J/m$^2$.h) nemlineáris kúszási törésmechanikai jellemző.




**1. ábra.** Az $\dot{a}(P)$ függvény lefutása kettős logaritmikus koordinátarendszerben (vázlat)

Kettős logaritmikus koordinátarendszerben az $\dot{a}(P)$ függvény középső, II. szakasza közel lineáris ezért közelíthető az ismert tapasztalati hatványfüggvénnyel:

$$\dot{a} = A \cdot P^n \quad \textbf{(1)}$$

Az $A$ együttható és az $n$ kitevő, amelyek adott feltételek között állandók, nem függetlenek egymástól. Az ln$A$ és az $n$ összetartozó értékei között lineáris korreláció van, amelynek érvényessége – egyesek, pl. [1], állításával ellentétben – nem korlátozódik egy adott ötvözetre. *Tóth László* és társainak kutatásai szerint [3] – alapvető anyagszerkezeti okok miatt – egyes ötvözetcsoportokra azonos korrelációs összefüggés érvényes, mégpedig függetlenül a hőmérséklettől és – fárasztás esetén – az igénybevétel aszimmetria tényezőjétől is. Ilyen ötvözetcsoportoknak bizonyultak a diszperz karbidokkal keményített ferrites acélok, az öntöttvasak, a nemesített Al-ötvözetek és a Ti-ötvözetek. Az $A$-és az $n$ anyagszerkezet-függését, például az [1] tanulmányban közölt szövetszerkezeti és fraktográfiai vizsgálatok is alátámasztják.

Az elmondottakat bizonyítja az is, hogy az említett ferrites acélok csoportjára a fáradás okozta repedésnövekedési adatokból ($\dot{a}$ mm/ciklus, $K$ MPa.m$^{1/2}$) meghatározott korrelációs egyenes [3], az

$$\ln A = -9{,}181 - 3{,}300.n,$$

párhuzamos a kúszás okozta repedésnövekedés [1]-ben közölt adataiból ($\dot{a}$ mm/h, $K$ MPa.m$^{1/2}$) meghatározott:

$$\ln A = -4{,}144 - 3{,}261.n$$

egyenessel. Vagy is, a párhuzamos eltolódás az különböző dimenzióiból ered.

A repedésterjedés (1) összefüggéssel közelített szakaszának hőmérsékletfüggését a folyamat termoaktivált jellege határozza meg.

*Schuchtár Endre* melegalakító (K13 – közepesen ötvözött, ferrites) szerszámacél izotermikus (20, 300 és 500 °C) fárasztóvizsgálata alapján a következő összefüggésre jutott [4]: