MODEL FOR TIME-TO-FRACTURE DETERMINATION OF LOW-ALLOYED STEEL UNDER CREEP CONDITIONS

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ABSTRACT

Structural integrity of power plant component is usually estimate by its residual service life. Input data for component residual life estimation exposed to high-temperature creep is the creep strength of material (time dependent strength), after particular service period. Creep strength is usually obtained by statistical treatment of experimental data, and it decreases during service time due to degradation of microstructure. Since the creep resistance decreases with time due to microstructure degradation, it is essential for practice to use microstructural parameters in order to define statistically the changes in the creep resistance, which in turn enables to follow the physical significance of these changes. Due to a considerable scatter in the creep data, treatment by statistical analysis of the physically significant parameters is a problem that, to date has not been satisfactorily solved. One of attempt to overcome the lack of all data, which were required for previous analyses and estimation of service resources/service life, is done in this paper. Kinetic process has to be defined by the rate of process or by time for which such process will cause the rupture, instead of creep strength. Thus, the time to fracture of the component subjected to a high-temperature creep is determined due the kinetic theory of strength (model Zurkov) and using the microstructural parameters as an input data.

1 INTRODUCTION

According to the literature (Saxsena [1]), a number of methods have been developed for residual life estimation of components exposed to high-temperature creep, from different aspects of view: time of exploitation, creep strength as well as different characteristic of material, but none of them is undertaken to criticizm. Each methodology has its own advantages and disadvantages. Numerous problems are appeared during the attempt of finding physics interpretation of such models. This is not criticising of methods as such, but the main accent is directed to point out lack, which is hard to exclude due to incomplete physical interpretation of the process, that describes parameter's methods resulted by mechanistical creep process approach. Generaly, development of great number of phenomenological models, and models in which different characteristics of material are considered responsible for its strength is a consequence of different approaches in consideration of material strength.

Statistical approach is based on determination of some "critical" value of strength (from the mechanical point of view), which has to be related to the experimental conditions, state and properties of the material. These "critical" values are yield point, creep strength, etc.

Kinetic approach, which is considered in this paper, is based on the fracture process that occurs during the service (as a consequence of damage accumulation process). However, it is not possible to define this kinetic process by some "critical" value. Rather, kinetic process has to be defined by the rate of process or by the time for which such process will cause the fracture (В.Р.Регель at al., [2]). Thus, the time to fracture of the solid body (τ) under the load as a property of the material can be considered as the time for which the material can resist load measured from the moment of loading to the final fracture (Bakic [3]). From the engineering point of view, "time to fracture " has to be equal to the service life of the material, associated with the such state of material which is not reliable for the further service (according to some criterions) rather then associated with final fracture of the material (Бугай et al. [4]).

2 BACKGROUND

According to the kinetic theory of strength, the base for creep failure model is the eqn (1), relating the time-to-fracture with the activation energy as a function of stress and temperature (Регель [2]), is:

$$\tau_f = \tau_0 \exp\left(\frac{U_0 - \gamma\sigma}{kT}\right) \tag{1}$$

where τ_o - constant with the same value for all materials because it depends on the energy of atom vibration; τ_f – time to fracture, U_0 – coefficient expressing the fracture activation energy and is a material characteristic; γ - the constant which depends on the microstructure of the material with dimension of atomic volume whose physical meaning is a local preloading in microstructure (exp. carbides), initiating the fracture; R – universal gas const.; T – absolute temperature; σ - applied stress. The eqn (1) is consisting the terms that actually define strength of the material, i.e. long - term strength.

Experimental results (Регель et al. [2]) show that it is necessary to introduce some correction in eqn (1) if we wont to get a correct result for the real alloys. In order to correct eqn (1) for the creep resistance in the case of low stresses and elevated temperatures, which are usually working conditions for most components on power plants (at the same time that is boundary conditions for the time-to-fracture), it was necessary to introduce the function for pre-exponential constant $\tau_0=\tau_a\varphi(T,\sigma)=10^{-13}\varphi(T,\sigma)$, and in the case of diffusion creep it becomes $\tau_o = A(T^2/\sigma^3)$ (Bakic et al [5]). However, some correction must be done for coefficients U_0 and γ also, due the effect of alloying in the case of multiphase alloys (Регель et al. [2], Крутасова [6], Kumanin [7], Chadek [8]). In that case it is necessary to find out correction coefficients, which have a strong influence on increase or decrease of U_o and γ in eqn (1), due to the changes in microstructure, so eqn (1) get a form:

$$\tau_f = AT^n\sigma^{-m}\left(\frac{U_0 \pm \sum_{i=1}^{n}\Delta U_i - (\gamma\sigma \pm \sum_{i=1}^{n}\Delta\gamma_i\sigma)}{RT}\right) \tag{2}$$

where: ΔU_i and $\Delta\gamma_i$ are coefficients reflecting the change in the fracture activation parameters. In this work an approach to establish the applicable model, based on data from NDT tests in situ for estimating the time-to-fracture based on the kinetic theory of strength for heat resistance Cr-Mo-V steel is described.

3 BASIC MICROSTRUCTURAL PARAMETERS FOR A MODEL

Numerous experimental results prove that the following microstructural parameters, have a strong influence on the time-to- fracture of heat-resistance CrMoV steels:

1. The content of chromium and molybdenum in matrix (Cr+Mo, %). After long term exploitation Cr and Mo atoms move from solid solution to a carbide phase (Bakic [3], Крутасова [6]) and therefore the amount of atoms which remain in solid solution is good indicator of the stage of degradation during time dependent diffusion controlled processes (Крутасова [6], Kumanin [7]). Thus, it is necessary to take into the consideration these influence through the activation energy correction factor (ΔU_1). Carbon is the necessary element for carbide phase formation and the influence of its migration from matrix will be concerned intermediary through Cr and Mo migration. The changes of other alloying elements content have minor effect on matrix strengthening (Bakic [3]).
2. Volume content of carbides phase. The changes in amount and morphology of carbide phase during time is well known (Saxsena [1], Крутасова [6], Cane [9], Pigrova [10]) and present the

basic parameters which could be track down during the process of material degradation in creep condition. It is showed that all carbide phase particles (some of them was in material before exploitation (Baird [11]) while other participate during the diffusion controlled processes) at one moment start to growth or coalescence and then precipitated at grain boundaries probably due to dissolution and migration of alloying elements. This influence on time-to-fracture are considering through the activation energy correction factor (ΔU_2).

The influence of Mo and Cr content in matrix and the size of carbide phase present the measure of changes in interatomic space in material, but there is no precise specified correlation between this two values, and it is necessary to considered their influence apart, although the fact that these values describe the same process and have similar effect on fracture activation energy.

3. Subgrain size. According to the literature (Chadek [8], Cane [9], Goto et al. [12]) cavities type of damage, and also the initial fracture usually occurs at the subgrain boundaries as well as on grain boundaries with participated hard secondary phase. Subgrain boundaries are consisted of dislocation network and present areas with local stress concentration. According to the Kinetic Strength Theory, the area with local stress concentration present the sites with the highest probability for fracture initiation. Local preloading caused by local concentration of applied stresses have to be considered as correction of preload coefficient of nonalloyed matrix γ ($\Delta\gamma_i$). During the long time creep, as a result of recovering process dislocation substructure disappeared, dislocations rearranged and subgrains size growth (Chadek [8]).
4. Interparticle spacing of secondary phase. Changes in dispersion of secondary phase particles is also one of the values for which is confirmed that have role in creep processes ([Saxsena, [1]). Carbide particles growth, their number decrease and as a consequence probability that particles would be a barrier for dislocation motion also decreased, as well as the probability of local stress concentration. The influence of interparticle spacing growth on preload coefficient has the negative sign because the number of stress concentration places decrease, and the preload coefficient have to be diminished by the value of $\Delta\gamma_2$. There has been evidence (Bakic [3], Крутасова [6], Kumanin [7]) that interparticle spacing have the greater influence on dislocation motion than the size of particles, as well as that the particle with greater size could be a good barrier for dislocations.

Apart from noted influence factors there are many other factors that are contributed in strengthening of material and times to fracture (Baird [11]), nevertheless their influence is minor or insufficiently explored in the view of kinetic theory and therefore not included in analysis. As a regression model point of view it is very important to choice the most influenced factors in the purpose of model simplification i.e. diminishing of unknown values number which lead to model in determination lowering.

4 MODEL DEVELOPMENT

Since the correction data for the activation energy ΔU_i and prestressing $\Delta\gamma_i$ are not known, direct application of the influencing parameters is possible by introducing the weighted coefficients for each of these parameters. Also, it was assumed that there is a linear relationship between the corrected values and dependability coefficients for each correcting factor in respect to the actual value of the influencing indicators.

After preparation for regression analyses, eqn (2) is in form:

$$\log \tau_f = \log A + n \log T - m \log \sigma + \log e \frac{U_o}{RT} \pm \log e \frac{\sum_{i=1}^{n} U_i}{RT} - \log e(\gamma \frac{\sigma}{RT} \pm \frac{\sum_{i=1}^{n} \gamma_i \sigma}{RT}) \quad (3)$$

or:

$$\log \tau_f = a + n \log T - m \log \sigma + \frac{b}{T} \pm \frac{\sum_{i=1}^{n} b_i}{T} - (c \frac{\sigma}{T} \pm \frac{\sum_{i=1}^{n} c_i \sigma}{T}) \quad (4)$$

after replacing: a = log A; n = 2 [2,4]; m = 3 [2,4]; b = 0,4343U_o/R – (Uo – fracture activation energy for pure iron is 504KJ/mol [2,7]; R – universal gas costant); b_i = 0,4343ΔU_i/R; (ΔU_i – correction coefficient for activation energy); c = γ/R; (γ - preloading coefficient for unalloyed metal and pure iron is 0,155 KJ/molMPa /4/, and coefficient c is 8K/MPa); c_i = $\Delta\gamma_i$/R; ($\Delta\gamma_i$ - coefficients which have influence on preloading coefficient).

After introducing the weighted coefficients in eqn (4) for each structural parameters (content Cr+Mo in matrix - λ_1; volume content of carbides phase - λ_2; subgrane size - μ_1; interparticle spacing of secondary phase- μ2), eqn (4) get final form:

$$\log \tau_f = a + n \log T - m \log \sigma + \frac{b}{T} \pm (\lambda_1 \frac{b_1}{T} + \lambda_2 \frac{b_2}{T}) - (c \frac{\sigma}{T} \pm (\mu_1 c_1 \frac{\sigma}{T} + \mu_2 c_2 \frac{\sigma}{T})) \quad (5)$$

4 EXPERIMENTAL INVESTIGATION

The heat-resistance beinitic CrMoV steel was investigated in the aim to determine the input data for the model (maximal content: Cr=1,25%, Mo=1,25% and V=0,3%). The X-ray analysis was used to determine the content of Mo and Cr in solid solution while the volume fraction of carbide phase, spacing between secondary phase particles and the subgrain size were determined on the samples removed from the service using light microscope at 1000x and 2000x and scanning electron microscopy. On Figure 1 is shown the typical microstructure of investigated samples after service with heterogeneous dispersion of carbides.

Table 1: Values of microstructural parameters required for the regression analysis.

Sample No	Applied stress, σ	Working temp. t	Cr+Mo in base (Δ)	Vol. fraction of carbide $f_v=(1.17r_{sr}/\lambda)^2$	Subgrain size $\bar{d}_{sub}$	Mean carbides interspacing λ_{sr}	Time to fracture, τ_{fi}
	~ MPa	K	%	%	μm	μm	h
1.	150	848	0.80	25.63	0.60	0.52	9525
2.	150	848	0.68	18.03	0.63	0.62	14570
3.	125	848	0.80	25.63	0.60	0.52	9525
4.	125	848	0.68	18.03	0.63	0.62	14570
5.	125	848	0.80	14.99	0.70	0.68	21850
6.	125	848	0.65	18.62	0.90	0.61	33180
7.	100	848	0.80	25.63	0.60	0.52	9525
8.	100	848	0.80	14.99	0.70	0.68	21850
9.	100	848	0.65	18.62	0.90	0.61	33180
10.	100	848	0.69	13.37	1.05	0.72	48158
11.	100	848	0.58	8.75	1.35	0.89	53000
12.	100	848	0.54	8.56	2.10	0.90	80040

Figure 1: Microstructure of sample 12 (SEM)

These results were then used to determine the values of the structural parameters appearing in the eqn (5) and are shown in Table 2. In Table 2 was also shown the time-to-fracture of each specimen determine by life fraction rule.

5 THE RESULTS AND DISCUSSION

The model, as expressed by eqn (5), can only provide adequate solution by "fixing" at least two regression parameters, that is, by eliminating these as variables from the regression analysis. Based on our calculations, it was found that the most acceptable results are obtained by fixing the coefficients representing the contributions from the volume fraction of carbide phase and subgrain size that are the least time-sensitive parameters during exploitation. If the experimental conditions of any steel under test do not provide a possibility to set two the least sensitive parameters to the changes occurring in the material, that is, to "fix" these, then further solution of the equation gives unreliable predictions (Буртаев et al., [13]). On the bases of the data shown in Table 1 regression coefficients in eqn (5) were determine and are shown in Table 2. The sign of values (λ_1, λ_2, μ_1 and μ_2) correspond to assumed physical interpretation from the point of view of decreasing or increasing fracture activation energy and preload coefficient.

Table 2: Values of regression coefficients obtained by solving eqn (5)

Coefficient	a	λ_1	λ_2	μ_1	μ_2
Value	-25,61	$-1{,}349 * 10^3$	12,786	$-2{,}239 * 10^3$	20,57

Table 3: Time to fracture on the basis of experimental results and according to regression model

Time to fracture	Sample N^o					
	1	2	3	4	5	6
τ_{fi} -experimental	9525	14570	9525	14570	21850	33180
$\tau_{f,}$ - model	6056	17709	11209	20398	20304	19301
Time to fracture	Sample N^o					
	7	8	9	10	11	12
τ_{fi} -experimental	9525	21850	33180	48158	53000	80040
$\tau_{f,}$ - model	26434	26434	25186	31854	76589	89778

The only drawback is associated with the coefficient *a* which is represented as a thermo-fluctuating period of iron atom (it was also accepted that atoms of other elements have the same value) whose value was experimentally determined as 10^{-13} s, that is, $2.77x10^{-16}$ h. Hence, the real value is of *a* is approximately -15.56 (Bakic, [3]). In the model, *a* has a value of -25.61 which is within an order of magnitude close to the value found in the literature. The difference can be indicative of not only the unavoidable errors in the model but also possible changes in the thermo-fluctuating period when dealing with a complex metallurgical system. Time to fracture experimentaly affirmed and obtained by model, eqn (5) and coefficients in Table 2 are presented in the Table 3. Results obtained by model correspond very well to experimental results.

6 CONCLUSIONS

Based on the kinetic theory of strength, a regression model for determining the time-to-fracture of low-carbon low-alloy Cr-Mo-V steels under creep conditions was developed. As the one of the most important and also frequently used material in power industry, low alloyed CrMoV steel was selected, e.g. one for whom exist enough data for different working conditions and times spend in exploitation. In order to apply the described model in engineering practice, following conditions have to be satisfy:

1. Model which truly describe the basic physic of the particular phenomenon, in this creep case, has to be chosen;
2. Determination of the temperatures and stresses range in which the long-time strength is studied. In the particular range the same deformation and fracture mechanism have to be responsible for the behaviour of the material as in the case of the real service conditions;
3. Experimental work has to be planed in the particular temperature and stress range taking into account minimal losses in the material and time and the maximal accuracy of the coefficients for mathematical model.

Verification of kinetic strength theory equation, in the case of complex – multiphase systems is possible to perform only by valid choice of input data in the purpose of proper residual lifetime calculation and comparison of obtained results with verified data from practice.

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