

## DISCUSSION

*L. Remy<sup>1</sup> (written discussion)*—You assume oxidation to form the basis of your high temperature fatigue model. Oxidation is a time-dependent phenomenon; this time dependency has been explicitly described in a recent fatigue crack growth model accounting for oxidation fatigue interactions [1]. I do not see such a time dependency in your model. How do you take into account frequency effects in high temperature fatigue?

*G. R. Romanoski et al. (authors' closure)*—This time dependency is recognized with regard to Eq 4. Time does not appear explicitly in the crack growth expression, because only tests of constant cyclic period were being considered. Introduction of a  $t^{1/2}$  factor to account for the time dependency of diffusion is suggested and is in agreement with the Reuchet-Remy model to which you referred [1] and an earlier one by Antolovich et al. [2]. The implication of a more general formulation which includes time is that the cohesive strength of the grain boundary would be degraded to a greater depth with increasing cyclic period. This could be represented in Fig. 3 of our paper by a series of curves for grain boundary cohesive strength which shift downward and to the right with increasing cyclic period. Sufficient data was not available to be more general at this time.

The formation of a physical oxidation product is not considered to be the necessary precursor to crack extension, but rather the environmental degradation of grain boundary cohesive strength ahead of the crack tip. Oxidation product was observed to form at the specimen surface, along crack flanks, and even at crack tips for cracks that had arrested in grain boundaries which deviated significantly from a favorable orientation. Grain boundary decohesion has been shown to precede the formation of bulk oxidation product [3]. Oxidation is acknowledged to play an important role in the initiation process for alloys and test conditions not giving rise to an intergranular fracture path and may play a role under any conditions where oxide wedging affects the mechanical driving force.

*C. T. Sims<sup>2</sup> (written discussion)*—The two superalloys you studied are similar in several ways: both are cast, they are of similar "type" chemical composition, and they both develop a grain boundary  $\gamma'$  film. Yet, the inelastic strain parameter versus  $N_f$  curve is much closer to the stress parameter versus  $N_f$  curve for one alloy than it is for the other. Can you attribute some physical significance in alloy behavior to this difference? Does it, for instance, relate to crack growth rate?

*G. R. Romanoski et al. (authors' closure)*—The greater separation of the two curves for IN100 is due to a significantly higher response stress for that alloy at any given imposed strain range. Since the scales are the same in Figs. 5 and 6 of our paper, they can be superimposed for consideration of the physical significance of these relative positions. In considering inelastic strain range versus cyclic life, the René 80 data exhibit longer cyclic lives by a factor of two to three when compared with the IN100 data. On the basis of environmental considerations alone, one would expect shorter lives for René 80 due to the higher temperature. However, the response stress for IN100 is greater than that for René 80 by approximately 50% for all values of inelastic strain range. This behavior is consistent with the proposed model. Higher values of the maximum stress result in faster crack growth rates and require a smaller critical crack size to bring about failure, hence, shorter cyclic lives.

If the stress parameter versus cyclic life is compared for these two alloys, the relative positions are inverted. The René 80 data fall to shorter cyclic lives by a factor of two to three. (This is also

<sup>1</sup>Centre des Matériaux, Ecole des Mines de Paris, B.P. 87, 91003 Evry Cedex, France.

<sup>2</sup>General Electric Co., Schenectady, NY.

true if the comparison is made on the basis of stress versus cyclic life.) When stress is recognized as the driving force for crack extension under these conditions, the René 80 data fall into a relative position consistent with an accelerated environmental degradation. A comparison of the two alloys tested at the same temperature would give a better indication of inherent resistance to HTLCF damage by this mechanism. The two governing factors would be grain boundary cohesive strength and response stress.

*B. Ilshner<sup>3</sup> (written discussion)*—This question relates to the depth of decohesion,  $\Delta a$ , introduced in one of your introductory slides. You associate this loss of cohesion to environmental interaction, probably oxygen. Is it your conclusion that this “zone of decohesion”—being a very convincing model by itself—depends on the assumption of environmental interaction? Or would you also agree to discuss a modified version stating that the decohesion zone may well be *extended* by environmental interaction *but* does exist even in vacuum due to acceleration cavitation which is produced by the local stress concentration ahead of the crack tip? See, for example, the work of Riedel. By the way, this effect may also be summarized within the context of environmental interaction, namely, as diffusion of vacancies from the outside vacuum along grain boundaries into the process zone. Finally, if  $a$  is mainly determined by oxygen diffusion, there should be a definite hold time effect,  $\sqrt{D_{gb} t_h}$ .

*G. R. Romanoski et al. (authors' closure)*—In examining specimens of the same materials tested in vacuum, we have also observed evidence of intergranular decohesion [3]. The mechanism of crack extension may be essentially the same, because grain boundaries are intrinsically weaker than the matrix at high temperatures. In vacuum, crack initiation was associated with a loss of material from grain boundaries at the intersection with the specimen surface, particularly for René 80 tested at 1000°C. The model was applied to these data and yielded a significantly lower correlation ( $R = 0.78$ ). This suggests that the number of cycles to crack initiation represents a significant fraction of cyclic life and would not be accounted for by the model in its present form. Testing of these materials at high temperatures in air ensures crack initiation at grain boundaries early in cyclic life, if not on the first cycle. Dramatic evidence for the role of environment in degrading the cohesive strength of grain boundaries is manifested in the significant reductions in HTLCF life observed for specimens tested after pre-exposure at high temperature. A hold time effect is observed in these alloys.

### Discussion References

- [1] Reuchet, J. and Rémy, L., *Metallurgical Transactions A*, Vol. 14A, 1983, pp. 141–149.
- [2] Antolovich, S. D., Liu, S., and Baur, R., *Metallurgical Transactions A*, Vol. 12A, 1981, pp. 473–481.
- [3] Romanoski, G. R., “Mechanisms of Deformation and Fracture in High-Temperature Low-Cycle Fatigue of René 80 and IN100,” M.S. thesis, University of Cincinnati, 1981; also NASA Contractor Report 165498.

<sup>3</sup>Department des Matériaux, Ecole Polytechnique Fédérale de Lausanne, 1007 Lausanne, Switzerland.