12

INTERACTIONS BETWEEN FATIGUE AND DEFORMATION IN STRUCTURAL MATERIALS

A. J. Kennedy

Professor of Materials, The College of Aeronautics
Cranfield, England

INTRODUCTION

Although both creep and fatigue have been subjected to intensive study in recent years, in both metallurgical and in engineering terms, it cannot be said that the degree of understanding gained is a fair return for the effort expended. There is still a great deal of straightforward empiricism about design, whether this be for fatigue or creep conditions, based on formulations which are rarely anything more than convenient expressions of experimental data. Occasionally one or other of such expressions may bear some relation to those which are thought to apply to the fundamental physical processes, but on the whole the problems of prediction, with which the designer is essentially concerned, and those of fundamental explanation, which fall within the province of the metallurgist and physicist, have created quite separate subjects, with very few useful cross-links.

This situation needs to be appreciated if we are to give any adequate assessment to the questions raised by the simultaneous, or consecutive, operation of both processes. The complexity of material behavior in such circumstances is such that the established methods of design, which are already hard-stretched when a realistic creep or fatigue history needs to be handled, simply break down altogether. The position is a classic illustration of the limitations which technology comes up against when it presses on too far beyond the range of basic knowledge.

As far as creep-fatigue interactions are concerned, many of the difficulties seem to be traceable to the fact that there is an inadequate appreciation of the effects that dynamic stresses exert on the nature of the substructure of metals and hence on those processes which are conditioned by this substructure. A so-called "fatigue" stress, initiates, in time, a microcrack: this is propagated under the action of the stress, and failure may result. But a number of other processes may be activated by the stress, some of which may have a strong bearing on the character and the extent of the deformation processes. Indeed, stresses at levels well below the fatigue limit exert pronounced effects on processes such as recovery, work-hardening and creep. In pure metals and in many alloys, provided the stress level is not so high that a finite plastic strain is imposed each cycle, an alternating stress accelerates the processes of recovery. A work-hardened material therefore softens more rapidly than it would do under a simple thermal anneal at the test temperature. For certain alloys where accelerated precipitation may be activated, a positive hardening may result. In any case, for a material in a (thermally) fully softened state, there is no evidence that dynamic stressing can reduce this any further: stressing now results in an increase in hardness, which generally continues until a fatigue fracture occurs. One important form of work-hardening is, of course, that developed by creep. In this case, under conditions in which the rate is controlled by recovery, as it is at about half the melting point absolute, one would expect the rate to be accelerated, provided no other complicating factors, such as precipitation, appear. This is indeed what is generally observed.

In detail, however, the interactions are clearly more complex. As the realization of a supersonic transport aircraft, operating in the region of Mach 2.3, raises very critical questions concerning the creep behavior of the skin materials, if these are to be light alloys, a review of the current position would appear to be timely. This will be attempted in the subsequent section, and in addition, some new results will be presented of experiments on a number of aluminum alloys designed to elucidate the character of some of the interactions which evidently occur.

METALLURGICAL INTERPRETATIONS OF ESTABLISHED INTERACTIONS BETWEEN FATIGUE AND DEFORMATION IN METALS

The effect of a superimposed fatigue stress on the creep rate has been examined by a number of authors, including Greenwood,[1] Lazan,[2] Kennedy,[3] Meleka and Dunn,[4] and Manjoine,[5] while the effects on the tensile behavior have been studied by Nevill and Brotzen,[6] Blaha and Langenecker,[7] and Fransson.[8] The accelerated resoftening of worked metals by an imposed fatigue stress was the subject of experiments by Polakowski and Paechoudhuri[9] who used hardness as an index, and also by Kennedy[3] using subsequent creep behavior as an index of the extent of the recovery. Kaufman and d'Appolonia[10] have studied the effect of tensile and torsional prestrain on the subsequent fatigue behavior of a titanium alloy, and the response to a variety of cyclic conditions of stress and temperature is covered by the papers which make up the ASTM 1954 Symposium on this subject.[11]

If we consider the case of creep under a varying load, it will be clear that if the stress variations are rapid enough, then fatigue may occur. Even if fatigue as such is not developed, the variations of stress may transform the problem significantly. Perhaps the simplest way to express this condition is to say that if the period of the stress cycle is of the same order of magnitude as the relaxation time for the creep process, or less, new factors arise which cannot be deduced from a study of the conventional creep data, however comprehensive this might be. If the period of the stress cycles (or stress changes—whatever form these may take) is long, then there are grounds for believing that the response may be calculated, although this may be laborious if the conditions are complex. We shall illustrate these points by reference to two particular sets of experiments.

In experiments on lead at 35°C, Kennedy showed that the creep rate exhibited under a stress ($\sigma_s \pm \sigma_f$) where σ_s is the static component and σ_f the amplitude of the alternating component was greater than that under a continuously applied *constant* stress of value ($\sigma_s + \sigma_f$). As the mean stress in the case of the superimposed fatigue stress was σ_s, it is obvious that, even allowing for the nonlinear relation between creep rate and stress, the result is not explicable in terms of the conventional data. On the other hand, if relatively slow changes are applied, so that the material may be regarded as being in a state of creep equilibrium with the applied stress, at least for the major part of the time, then a more direct analysis is possible. The work of Dorn and his associates[12] on aluminum illustrates this point very well. This demonstrated that the steady-state creep rate achieved under a given stress is not dependent on previous stress history. A decrease of stress in the steady-state region results in an abrupt reduction, followed by a period of *decreasing* creep rate, and then finally by a gradual increase to the steady-state rate associated with the new stress. The reason that the creep rate depends upon the pattern of dynamic stressing (stepwise stress increments, alternating components, or increments of plastic strain) derives from the fact that in the range of temperatures of technological interest, creep is controlled by the nature of the substructure and by the rate at which this substructure can recover, in the metallurgical sense. During creep, an equilibrium substructure is established (the term *substructure* being used here to include features on the atomic scale, such as dislocation arrays and piled-up groups, and also vacancies and interstitial atoms), further deformation being achieved only by the release of dislocations through the operation of diffusional processes. In the terms of metal physics, creep in this range is predominantly controlled by the process of dislocation climb. We need not pursue these matters in detail in order to appreciate that any operation, mechanical or thermal, which alters the substructure, or the rate of diffusion in the lattice, or both, must affect the creep rate.

The rate of diffusion in a metal lattice, say the rate of self-diffusion, D, of the base element, depends on temperature according to the well-known exponential function $D = D_0 \exp(-Q/RT)$, where D_0 and Q are constants, R the gas constant, and T the absolute temperature. The relevance of creep functions which embody temperature in this form, and particularly the approach adopted by Dorn,[16] will be apparent. But D also depends on the concentration of point defects (vacancies and interstitial atoms), a quantity which can be increased by plastic deformation, or by dynamic stressing, because a certain type of dislocation intersection gives rise to a feature which can advance in the lattice only by generating point defects of one kind or another.

It is now possible to see the kinds of effects that alternating stresses might have on particular materials. If the level of stress is so high that significant work-hardening occurs in each cycle, although the diffusion rate may be significantly increased, its effects may be more than compensated by the increased concentration and interaction of the dislocations themselves. At lower stresses, the dislocation network will be little affected, although the diffusional processes could be positively modified. In the case of alloys which achieve their resistance to creep by some form of continuous atomic migration, whether this be towards dislocations as such, or to other features in the substructure, then (depending on the temperature) the increased rate might well be beneficial. Indeed, it is a fact that many of the creep resistant alloys have greater lives under conditions of superimposed fatigue. A good example of this is provided by the cobalt and nickel alloys at elevated temperatures. On the other hand, if no such mechanism of creep resistance is operating, then the increased rate of diffusion must directly increase the rate of recovery, and hence that of creep itself in the range where it is recovery-controlled.

The converse effect, namely, the influence of plastic deformation on fatigue, is less well established, and open to less fruitful fundamental speculation. This derives mainly from the fact that the controlling mechanisms of fatigue-crack initiation and propagation are not well understood. Although there are now very strong grounds for associating the susceptibility of metals to fatigue cracking with the extent to which a particular dislocation mechanism can occur (the mechanism being that of cross-slip), the development of a realistic treatment of polycrystalline behavior, based on this fundamental model, is fraught with difficulty. Some of the reasons for this are directly associated with the structural complexity of the more useful engineering materials, while others reflect our ignorance of many of the fundamental quantities involved: stacking fault energies for example.

FATIGUE-ACTIVATED CREEP STUDIES ON AN ALUMINUM-MAGNESIUM ALLOY

Experiments have been conducted on thin-walled tubes of an aluminum alloy (3.08 percent Mg, 0.3 Fe, 0.44 Mn, 0.18 Si, 0.06 Cu) under conditions of simple shear. The elements of the experimental assembly are shown in Fig. 1. Two separate tubes are used, these being clamped at A and A', and connected at B and B' to a disc which transmits the torque imposed by the spring loading device C and C'. A simple electrical system maintains the imposed torque constant to within one part in 10^4. The double-specimen is mounted in a thermostatically controlled air enclosure, capable of maintaining temperatures up to about 250°C. The torsional creep strain is detected optically using a light beam L reflected from a small mirror M attached to the torque disc shaft. The reflected spot of light L' is directed onto a photoelectric follower, and the creep recorded as a continuous trace (the actual instrument used is a "Graphispot"). The imposition of a fatigue stress is achieved by using an electromagnetic vibrator. This vibrates the wire connecting the torque disc to the springs, the oscillatory displacement being normal to the line of the wire. No restraint is imposed by this device on the free lateral movement of the wire. The amplitude of the alternating stress can be measured by directly recording the vibration of the specimen using the equipment described above. During creep under a superimposed fatigue stress, the width of the trace enables the fatigue amplitude to be calculated. In these particular experiments the frequency used was 50 cps, and the fatigue stress amplitude only about 1 percent of the static creep stress.

The form of the experimental creep curves is illustrated in Fig. 2, the stress being 7.9×10^3 psi, and the temperatures as indicated. The stress was removed in all cases after 100 min creep, and the form of the creep recovery curve recorded. For clarity in this diagram, only the recovery for the 160°C test is shown.

The creep recovery curve represents a component of creep which should behave in accordance with Boltzmann superposition: there is, at least, good evidence of this for very small strains (see, e.g., Henderson[13]). The proposition can be tested fairly rigorously: unloading after various creep times, and allowing recovery to take place for a prolonged period, enables the residual plastic strain after the creep periods selected to be established. Direct subtraction from the continuous experimental creep curve then gives the anelastic, or recoverable, creep. In addition, if the creep test is continued for a time sufficiently long for the anelastic creep contribution to have become negligible, then the form of the creep recovery curve is simply a reverse replica of that for forward creep. We thus have two methods of establishing the validity of the separation, and, in particular, we

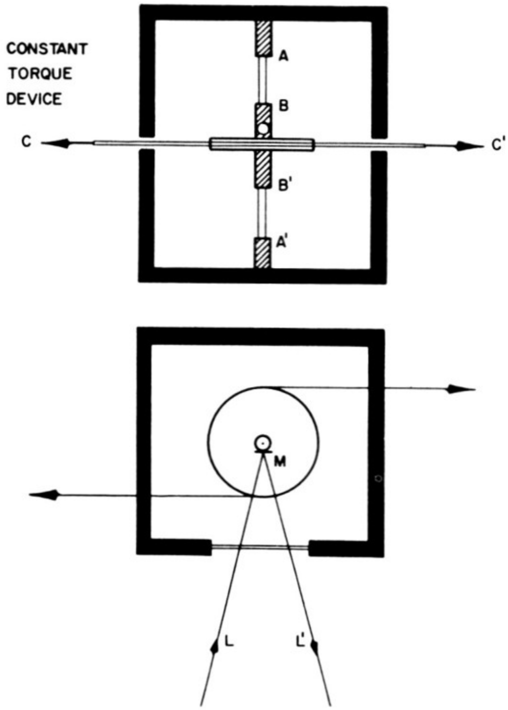

Fig. 1. Diagram of the torsional creep testing assembly.

may check whether the duration of the creep test (100 min in the present series) is sufficiently long for the creep recovery curve to be accepted as being representative of the anelastic forward creep component. This check was made and the curve deduced from a series of unloading experiments, after different previous creep times, was found to be in very close agreement with that for creep recovery after 100 min.

We are now in a position to separate, in an unambiguous way, the recoverable and nonrecoverable (plastic) components. Taking the curve for 160°C again as an example, we find that the creep strain separates into the two (additive) curves shown, one of which is closely logarithmic in form ($\epsilon \sim k \log t$, where ϵ is strain, t the time and k a constant) and the other linear over the greater part of the creep time—from at least 20 to 100 min for the 160°C case. This plastic component has a small transient stage, and the best fit in this range is obtained by using a power function of the time with a small exponent (say 0.1 to 0.2). We shall not pursue this question here as it lies outside the scope of the present work.

The closeness of fit of the anelastic component to a logarithmic function is demonstrated by Fig. 3. According to the standard physical treatment of logarithmic creep (Mott and Nabarro,[14] and Wyatt,[15] for example), temperature appears as a simple factor in k (Eq. 1), that is, $k = k'T$. Figure 4 shows how well this is obeyed in the present experiments: agreement with the theory is evidently good. We might note that, in this same temperature range, the rate of plastic

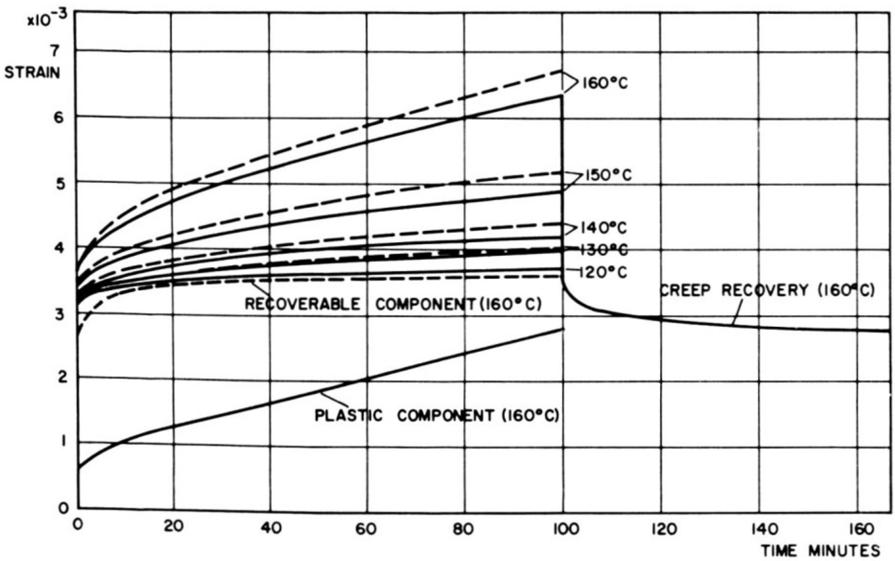

Fig. 2. Torsional creep curves, and the form of the creep recovery for the test at 160°C. The result of analyzing the creep at this temperature into a recoverable and a plastic component is demonstrated by the curves indicated. The dashed lines show the effect of superimposing a fatigue stress with an amplitude about 1 percent of that of the static stress which was 7.9×10^3 psi in all cases.

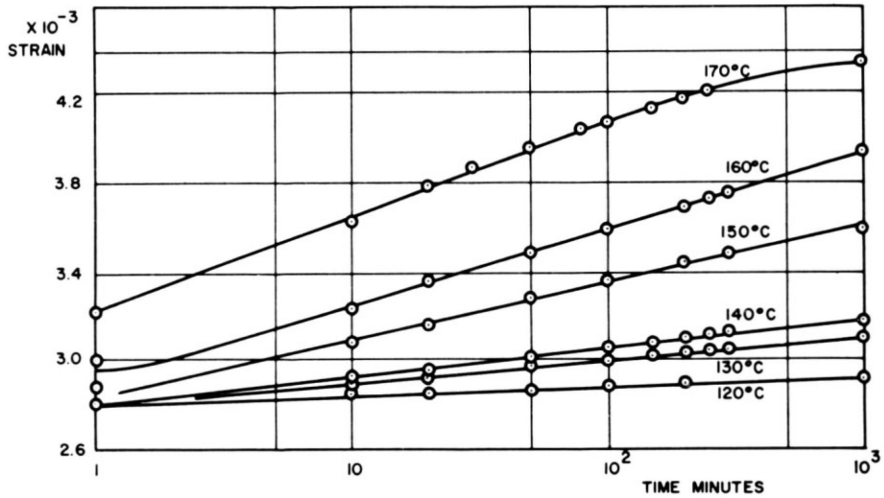

Fig. 3. Demonstrating the logarithmic form of the recoverable creep component for the temperatures investigated.

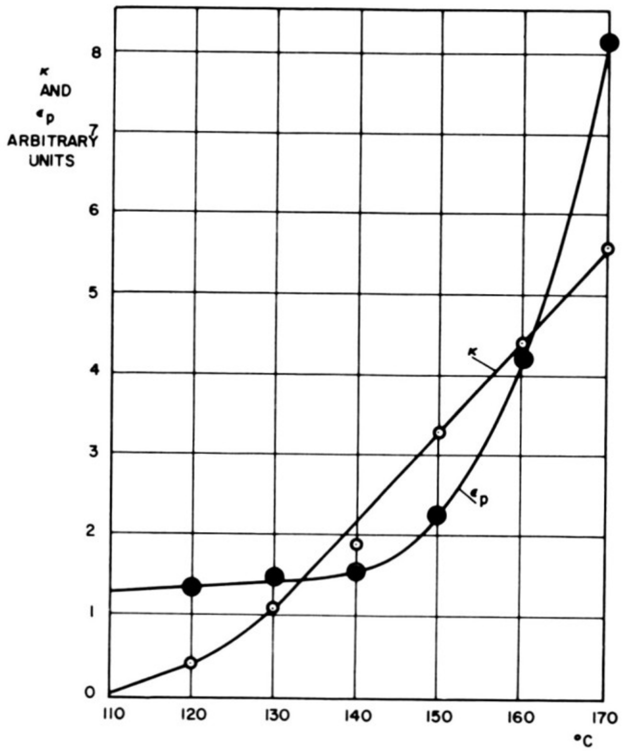

Fig. 4. The dependence of k, the logarithmic coefficient, and ϵ_p the residual plastic strain on test temperature.

creep increases by a factor of $\sim$100, compared with the tenfold increase in the coefficient of logarithmic creep. The temperature-dependence of the plastic creep component is illustrated by Fig. 5, in which the creep rate (deduced from the linear rate of creep—see Fig. 2) has been plotted against the reciprocal of the absolute temperature for all the tests conducted under a static stress of 7.9×10^3 psi. For a simple thermally-activated process, for which the rate is given by $\dot{\epsilon} = af[t \exp(-Q/RT)]$ (an approach which is particularly associated with the work of Dorn,[16] a linear relationship between log $\dot{\epsilon}$ and T^{-1} should be obtained, the slope being $-Q/R$. As Fig. 4 demonstrates, in the temperature range covered by these experiments, the slope changes markedly, and quite critically between 150 and 160°C. Above 160°C, the slope yields an activation energy of 63.5 kcal/mole, while below 145°C, the energy deduced from the data is 15.3 kcal/mole. These values are markedly different from those deduced by Dorn on a similar material. The *mean* slope derived from the data is associated with a Q-value of 28.5 kcal/mole, and this is, in fact, within the range of values which Dorn obtained.

The question of the derivation of activation energies is a complex one. It follows from the expression $\dot{\epsilon} = af[t \exp(-Q/RT)]$ that unless the form of the function f is established, Q cannot be derived from data of the kind represented

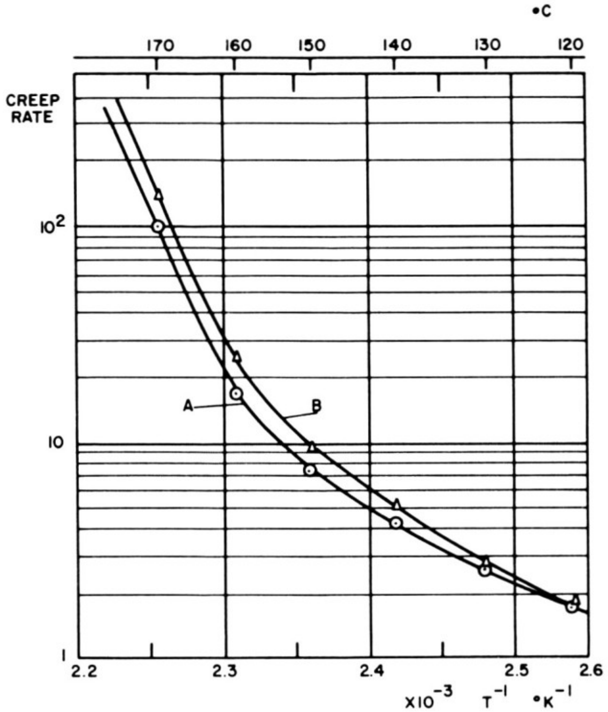

Fig. 5. The dependence of the steady-state plastic creep rate on temperature. The curve A represents results at constant stress, while curve B represents the results at the same static stress, but with a 1 percent fatigue stress superimposed.

by Fig. 2. In the present case we have assumed that linear creep only is under consideration, and also that the constant a is not temperature-dependent. Most of the theories relating to diffusion-controlled mechanisms lead to expressions which involve terms of the general form sinh $[af(\sigma)/kT]$ where $f(\sigma)$ is some function of the stress, usually of the form $b\sigma^m$, where m is approximately 2 or 3. In the present instance no more detailed analysis need be attempted, as the relation between $\dot{\epsilon}$ and T^{-1} may be taken as a reference against which the data for fatigue-accelerated creep may be compared. However, it might be noted that no adjustment of the curve, based on the introduction of a sinh $[af(\sigma)/kT]$ term into the analysis, will yield a straight line, unless a slope of less than the minimum is accepted. As a consequence, any recalculation on this basis will lead to a value for the activation energy less than 15.3 kcal/mole.

The higher activation energy derived from Fig. 5 (63.5 kcal/mole) is, in fact, almost exactly the energy for recrystallization in aluminum, and has been associated with "recovery" by Dorn. Certainly this is a much higher value than the energy for self-diffusion ($\sim$ 33 kcal/mole), which one would expect to approximate to that of recovery through thermal climb.

We shall now report briefly on some work concerned with the effect of creep deformation on fatigue properties, and finally discuss the results as a whole.

THE EFFECT OF CREEP DEFORMATION ON FATIGUE LIFE

The work of Kaufman and d'Appolonia[10] on a titanium alloy (RC-55) demonstrated the striking increase in fatigue strength that could be obtained by previous cold work. In their experiments, fatigue was imposed under rotating-bend conditions. Both tensile and torsional prestrain improved the fatigue life, the torsional strain (30 percent) raising the fatigue limit from 47.5×10^3 psi to 63.5×10^3 psi, while the tensile strain increased the limit to only 55×10^3 psi. In the case of notched specimens, either type of pre-strain actually reduced the fatigue strength, which is somewhat at variance with other results on plastically deformed notched test pieces. If the primary crack-forming mechanism in fatigue is indeed dependent on the mobility of dislocations (and screw dislocations in particular) then it might not be unreasonable to expect such an effect, as the interlocking of dislocations, and their increased concentration brought about by plastic deformation would produce this effect. We exclude, for the moment, that class of material in which plastic deformation may actually release dislocations from a locked state (i.e., strain-aging materials).

In fact, no general rule can be drawn up. Experiments have been conducted at The College of Aeronautics by Loriston-Clarke[17] on two aluminum alloys [Al 4Cu (L65), and Al 2.5Cu 1.5Mg 1.2Ni 1.0Fe (RR58)], the creep deformation being imposed by torsion, subsequent fatigue stressing being carried out under rotating-bend conditions. The full details of the work cannot be entered into here, but the results set out in Fig. 6 summarize the essential findings. Under high-stress fatigue conditions (56×10^3 psi for L65 and 45×10^3 psi for RR58), both alloys show an average life of 8,000–10,000 cycles. Deformation applied before the fatigue test, however, has a striking effect, increasing the life of RR58 by a factor of nearly 2 (taking averages), while the life of L65 is reduced, also by a factor approaching 2. From the results it will be evident that the conditions under which the deformation is imposed appear to be of secondary importance, as both slow creep and rapid deformation (both imposed at 150°C) have the same effect on the subsequent room temperature fatigue life. Nor does annealing restore the material to its original state.

Both these alloys are, of course, precipitation hardening, and there would be a finite rate of precipitation in both at 150°C. However, if the precipitate dispersion is in some way modified during creep, as one would expect, then in the one case this is certainly beneficial, at least as far as fatigue is concerned.

GENERAL CONCLUSIONS

Interactions of the type discussed above could produce effects which would need to be taken into account in design. The experiments described will have demonstrated that some idea of the possible importance of these effects can be gained without extending the experiments to the prolonged loading times encountered in practice. Such measurements on the response of materials to combined creep and fatigue conditions are valuable in establishing the significance of certain metallurgical changes which occur and also in designing more realistic long-term programs. The degree to which creep may be accelerated by

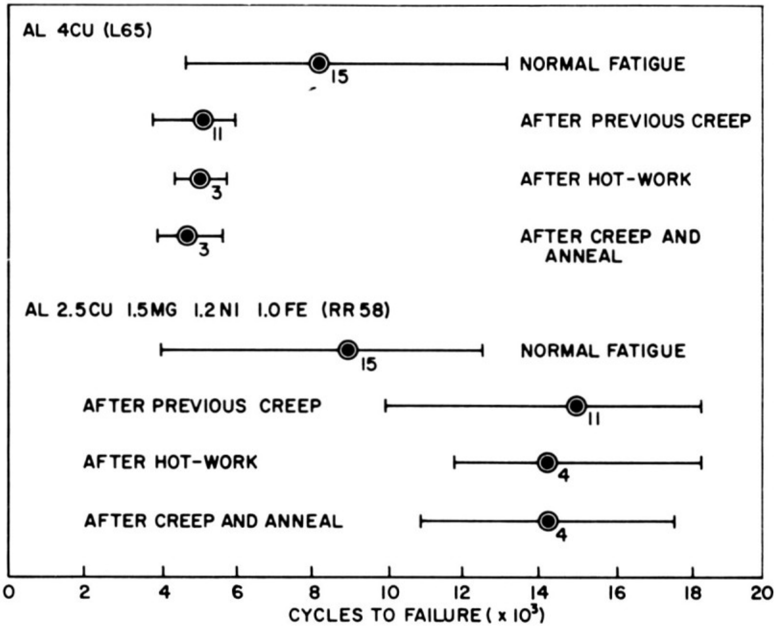

Fig. 6. The effect of various thermal and mechanical treatments on the fatigue lives of specimens of two aluminum alloys.

dynamic stresses, and the fact that plastic deformation may be either beneficial or deleterious to the fatigue strength, are clearly of primary importance in materials selection for supersonic transport aircraft. Some later results obtained with the test equipment described in this paper will be reported at the meeting.

REFERENCES

1. Greenwood, J. N., *Proc. Amer. Soc. Test. Mat.*, 49, 834, 1949.
2. Lazan, B. J., *Proc. Amer. Soc. Test. Mat.*, 49, 757, 1949.
3. Kennedy, A. J., International Conference on the Fatigue of Metals, *Inst. Mech. Engrs.*, 1956.
4. Meleka, A. H., and G. B. Dunn, *J. Inst. Metals*, 88, 407, 1960.
5. Manjoine, M. J., *Proc. Amer. Soc. Test. Mat.*, 49, 788, 1949.
6. Nevill, G. E., Jr., and F. R. Brotzen, *Proc. Amer. Soc. Test. Mat.*, 57, 751, 1957.
7. Blaha, F., and B. Langenecker, *Naturwissenschaften*, 42, 556, 1955.
8. Fransson, A., *Proceedings of a Colloquium on Fatigue*, Stockholm, Springer, 1956.
9. Polakowski, N. H., and A. Palchoudhuri, *Proc. Amer. Soc. Test. Mat.*, 54, 701, 1954.
10. Kaufman, J. G., and E. D'Appolonia, *Proc. Amer. Soc. Test. Mat.*, 55, 999, 1955.
11. Symposium on Effect of Cyclic Heating and Stressing on Metals at Elevated Temperatures, *ASTM Special Publication No. 165*, 1954.
12. Raymond, L., W. D. Ludemann, N. Jaffe, and J. E. Dorn, *WADD Technical Report 60–326*, 1960.
13. Henderson, C., *Proc. Roy. Soc. A.*, 206, 72, 1951.
14. Mott, N. F., and F. R. N. Nabarro, Bristol Report on the Strength of Solids, Physical Society, London, 1945.
15. Wyatt, O. H., *Proc. Phys. Soc., B*, 66, 459, 1953.
16. Dorn, J. E., "Creep and Recovery," *ASM*, 1957.
17. Loriston-Clarke, A. G., Diploma Thesis, College of Aeronautics, Cranfield, 1961.