

Pa., who conducted the five-ball fatigue tests under contract NAS 3-11617 and the personnel from General Electric Co., Cincinnati, Ohio, who conducted the 120-mm bore bearing fatigue tests under contract NAS 3-11148.

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DISCUSSION

L. Leonard²

The authors have very appropriately focused upon the need for truly comparative rolling contact fatigue data for a range of candidate materials. In line with this they have presented information pertinent to the selection of appropriate materials for specific rolling contact applications.

In reviewing the authors' discussions and conclusions I feel there are some metallurgical aspects which merit further consideration at this time. In standardizing their testing program,

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the authors chose the parameters of retained austenite, grain size and hardness as the prime metallurgical controls for their test materials; comprehensive metallurgical characterizations apparently were not performed. Accordingly each material was tested in one specific heat treat condition which may not correspond to the optimum metallurgical state for maximum rolling contact fatigue resistance. For example, the factors of composition; grain size and shape; grain boundary segregation (elements, carbides, etc.); uniformity and fineness of microstructure in general; morphology and distribution of phases, particularly carbides; substructure; stability of microstructure under cyclic loading; ductility; fracture toughness; and residual stresses all must play interacting roles in defining a metal's behavior. Hardness, which reflects the combined effects of many metallurgical parameters does not uniquely represent any single metallurgical state. Therefore it is not surprising that the authors' data show that a hardness specification cannot be arbitrarily selected as a criterion by which to judge rolling contact performance. This points out the need for *more complete characterization* of metals in studies such as that presented here in order to provide a more comprehensive interpretation of the test data.

The reduction of hardness with temperature may not be a good criterion by which to critically judge the elevated temperature rolling contact capabilities of the materials tested. As long as hardness does not drop to such an extent that brinelling rules out that material condition, enhanced ductility or toughness may more than offset the reduced strength. This could result in significant differences in the relative behavior of the metals tested. Furthermore, some other metallurgical factor, such as primary carbide size and distribution, may dominate in controlling the rolling contact behavior at all temperatures and thus overrule a hardness stability criterion.

In brief, since fatigue failure involves both crack initiation and propagation, parameters such as ductility, fracture toughness, microyield strength (or ease of dislocation motion), and microstructural stability should all be more critical parameters than hardness in defining both the alloys most likely to exhibit good rolling contact fatigue behavior and the proper heat treatments to be employed.

The authors' correlation of fatigue life with alloy content can be misleading, since the alloy content itself influences so many other factors, which can play a role in any given steel in determining fatigue life. If the authors had included a 1080 plain carbon steel at 60 Rc, the alloy content would have been essentially nil, yet life would probably have been much lower than that of the AISI 52100 alloy steel. While analysis of carbide composition, size and morphology may help explain the alloy effect, matrix properties are also undoubtedly influenced by composition, and they in turn will influence fatigue behavior.

Another aspect on which I would like to comment is the authors' observation that all failures appeared to be "classical subsurface" in origin. In recent scanning electron microscopy, SEM, work³ on deep-grooved ball bearing inner races the discussor was able to show that spalls can grow both in and opposite to the rolling direction and thus remove all evidence of surface initiated failure. Furthermore in analyzing a series of tests on the same steel, all run under identical loading but different lubrication conditions, a very wide range in L_{10} lives was measured. Among the tests the variation in the subsurface stresses should have been small compared to that in the surface stresses. Thus surface initiated failures most likely accounted for the L_{10} differences. However, careful SEM study could reveal no differences between the early spalls of a low L_{10} test group compared to spalls formed late in a high L_{10} test group. Also, the SEM study revealed that significant surface morphology differences are not detectable by customary surface evaluation techniques and that significant surface pitting or microspalling occurred under all lubrication

³ Leonard, L., Cotellesse, G., and Erhardt, K., Final Report on Scanning Electron Microscopy Study of 6309 Size Bearings, SKF Industries, Inc., Contract N00014-70-C-0229, AD 717959, Jan. 1971.

conditions. Accordingly, it is entirely possible that the L_{10} lives reported by the authors may be due in part to surface finishing effects, including both morphological and residual stress. Indeed, as pointed out by the authors, the high alloy steels are more difficult to grind and finish than AISI 52100. It would be most informative therefore, to study representative unrun balls (if still available) to determine if surface effects contributed to the authors' data.

R. E. Maurer

It is pointed out by the authors that there is a correlation of the lowest fatigue lives with the materials containing the greatest weight percentage of carbide forming alloying elements, and that the size and distribution of the carbides, which vary with alloy content, appear to be primary factors in rolling contact fatigue life. It is believed by this discussor, that rolling contact fatigue life is affected by carbide shape, size, distribution, and location, not only of the primary carbides to which the authors refer, but also carbides which precipitate during heat treatment.

It is a universally accepted fact that the mechanical properties of steel can be drastically reduced by the presence of grain boundary carbides. Generally, the greater the percentage of carbide forming elements such as molybdenum, chromium, tungsten, and vanadium, the more care must be exercised in the heat treatment to minimize deleterious carbide precipitation. In addition, overheating of a steel containing these elements in forging, heading, or heat treating may result in "burning" or localizing melting of the steel. This occurrence results in the formation of skeletal or angular shaped primary carbides in the grain corners and boundaries and has an embrittling effect on the material. Microporosity is often associated with burning and will act as a potent stress raiser and cause premature failure in rolling contact.

Figs. 3 and 4 of the authors' reference [15] (Rolling-Element Fatigue Lives of Four M-Series Steels and AISI 52100 at 150°F) are photomicrographs of hardened M-1 and M-10, respectively, representative of the microstructures of the ball specimens which were tested. These photomicrographs show indications of overheating and concentration of the primary carbides in the grain boundaries, particularly in the M-10 steel. It also appears that there was rather heavy carbide precipitation in the grain boundaries during heat treatment (as evidenced by the darkly etched grain boundaries). A very severe condition of grain boundary carbide precipitation is shown in Fig. 4 of the authors' reference [13] (Fatigue Lives at 600°F of 120-Millimeter-Bore Ball Bearings of AISI M-50, AISI M-1, and WB-49 Steels) for the hardened WB-49 bearing rings. This condition is a result of air cooling from the 2200 deg F austenitizing temperature which is indicated as the quench cycle in Table II of reference [13]. An intermediate salt quench such as that used for M-50 is generally used for both M-1 and WB-49 to minimize grain boundary carbide precipitation. Both overheating and heavy grain boundary carbide precipitation have been observed as causes for premature failure in rolling contact.

Endurance tests performed at the discussor's laboratory at 600 deg F with bearings having rings made of M-1 and WB-49 steels, which were manufactured and heat treated to avoid burning and excessive grain boundary carbide precipitation, yielded lives for these materials which are comparable to that of M-50 steel. Therefore it is quite possible that microstructure played a significant role in determining the endurance characteristics of the materials tested in the authors' investigation.

Authors' Closure

The authors would like to thank the discussors for their comments. It must be pointed out that the heat treatment given to each of the materials tested in the five-ball tester was generally the standard heat treatment for that given material which was provided by the ball supplier. In other words, unless one specifies

a special heat treatment, a bearing manufacturer would generally use these heat treatments for bearings manufactured from each material. The authors purposely chose no special heat treatments in order to achieve a fair comparison of the materials in a condition that could be readily duplicated by a bearing manufacturer or user. The determination of whether or not the heat treatment used was optimum was not within the scope of this program.

The factors that Dr. Leonard lists may indeed interact to determine a materials rolling-element fatigue characteristics. However, most, if not all, of these factors are determined by the material alloy composition, heat treatment, and specimen size. All of these were known quantities in this program. Slight variations in these factors may well account for the differences in fatigue life for the different lots of each material.

It has been established [7, 11]⁴ that for a given rolling-element steel, the fatigue life increased with increased hardness. Further research [20] indicated that life was a function of the difference in hardness between two rolling elements wherein the rolling-element receiving the greater number of stress cycles should be one to two points softer than the mating element. Further research indicated that the cause for the increased life was due to compressive residual stress induced during running [21]. Work performed on the changes in microstructure with rolling-element stressing revealed no interrelation between changes in subsurface microstructure with cycling load and fatigue life.

In the instant investigation, the authors methodically controlled all the known variables mentioned in the foregoing and by Dr. Leonard as closely as the state-of-the-art would allow. The only factors which were not controlled and which are both functions of material chemistry and heat treatment were carbide size and distribution. The results of this investigation showed a correlation between percent alloying element present and rolling-element fatigue. As discussed by the authors in the paper, this would imply an interrelation between carbide size and distribution and fatigue life. Further investigation on the interrelation of carbide size and distribution is warranted.

Dr. Leonard's example of the 1080 carbon steel requires an extrapolation of at least an order of magnitude beyond the range of the present data. Such an extrapolation can, of course, be misleading. The authors are grateful to Dr. Leonard for raising this point, and take this opportunity to state that the relation between life and percent alloying elements should be considered only within the range of the existing data.

It is the nature of the failure detection and shut down system on the five-ball fatigue testers to detect a fatigue spall early in its formation and shut down the tester before significant growth has occurred. The authors' experience indicates that in full scale bearing tests, evidence of surface initiated failure remains, if the bearing is removed from testing early enough. The authors believe that from the appearance of the spalls (size, as well as depth) the failures were subsurface in origin.

The authors do not dispute Mr. Maurer's contention that the mechanical properties of steels are affected by the presence of grain boundary carbides. However, the authors are not aware of any published data that show a reduction in rolling-element fatigue life related to grain boundary carbide precipitation. Further, a more detailed examination with optical microscopy of the test materials in Table 2 has shown no evidence of overheating and heavy carbide precipitation in the grain boundaries. To detect the presence of grain boundary precipitation, transmission electron microscopy is preferred. Such an examination was not performed. It is improbable that the presence of these microstructural defects was significant, or had a significant effect on the fatigue test results.

The evidence of the grain boundary carbide effect offered by Mr. Maurer is not relevant. It is assumed that Mr. Maurer refers

⁴ Numbers in brackets designate references at end of paper and Additional References at end of authors' closure.

to the 25-mm bore ball bearing tests of NASA contracts NASw-492 and NAS3-7912 reported in [19, 22 to 24] in his final comment. The lives of the M-1, M-50, and WB-49 bearings at 600 deg F were not comparable in these tests, contrary to Mr. Maurer's contention. These bearings were run to only twice AFBMA (catalog) life where the M-1 bearings showed no failures and the few M-50 and WB-49 bearings that failed showed surface distress from marginal lubrication conditions (glazing and smear-ing). The failures were not rolling-element fatigue type failures. Fig. 10 shows that the 120-mm bore bearings of M-1 and M-50 (tested with the same lubricant as the 25-mm bearing tests) had 10 percent lives exceeding AFBMA life by factors of 6 and 13, respectively. These 120-mm bore bearing failures were classical subsurface initiated rolling-element fatigue failures [13] as were the WB-49 failures which bearings gave lives less than AFBMA life. Had the 25-mm bore bearings been run at more favorable lubrication conditions where fatigue is the criterion of failure, lives much greater than twice AFBMA life would be expected. These tests with the 25-mm bore bearings where lubrication distress was the predominant mode of failure cannot be used to

draw conclusions relating to the effect of microstructure (grain boundary carbide precipitation) on rolling-element fatigue.

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