Fracture and Fatigue Crack Propagation in Bearing Steels

B. L. Averbach

Professor of Materials Science
Massachusetts Institute of Technology
Cambridge, Massachusetts 02139 U.S.A.

ABSTRACT

The critical stress intensity factors of the steels used on the mainshafts of aircraft gas turbines are shown to be in the range 17-22 $MPa \cdot m^{1/2}$ (15-20 $ksi \cdot in^{1/2}$). The rates of fatigue propagation are quite high and do not exhibit a threshold value. These data suggest that bearings should be limited to DN = 2×10^6, where D is the diameter of the bore in mm and N is the number of revolutions per minute, in order to avoid tension fractures arising from the hoop stresses.

Introduction

The mainshafts of aircraft gas turbines are supported by ball and roller bearings which operate at high speeds. The Hertz stresses are not considered to be high, with both radial and thrust loads involved in normal operation, but tensile hoop stresses are present, and bending and vibrational forces may also be added on occasion. The bearing technology has evolved along with engine developments, and the present day bearings are larger and lighter and rotate faster than the ones used in the early production engines made in the 1950s. The composition of the steels, and the methods of making these steels have also changed, and techniques of non-destructive testing have improved to the point where failures in mainshaft bearings are rare.

As speeds and sizes have increased and section sizes decreased, however, the hoop stresses on the rings have increased to the point where we have approached a limitation in bearing design. At high shaft speeds, hoop stresses can introduce enough tension in the rings, so that a flaw, which starts as a small fatigue crack, can reach a critical size and propagate into a catastrophic failure of the bearing. A convenient parameter for bearing speed is expressed in the product, DN, where D is the diameter of the bore in mm and N in the number of revolutions per minute. The hoop stresses are proportional to $(DN)^2$. The new engines operate at about 2 million DN, with the most advanced engine at about 2.3 million DN. Beyond that there are questions of reliability, and in this paper we consider the metallurgical factors which limit the fatigue and fracture properties of the current bearing steels.

In our investigation we used compact tension specimens, 2.5 x 2.5 x 0.25 in., of the type described in ASTM E399, and shown in Figure 1. A fatigue crack was started at the notch by repeated tension-unload cycles. After the crack had progressed beyond the initial region of plastic flow, the rate of fatigue crack propagation, dC/dN, was measured. C is the total crack length, measured from the tension axis, and N is the number of cycles. After sufficient data on fatigue crack growth rates had been accumulated, the specimens were pulled to fracture and the critical stress intensity factors, K_{Ic}, were determined. All of our data met the criteria set by ASTM for valid plane strain fractures.

Bearing Steels

The three bearing steels now in use were investigated, and the compositions are given in Table 1. Type 52100 is the traditional bearing steel and is still in use in a few of the oldest engines. It is also of interest in non-aircraft turbines where service conditions are not as severe. The material which we tested came from an electric furnace heat and was vacuum-degassed. The inclusion count was very low and met aircraft bearing specifications requirements. M-50 is a low-alloy high speed steel and is used in all of the aircraft gas turbines now produced in the U.S. This steel was vacuum-induction melted and vacuum-arc remelted (VIM-VAR), in accordance with the best American practice. 18-4-1 is the traditional high speed steel and is now used in all of the high performance British engines. It was made in England by the electroslag process in accordance with their preferred practice. Both high speed steels exhibited very few inclusions, and the carbides were well-spheroidized and uniformly distributed.

Fracture Toughness

Our fracture toughness data on 52100 are summarized in Figure 2, where we show how the fracture toughness varies with heat treatment. Two effects are apparent. The fracture toughness is improved by austenitizing at 815°C (1500°F) in comparison with the use of higher austenitizing temperatures. At a given austenitizing temperature, there is a decrease in fracture toughness on tempering at 230°C (450°F) in comparison with the maximum values observed on tempering at 170°C (340°F). The tempering effect at 230°C is probably associated with the almost complete transformation of the retained austenite at this temperature. In aircraft practice, however, it is necessary to eliminate the retained austenite in order to achieve dimensional stability, and for this reason, these bearings are usually tempered at 230°C.

The influence of heat treatment can also be seen in Figure 3 where we show the fracture toughness as a function of hardness. At a given hardness, such as 62 R_C, it is possible to achieve about 10% more fracture toughness by using the low austenitizing temperature. An explanation of this effect may be in the carbon contents of the martensites. Table 2 shows the volume percentage of undissolved carbides and the corresponding carbon contents in solution for each austenitizing temperature. The carbon content at the lowest austenitizing temperature, 0.49% C. is substantially lower than at better fracture toughness of the former to the better ductility of the lower carbon martensite. Examinations of the fracture surfaces show that the fracture proceeds through the martensite and around the carbides. The fracture toughness is thus influenced primarily by the nature of the martensite and it appears that the toughness of low carbon martensite is superior.

The fracture toughness of the high speed steels is shown in Figure 4 as a function of tempering temperature. The M-50 specimens were austenitized in an atmosphere furnace at 1095°C (2000°F) and tempered 3 times for two hours at each of the temperatures indicated. The usual tempering temperature in practice is 540°C (1000°F). The 18-4-1 specimens were austenitized at 1250°C (2280°F) and triple tempered for two hours at each temperature. The normal practice calls for tempering at 560°C (1040°F). The open points in Figure 4 refer to specimens which were refrigerated at -80°C (-110°F) for one hour between the first and second tempers. The retained austenite contents of the refrigerated and the corresponding non-refrigerated samples were well below one percent. It appears that refrigeration was not a factor in either the retained austenite contents or the fracture toughness under these heat treating conditions.

If we compare the fracture toughness values of the high speed steels at the tempering temperatures usually used (the hardness values are the same, at about 62 R_C) it is evident that the fracture toughness of 18-4-1 is about 25% higher than that of M-50. This seems unusual, since the volume of carbides in 18-4-1 is much greater than in M-50, and here also, the explanation may be in the composition of the matrix. The various types of carbides were identified in a scanning electron microscope and the volume fractions determined. The compositions of the martensites were calculated, and these are shown in Table 3. It is interesting to note that the 18-4-1 steel has a much lower carbon in solution (0.42%) than the M-50 (0.53%). Furthermore, if we compare the toughness of all three bearing steels as a function of hardness, Figure 5, it is evident that the toughness shows the same rough correlation with the carbon content of the martensite. The lower the dissolved carbon, the higher the toughness.

There is another interesting feature in the alloy contents of the two high speed steel martensites. Both steels have about the same chromium and vanadium contents. The 18-4-1 martensite contains 5.0 W, and if we take the equivalent molybdenum value as one-half of the tungsten content, it is evident that both steels have very similar molybdenum/tungsten equivalents. Thus, both high speed steel martensites have equivalent alloy contents, and any differences in fracture toughness can be ascribed to the difference in carbon content.

Fatigue Crack Propagation

The rates of fatigue crack propagation are summarized in Figure 6 for the three bearing steels. These values refer to tension-release fatigue, and at low values of the applied stress intensity, ΔK, the crack propagation follows the law,

$$\frac{dC}{dN} = A(\Delta K)^m \tag{1}$$

where A is a constant, and m is the growth rate exponent. At higher values of ΔK, the crack growth rate is very rapid since it is approaching the critical stress intensity for plane strain fracture, K_{Ic}.

The rates of fatigue crack propagation are similar for the three steels. The growth rate exponent is three, and this is similar to that observed for many high strength steels. Within the band, 18-4-1 appears to have somewhat lower fatigue crack growth rates than M-50, but these differences are probably not significant.

Critical Crack Sizes

A critical crack size, a_c, is defined as one which will lead to rapid fracture at the given operating stress. A crack which is smaller than a_c will not lead to immediate fracture. The critical crack size may be calculated if the critical stress intensity factor, K_{Ic}, and the operating stress are known. The hoop stresses developed in a typical 150 mm bore bearing have been calculated by John Clark[1] and these are listed for various values of DN in Table 4. Using these hoop stresses, our measured values of K_{Ic} and the expression for an approximate value of critical crack size,

$$a_c = \frac{K_{Ic}^2}{\pi\sigma^2} \tag{2}$$

where a_c is the critical crack size and σ is the hoop stress, we have calculated the critical crack sizes for each of the steels at DN values from 2 to 4 million, and these are listed in the table.

These calculations show that the critical crack sizes for the three steels are similar and that the small differences in fracture toughness between these steels are not significant. The critical crack size at 2 million DN is about 1/4 in., (0.63 cm) which is somewhat larger than the depth of the spalls observed in the few spalling fatigue failures which have been found. At 3 million DN, however, the critical crack size is of the order of 1/16 in. (.16 cm) and at 4 million DN, the critical crack size is about 1/32 in. (.08 cm).

The situation is further complicated by the ability of a very small crack at high DN to grow rapidly by fatigue to reach the critical crack size before spalling can occur. For example, in Figure 6, we show that a crack will grow at a rate of 10^{-7} in/cycle at a stress intensity of $\Delta K = 5$ ksi·in$^{1/2}$. At 2 million DN the hoop stress in 20 ksi, and using equation (2), this value of ΔK corresponds to a crack about 0.02 in. (.05 cm) long. For an approximate calculation we assume that the crack grows at a constant rate, 10^{-7} in/cycle. At this rate, only 2.5 million cycles will be required to reach the critical crack size, which is about 0.25 in. This seldom happens, because the stress concentration is first relieved by spalling. At 3 million DN, however, $\Delta K = 5$ ksi·in$^{1/2}$ corresponds to a crack only .005 in. (.012 cm) long, and this would reach the critical crack size, .062 in., in about 600,000 cycles. At 4 million DN, the corresponding crack is .002 in. (.005 cm) and it would reach critical size (.031 in.) in about 300,000 cycles.

These calculations indicate the difficulties which would be associated with the use of these bearing materials at high DN values. There is the danger that a small fatigue crack could grow very rapidly to the critical crack size and then fail catastrophically. A few bearing tests have been run at 3 and 4 million DN, and these predictions have been verified. The lifetimes were only a few hours and the rotating rings broke up into fragments. Athigh DN, failure occurs by fracture before it occurs by spalling fatigue, and it is evident that prudent design will not go far beyond 2 million DN with the current bearing steels.

Acknowledgements

The data on fracture toughness and fatigue crack growth rates are taken from a series of 3 papers with J. A. Rescalvo[2,3,4] which were the basis of his doctorate thesis at M.I.T. He was supported by a fellowship from the Instituto Tecnologico Para Postgraduados (ITP) of Spain. The research was sponsored by the Fafnir Bearing Division of Textron, Inc. and we are grateful to Robert Fairchild, P. K. Pearson and Henry Hubbell for their active help and encouragement. We would also like to acknowledge several interesting discussions with John C. Clark and Eric Bamberger of the General Electric Company.

References

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2. J. A. Rescalvo and B. L. Averbach, Fracture Toughness and Fatigue Crack Growth Rate in 52100 Bearing Steel. Submitted for publication.

3. J. A. Rescalvo and B. L. Averbach, Fracture and Fatigue in M-50 and 18-4-1 High Speed Steels. Metallurgical Transactions A, Vol. 10A, 1265-1271, September 1979.

4. J. A. Rescalvo and B. L. Averbach, SAM Analysis of Intergranular Fracture in 52100 steel. To be published in Met. Trans.

Table 1. Compositions of Bearing Steels

type	C	Cr	Mo	W	V	MN
52100	1.06	1.54	----	-----	----	0.40
M-50	0.80	4.14	4.26	-----	0.99	0.27
18-4-1	0.74	4.41	0.30	17.95	1.20	0.31

Table 2. Undissolved Carbides and Carbon Contents of Martensite in 52100 Bearing Steel

austenitizing temperature	undissolved carbides volume %	% C in martensite
annealed	16	----
815°C (1500°F)	8	0.49
845°C (1550°F)	6	0.63
870°C (1600°F)	5	0.67

Table 3. Compositions of Martensites in Hardened Bearing Steel

steel	austenitizing temperature	undissolved carbides vol. %	composition of martensite, wt. % C	Cr	Mo	W	W
52100	815°C(1500°F)	8	.49	.82	----	----	----
M-50	1095°C(2200°F)	2.6	.53	4.00	2.95	----	.35
18-4-1	1250°C(2280°F)	15.2	.42	3.60	----	5.00	.65

Table 4. Critical Crack Sizes

$$a_c = K_{Ic}^2/\pi\sigma^2$$

steel	K_{Ic} ksi·in$^{1/2}$	DN 10^6	hoop stress ksi	a_c in
52100	18	2	20	.26
		3	38	.07
		4	64	.025
M-50	16	2	20	.20
		3	38	.06
		4	64	.02
18-4-1	19	2	20	.29
		3	38	.08
		4	64	.03

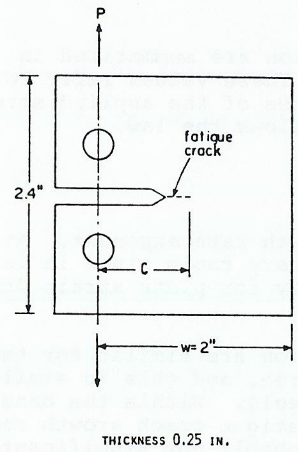

Figure 1. Fracture Toughness Specimen

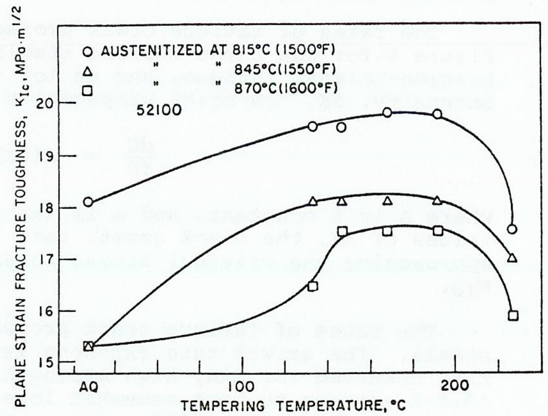

Figure 2. Plane strain fracture toughness, K_{Ic}, in hardened 52100. The maximum scatter was ± 0.5 units.

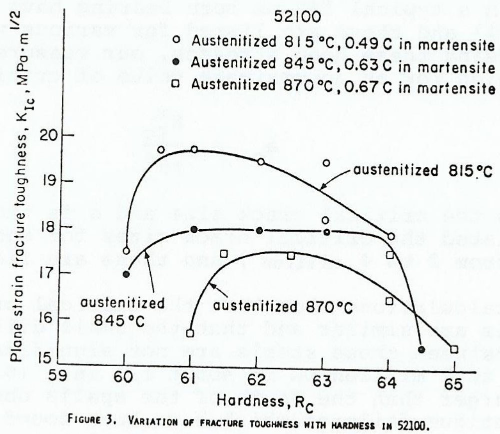

Figure 3. Variation of fracture toughness with hardness in 52100.

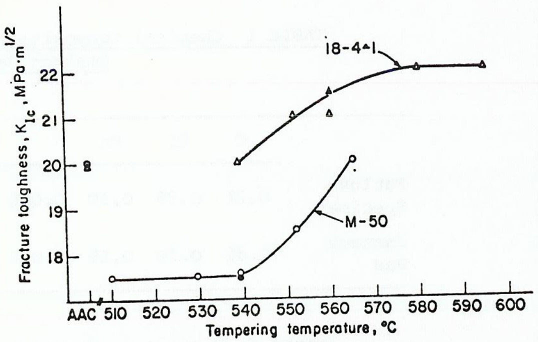

FIGURE 4: FRACTURE TOUGHNESS OF M-50 AND 18-4-1. SOLID POINTS, REFRIGERATED AFTER FIRST TEMPER.

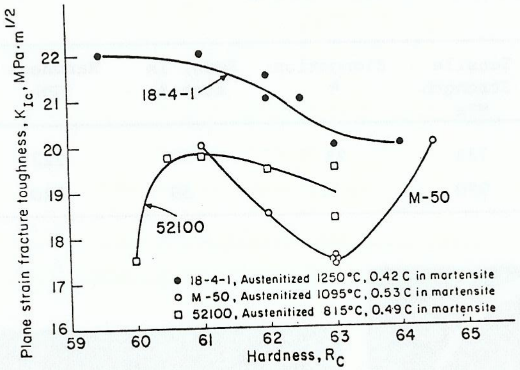

FIGURE 5. A COMPARISON OF THE FRACTURE TOUGHNESS OF THE THREE BEARING STEELS.

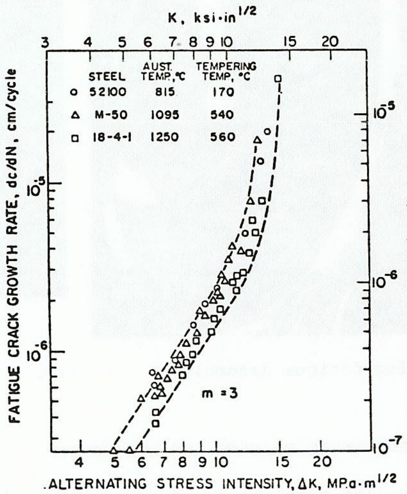

FIGURE 6. FATIGUE CRACK PROPAGATION RATES IN THE THREE BEARING STEELS.