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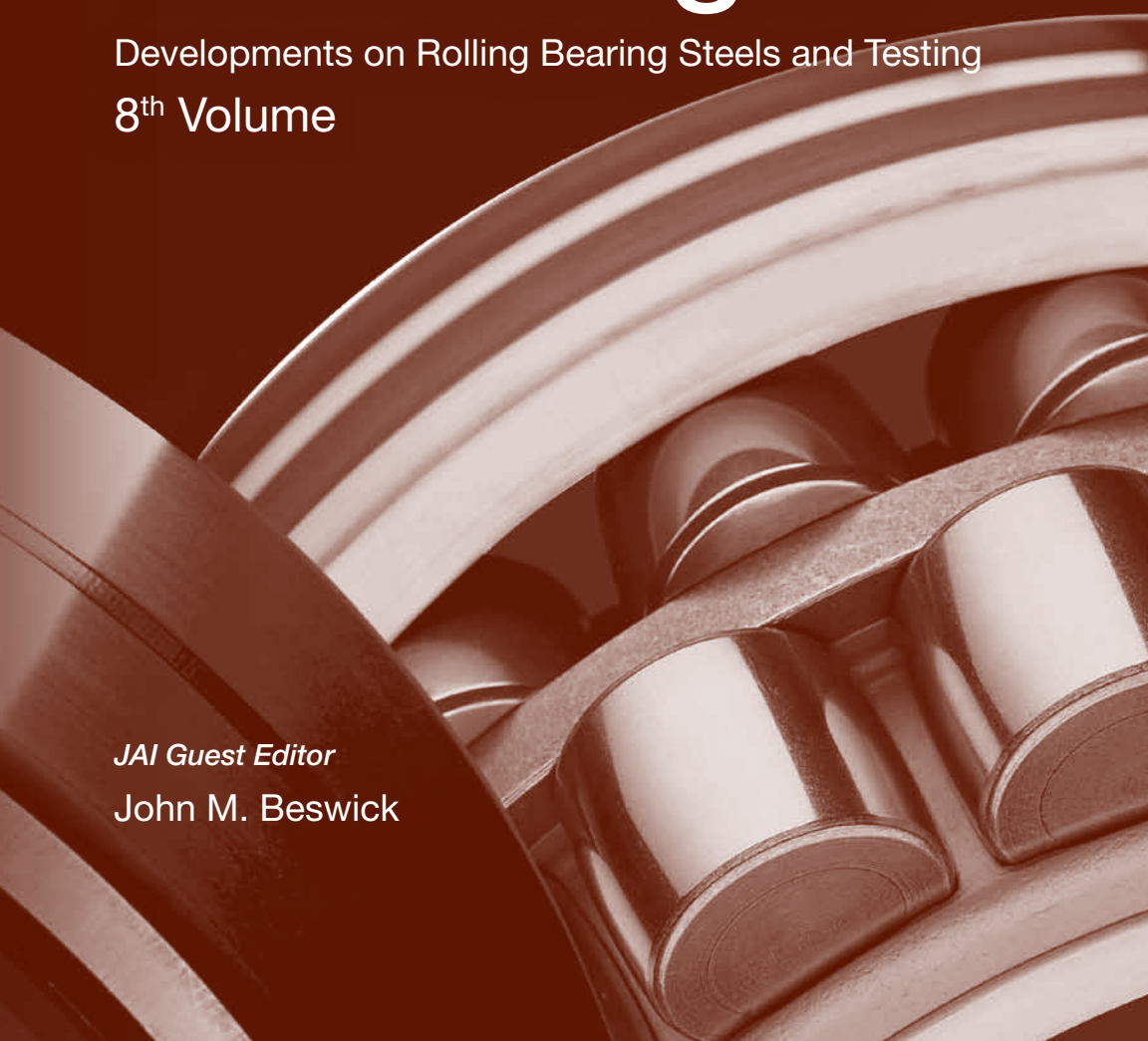
STP 1524

Bearing Steel Technologies

Developments on Rolling Bearing Steels and Testing
8th Volume

JAI Guest Editor

John M. Beswick



**Journal of ASTM International
Selected Technical Papers STP1524
Bearing Steel Technology:
Developments in Rolling Bearing Steels
and Testing—8th Volume**

JAI Guest Editor:
John M. Beswick



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Foreword

THIS COMPILATION OF THE *JOURNAL OF ASTM INTERNATIONAL (JAI)*, STP1524, on *Bearing Steel Technologies, 8th Volume: Developments on Rolling Bearing Steels and Testing*, contains only the papers published in JAI that were presented at a symposium in Vancouver, BC, Canada on May 21–22, 2009 and sponsored by ASTM Committee A01 on Steel, Stainless Steel, and Related Alloys.

The JAI Guest Editor is John Beswick, SKF Business and Technology, Nieuwegein, The Netherlands.

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Overview

Bearing steel technology is a seemingly all-encompassing term to describe the metallurgical know-how on steels and processes for the production and usage of rolling bearing steels. In the pursuit of efficiency, the rolling bearing industry has standardized the steels and testing methods and reduced the costs of the metallurgical processes. As time elapses, the knowledge of why and how the standards were prepared fades into the past, i.e. it is forgotten. Much has been published in the open literature on the subject for specialists (fellow steel technologists) and the first ASTM International Symposium on Bearing Steel, sponsored by ASTM Committee A01 and its Subcommittee A01.28, was held in Boston in 1974. Since then, bearing steel symposia have been held at regular intervals and the program for the ASTM Eighth International Symposium on Bearing Steel, in Vancouver on May 21–22, 2009, contained papers on the subject of bearing steel technologies. In particular, the subject of micro cleanliness assessment methods in bearing steels was revisited 35 years after the 1974 Boston symposium on the subject.

Knowledge of what is important in bearing steel steelmaking and processing is of utmost relevance to efficient steel and component sourcing and steel usage in rolling bearing components. Representatives from many of the top bearing steel steelmakers, rolling bearing producers, and research and development institutes presented papers. The presenters originated from: eight countries, seven bearing steelmakers, six rolling bearing producers, and seven research and development institutes.

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**BEARING STEEL STEELMAKING
AND SEMI-FINISHED
PRODUCT MANUFACTURING
TECHNOLOGIES**

Kwanho Kim,¹ Kyungshik Oh,¹ Joodong Lee,¹ and Duklak Lee¹

Quantitative Relationship between Degree of Center Segregation and Large Carbide Size in Continuously Cast Bloom of High Carbon Chromium Bearing Steel

ABSTRACT: One of the disadvantages of the continuous casting process, compared to ingot casting, is the center segregation, which causes the formation of large carbides in blooms of high carbon chromium bearing steel. Many activities have been performed to minimize the center segregation by steel manufacturers, and until now, the soft reduction is chosen as the best way to control it. Large carbides formed during casting, detrimental to the rolling contact fatigue life of bearing components, can just be eliminated by holding blooms at high temperatures for a long time before hot rolling, which is called the soaking process. Therefore it is necessary to examine the relationship between the degree of center segregation and large carbide size in continuously cast blooms for a more efficient soaking process. The aim of this research is to describe the relationship quantitatively. Continuously cast blooms of high carbon chromium bearing steel, AISI 52100, were investigated, and the degree of center segregation was not defined as the ratio of carbon concentration in the segregated region to that of nominal composition (C/C_0), as it is defined conventionally, but evaluated with a discrete index by comparing the macrograph of a bloom with the standard one settled arbitrarily for the study. The higher was the degree of center segregation, the bigger was the large carbide, and the quantitative relationship between the degree of center segregation and the maximum size of the large carbide was well fitted linearly with a reliability of 95.9 %. In order to apply soft reduction adequately during casting for the study, an in situ equipment to measure the real thickness of a solidifying bloom was installed in front of the soft reduction zone, and both the degree of center segregation and the large carbide size were improved.

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KEYWORDS: continuous casting, center segregation, large carbide, soft reduction, soaking

Introduction

Continuous casting has become a widely accepted process to manufacture high carbon chromium bearing steels by both steel producers and bearing industry around the world, even though ingot-cast materials are still required for special purposes. It is a well-known fact that, however, the continuous casting process has a fatal disadvantage compared to ingot casting; the center segregation in cast blooms or billets [1], causing the formation of large (primary or massive) carbides. It was reported that the formation of large carbides in high carbon chromium bearing steels was at a maximum in the center of a continuously cast bloom [2].

The rolling contact fatigue life of bearing steels is strongly dependent on some metallurgical factors, such as chemical compositions, nonmetallic inclusions, large carbides, hardness after final heat treatment, and so on [3,4]. From this viewpoint, the formation of large carbides by the center segregation in continuously cast blooms was proven to be detrimental to the fatigue life of bearing steels [5]. Many activities have been performed, therefore, to minimize the center segregation by steel manufacturers, and until now, the soft reduction during solidification is chosen as the best way to control it as shown in Fig. 1.

In fact, the formation of large carbides cannot be completely inhibited even in the case that the soft reduction is applied in a proper manner during casting. The actual way to eliminate them entirely is holding continuously cast blooms at high temperatures for long enough times to distribute the segregated alloying elements homogeneously before hot rolling. This kind of homogenization is called the soaking process. Okamoto et al. defined the relative soaking time requirements for continuously cast high carbon chromium bearing steel based on the dissolution of large carbides, which result from the enrichment of carbon and chromium ahead of the solidification zone during casting [2,6]. Similarly, Malinovskaya et al. defined the soaking time requirements for dendritic segregation homogenization in 1C-1.5Cr steel [7].

From the above reports, it can be considered that if the soft reduction can minimize the center segregation during casting, thus leading to the minimal formation of large carbides, the load of the soaking process to remove them can also be reduced at the same time. Accordingly, it is necessary to examine the relationship between the degree of center segregation (DCS) and large carbide size (LCS) in continuously cast blooms for a more efficient soaking process. This paper focused on the relationship and made an effort to describe it quantitatively.

Materials and Experimental Method

The high carbon chromium bearing steel used for the study was AISI 52100, and five heats (500 tons) of the steel were melted. Their chemical compositions are shown in Table 1, and it can be considered that there is no difference in

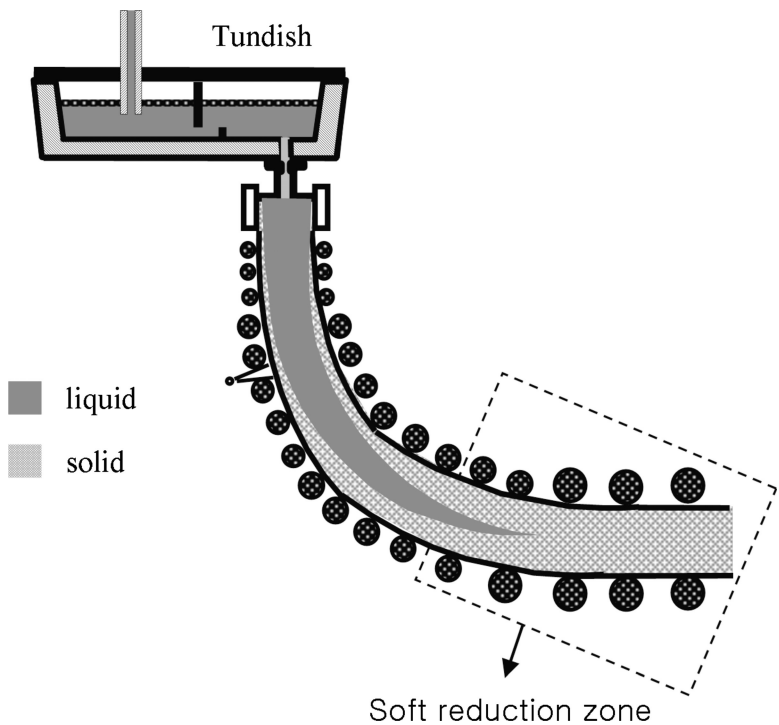


FIG. 1—Soft reduction zone in a curved continuous caster.

chemical compositions among their continuously cast blooms. This is important because the relationship between the DCS and LCS could be distorted. If there was a difference in chemical compositions among the cast blooms, not only their DCS but also their LCS might be varied even for the same casting condition.

The continuously cast blooms of the size of 400 mm in width and 300 mm in thickness were manufactured by varying casting conditions so that they had different DCSs and their LCSs from one another. The amount of soft reduction was varied 6.0~9.6 mm, and the casting speed was 0.65~1.05 m/min in order to obtain different DCSs. The main parameter that had an effect on seg-

TABLE 1—Chemical compositions of AISI 52100 steels continuously cast for the study (wt %).

Heat Number	C	Si	Mn	P	S	Al	Cr	Ti	O
1	1.02	0.26	0.35	0.016	0.006	0.009	1.46	0.005	0.0006
2	1.00	0.25	0.34	0.016	0.005	0.010	1.47	0.006	0.0006
3	1.02	0.26	0.34	0.015	0.008	0.011	1.44	0.008	0.0005
4	0.99	0.24	0.34	0.018	0.004	0.009	1.46	0.005	0.0007
5	1.01	0.25	0.36	0.016	0.006	0.013	1.44	0.005	0.0008

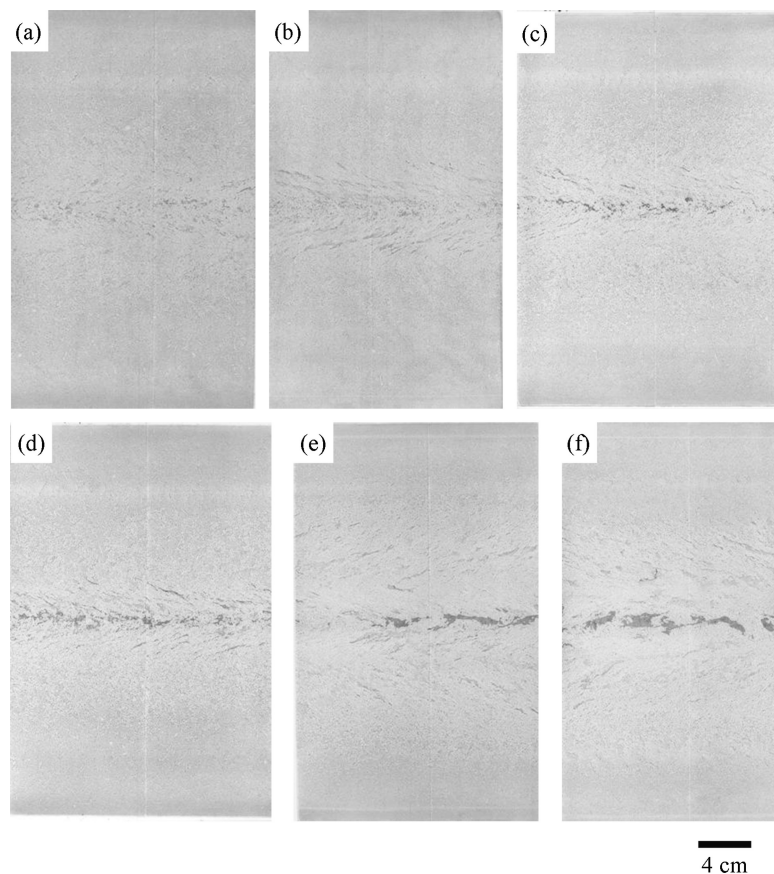


FIG. 2—Standard macrographs of AISI 52100 continuously cast blooms having DCSs of (a) 0, (b) 1, (c) 2, (d) 3, (e) 4, and (f) 5, respectively. The macrographs were compared to determine the DCS of blooms investigated.

regation reduction was the amount of soft reduction. After casting, each bloom was macroetched, and then the DCS was determined by comparing its macrograph with the standard ones in Fig. 2. The DCS had been divided from 0 to 5, and it was the worst degree with the highest index. It should be noted that the DCS, for the convenience of the investigation, was not defined as the ratio of carbon concentration in the segregated region to that of nominal composition (C/C_0), as it is defined conventionally, but evaluated with a discrete index by comparing the macrograph of a bloom with the standard one. For reference, it was estimated that a DCS of 2 was approximately equal to 1.3 of C/C_0 .

The specimens for microstructural analysis were of $100 \times 100 \text{ mm}^2$ and obtained from the segregated region of blooms, and, at least, 30 large carbides per each specimen were examined to determine the LCS by measuring their maximum thicknesses. Then the average size of the large carbide in each bloom was calculated.

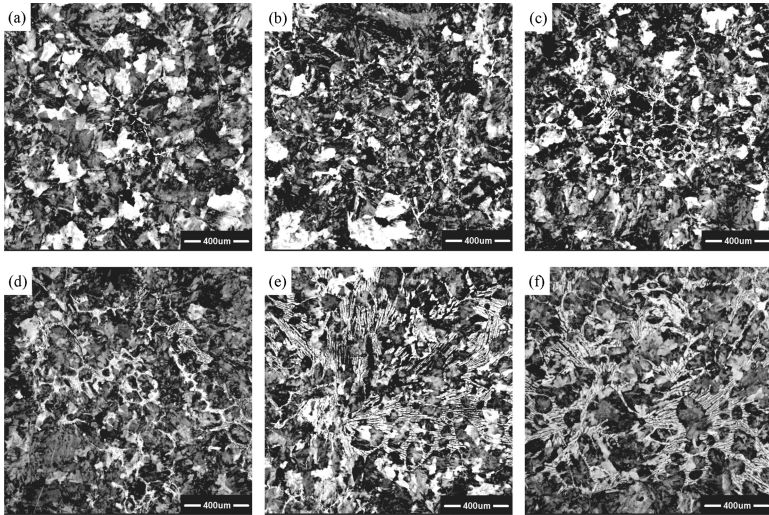


FIG. 3—Micrographs showing the large carbides in AISI 52100 continuously cast blooms having DCSs of (a) 0.5, (b) 1, (c) 1.5, (d) 2, (e) 3, and (f) 4, respectively.

Test Results and Discussion

Continuously cast blooms were produced to have different DCSs by varying casting conditions. The microstructures within the center segregation region were shown in Fig. 3. In the cast bloom of the DCS of 0.5 (Fig. 3(a)), the large carbide was almost not formed. With the increase in the DCS, however, not only the amount of large carbide formed but the thickness of it have increased, over 100 μm in the cast bloom of the DCS of 4 (Fig. 3(f)). Thus it can be assured that an increase in the DCS promotes the precipitation of large carbides in continuously cast blooms.

At least 30 large carbides per specimen were examined to measure their thicknesses, and the result is shown in Table 2. It is confirmed again that the higher the DCS of a continuously cast bloom is, the bigger the large carbide is. Figure 4 illustrates the relationship between the DCS and the average size of large carbide in continuously cast blooms, and it is well fitted linearly as follows:

$$\text{average LCS} = 2.44 \times \text{DCS} + 30.66 \quad (1)$$

where:

LCS=large carbide size and

DCS=degree of center segregation.

The above quantitative relationship had a reliability of 88.9 %, and the standard errors of the slope and the intercept were 0.302 96 and 0.717 64, respectively.

In order to eliminate the large carbides completely, the soaking condition for the large carbides should be focused not on the average but on the maximum size. Therefore, it is necessary to examine the quantitative relationship

TABLE 2—Average and maximum sizes of large carbides in continuously cast blooms.

DCS	LCS (μm)	
	Average	Maximum
0.5	30.16	48
1	32.85	59
	34.14	63
	34.66	69
	34.37	67
1.5	34.85	81
2	38.24	93
4	39.25	105
	41.35	111

between the DCS and the maximum size of the large carbide. The result is shown in Fig. 5, and the relationship is also well fitted linearly as follows:

maximum LCS = 15.76 × DCS + 45.82 (2)

This quantitative relationship had a reliability of 95.9 %, higher than Eq 1, and the standard errors of the slope and the intercept were 1.153 94 and 2.733 43, respectively. The statistics on Eqs 1 and 2 is summarized in Table 3. Equation 2 had more reliability than Eq 1, which means that the maximum size of the

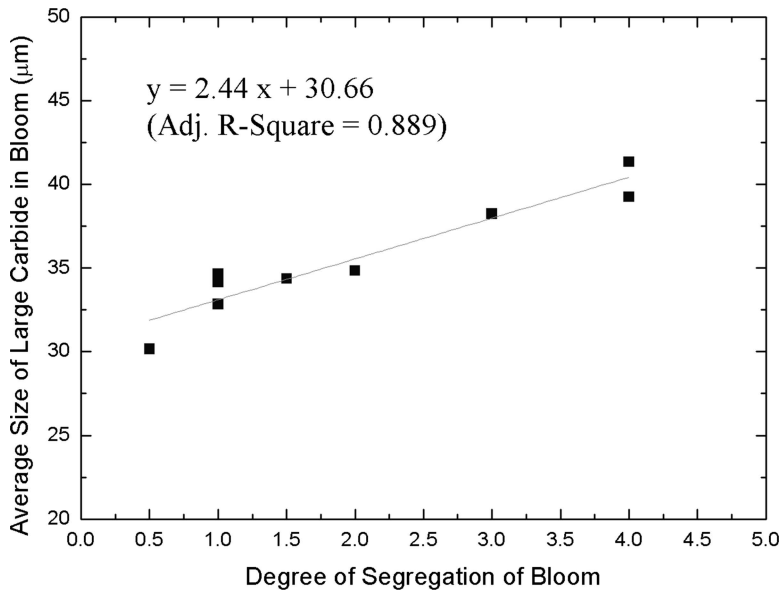


FIG. 4—The relationship between the DCS and the average size of the large carbide in continuously cast blooms. Adj. R-square means the reliability of the linear fitting.

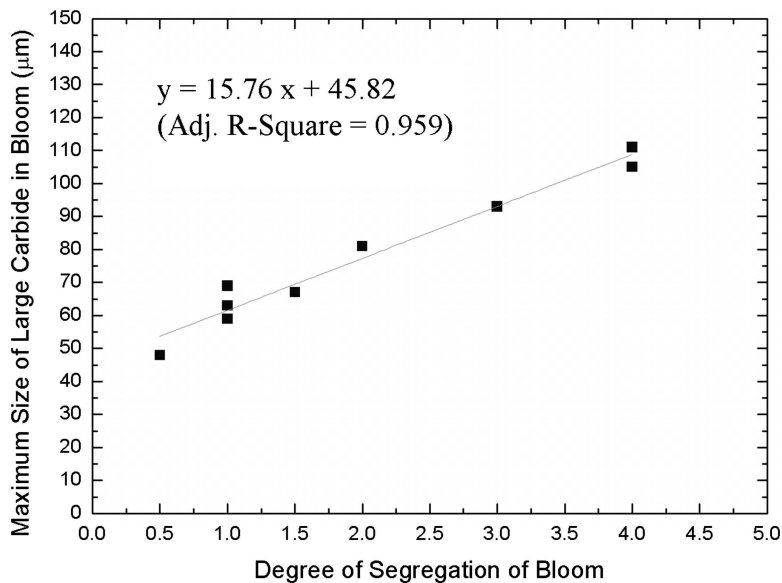


FIG. 5—The relationship between the DCS and the maximum size of the large carbide in continuously cast blooms.

large carbide would be more closely related with the DCS than the average size.

Regarding Figs. 4 and 5, it should be noted that the average or maximum sizes of large carbides in continuously cast blooms of the same DCS of 1 and 4 were different from one another. This implies that the LCS is strongly dependent on the DCS but not entirely. Other factors can affect the formation of large carbides during casting, such as the casting start temperature, the casting speed, the cooling rate, etc.

The relationship between the DCS of cast blooms and the average size of the large carbide in hot rolled wires was shown in Fig. 6. The diameter of wires was 10 mm. It seems that there is no distinct relationship between them. Possibly it results from the disappearance of the effect of the DCS due to the soaking process and hot rolling.

Okamoto and co-workers defined the relative soaking time requirements for continuously cast high carbon chromium bearing steel to dissolve large

TABLE 3—Standard errors and R-square values (reliability) for the quantitative relationship shown in Figs. 4 and 5.

Size of Large Carbide	Slope		Intercept		Statistics
	Value	Standard Error	Value	Standard Error	Adj. R-Square
Average	2.438 97	0.302 96	30.663 18	0.717 64	0.8886
Maximum	15.758 62	1.153 94	45.816 09	2.733 43	0.958 66

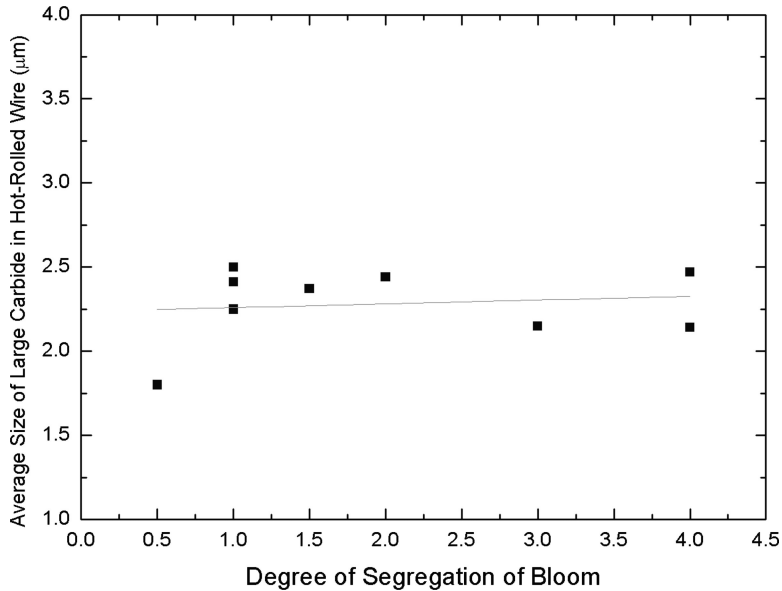


FIG. 6—The relationship between the DCS of cast blooms and the average size of the large carbide in hot rolled wires.

carbides which result from the center segregation as follows [2,6]:

$$\ln t_h = 86\,300/T - 44.56 + \ln(2.5a^2/4) \tag{3}$$

where:

- t_h =soaking time in h,
- a =LCS in cm, and
- T =soaking temperature in K.

According to Eq 2, even if a continuously cast bloom of high carbon chromium bearing steel is manufactured with the best quality in center segregation, there can be always large carbides larger than $45.82\ \mu\text{m}$, as stated earlier. By calculating with Eq 3, it is necessary to soak a continuously cast bloom at 1200°C for 16 h in order to dissolve those large carbides. However, if the maximum size of large carbide in blooms is reduced to the level of the average size of it in this study, $30\ \mu\text{m}$, then the soaking time can be diminished to 7 h at the same temperature. This suggests that it is possible to reduce the soaking time at high temperatures dramatically with the decrease in the LCS by enhancing the DCS, which is believed to be achieved by the soft reduction so far.

Actually, during casting high carbon chromium bearing steels, the soft reduction is carried out by controlling the roll gap between upper and lower soft reduction rolls. For curved continuous casters, it is apt to be a narrow space between upper and/or lower soft reduction roll and a solidifying bloom, as shown in Fig. 7 by arrows. It means that soft reduction rolls do not get in direct touch with a solidifying bloom, and then the soft reduction could not be performed adequately. In other words, the amount of soft reduction on a casting

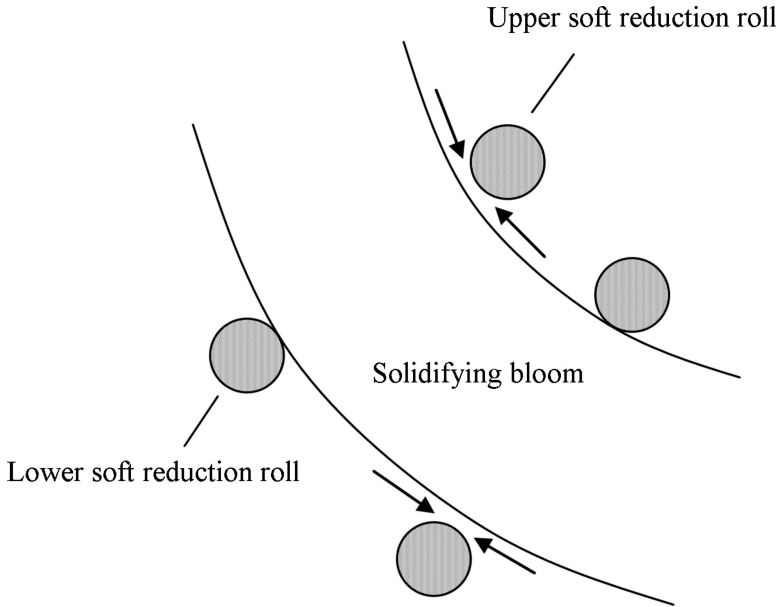


FIG. 7—Schematic illustration showing the presence of narrow space between an upper and/or lower soft reduction roll and a solidifying bloom in a curved continuous caster.

bloom by soft reduction rolls will be insufficient as much as the narrow space, resulting in little improvement of center segregation. One of the probable solutions to this problem is establishing the amount of soft reduction not by the roll gap between upper and lower soft reduction rolls but by the real thickness of a solidifying bloom.

For this study to accomplish the above solution, an in situ equipment to measure the real thickness of a solidifying bloom was installed in front of the soft reduction zone, and it was confirmed that both the DCS and the LCS were improved. Figure 8 shows the macrographs of continuously cast blooms before and after the installation of the in situ equipment. The average DCS was improved from 2.25 to 0.5 by the installation of it. Correspondingly, the LCS would be decreased so that the efficiency of the soaking process was evidently increased. As shown in Fig. 9(d), a large carbide did not remain in the bloom in Fig. 8(b) after soaking at 1200°C for 2 h.

Conclusions

Continuously cast blooms of high carbon chromium bearing steel, AISI 52100, were investigated to reveal the relationship between the DCS and the LCS in them. The DCS was not defined as the ratio of carbon concentration in the segregated region to that of nominal composition (C/C_0), as it is defined conventionally, but evaluated with a discrete index by comparing the macrograph

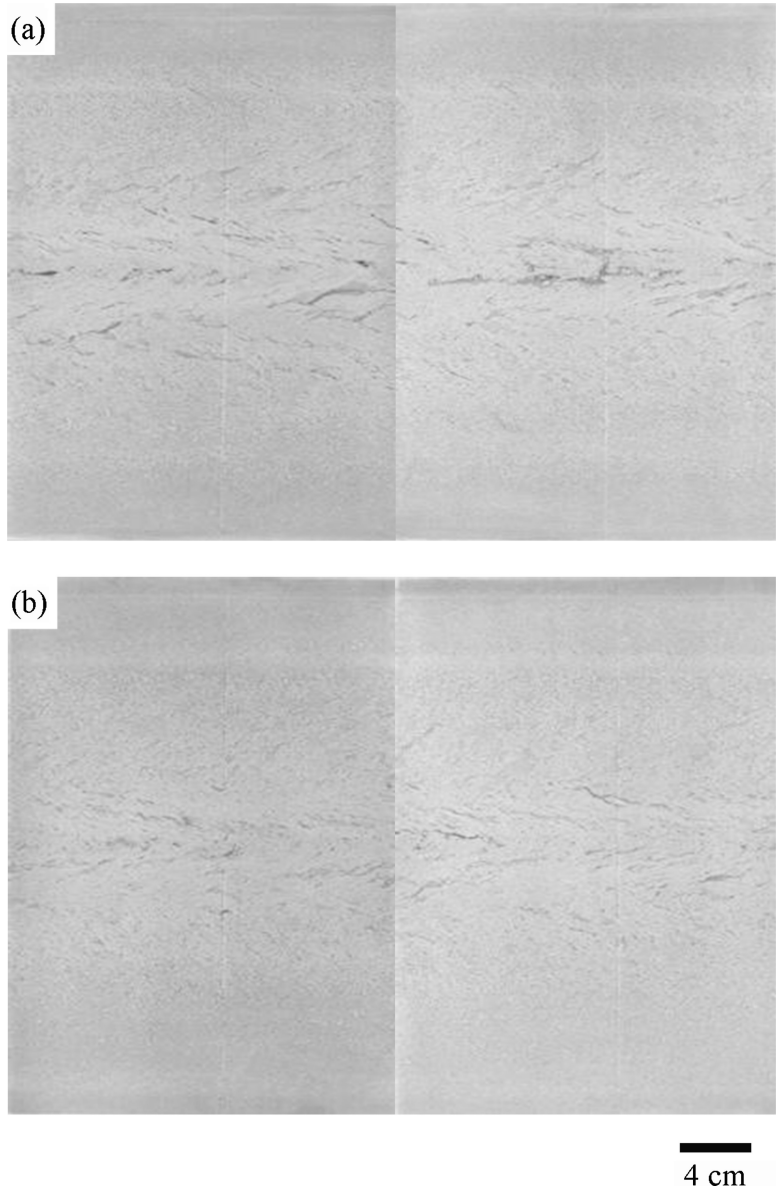


FIG. 8—Macrographs showing the improvement of the average DCS from (a) 2.25 to (b) 0.5 by installing an in situ equipment to measure the real thickness of a solidifying bloom.

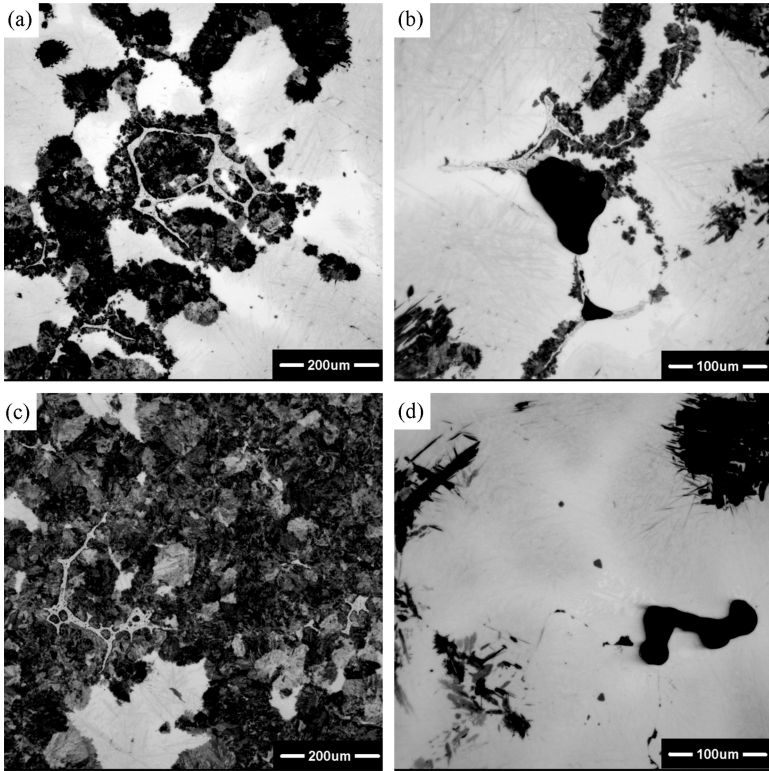


FIG. 9—Micrographs showing the effect of the in situ equipment on soaking continuously cast blooms in Fig. 8 at 1200°C: (a) Fig. 8(a) soaked for 1 h, (b) Fig. 8(b) for 1 h, (c) Fig. 8(a) for 2 h, and (d) Fig. 8(b) for 2 h.

of a bloom with the standard one settled arbitrarily for the study. The following conclusions were achieved.

- (1) The higher the DCS was, the bigger the large carbide became.
- (2) The quantitative relationship between the DCS and the maximum size of the large carbide was well fitted linearly as follows:

$$\text{maximum LCS} = 15.76 \times \text{DCS} + 45.82$$

The quantitative relationship had a reliability of 95.9 %, and the maximum size of the large carbide was more closely related with the DCS than the average size.

- (3) After hot rolling, there was no distinct relationship between the DCS of cast blooms and the average size of the large carbide in hot rolled wire possibly due to the soaking process and hot rolling.
- (4) In order to apply soft reduction adequately during casting for the study, in situ equipment to measure the real thickness of a solidifying bloom was installed in front of the soft reduction zone, and both the DCS and the LCS were improved.

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**MATERIAL AND HEAT
TREATMENT DESIGN
CONSIDERATIONS, INCLUDING
FRACTURE MECHANICS
AND STRUCTURAL STRENGTH,
FOR ROLLING BEARINGS**

J. Dong,¹ H. Vettters,² F. Hoffmann,² and Hans W. Zoch²

Microstructure and Fatigue Strength of the Bearing Steel 52100 after Shortened Bainitic Treatment

ABSTRACT: Quenching to obtain martensite is the mostly applied process for standard rolling element bearings. Isothermal treatment in the lower bainitic range is used as an alternative method to generate favorable compressive residual stress on the surface of components, e.g., in spherical roller bearings. The duration of the bainitic treatment, however, is much longer than that of a martensitic treatment because more or less a complete transformation of austenite to bainite is usually requested. This causes higher energy consumption and a longer production period. Therefore it is desirable to perform bainitic treatment with a shortened process duration. In the present work possible processes for shortening the bainitic treatment of the bearing steel 52100 were primarily investigated by dilatometric experiments. Some selected processes were carried out in an industrial salt bath. The microstructures of bainite were observed by optical microscope, transmission electron microscope, and field emission scanning electron microscope. These were compared to martensitic microstructures. The cyclic fatigue strength of the steel after shortened bainitic treatments was examined using a rotating-bar fatigue test. The results show that the fatigue resistance while maintaining the requested minimum hardness of 58 HRC was even enhanced significantly through the shortened treatments particularly by means of a two-step bainitic treatment. The process duration was only about 25 % of the conventional time. The influence of the bainitic microstructure on the fatigue strength of the steel is discussed.

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KEYWORDS: lower bainite, dilatometry, fatigue strength, bainitic microstructure, bearing steel

Introduction

High fatigue strength and dimensional stability are the most important requirements of precision machine components. In the case of rolling element bearings made from the steel 52100, martensitic hardening followed by tempering at low temperature is often applied as the final heat treatment method. Yet the kinetic of microstructural transformation generates unfavorable tensile residual stresses at the surface of martensitic through-hardened rolling bearings. Retained austenite furthermore can decompose during application and cause dimensional change [1]. The well-known isothermal treatment in the lower bainitic range allows us to obtain suitable mechanical properties of machine components and favorable compressive residual stresses at the surface. Additionally the dimensional change of machine components during operation caused by the transformation of retained austenite can be greatly reduced [2]. A main disadvantage of the isothermal bainitic treatment in salt baths however is the long duration of the treatment, which exceeds that of martensitic hardening by hours. This will result in higher costs.

Therefore it would be desirable to shorten the transformation duration in the lower bainitic range while maintaining the beneficial properties. It has already been reported [3,4] that a short pre-quenching in the martensitic range could shorten the duration of bainitic transformation. An excellent combination of strength and toughness could be obtained by a suitable mixture of bainite and martensite in the microstructure of the steel [5]. A cyclic heat treatment process could be another possible method to accelerate bainitic transformation [6].

In the present work the transformation behavior of the bearing steel 52100 in the temperature range between 210 and 300 °C was investigated by means of dilatometry. Some selected processes were carried out in industrial salt baths. The cyclic fatigue strength of the steel after shortened bainitic treatments was examined using a rotating-bar fatigue test. The microstructures were observed by optical microscope (OM), transmission electron microscope (TEM), and field emission scanning electron microscope (FESEM). The influence of the different microstructures on the fatigue strength of the steel is discussed.

Experimental

Material and Shape of Specimens

The bearing steel 52100 was used as specimen material. The steel bars of 60 mm in diameter were produced by continuous casting, hot-rolling, and spheroidization. Specimens were machined from the bars with axial direction parallel to the rolling direction. To exclude the influence of primary segregation bands, the core of the bars within a diameter of 15 mm was rejected. Cylindrical specimens with diameter of 4 mm and length of 10 mm were made for

dilatometric investigation. Disk specimens of 5 mm in thickness and 30 mm in diameter were used for metallographic inspection, hardness test, and X-ray diffraction analysis. Specimens for fatigue test, after machining and heat treatment, were finished by grinding and polishing to a roughness of $R_t=0.6 \mu\text{m}$, as described in Refs 7 and 8. The surface layer of the specimens showed a compressive residual stress of about $-770\pm110 \text{ MPa}$ in longitudinal direction caused by manufacturing procedures. These stress values at the surface of the specimens were analyzed by the X-ray diffraction method using Cr radiation. The specimens with similar surface residual stress grades were selected for the fatigue test. The influence of the surface residual stress caused by manufacturing procedures was then kept as a constant for all fatigue experiments.

Heat Treatment Processes

Several heat treatment processes to accelerate the bainitic transformation were studied first by dilatometry. During the dilatometric experiments, the specimens were inductively heated in vacuum ($7.7\times10^{-3} \text{ mbar}$) and quenched or cooled in nitrogen gas. The bainitic transformation follows as an autocatalytic process that can be described by dilatometric measurements [4]. The increase in length gives the portion of the volumetric increase due to the transformation process. The relationship between the amount of transformed bainite and the duration of the isothermal treatment can be expressed by the Johnson-Mehl-Avrami equation [9,10].

The heat treatments of the specimens for fatigue test were conducted in industrial salt baths and quenched in oil or water at ambient temperature.

Retained Austenite and Hardness

The amount of retained austenite was determined by the X-ray diffraction method with Cr K_α radiation. The two-peak method of $(220)_\gamma$ and $(211)_\alpha$ was used to calculate the volume percentage of retained austenite [11]. For a reliable accuracy the measurement was repeated ten times, and the data were evaluated statistically. The three standard deviations lie within two percentage points for the quantities of retained austenite. The two-peak method was acceptable because of the absence of preferred orientation in the specimens [12]. The hardness was measured by the Rockwell C test. The mean values were achieved from five measurements for each specimen and the standard deviation was less than 1 HRC unit.

Rotating-Bar Fatigue Test

To investigate the cyclic fatigue strength (S_w), rotating-bar fatigue tests have been performed at four load levels. Each level has been certified with at least seven specimens. The tests have been carried out on two test equipments³ with the load ratio $R=-1$, the loading frequency of 120 Hz, and the ultimate load cycle of 1×10^7 . For statistical evaluation it has been assumed that the cyclic

³Schenck PUNZ.

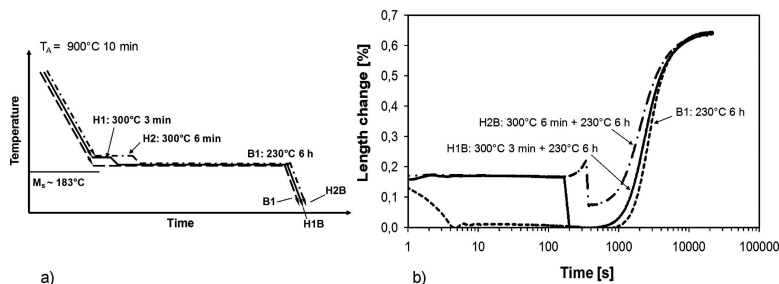


FIG. 1—Dilatometric experiments: (a) Processes and (b) dilatation curves of the bainitic transformation with pre-holding (origin: $T=300^\circ\text{C}$, $t=0$, and $\Delta L=0$).

fatigue limit follows a two-parametric Weibull distribution [13]. The parameters of the Weibull distribution have been determined by regression of detected mean values of fracture probability. The cyclic fatigue strength has been evaluated according to the fracture probability (P_B) of 50 %.

Results

Dilatometric Experiments

A pre-holding stage at elevated temperature, as illustrated in Fig. 1(a), may accelerate the diffusion process and therefore benefit bainite nucleation. Two experiments, H1B and H2B (Fig. 1(a)), were conducted and compared to a conventional full bainitic treatment B1, which served as reference. In these experiments the specimens were austenitized at 900°C 10 min, quenched at 300°C , and then held at 300°C for 3 and 6 min, respectively, before the isothermal treatment was carried out at 230°C over a 6 h period. The austenitizing temperature was chosen to ensure a sufficient incubation time prior to the bainite formation. A higher austenitizing temperature leads to more dissolution of carbides, more stable austenite, and therefore a larger delay of bainitic transformation [14].

The dilatometric curves of the three processes were compared in the range of bainitic transformation from the temperature at 300°C (Fig. 1(b)). From this point of origin, the duration and the length change were calculated. The contraction of curve B1 at the beginning was due to the temperature change from 300 to 230°C . The increase of specimen length after an incubation period of about 1000 s was caused by the bainite formation. The pre-holding period (H1) of experiment H1B (Fig. 1(a)) remained within the incubation range as the length of the specimen remained constant within 180 s at 300°C (Fig. 1(b)). However, the pre-holding period (H2) of experiment H2B (Fig. 1(a)) already entered the bainite range, i.e., the bainitic transformation has started. This could be seen from the dilatation curve H2B (Fig. 1(b)), on which a small peak at 360 s was attributed to the bainitic transformation. The contraction of the curves H1B and H2B at 180 and 360 s was due to the temperature change from

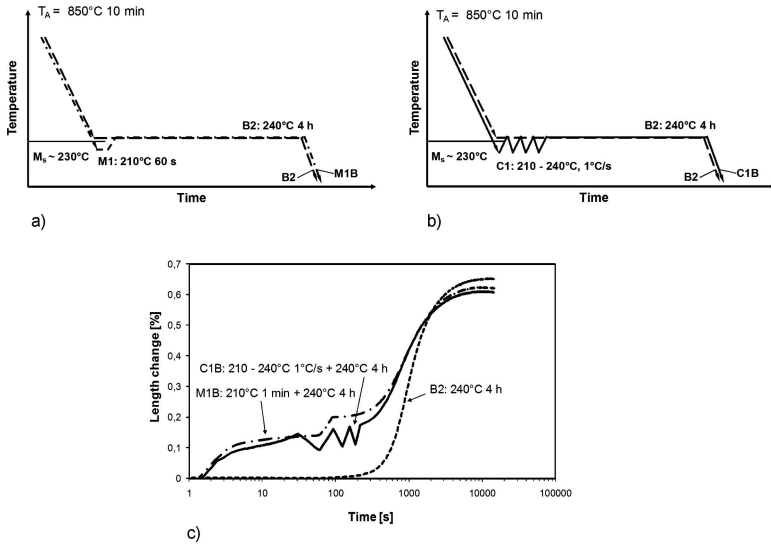


FIG. 2—Dilatometric experiments: [(a) and (b)] Processes and (c) dilatation curves of the bainitic transformation with pre-quenching or cyclic pre-austempering (origin: $T=230^\circ\text{C}$, $t=0$, and $\Delta L=0$).

300 to 230°C . After the contraction, a length change value of 0.07 % remained on the curve H2B at 360 s. This attributed to bainite formation at 300°C within the pre-holding period of H2B as mentioned above. The two experiments, H1B and H2B in Fig. 1(b), showed a higher length increase at the beginning of the isothermal treatment at 230°C , compared to reference B1, and this increased with the extended pre-holding period. The rate of the length increase slowed down gradually with the time. The transformation duration has been reduced by about 13 % for the process of H2B to reach 90 % of the total length increase caused by bainite formation, but no time has been saved for H1B.

Martensitic pre-quenching and cyclic pre-austempering (Fig. 2(a) and 2(b)) were two other candidates considered for shortening the duration of bainite transformation. The short pre-quenching of austenite into martensite range leads to formation of a small amount of martensite plates, which may work as nuclei or enhance nucleation for the subsequent bainitic transformation. In addition a cyclic temperature change may provide a beneficial stress state.

The dilatometric curves of the two experiments showed (Fig. 2(c)) the increase of bainitic transformation rate at the beginning, similar to the case of pre-holding (Fig. 1(b)). It is remarkable that the length increase of about 0.14 % (20 % of the total length increase) at the beginning is attributed to martensite formed by pre-quenching. The transformation rate slowed down gradually during the progress of the bainitic transformation. The transformation duration has been reduced about 18 % and 22 % for pre-quenching M1B and cyclic pre-austempering C1B, respectively, to reach 90 % of the total length increase caused by the formation of martensite and bainite.

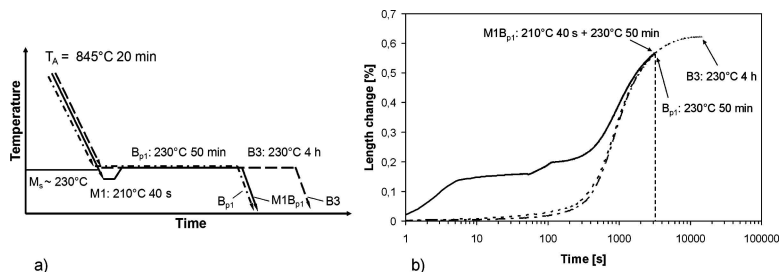


FIG. 3—Dilatometric experiments: (a) Processes and (b) dilatation curves of the bainitic transformation with post-quenching or combined treatment (origin: $T=230^\circ\text{C}$, $t=0$, and $\Delta L=0$).

The phase transformation from austenite to bainite in the final stage for the last 10 % (volume percentage) of austenite takes about 80 % of the entire transformation duration. Post-quenching (B_{p1} in Fig. 3(a)) could be considered as a measure to shorten the process duration by quenching the remaining austenite after the partial bainitic transformation into the martensite range and to obtain a duplex microstructure of bainite and martensite. (The quenching in the final stage was used to remove salt from specimens. It could be replaced by cooling in air for possible martensitic transformation.) Furthermore a combined treatment of pre-quenching and post-quenching could be taken into account ($M1B_{p1}$ in Fig. 3(a)). The austenitizing conditions were adapted to the industrial salt baths, which were used later for the heat treatments of the fatigue specimens.

The specimen was held at 230°C for 50 min and then quenched to ambient temperature by process variant B_{p1} (Fig. 3(b)). Thereby the percentage of transformed bainite was estimated to be about 87 % (volume percentage) from the length increase of the dilatation curve. The amount of retained austenite after quenching to ambient temperature was determined to be about 13 % by means of X-ray diffraction analysis. This means that martensite has not formed. This may be explained by the stabilization of the retained austenite. Because the residual austenite was gradually enriched with carbon during the partially bainitic transformation and the M_s -temperature of the residual austenite was consequently lowered [15,16] the residual austenite could not be transformed to martensite any more by quenching to ambient temperature.

The combined treatment showed a similar result as the post-quenching. The dilatation curve of $M1B_{p1}$ (Fig. 3(b)) reached nearly the same point as curve of B_{p1} after 50 min bainitic transformation. The retained austenite was determined to be 11 % and a bit less than that in the case of B_{p1} .

To accelerate bainitic transformation in the final stage, a two-step bainitic treatment [17–19] could be an alternative to post-quenching. For the confirmation purpose two experiments of a two-step process were carried out (Fig. 4(a)). The dilatation curves (BB1 and BB2) in Fig. 4(b) were compared with that obtained by the conventional one-step process ($B3$). The length of the specimens increased rapidly during the short holding period of the second step at a raised temperature. This result indicates an accelerated transformation of re-

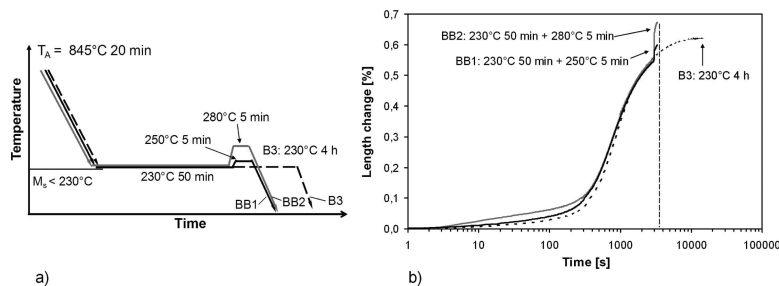


FIG. 4—Dilatometric experiments: (a) Processes and (b) dilatation curves of the two-step bainitic transformation (origin: $T=230^\circ\text{C}$, $t=0$, and $\Delta L=0$).

sidual austenite into bainite. The retained austenite after the heat treatments was determined to be 9 % for BB1, whereas it was not detectable (<3 %) for BB2. The two specimens had the same hardness of about 61 HRC. It was proven that bainitic transformation could be accelerated through a two-step process with suitable settings of temperature and holding time to obtain a full bainitic microstructure without loss of hardness.

Rotating-Bar Fatigue Tests and Hardness

Two shortened processes, namely, bainitic transformation with post-quenching (B_p) and two-step bainitic transformation (BB), were selected from the dilatometric experiments for the fatigue test. In addition a martensitic treatment (M_{qt}) and a conventional full bainitic treatment (B) were introduced as references. These four heat treatments were carried out in industrial salt baths and applied to the fatigue specimens (Table 1). The retained austenite and the hardness of the four experiments are given in Table 1. The three bainitic microstructures had lower hardness than the martensitic microstructure but achieved the requested minimum hardness of 58 HRC for rolling bearings.

The curves of the fracture probability versus stress amplitude and the values at $P_B=50$ % within the diagram (Fig. 5) showed that even the fatigue strength was increased by the shortened processes of bainite transformation with post-quenching (B_p) and the two-step bainitic treatment (BB) despite higher amounts of retained austenite. Under these test conditions the fatigue

TABLE 1—Heat treatments, retained austenite, and hardness.

Process	Heat Treatment Conditions	RA (%)	Hardness (HRC)
	Austenitization 845°C 20 min in salt bath		
M_{qt}	Quenched in oil to RT/ 170°C , 2 h	14	64
B	230°C , 4 h	<3	60
B_p	230°C , 50 min/water RT	13	61
BB	230°C , 50 min/ 250°C , 5 min/water RT	9	61

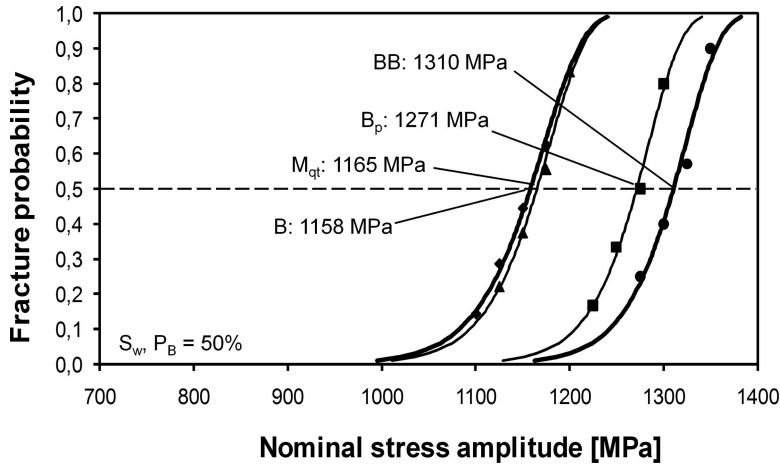


FIG. 5—Fatigue strength of the steel after different heat treatments with a fracture probability of 50 %.

strength value of the complete bainite transformation (B) was nearly at the same level as that of the martensitic quenching and tempering (M_{qt}). These results were evidently worse than those from the shortened heat treatments. At the same time it can be assumed that heat treatment related residual stresses are of lower influence because the small cross section of the specimens hampers formation of significant residual stress patterns.

A specimen with heat treatment of the two-step bainitic treatment (BB) passed through the fatigue test under the nominal stress amplitude of 1325 MPa for 10 million loading cycles. The retained austenite in the microstructure of the specimen after the fatigue test was determined to be about 9 % by means of X-ray diffraction analysis. The retained austenite remained unchanged by the cyclic stress, which implied a very high mechanical stability.

Microstructure

The microstructure of the steel after quenching and tempering (M_{qt}) consisted of tempered martensite and retained austenite (Fig. 6(a)) of about 14 % measured by X-ray diffraction. The bainitic transformation with post-quenching (B_p) led to a bainitic microstructure with 13 % retained austenite (Fig. 6(b)). The fine globular carbides in both microstructures remained unchanged. Yet the detailed characteristics of martensite and bainite in the fine grained microstructure could not be distinguished by the use of an optical microscope.

The microstructure of the tempered martensite (Fig. 7(a)) is showed by TEM as fine plate-shaped martensite with precipitated very fine carbides. The bainite (Fig. 7(b)) appeared in plate shape too, but the precipitated carbides in a preferred orientation within the plates clearly distinguished bainite from tempered martensite. In comparison with the parallel arranged carbides in bainite,

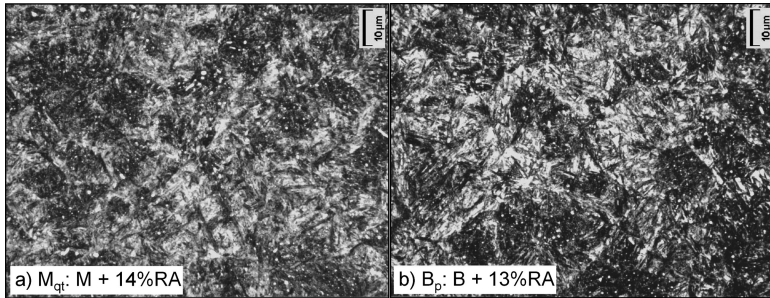


FIG. 6—Microstructures (OM) of the steel after (a) martensitic and (b) bainitic treatments (etching: 3 % Nital).

the carbides in tempered martensite were much finer and orientated in different directions. The retained austenite could be located between the plates of martensite and bainite in both microstructures.

The microstructures, as shown in Figs. 6 and 7, were further observed by FESEM (Fig. 8). The retained austenite was clearly presented in the matrix of both microstructures (arrow 1). There were no carbides precipitated within the domains of retained austenite. This means that the precipitation of carbides was not the leading reaction of bainitic transformation under the given experimental conditions. In contrast the formation of a narrow ferrite spine (arrow 2) seemed to be the leading reaction. The “secondary plates” of ferrite appeared parallel on one side of the initiating ferrite spine and had an angle of approximately 55° to 60° to the spine (arrow 3). Between the secondary plates of ferrite, the carbides precipitated. These observations agreed well with the mechanisms of bainite formation described by Spanos [14].

Discussion

The bainitic reaction is controlled by diffusion processes, which can be influenced not only by the austenitization temperature and time as well as the bai-

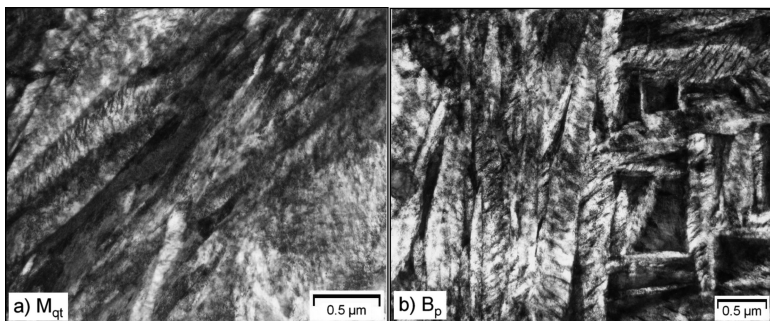


FIG. 7—Microstructures (TEM) of the steel after (a) martensitic and (b) bainitic treatments.

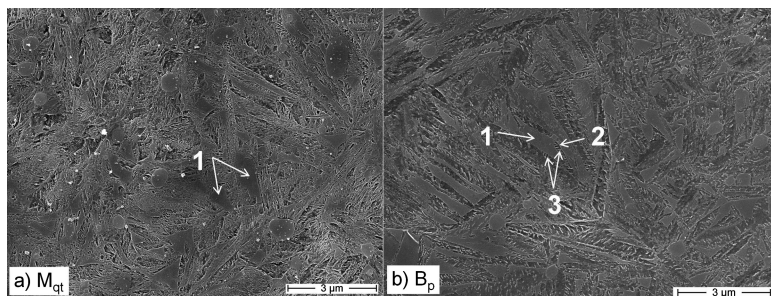


FIG. 8—Microstructures (FESEM) of the steel after (a) martensitic and (b) bainitic treatments.

nitic transformation temperature [20] but also by varying procedures of the heat treatment under given temperatures of austenitization and isothermal transformation. The two-step bainitic treatment could be used to shorten the duration of the isothermal treatment in the lower bainitic range and to reduce the amount of retained austenite to be negligible low. Other process variations, like pre-holding and pre-quenching as well as cyclic pre-austempering, only had a limited acceleration effect, whereas the post-quenching after partial bainite transformation up to 87 % had no effect on transformation of the residual austenite to martensite.

The retained austenite in bainitic microstructures, with an amount of up to 13 % after the two-step bainitic treatments or the shortened bainitic treatment by post-quenching, did not lead to loss in hardness compared to that of full bainite transformation. The fatigue strength of the steel with the two different heat treatment processes, namely, B_p and BB, was also enhanced significantly, even though the retained austenite was presented in the microstructures with an amount of about 13 % and 9 %, respectively. The two-step process achieved the best fatigue result.

In order to understand the results of the fatigue test the martensitic and bainitic microstructures were observed by microscopes as described above. There are essential differences between martensitic and bainitic microstructures considering their formation conditions as listed in Table 2. The retained austenite in bainitic microstructure could be more stable than the retained austenite in martensitic microstructure. This was observed by an additional experiment in which three cylinder specimens ($\varnothing 10 \times 25 \text{ mm}^2$) were treated by the three processes as given in Table 1, namely, M_{qt} , B , and B_p , and subsequently tempered for 100 h at 180°C . The length change was determined by the use of a micrometer gauge and was listed in Table 3, in which the amount of retained austenite prior to the temperature exposure was given. The large length increase of the specimen with martensitic treatment (M_{qt}) could be attributed to the transformation of retained austenite to martensite. The slight length increase of the specimen treated by shortened bainite treatment (B_p) implied the higher stability of retained austenite in bainitic microstructure.

It should be pointed out that earlier investigations describing the fatigue crack propagation (FCP) behavior of metastable austenite in steels [21,22]

TABLE 2—Differences between martensitic and bainitic microstructures.

Microstructure	Martensitic	Bainitic
Transformation temperature	Athermal	Isothermal
Diffusion	Diffusionless	Diffusion controlled
Chemical composition	Nearly homogeneous	Carbon redistributed
Residual austenite stabilized mainly by	Transformation stress	Carbon enrichment
Stability up to bainite transformation temperature	Martensite decomposed	Bainite unchanged
Stability of retained austenite	Impaired by martensite decomposition	Unchanged as long as bainite unchanged

TABLE 3—Length change of the steel after temperature exposure (180°C, 100 h).

Heat Treatment	M_{qt}	B_p	B
Retained austenite, %	14	13	<3
Length change, %	0.216	0.060	0.004

showed different results on the role of metastable austenite. Chanani et al. reported [21] that poorer FCP properties were found in the trip steel in which the austenite had the highest stability. In contrast the trip steel with less stable austenite showed better FCP properties. This was attributed to a beneficial energy-absorbing effect of the strain induced martensite transformation. In this work the microstructure was full austenitic, and this is not comparable with the case of the bainitic microstructure with retained austenite of less than 13 %. However, Liu et al. reported [22] a positive role of retained austenite in bainitic microstructures of a Si-Mn steel with respect to FCP properties. This work shows that the FCP threshold of the steel increased with an increase in the volume fraction of carbon-saturated austenite. The behavior of crack growth indicated that the deformation strengthening ability of the austenite had a significant beneficial effect on the FCP in the threshold region. This is in good agreement with the results of the present work that the very stable retained austenite had a positive influence on the resistance against the formation and growth of fatigue cracks. A relaxation of existing concentrated stresses can be obtained by local plastic deformation of the retained austenite in the microstructures. This local plastic deformation causes not only strengthening of the retained austenite but also crack stopping effects [23]. As shown by the experiments, the retained austenite did not transform at ambient temperature by the cyclic loading. Therefore dimensional changes exceeding manufacturing tolerances are not expected under conventional operating conditions.

The experimental work had demonstrated that the bainitic transformation with post-quenching and the two-step bainitic treatment (B_p and BB) could lead to a significant shortening of the heat treatment duration. If the duration of the conventional full bainitic transformation (B) was normalized to 100 %, the corresponding duration of the two short-term bainitic treatments could be reduced to 25 % (Fig. 9), which allows cost savings in production.

Conclusion and Outlook

From the experimental work it was demonstrated that the short time bainitic transformation with post-quenching and the two-step bainitic treatment (B_p and BB) could lead to a significant shortening of the heat treatment duration while maintaining the required hardness for steels and avoiding dimensional changes by later retained austenite decomposition. The duration of the shortened heat treatment in the lower bainitic range could be reduced to about 25 % of that of a full bainitic transformation. The fatigue strength of the steel was enhanced significantly through the shortened treatments particularly by means of a two-step bainitic treatment. The retained austenite with an amount up to 13 % within the bainitic microstructure did not impair the fatigue strength. It

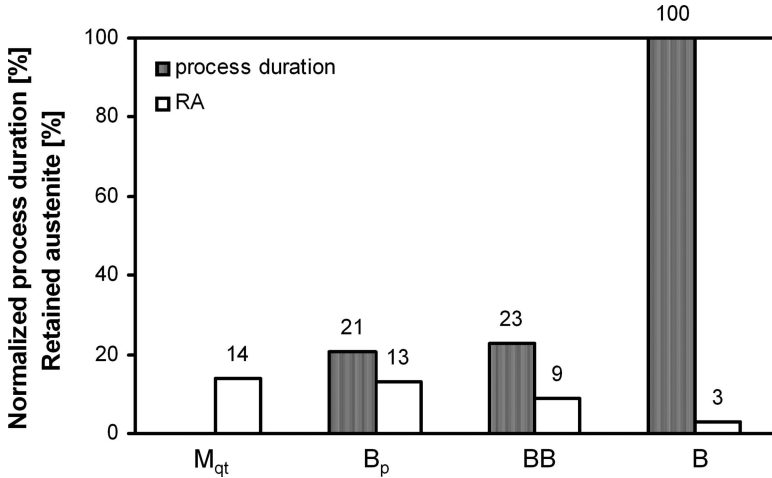


FIG. 9—Duration of the heat treatments at the lower bainitic range and retained austenite in the microstructures.

could be attributed to the high stability of the retained austenite, which was continuously enriched with carbon during bainite transformation and maintained unchanged under the cyclic loading at ambient temperature. The retained austenite in bainitic microstructure could be distinguished from that in martensitic microstructure.

It should be noticed that the rotating-bar test has different stress conditions compared to those of a rolling bearing, so the results of the present work could not directly be transferred to rolling bearings. Further investigations will compare the bearing life of full bainitic and two-step bainitic treated specimens; the latter will contain a certain amount of retained austenite. The question will be answered as to which role the special retained austenite plays in the bainitic microstructure under elasto-hydrodynamic and contaminated lubrication conditions.

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Case Depth and Static Capacity of Surface Induction-Hardened Rings

ABSTRACT: Large-size slewing bearings are usually surface hardened by means of induction heating. The load-carrying capacity of the bearing is dependent on, among others, the depth of the hardened layer, i.e., case depth (CD). It is of crucial importance for bearing manufacturers to ensure that sufficient CD is produced to meet the required bearing capacity for the applications. Compared to through-hardened bearings, the calculation method for the load-carrying capacity of surface-hardened bearings, especially the induction-hardened bearings, is not well established. This paper reports on a new calculation method for the static capacity of induction-hardened rings. The method is based on consideration of both the plastic indentation on the raceway and the damage tolerance in the subsurface region. The models for evaluating plastic indentation and subsurface damage have been validated with the standing contact fatigue testing.

KEYWORDS: surface induction hardening, case depth, static capacity, slewing bearings, plastic indentation, core crush, cracks, defect tolerance

Introduction

Large-size slewing bearings are usually surface hardened by means of induction heating. The load-carrying capacity of the bearing is dependent on, among others, the depth of the hardened layer, i.e., case depth (CD). It is of crucial importance for bearing manufacturers to ensure that sufficient CD is produced

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to meet the required bearing capacity for the applications. However, a hardened layer deeper than required should be avoided not only because of the cost involved in induction hardening, but also because of the fact that producing a too deep case increases the risk of surface cracking during induction hardening. In order to design and dimension the bearing that fits to the application, the bearing designers need to be able to calculate correctly both the static capacity and dynamic capacity.

For through-hardened bearings, the calculations for static capacity and dynamic capacity have been well established and accepted in the ISO. The static capacity was referred to as the static load applied to a non-rotating bearing that will result in a permanent raceway indentation of $1 \times 10^{-4} D_w$ (with D_w being the rolling element diameter) at the weaker of the inner or outer raceway contacts occurring at the position of the maximum loaded rolling element [1]. Later, the maximum contact pressures of 4000 MPa for line contact and 4200 MPa for point contact were introduced in the ISO [2] for the calculation of the static capacity of rolling bearings.

Compared to through-hardened bearings, the calculation method for the load-carrying capacity of surface-hardened bearings, especially the induction-hardened bearings, is not well established. This is due to the complexity that the capacity of a surface-hardened bearing is also dependent on the CD and the strength of the core material. Insufficient CD may result in the so-called core crushing, a severe failure in form of cracking and flaking of the hardened layer due to excessive plastic flow in the core. The present study focuses on the static capacity of induction-hardened bearings.

There have been some published criteria [3,4] for determining the static capacity of surface induction-hardened rings. The major difference among the existing criteria lies in whether or not allowing plasticity in the core of the bearing. Sague and Rumbarger [3] postulated that in order to avoid core crushing, the maximum shear stress at the case-core interface, arising from a static load, should be lower than the shear yield strength, which was assumed to be 0.425 times the ultimate tensile strength (UTS) of the core material. Zwirlein and Wieland [4], however, stated that the equivalent von Mises stress at the case-core interface, resulting from a static load, could be as high as 1.25 UTS for ball bearings and 1.0 UTS for roller bearings. It was further stated in Ref 4 that a raceway plastic indentation of $5 \times 10^{-4} D_w$ caused by a static load had no negative effect on the fatigue strength of the hardened case under the cycling speed usual in slewing bearing applications.

Allowing no plasticity in the core [3] eliminates the risk of the core-crushing failure. However, it normally requires a thick case layer with respect to the rolling element diameter, which is costly to produce and may even be beyond the capacity of the induction heating facility if the bearing is large and exceeds a certain size. The rules allowing plasticity in the core, such as those published in Ref 4, were mainly based on experience, which might be correct for a certain range of bearings with specific material and heat treatment. Their validity to other ranges of induction-hardened bearings remains questionable. It is thus of technical significance to understand and, desirably, to be able to determine the limit of plasticity caused by a static load, which will not endanger the raceway integrity and the fatigue strength of the bearings. Such a limit

must be related to bearing geometry, CD, and the mechanical properties of the core material.

Summarizing the aforementioned arguments, we can conclude that in order to determine the static capacity of a surface-hardened bearing, two aspects resulting from the applied static loading have to be considered: The permanent raceway indentation and the subsurface damage. The former is to guarantee the smoothness of bearing motion, such as blade pitching in wind turbines, whereas the latter is to ensure the integrity of the bearing raceway or to avoid the core-crushing failure.

The objective of the present study is to develop a calculation method for the static capacity of induction-hardened bearings, which is based on consideration of both the limit for raceway permanent indentation and the tolerance for subsurface damage. The development of the static capacity model involves the following:

- The description of the plastic indentation and subsurface residual stress, each being related to bearing geometry, CD, and material properties;
- The damage tolerance for core to exclude the risk of core-crushing failure; and
- Model validation by experiment.

Finite Element Analysis

Elasto-plastic finite element analysis was performed to study the plastic deformation of an induction-hardened surface and subsurface stress, resulting from indentation of a ball or roller subjected to static loading. The general-purpose finite element (FE) package ABAQUS [5] is employed for this study. In the FE model (Fig. 1), the indenter (ball or roller) was modeled as a through-hardened component, and the ring was modeled as a layered body containing a hardened case layer of depth CD, a soft core, and a transition layer with a thickness of 0.1 CD between the case and the core, as shown in Fig. 2(a). The stress-strain curves were measured from compression tests [6] of several medium-carbon steels both in soft and hardened states. The soft specimens were machined from blocks cut from forged rings with hardness ranging from 190 Hv to 308 Hv. The hard specimens were martensitically heat treated to hardness of about 670 Hv, corresponding to the required hardness of the hardened cases in slewing bearings. The measured stress-strain curves for the soft and hardened materials were used for the case layer and the core in the FE model. The constitutive (elasto-plastic) behaviour of the transition layer is assumed to be a result of linear interpolation between the end of the case and the start of the core, as schematically shown by Fig. 2(b). The von Mises yield criterion and isotropic hardening were employed for the description of the elasto-plastic behaviour of the materials. Simulation was done for one loading followed by unloading.

Parametric study was made to calculate plastic indentation δ for two extreme contact cases: Circular point contact (CPC) and cylindrical line contact (CLC). The parameters considered include (i) load in terms of the maximum Hertzian pressure p_0 , (ii) rolling element diameter D_w , and (iii) CD.

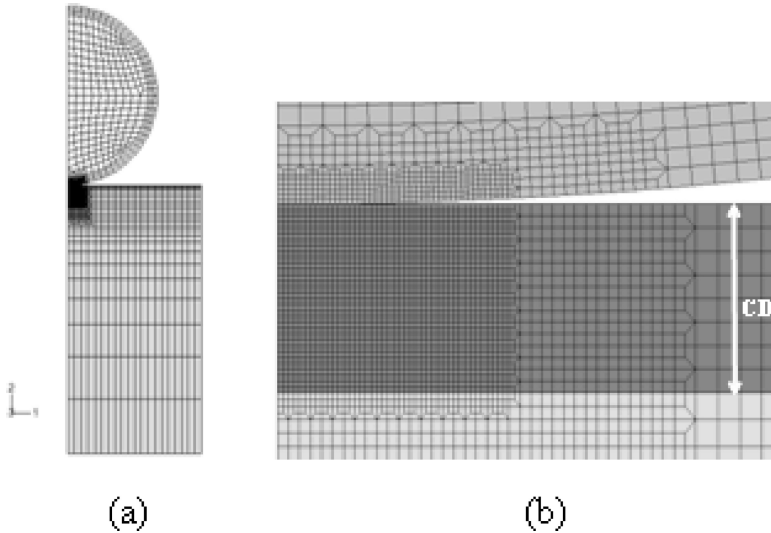
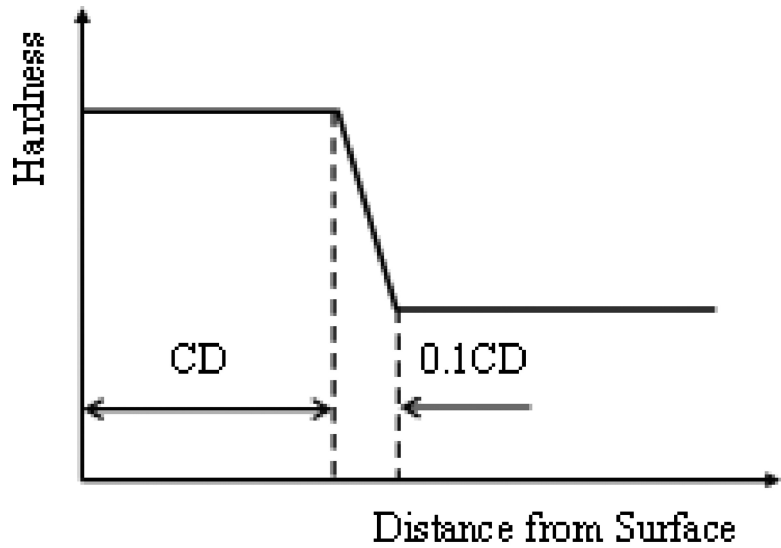


FIG. 1—FE model for indentation of an induction-hardened surface by a ball or roller. (a) Global model; (b) local mesh near the contact.

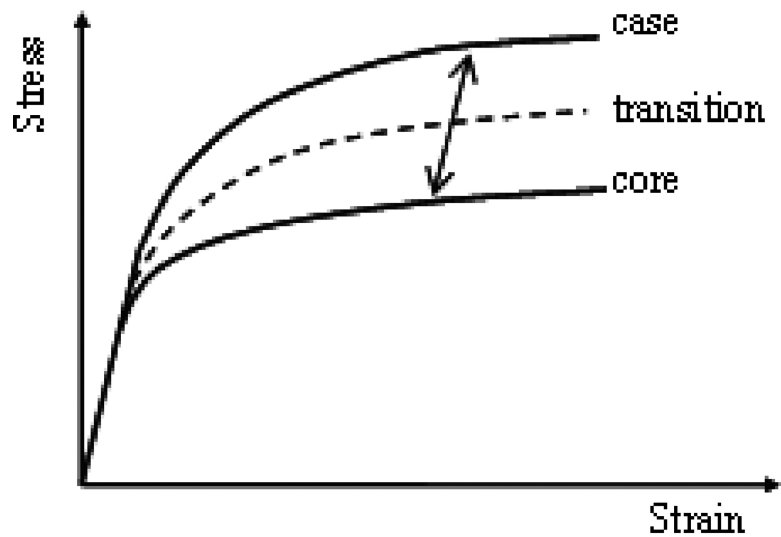
Figures 3 and 4 show the plastic indentation δ resulting from indenting a ball and roller, respectively, onto a flat surface induction hardened with different CDs. In the graphs both δ and CD are normalized by the indenter diameter D_w . Calculations were made also for different applied loads in terms of the maximum Hertzian pressure p_0 normalized by the yield strength of the core material σ_y . Obviously, plastic indentation decreases with an increase in CD. Under a specific load level, the plastic indentation ceases to decrease with CD if CD exceeds a certain value. In this situation, the stress in the core is below its yield strength, and plasticity occurs only in the hardened case, a situation equivalent to a through-hard component. It can also be seen that the “through hard” can be achieved at a lower load and less deep CD (Fig. 3) under CPC, compared to the CLC (Fig. 4). This is owing to the difference in subsurface stress distribution between the two contact conditions. The maximum von Mises stress in CPC is located at a shallower depth than that in the CLC.

In order to gain some insight into the core crushing in surface induction-hardened rings, the subsurface response in terms of plasticity and residual stress was investigated. If the stress resulting from a static load exceeds the yield strength of the core material, the core undergoes plastic flow. The plastic flow causes damage in the subsurface in the form of residual stress. Consider, for example, a situation of a shallow case ($CD/D_w = 2\%$) and applied contact pressure p_0 is 5.4 times the yield strength of the core material. It can be seen from Fig. 5 that if the applied load is high and/or the CD is shallow, a high tensile residual stress will be generated in the case-core transition region, which may cause cracking or delamination at the case and core interface.

Severe plasticity in the core also weakens the support of the core to the case layer and, as a result, the case will be subjected to severe bending by the load.



(a)



(b)

FIG. 2—Schematics of hardness profile of a surface induction-hardened component (a) and the stress-strain curves for the material in the hard case, soft core, and the transition zone between the case and the core (b).

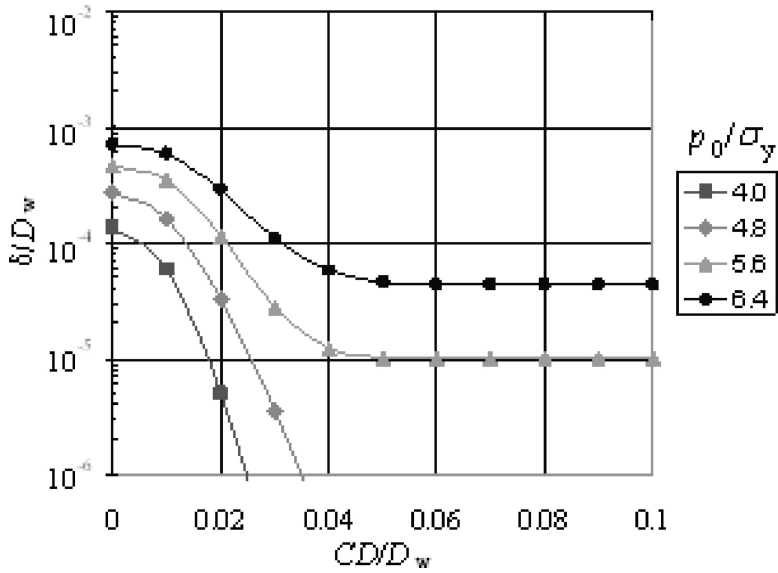


FIG. 3—FE results of surface plastic indentation (δ) as a function of CD in CPC.

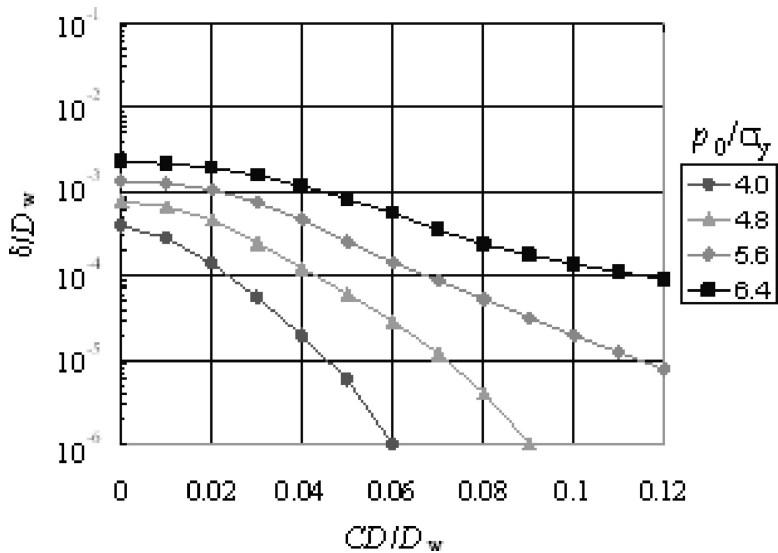


FIG. 4—FE results of surface plastic indentation (δ) as a function of CD in CLC.

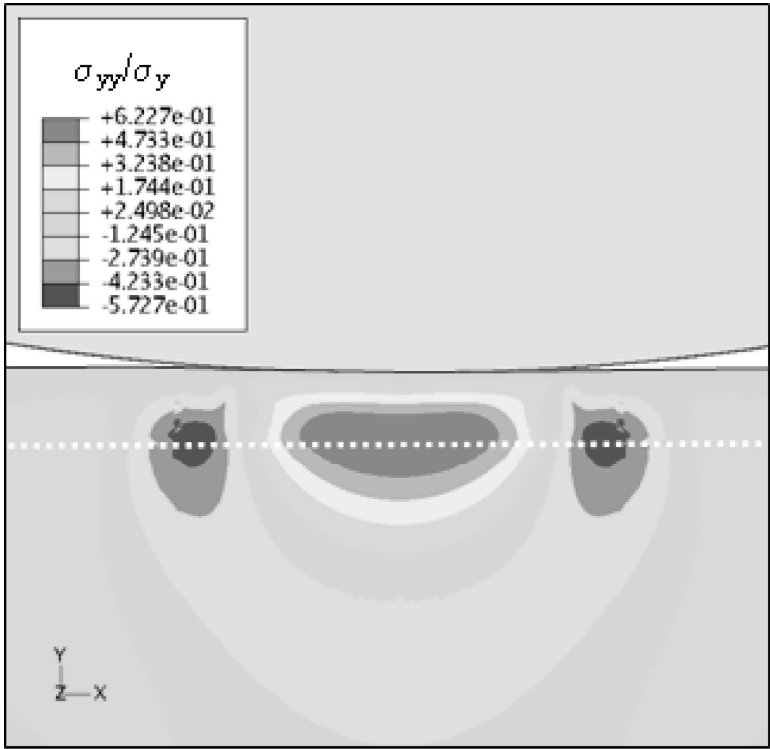


FIG. 5—Subsurface damage in form of plasticity-induced residual tensile stress (perpendicular to surface) due to a static load. $CD/D_w=0.02$, $p_0=5.4\sigma_y$.

The bending of the case may be significant if the case layer is shallow, as shown in Fig. 6. The bending of the case layer may lead to cracking of the case if the bending stress in the case is too high. Core crushing is actually a consequence of deterioration of the core due to plastic flow, which weakens the support to the case layer.

The tensile stress in the subsurface, resulting from a high load, is an indication of potential failure of core crushing, which must be accounted for in the static capacity model.

Experimental Study

Testing Methodology

The standing contact fatigue (SCF) testing was employed to experimentally study the relevant failure mechanisms and to generate data for validating the proposed static capacity model. The SCF testing involves cyclically indenting a flat specimen with a ball or roller (see Fig. 7(a)) in which the load applied on the indenter pulsates between P_{min} and P_{max} (Fig. 7(b)). A minimum load P_{min} was

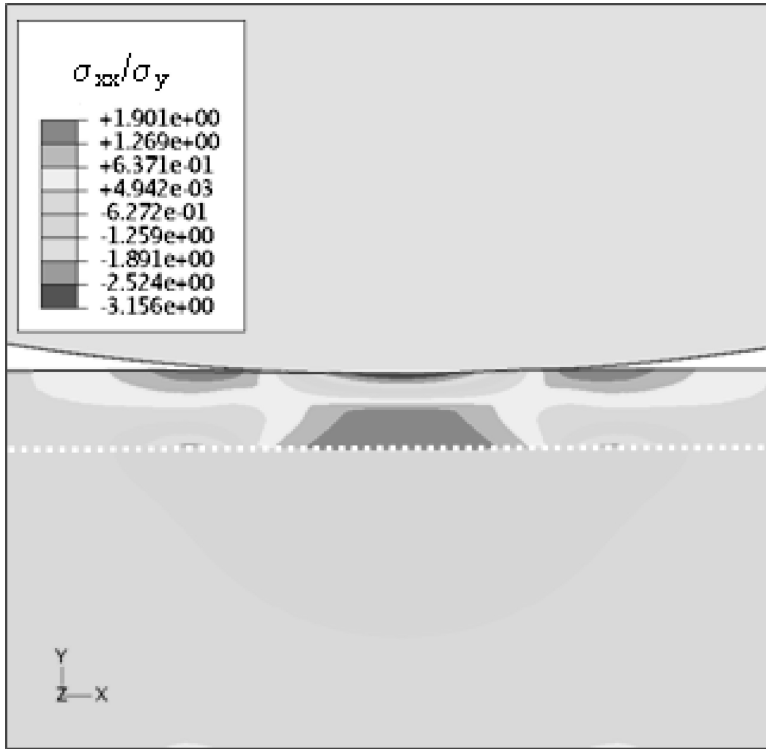


FIG. 6—Subsurface damage in the form of plasticity-induced residual tensile bending stress (parallel to the surface) due to a static load. $CD/D_w=0.02$, $p_0=5.4\sigma_y$.

chosen to keep the load applied on the same spot on the specimen. The methodology of the SCF was proposed and published by Alfredsson and Olsson [7] from the Royal Institute of Technology.

The reason for choosing the SCF testing was twofold. It may reproduce the damaging cracks relevant to the core-crushing failure of induction-hardened rings. Moreover, the standing contact load, i.e., standing-still roller or ball pressed cyclically to the specimen, is also a loading condition relevant for slewing bearings in some slewing bearing applications.

Test Specimens and Equipment

The material used for the testing specimen is 42CrMo4. The chemical composition of the material is given in Table 1.

The testing specimens were cut from a forged ring and machined to blocks with a dimension of $120 \times 130 \times 50 \text{ mm}^3$. The forged ring was tough tempered. The specimens were surface hardened by induction heating with two CDs, namely, a shallow CD of 0.5 mm and a deep CD of 1.1 mm. The surface hardness was around 670 Hv.

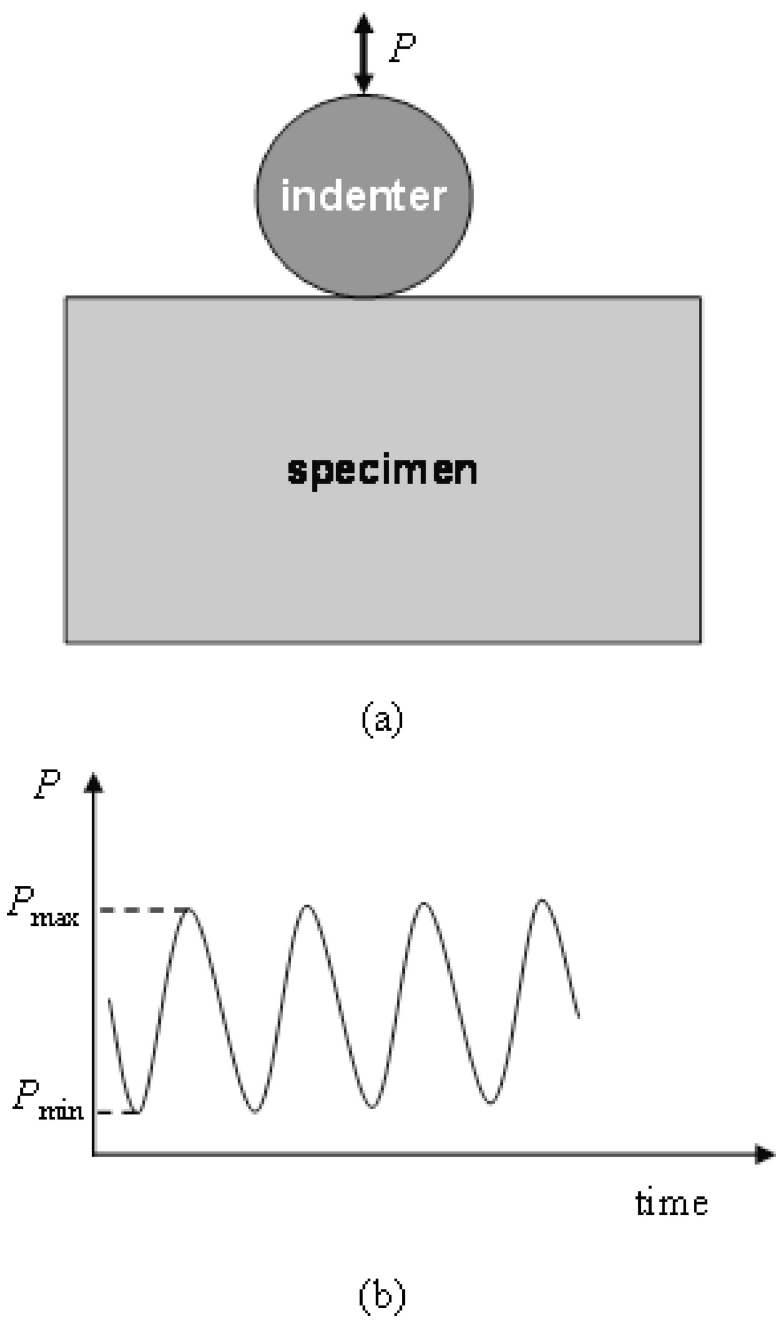


FIG. 7—Schematic of the SCF testing: (a) Test set-up; (b) load history.

TABLE 1—Chemical composition (wt %) of 42CrMo4.

C	Mn	P	Si	S	Cr	Ni	Mo	Cu	V	Al	Ti	Nb
0.404	0.83	0.01	0.25	0.006	1.02	0.19	0.22	0.21	0.003	0.033	0.0043	0.003

The indenter used for the SCF testing was a through-hard crowned roller of a diameter of 10 mm and a crowning radius of 98 mm. For each test, a new roller was used. The indentation of the roller onto the testing block forms an elliptical (point) contact with b/a ratio of 6.84 (b and a are the semi axes of the contact ellipse).

The tests were conducted on a servo-hydraulic testing machine, MTS 100 kN, equipped with digital controllers.

As shown in Fig. 7, the specimen was positioned and fixed to a support plate. A cyclic load varying as a sine function with time was applied on the indenter. The minimum load was constant at 0.05 kN in order to keep the roller at the same contact location.

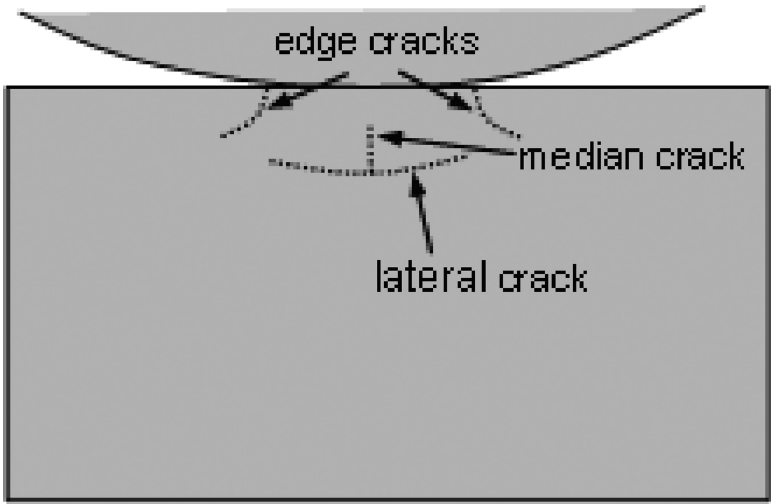
Results

In the SCF tests on both shallow- and deep-case specimens, three types of cracks were observed and identified as the lateral crack, the median crack, and the edge crack, as schematically shown in Fig. 8(a). The lateral crack developed at the case-core transition region; the edge occurred at the edges of the contact, whereas the median crack initiated from the upper surface of the lateral crack and grew vertically towards the surface. Figure 8(b) shows a fully developed lateral crack, whereas no edge and median cracks are formed. The location and the shape of the lateral crack shown in Fig. 8(b) correlate well with the plasticity-induced tensile residual stress calculated from FE analysis (see Fig. 5). The edge and the median cracks indicated in Fig. 8(a) seem also to coincide with the predicted damage zones shown in Fig. 6.

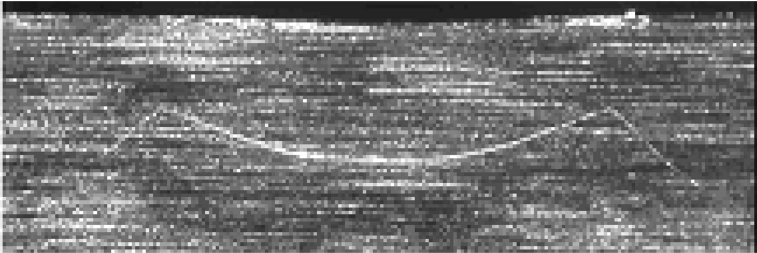
It was found in the present experiments that the required load to develop the edge cracks and the median cracks was substantially higher than the load to develop the lateral cracks. Furthermore, the edge cracks were formed later than the lateral cracks. The median cracks could only be formed if lateral cracks were present.

The phenomenon that the required load for generating the lateral cracks is lower than the required load for generating the edge cracks and the median, can be understood from the FE analysis. The stresses in the positions of the edge crack and the median crack, as shown in Fig. 6, vary slightly around a high residual (static) stress, when an alternating load is applied. In other words, in those regions, the stress amplitude is pretty low, but the mean stress is high. In the case-core transition region where the lateral crack develops, the stress varies from compression to tension, thus with a big range/amplitude upon application of alternating loading. Therefore, the load required to trigger the lateral crack is lower than that to initiate the edge and median cracks.

In view of the fact that the lateral crack can be generated at a lower load than the other two types of cracks, it is reasonable to consider the load for the



(a)



(b)

FIG. 8—Indication of three types of cracks observed in the specimens in the SCF testing: The lateral crack, the edge cracks, and the median crack (a) and a picture of a well developed lateral crack at the case-core transition zone of the specimen (b).

formation of the lateral cracks as the fatigue load limit. For the shallow-case specimen the fatigue load limit is 7 kN, corresponding to a nominal Hertzian contact pressure of 3.98 GPa, while for the deep-case specimen, the fatigue load limit is 18 kN, corresponding to a nominal Hertzian contact pressure of 5.46 GPa.

Residual surface deformation after the SCF testing, as well as the permanent plastic indentation from a single load, was measured using the Talysurf equipment.

Formulation of Surface Indentation and Subsurface Damage

Surface Permanent Indentation

The permanent indentation of the raceway due to static loading is an important aspect to be considered for the static capacity of bearings. The evaluation of the plastic indentation, however, is not a trivial task. It relies on accurate measuring equipment and/or a theoretical model that accounts for the subsurface stress resulting from applied load and the resilience of the material to plastic straining. Calculation of plastic indentation is even more difficult for surface-hardened bearings in that the depth of the case layer as well as the material properties of both core and case can influence the magnitude of the surface indentation.

An early published model for estimating plastic indentation in bearings was due to Palmgren [9]. Based on empirical data for bearing quality steel through hardened between 63.5 and 65.5 Rockwell, such as those in Ref 8, Palmgren [9] developed the following formula to describe the total permanent indentation δ_t (for both contact bodies) for point contact:

$$\delta_t = 1.3 \times 10^{-7} \frac{P^2}{D_w} (\rho_{I1} + \rho_{II1})(\rho_{I2} + \rho_{II2}) \quad (1)$$

where:

P = applied load,

D_w = diameter of indenter (body I),

ρ_{I1} = curvature of body I, and so on.

The units in Eq 1 are newtons and millimetres.

It should be pointed out that Palmgren's data [9] were based on indentation tests carried out in the 1940s, and the accuracy of the data was dependent on the measurement devices available then [1]. An obvious drawback of Eq 1 is that the calculated plastic indentation starts from any load level no matter how low the load is. This is physically incorrect, as plastic deformation occurs only if the stress exceeds the yield limit of the material.

More test data of plastic indentation, such as those published in Ref 10, were discussed in the review paper by Yhland [11]. It was indicated in Ref 11 that the plastic indentation data from various sources exhibited conformance at high loads (say, above 4000 MPa of contact pressure) or large values of plastic indentation but deviated from each other at relatively low load. This might be attributed to the accuracy of measuring devices, which was not so reliable if the magnitude of the plastic indentation was small.

With improvement of measuring devices, the accuracy of simple formulas like Eq 1 for estimating plastic indentation, in through-hardened bearings may be modified based on reliable indentation testing data. For induction-hardened bearings, however, a simple description of plastic indentation is not available. Establishment of such a model based on indentation testing data would be very difficult if not impossible. This is because the plastic deformation in induction-hardened bearings depends on the strength of both case and core material and also on CD.

In this section we describe how to formulate the plastic indentation of induction-hardened rings based on the results of FE analysis.

Let us first consider the plastic indentation between two contact bodies made of homogenous materials. By homogenous we mean the hardness and microstructure of the steel are uniform through depth, such as a through-hardened ring or a non-hardened ring. The plastic indentation δ due to a static CPC can be expressed as

$$\frac{\delta}{D_w} = k_{\text{CPC}} \left\langle \frac{0.62p_0}{\sigma_y} - 1 \right\rangle^2 \quad (2)$$

where:

D_w = diameter of the rolling element,

σ_y = (compressive) yield stress of the material, and

p_0 = maximum (Hertzian) contact pressure.

In the above equation, the McCauley bracket notation $\langle \dots \rangle$ is used, i.e., the term in the bracket is set zero if the quantity enclosed is negative. The coefficient 0.62 in Eq 2 is due to the relationship between the contact pressure p_0 and the maximum von Mises stress $\sigma_{e,\text{max}}$ in the subsurface for CPC, i.e., $0.62p_0 = \sigma_{e,\text{max}}$. The coefficient k_{CPC} in Eq 2 is a constant dependent on the yield strength or the hardness of the material and can be determined by fitting the equation to FE calculation data.

The expression in the McCauley bracket means that if the maximum von Mises stress resulting from the applied load exceeds the yield strength of the material, the subsurface will yield and undergo plastic flow, leading to a plastic indentation δ on the surface; otherwise, the subsurface remains in the elastic regime, and no plastic indentation will be generated by the load. Furthermore, Eq 2 indicates that under the same load (p_0), the steel with higher strength (σ_y) will have a smaller plastic indentation.

In a similar manner, the plastic indentation δ due to CLC can be described by

$$\frac{\delta}{D_w} = k_{\text{CLC}} \left\langle \frac{0.56p_0}{\sigma_y} - 1 \right\rangle^2 \quad (3)$$

in which the coefficient 0.56 is due to the relationship between the between the maximum contact pressure p_0 and the maximum von Mises stress in the subsurface for CLC, i.e., $0.56p_0 = \sigma_{e,\text{max}}$. The coefficient k_{CLC} in Eq 3 is a constant dependent on the yield strength or the hardness of the material and can be determined by fitting the equation to FE calculation data.

If the surface is induction hardened, the plastic indentation δ is a result of plastic deformations from both the case and the core. The partition between the two contributions is dependent on CD. Considering this, we propose the following partitioning rule:

$$\frac{\delta}{D_w} = \rho \left(\frac{\delta}{D_w} \right)_{\text{core}} + (1 - \rho) \left(\frac{\delta}{D_w} \right)_{\text{case}} \quad (4)$$

where:

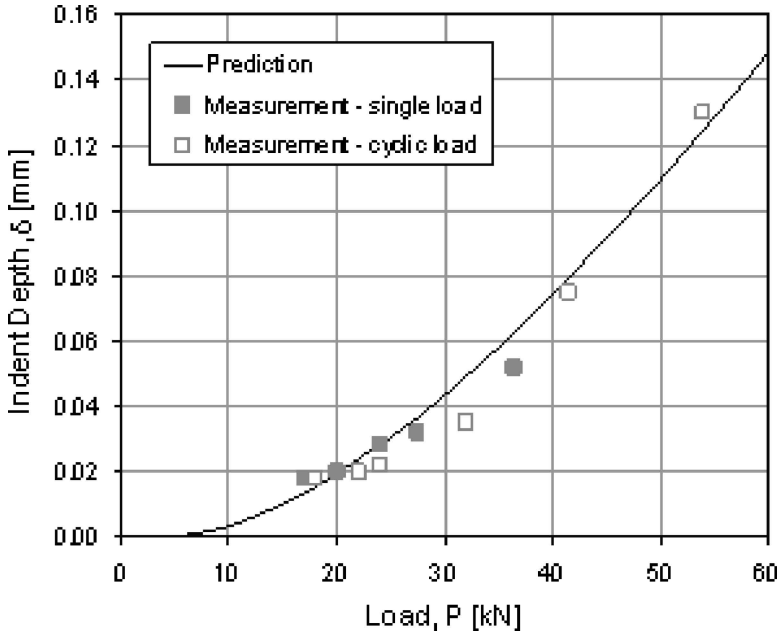


FIG. 9—Comparison between the predicted and measured depths of the surface residual indentation on the SCF testing specimens ($CD=1.1$ mm).

$(\delta/D_w)_{\text{case}}$ and $(\delta/D_w)_{\text{core}}$ =plastic indentation for case and core materials, respectively, which can be valuated using Eq 2 or Eq 3.

The partition parameter ρ in Eq 4 is a function of CD and contact pressure p_0 . Such a relationship can be represented by

$$\rho = \exp \left[-C \left(\frac{CD}{D_w} \right)^m \left(\frac{p_0}{p_{\text{ref}}} \right)^n \right] \quad (5)$$

where:

p_{ref} =reference pressure set to 1 GPa.

The constants C , m , and n can be determined by fitting Eq 4 together with Eq 5 to the plastic indentation data obtained from FE calculations.

So far we have considered two extreme contact situations, i.e., the CPC and CLC. For ball bearings in which the contact region is of elliptical shape with semi axes a and b , the plastic indentation can be approximated through a linear interpolation between the two extreme cases

$$\frac{\delta}{D_w} = \left(\frac{\delta}{D_w} \right)_{\text{CPC}} \left(\frac{a}{b} \right) + \left(\frac{\delta}{D_w} \right)_{\text{CLC}} \left(1 - \frac{a}{b} \right) \quad (6)$$

in which the subscript CPC means CPC with $b/a=1$ and CLC means CLC with $b/a=0$.

Using Eqs 2–6, we can also calculate the residual surface indentation of the SCF specimens described in the previous section. Figure 9 shows the compari-

son between the calculations based on Eqs 2–6 and the measurements in which the data from both single load and from cyclic load are included. It seems that the residual surface profile was controlled by the load in the very first cycle regardless of the value of the subsequent cyclic loading. A good correlation can be seen from Fig. 9 between the predicted depth of residual surface indentation and that measured from the SCF testing.

Subsurface Damage

It has been demonstrated by the SCF testing that severe standing contact loading may cause subsurface delamination at the case-core interface and surface cracking at the edge of the contact. These damages are believed to be representative of the core-crushing failure in induction-hardened rings. The SCT test also showed that for induction-hardened specimens, the fatigue limit was dictated by the subsurface damage in the form of lateral cracks. Therefore, we concentrate on the description of the subsurface damage resulting from static loading and the tolerance for such damage.

Figure 5 shows the tensile residual stress induced by the plastic damage in the case-core transition. The tensile residual is responsible for the lateral crack formation and development, and is thus representative of the subsurface damage caused by the static loading.

The tensile residual stress S in the case-core transition that is responsible for the formation of the lateral cracks can be expressed as

$$S = C_1 \sigma_y \left[\tan^{-1} \left(C_2 \left\langle \frac{\sigma_e}{\sigma_y} - 1 \right\rangle - C_3 \right) + \tan^{-1}(C_3) \right] \quad (7)$$

where:

σ_y = yield strength of the core material,

σ_e = equivalent von Mises stress at the case-core interface, and

C_1 , C_2 , and C_3 = constants, which can be determined by fitting Eq 7 to the FE results.

In the above equation the Macauley bracket notation is used, i.e., the term $\langle \dots \rangle$ is set to zero if the quantity enclosed is negative. Figure 10 shows the correlation between tensile residual stress S and the equivalent von Mises stress σ_e . It should be noted that σ_e is corresponding to the linear elasticity solution, which is a measure of the severity of the subsurface stressing due to the applied load. Moreover, the use of the von Mises stress allows accounting for the superimposed hoop and internal stresses.

As the weakest link in the material is the pre-existing defects such as inclusions and pores, cracks will be initiated first from the defects. If the defect size is $2c$, the critical stress S_c at the location of the defect can be determined by considering the fatigue threshold condition for a penny-shaped crack of the same size, i.e.,

$$S_c = \frac{\Delta K_{th}}{2} \sqrt{\frac{\pi}{c + c_0}} \quad (8)$$

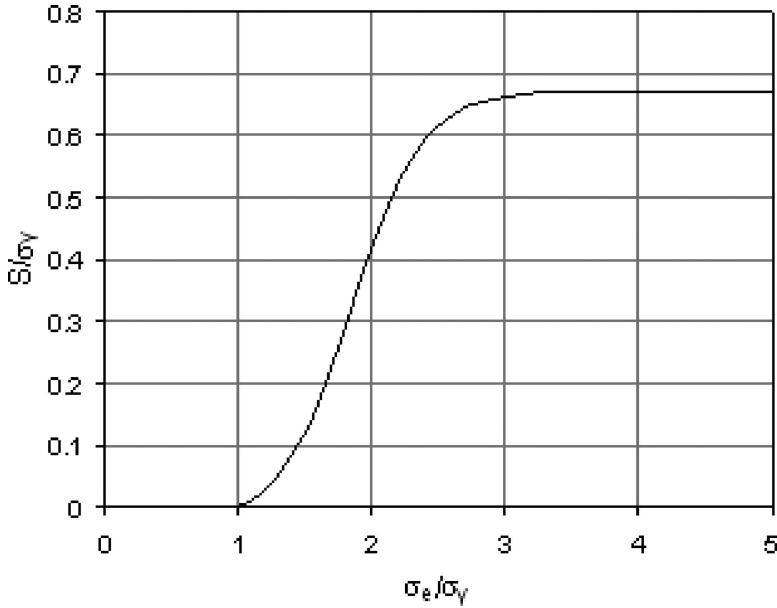


FIG. 10—Correlation of the residual stress responsible for case-core delamination with the equivalent von Mises stress at the case-core transition. The residual stress here is determined from FE analysis of CPC.

where:

ΔK_{th} = fatigue threshold of the core material and
 c_0 is determined from

$$c_0 = \frac{\pi}{4} \left(\frac{\Delta K_{th}}{\sigma_w} \right)^2 \quad (9)$$

where:

σ_w = fatigue (endurance) limit of the core material.

Derivation of Eqs 8 and 9 is based on the stress intensity factor solution for a penny-shaped crack [12] and the El Haddad parameter for small cracks [13].

Equation 8 defines the damage tolerance in terms of the tensile residual stress induced by the plasticity at the case-core interface. This tolerance is dependent on the size of the pre-existing defects in the damage zone.

The SCF testing was also employed to verify the damage tolerance model. The fatigue load limit is calculated using Eqs 7–9, below which no (subsurface) lateral crack will be formed. In the calculations different defect sizes are assumed, namely, 0, 60, and 150 μm .

The predicted fatigue load limits are compared with those determined from the SCF testing in Fig. 11. As can be seen from Fig. 11, the model prediction would stay on the conservative side if a larger inclusion was considered in the

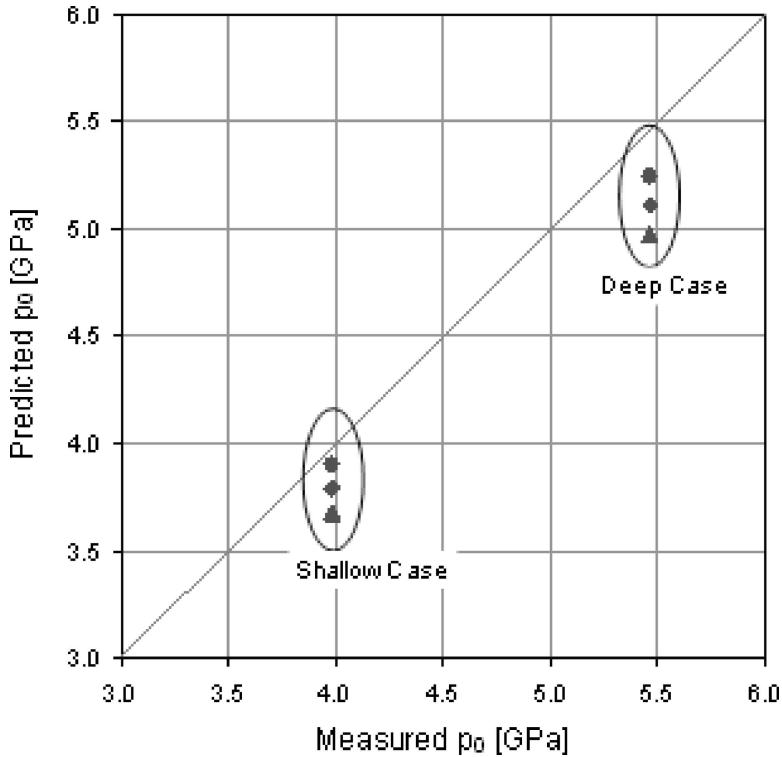


FIG. 11—Comparison between the predicted and measured standing contact load limit for subsurface lateral crack initiation for the shallow-case specimen ($CD=0.5$ mm) and the deep-case specimen ($CD=1.1$ mm) in the SCF testing.

calculation. It can be concluded that the present model can reasonably predict the occurrence of subsurface delamination caused by the standing contact loading.

Case Depth and Static Capacity

Using the models described in the previous sections, it is possible to evaluate the plastic indentation and subsurface damage caused by applied static load. Based on consideration of both plastic indentation on raceway and subsurface damage at the case-core interface, a new model is proposed for calculating the static capacity of induction-hardened rings.

Let us first demonstrate the dependence of the allowable loads based on consideration of surface indentation and subsurface damage tolerance, respectively, on CD and material strengths. A four-point contact bearing with the following geometry is chosen for the calculations:

- Pitch diameter $d_m=615$ mm;
- Ball diameter $D_w=34.925$ mm;

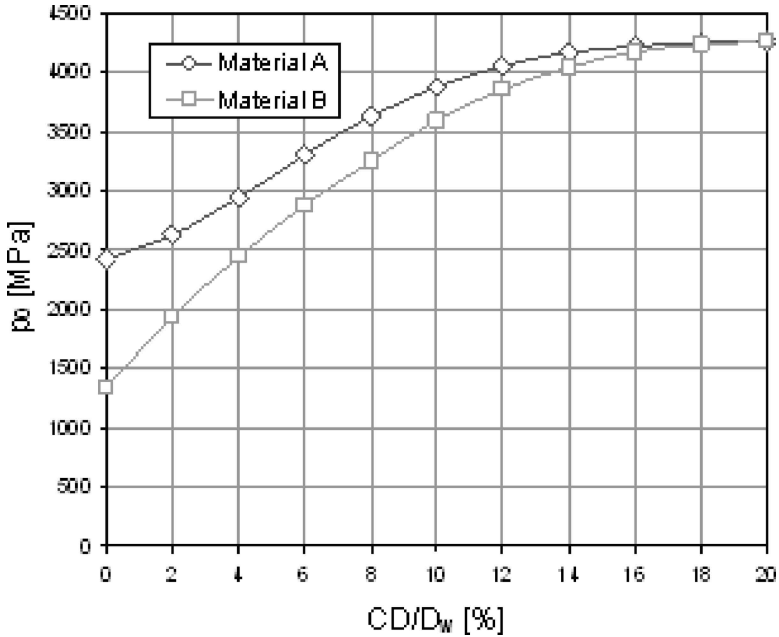


FIG. 12—Calculated static load in terms of contact pressure p_0 corresponding to permanent surface indentation of $10^{-4}D_w$ surface induction-hardened rings made of two different materials with different CDs.

- Osculation $f=0.52$; and
- Contact angle $\alpha=45^\circ$.

Furthermore, two core materials are considered: Material A with yield strength of 740 MPa and material B with yield strength of 330 MPa. It is assumed that the surface of the bearing is induction hardened with the case material strength of 1900 MPa, which is a typical value for hardened bearing steels.

Calculations have been made of the maximum contact pressure p_0 for generating a plastic indentation of $10^{-4}D_w$ for different CDs. It can be seen from Fig. 12 that the contact pressure increases with CD and approaches a constant level at large CD. Independence of the allowable contact pressure of the CD is an indication of equivalence to the through-hard situation. As can be seen from Fig. 12, a CD larger than $0.18D_w$ is almost equivalent to the through-hard situation for both materials. The present model predicts that contact pressure for causing plastic indentation of $10^{-4}D_w$ in the through-hard raceway is 4260 MPa, which is conformable with the ISO static capacity of 4200 MPa for ball bearings. In view of this, as well as the good validation results, we can conclude the present model gives a reliable estimation of plastic indentation for induction-hardened rings.

Also obvious in Fig. 12 is that the plastic indentation is also dependent on the material chosen for the bearing. The effect of strength difference between

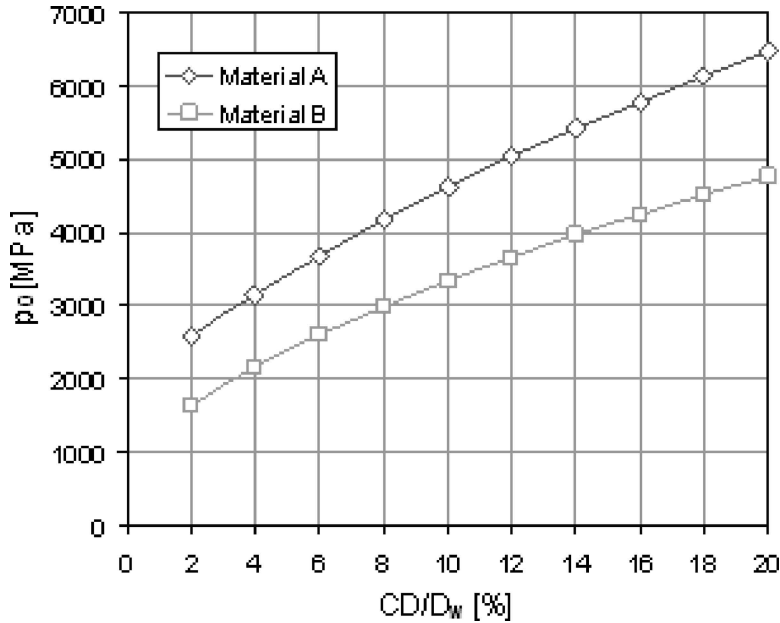


FIG. 13—Calculated allowable static load in terms of contact pressure p_0 that will not subsurface crack for surface induction-hardened rings made of two different materials with different CD's.

the two materials is quite significant for a CD less than $0.12D_w$: A much lower contact pressure is needed for material B than that for material A to generate the same amount of plastic indentation. The difference between the two materials decreases with CD. This is because the plastic indentation on the raceway is due to the plastic deformations from both the core and the case. With increasing CD, the contribution from the core decreases, whereas the contribution from the case increases.

With the model for description of subsurface damage, one can estimate the load limit below which the risk of core crushing due to standing contact load can be eliminated. Figure 13 shows the calculated standing contact load that may trigger subsurface cracking as a function of CD. This standing contact load limit is referred to as allowable load hereafter. Obviously, the allowable load increases with CD. The allowable load is also dependent on the cleanliness (defects or inclusions size) of the steel. In the calculations here, a defect size of $150\text{ }\mu\text{m}$ is assumed.

Figure 13 shows clear differentiation of material A from material B in terms of the allowable load. The difference displayed in Fig. 13 implies that the bearing made of a higher-strength core material (e.g., material A) can endure a higher static load than the bearing made of a lower-strength core material (e.g., material B).

The above calculation examples demonstrate that definition of static capacity of induction-hardened bearings has to consider both damage tolerance for

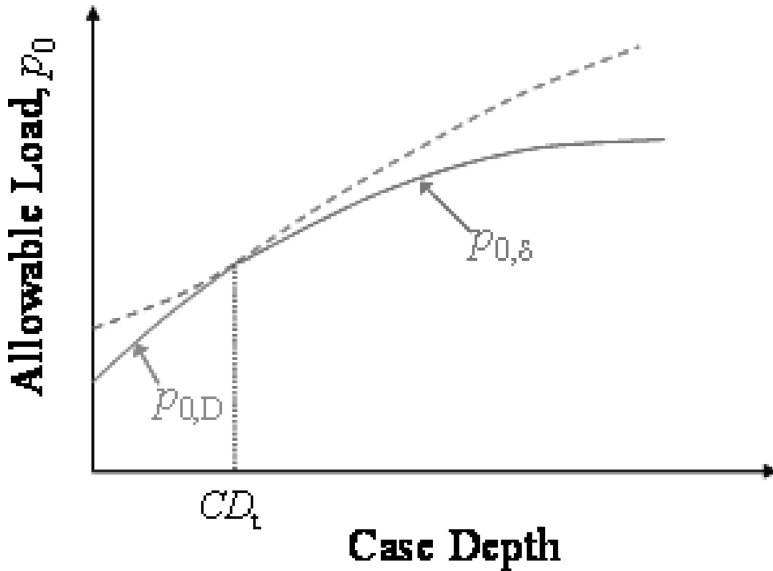


FIG. 14—Schematic definition of static capacity of an induction-hardened bearing based on consideration of both damage tolerance for subsurface and plastic indentations for raceway. The solid lines represent the allowable bearing load.

subsurface and plastic indentation for raceway. Evaluation of the static capacity based on specified plastic indentation may involve the risk of core crushing or case-core delamination if the CD is shallow or the specified plastic indentation is large. Calculation based only on the damage tolerance, on the other hand, may ignore the requirement for smooth operation of bearing since too deep surface indents would result from the calculated load for the case of relatively large CD. In order to determine the static capacity of an induction-hardened bearing, one has to calculate the allowable load for the specified plastic indentation, $p_{0,\delta}$, and the allowable load for the damage tolerance, $p_{0,D}$. The lower of the two calculated loads should be used as the allowable static load that the bearing can endure. Such a definition is illustrated schematically in Fig. 14. The two curves representing $p_{0,\delta}$ and $p_{0,D}$, respectively, intersect at $CD = CD_t$. For calculation of allowable load for CD less than CD_t , the $p_{0,\delta}$ curve prevails, whereas the $p_{0,D}$ curve should be taken as the allowable load for CD larger than CD_t . The solid lines in Fig. 13 represent the allowable load for the bearing, which meets both the requirements for limiting the plastic indentation within specification and for the subsurface damage tolerance to avoid core crushing. With these two curves, we can also calculate the required CD for the bearing to carry the specified static load. Suppose that the CDs corresponding to the $p_{0,\delta}$ curve and the $p_{0,D}$ curve are CD_δ and CD_D , respectively. The smaller between CD_δ and CD_D , is the required CD for the bearing.

It is now possible to evaluate the static capacity of induction-hardened rings based on specified plastic indentation and subsurface damage tolerance.

For material A, for example, the CD_t value is $0.02D_w$ (as can be seen from Figs. 12 and 13). This means that for a bearing made of material A, the allowable load based on plastic indentation ($1 \times 10^{-4}D_w$ in this case) should be used for the static capacity for the bearing with $CD > 0.02D_w$; for $CD < 0.02D_w$, however, the static capacity should be calculated from the damage tolerance. As can be seen from Figs. 12 and 13, for material B, the CD_t value is $0.14D_w$, meaning that for bearing made of material B, the static capacity should be calculated based on the damage tolerance for CD up to $0.14D_w$; for $CD > 0.14D_w$, the static capacity should be calculated from the specification for surface indentation.

It becomes clear that for induction-hardened rings, it is not necessarily required that the stress in the core, arising from static load, should stay below the yield strength of the material. Certain plasticity in the core is permitted as long as the amount of plasticity is below a certain limit. The corresponding load corresponding to the allowable amount of the plasticity in the core can be evaluated using the models developed for plastic indentation and subsurface damage tolerance. Based on the findings from the present study, a new way of calculating the static capacity of induction-hardened bearings is proposed, which involves the following steps:

- (1) Calculate the load (in term of contact stress) corresponding to plastic indentation of $1 \times 10^{-4}D_w$;
- (2) Calculate the load corresponding to the subsurface damage tolerance; and
- (3) The lower of the two calculated loads is the static capacity of the bearing.

The reason for choosing plastic indentation of $1 \times 10^{-4}D_w$ for calculation of static capacity is based on the consideration that plastic indentation of $1 \times 10^{-4}D_w$ is consistent with the definition of static capacity for through-hardened bearings. As can be seen from Fig. 12, the allowable load for plastic indentation of $1 \times 10^{-4}D_w$ approaches 4200 MPa, i.e., the ISO definition for static capacity of through-hardened ball bearings, with increasing CD .

The model can also be used to calculate the required CD for the surface induction-hardened ring to carry the specified static load. The model for calculating the static capacity and CD is implemented in an in-house developed software tool Slewind.

Concluding Remarks

When determining the static capacity of a surface-hardened bearing, two aspects resulting from the applied static loading have to be considered: The permanent raceway indentation and the subsurface damage. The former is to guarantee the smoothness of bearing motion, whereas the latter is to ensure the integrity of the bearing raceway or to avoid the core-crushing failure.

FE analysis of induction-hardened rings indicated that damage resulting from severe static loads might lead to delamination between case and core, and surface cracking. The SCF testing confirmed the occurrence of these cracks and

showed that the lateral crack at the case-core interface occurred earlier and at lower load than the edge cracks from the surface.

Based on FE results, two models have been developed for evaluating the plastic indentation and subsurface damage in induction-hardened bearings. These models relate the allowable load to CD, mechanical properties for case and core materials. The models are validated with SCF testing, and good correlation has been found between the model predictions and the measurement.

A new calculation method is proposed for the static capacity of induction-hardened rings. It is based on considerations for plastic indentation of $1 \times 10^{-4} D_w$ and subsurface damage tolerance. The damage tolerance, which is dependent on material fatigue strength and steel cleanliness, is introduced in order to eliminate the risk for the core-crushing failure.

The present paper is concerned with the static capacity of surface induction-hardened rings. It should be stressed the static capacity model is meant for evaluation of the bearing for carrying static load. By static load, we mean either a constant or an alternating load applied on a non-rotating bearing. Any load applied to a rotating bearing, no matter how short the duration is, must be treated as dynamic load. The required CD for the induction-hardened rings is also dependent on dynamic capacity, which remains to be investigated in the future.

Acknowledgments

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MICROSTRUCTURE BEHAVIOUR IN ROLLING CONTACT

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and Nelson H. Forster²

Stress Field Evolution in a Ball Bearing Raceway Fatigue Spall

ABSTRACT: The governing mechanisms of fatigue spall propagation in ball bearing inner raceways are investigated through the use of elastic-plastic finite element modeling, X-ray diffraction, and the visual inspection of fatigue spall cracks. The model simulates multiple ball impacts with a fatigue spall's edge in a 208 size ball bearing operating at 10,000 rpm. Ball impacts are shown to cause severe plastic deformation within the spall edge and induce tensile residual stresses. The finite element results are supported by X-ray diffraction measurements and the locations of cracks observed around the edge of a spall.

KEYWORDS: rolling element bearings, rolling contact fatigue, spall propagation, contact mechanics, bearing steels, gas/jet turbines, impact wear, elastic-plastic finite element analysis, residual stresses, X-ray diffraction

Introduction

The ever increasing demand for safer and more efficient military and commercial jet aircraft engines has encouraged the design of high performance bearings that have better corrosion resistance, longer rolling contact fatigue (RCF) life, and higher thrust load capacity. Hybrid bearings that use silicon nitride balls and case hardened metal raceways can meet these demands and substantially out-perform conventional through-hardened steel bearings [1–3].

However, even high performance bearings are not immune from the delete-

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rious effects caused by improper lubrication, foreign object debris, and excessive loading, all of which can cause surface fatigue failure [4]. These surface fatigue failures occur in the form of surface pitting or spalls on a ball or raceway surface. This type of surface fatigue failure is also observed in gears, cam-follower, and rails [5,6].

Conventional bearing fatigue life is usually defined as the amount of time or number of load cycles needed to initiate a surface spall. Impending bearing failure from RCF cannot be accurately predicted or detected until the bearing has started to liberate material in the form of fatigue spalls. Thus, the period of time from spall initiation to total failure becomes quite important. More robust materials with a high resistance to fatigue crack initiation and low spall propagation rate would help extend the period of detection and improve engine reliability.

Past and Current Research

Understanding the effects of plasticity, contact stress, fatigue, material microstructure, rate dependency, and residual stress formation in the vicinity of a spall is essential to designing safer bearings. However research of the governing mechanisms of spall propagation is limited. Probabilistic bearing life prediction models that are based on equations developed by Lundberg and Palmgren [7] are often used to estimate the life of a spalled bearing but neglect important effects such as localized plasticity. A comprehensive review and comparison of these bearing life theories are presented by Zaretsky et al. [8] and Sadeghi et al. [4]. Kotzalas and Harris [9] studied spall progression on chemical vapor deposition (CVD) American Iron and Steel Institute (AISI) 52100 steel balls that were driven by vacuum induction melted, vacuum arc remelted (VIM VAR) M50 steel V-Ring raceways and extended the bearing life prediction methods of Ioannides and Harris [10] to predict the remaining useful life of spalled bearings. Xu and Sadeghi [11] implemented damage accumulation laws within representative volume elements undergoing RCF to model the transformation of a dent into a progressed spall.

The spall propagation experiments on tapered roller bearings by Hoeprich [12] highlighted the randomness inherent to spall propagation and its unknown governing mechanisms. In a recent three-part series, an experimental and numerical investigation of spall propagation was presented [13–15]. Part I by Rosado et al. [13] showed the dependency of spall propagation rate on material selection and contact stress. Part II by Arakere et al. [14] presented the static elastic-plastic stress fields around an initial fatigue spall, the likely conditions that contribute to initial spall growth, and the importance of including plasticity effects in bearing fatigue life calculations. Part III by Forster et al. [15] investigated the effects of alloy content, heat treatment, and residual stresses on the stability of the microstructure and alterations in residual stress profile ultimately impacting bearing fatigue life and propagation rate.

The analysis presented here is unique because it uses finite element models that include the effects of plasticity to calculate the critical stresses and strains that develop within a spall edge during and after successive ball impacts. The

results are supported by the residual stresses calculated by X-ray diffraction (XRD) techniques and the location of cracks on an impacted spall edge. This information will support a plausible scenario of why fatigue spalls propagate. This analysis represents a first step toward identifying how material properties, bearing geometry, and operating conditions contribute to spall propagation and eventual bearing failure.

Nature of Spall Propagation

Spalls propagate predominately in the direction of ball motion during surface fatigue failure in bearings [12,13]. Figure 1 shows an example of damage progression along a bearing inner raceway. In the experiments explained in Ref 13, spalls are initiated naturally from material fatigue or by Rockwell C indentations, which act as stress risers and accelerate the spall initiation process. During bearing operation cracks develop around the indent, liberate surface material, and form a small initial spall. The initial spall will widen during operation as described in Ref 14 and eventually allow the ball to descend into the spall and impact the trailing edge. The spall will then progress around the raceway in the direction of ball motion until catastrophic failure occurs (Fig. 1(c)). Usually the catastrophic failure mode is fracture of the bearing cage leading to bearing seizure.

The spall's trailing edge will be defined as the edge that deteriorates with continuous bearing operation, whereas the spall's leading edge is a portion of the initial spall and does not liberate significant material during operation.

The numerous impacts that occur between balls and the trailing edge of the spall are believed to be the main driving forces of spall propagation (Fig. 2). The trailing edge is the only spall edge that is subjected to ball impacts *and* deteriorating; clearly there is a relationship between the two.

Also, significant cracks form only on the spall's trailing edge (Figs. 3 and 4) as the spall is propagating. This is another indication that more damage is occurring on the impacted edge in the form of cracks and not on the leading edge. Continuous ball impacts are believed to encourage these edge cracks to grow and cause fragments of material to liberate from the raceway's surface. The fragments collected by the Oil Debris Monitor in [13] were typically the same size as the edge of the spall.

This continuous loss of surface material increases the clearance between the bearing's inner and outer raceways, which creates enough space for the engine shaft to misalign. Severe vibrations, heat, noise, and eventual catastrophic bearing failure are then likely to occur.

Procedure and Finite Element Model

The finite element model must capture the geometry and physics of bearing operation as accurately as possible. However, this analysis will not model the entire inner and outer rings, all balls, lubrication, temperature effects, and interactions with the bearing cage. For this purpose, only a segment of the inner ring will be modeled in the interest of computational efficiency (Fig. 5).

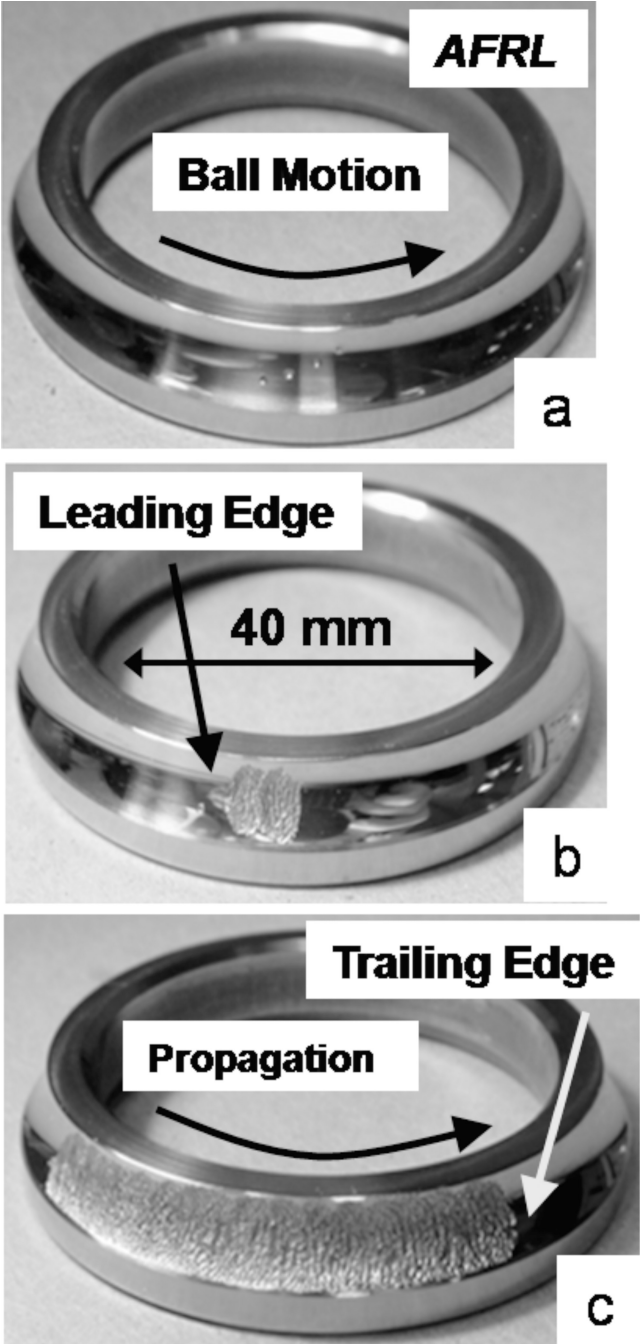


FIG. 1—(a) Virgin raceway. (b) Small spall. (c) Progressed spall.

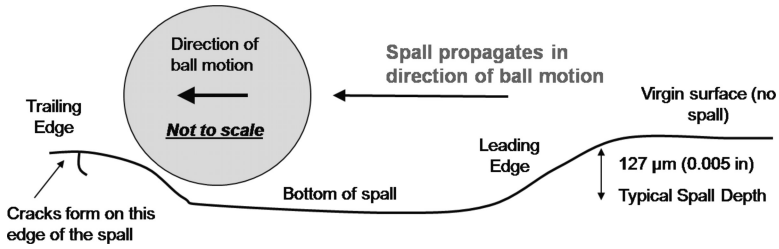


FIG. 2—Ball impact with spall's edge influences propagation direction.

The raceway segment will include the trailing edge of the spall since we are investigating the types of damage this specific spall edge is likely to develop during ball impact.

To best capture the geometry of the spall's edge, profilometer tracings were taken of the spall's trailing edge on four different raceways made of Pyrowear 675 (P675) case hardened and M50 through-hardened bearing steels (Fig. 6(a)). An average spall edge slope of 45° is determined from the four profile measurements and is applied to the finite element (FE) model geometry (Fig. 6(b)). This edge geometry is constant during spall propagation (Fig. 6(b)) and suggests that the failure mode controlling material loss from the spall's edge is the same regardless of spall length.

The finite element model will simulate three successive rigid ball impacts on the same spall edge at a ball velocity of 15.24 m/s (50 ft/s) (Fig. 7). The ball velocity is determined from the known rotational speeds of the cage and inner raceway of the 40 mm bearing used in Refs 13–15. Three impacts were chosen because there was no significant change in the stresses and strains within the spall edge after the second and third impacts.

Rigid balls do not require a mesh and reduce the number of elements in the model. All three of the balls have the density of steel (8 g/cm^3) and are expected to impart more energy into the spall than compared to a silicon nitride ball. The ball is not allowed to spin and can only translate within the plane of symmetry. Since contact is frictionless, the rotational kinetic energy of the ball will not significantly affect the nature of impact between ball and spall edge.

In house compression tests were performed on M50 steel specimens (Fig. 8) to obtain its stress strain curve up to 0.013 plastic strain and compared well with data available from Carpenter [16]. Since the plastic behavior of this material after 1.3 % plastic strain is unknown or unobtainable from other sources, the material is assumed to behave in a perfectly plastic manner after this strain is reached. Typically, very hard materials do not have a capacity to strain harden like copper or 303 stainless steel, so a perfectly plastic response is a reasonable assumption for this bearing steel and this analysis.

A very fine linear hexahedral mesh is designed around the spall's edge because the most critical stresses and strains are forming within this small region (Fig. 7). The Explicit solver within ABAQUS v6.8-2 was used since this is a transient analysis and has time dependent properties (the velocity of the ball) [17]. The von Mises (J2) failure criterion, associative flow rule, and isotropic hardening are appropriate for bearing steels and implemented here.

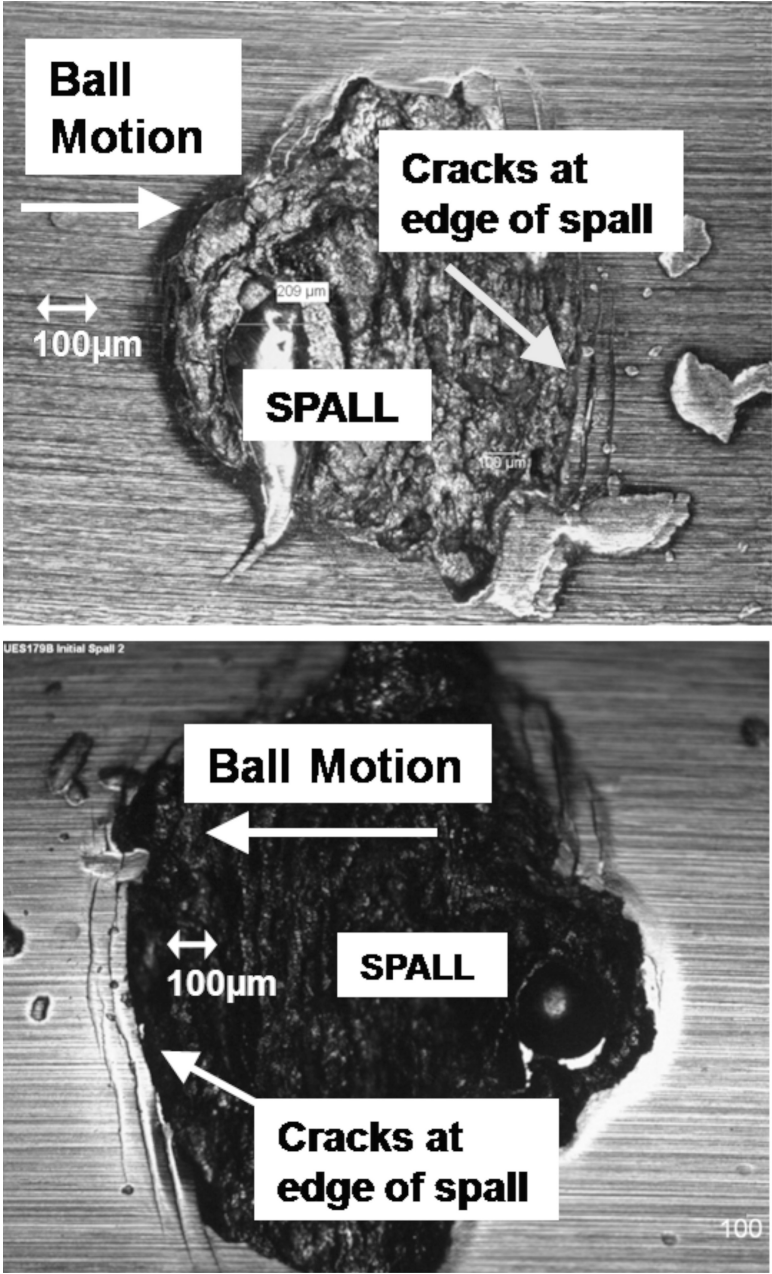


FIG. 3—Two examples of cracks on trailing edge of propagating spall.

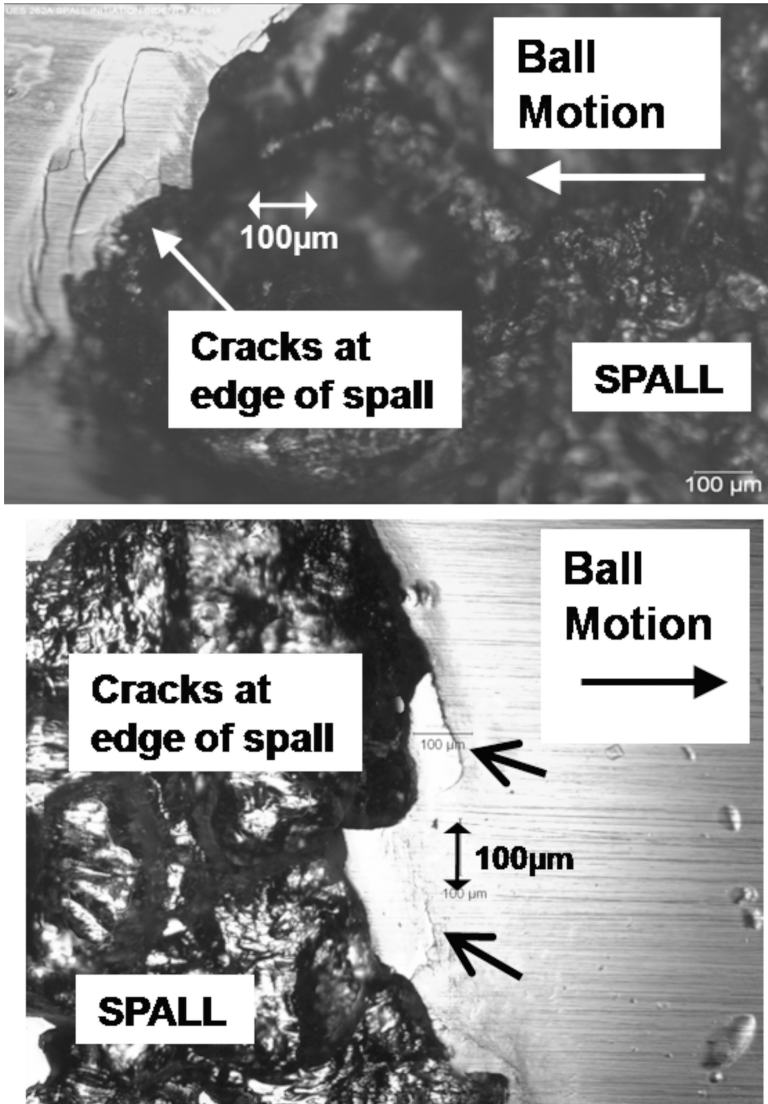


FIG. 4—Two examples of close up images of spall's edge cracks.

Results and Discussion

All plots of the finite element model results will be close up images of a spall edge's cross section (Fig. 9). The radial and hoop stresses within the spall *during* impact were calculated (Fig. 9) and are mostly compressive as expected. The maximum contact pressure between a ball and spall edge can reach 7 GPa, which is sufficient to cause subsurface yielding. Classical Hertzian contact so-

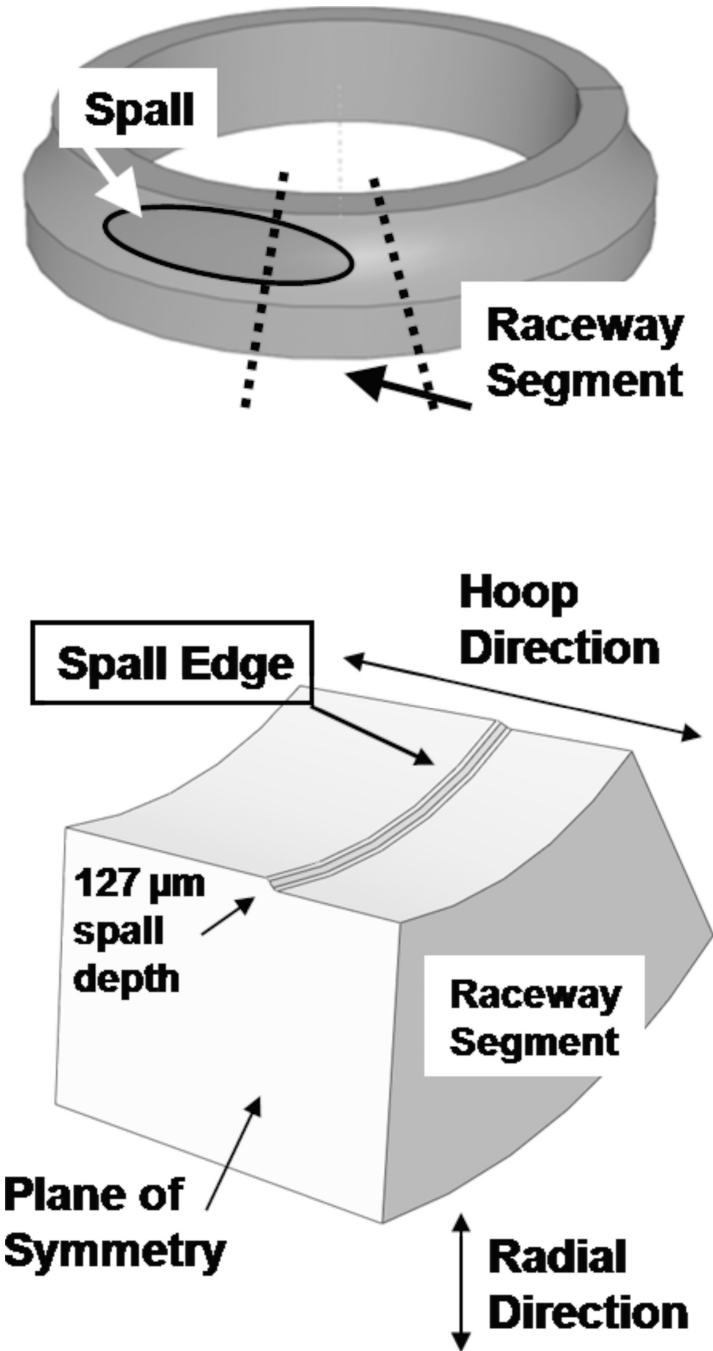


FIG. 5—Finite element model geometry.

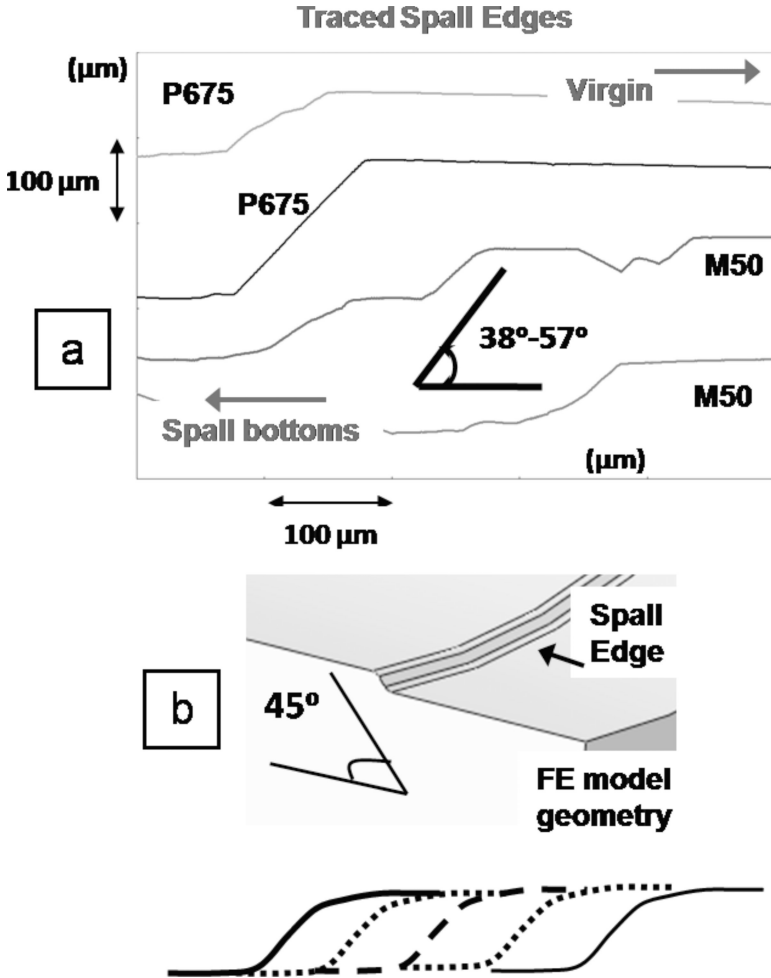


FIG. 6—(a) Tracings of spall edges from M50 and Pyrowear 675 bearing steels. (b) Consistent spall edge during propagation and used in FE model.

lutions do not apply here since the spall edge is very sharp and is undergoing considerable deformation during impact.

Of greater interest are the locations and distributions of *residual* tensile stresses around the edge of the spall. It is well understood that tensile residual stresses are detrimental to the fatigue strength of a material because they encourage fatigue crack initiation and growth. The finite element model calculates tensile residual hoop stresses to occur on the surface of the spall's edge after successive ball impacts (Fig. 10). This is significant because this location of residual tensile stress agrees with where cracks are seen around the spall's edge in the bearings from Refs 13 and 15 (Figs. 3 and 4).

The locations and magnitudes of residual hoop stresses within the spall's

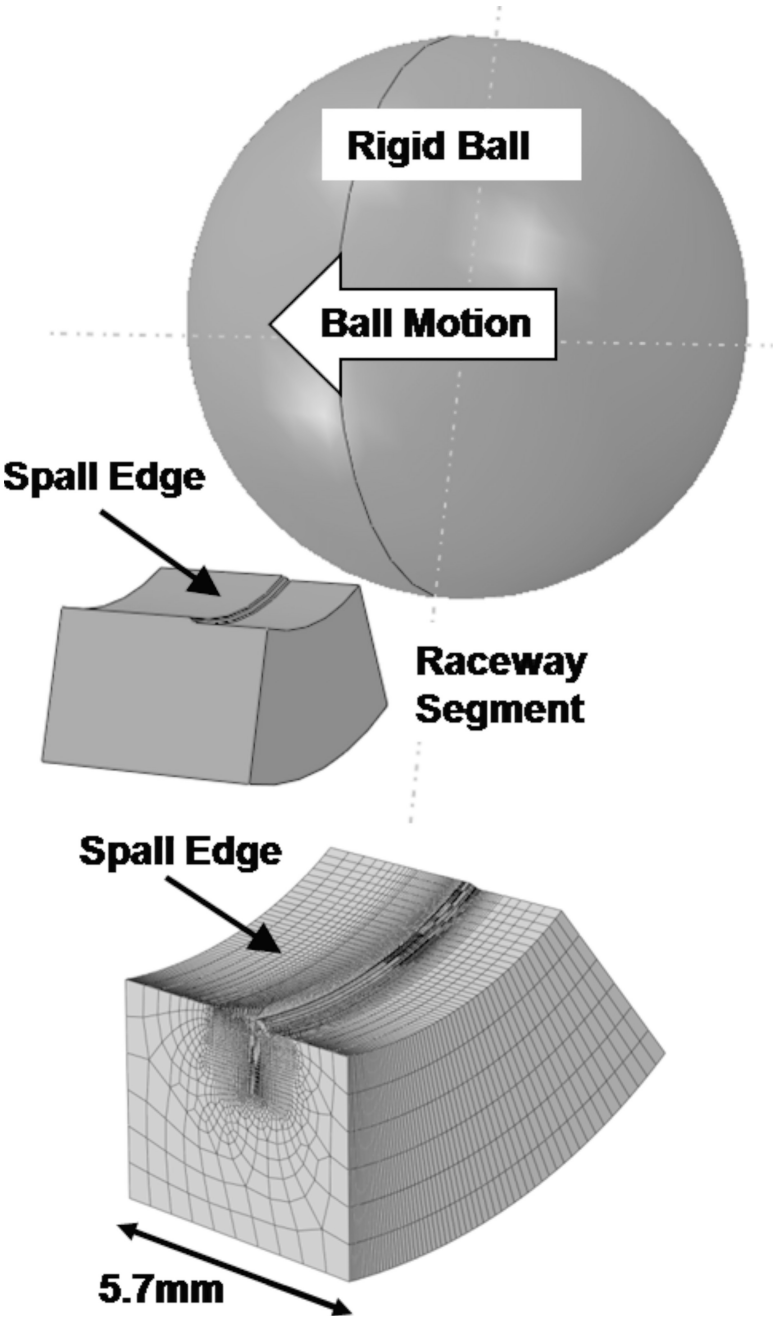


FIG. 7—FE model of ball impacting spall edge and finite element mesh.

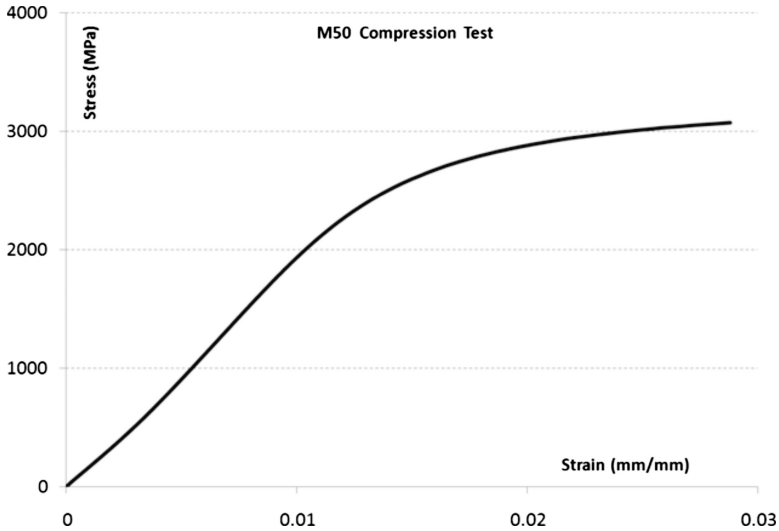


FIG. 8—Flow curve, M50 steel.

edge from the FE results are shown in Figs. 10 and 11. Locations D and E best capture the residual tensile stress at the surface mentioned earlier and the residual compressive stresses just beneath the spall's edge. The severity of the residual stresses decreases going away from the spall's edge as shown in profiles A, B, and C.

Also, locations A, B, and C of the finite elements results match closely to the profiles calculated from XRD techniques (Fig. 12) and described in more detail in Ref 15. The results of Forster et al. [15] showed that the changes in residual stress were cumulative over millions of cycles in AISI 52100, AISI M50, and M50 NiL bearings. XRD was done on bearing raceways that had experienced 10^6 – 10^{10} stress cycles, whereas the model is only simulating three impacts. Therefore, only qualitative comparisons can be made here.

For example, the residual stress profile in Fig. 12(a) [15] is from a spall initiated from surface indents and is quite different from the spall initiated from material fatigue without surface indents (Fig. 12(b)). This is not captured in the bearing model and may be a result of microstructure decay [15]. Also, the location of the experimental XRD measurement was about 3.2 mm (0.125 in.) from the edge of the spall in all cases. The spot size of the irradiated area is $1.3 \times 1.3 \text{ mm}^2$.

The XRD residual stress profiles are not a close match to profiles D and E because the location where XRD was performed was not close enough to the spall's edge to measure the residual tensile stresses shown in profiles D and E. Even if XRD was taken on locations D and E, unknown cracks within the spall's edge could have relieved residual stresses that XRD could not have picked up. This helps explain why the data is not an exact match for every location, but profiles A, B, and C are within the ballpark of what is determined from XRD.

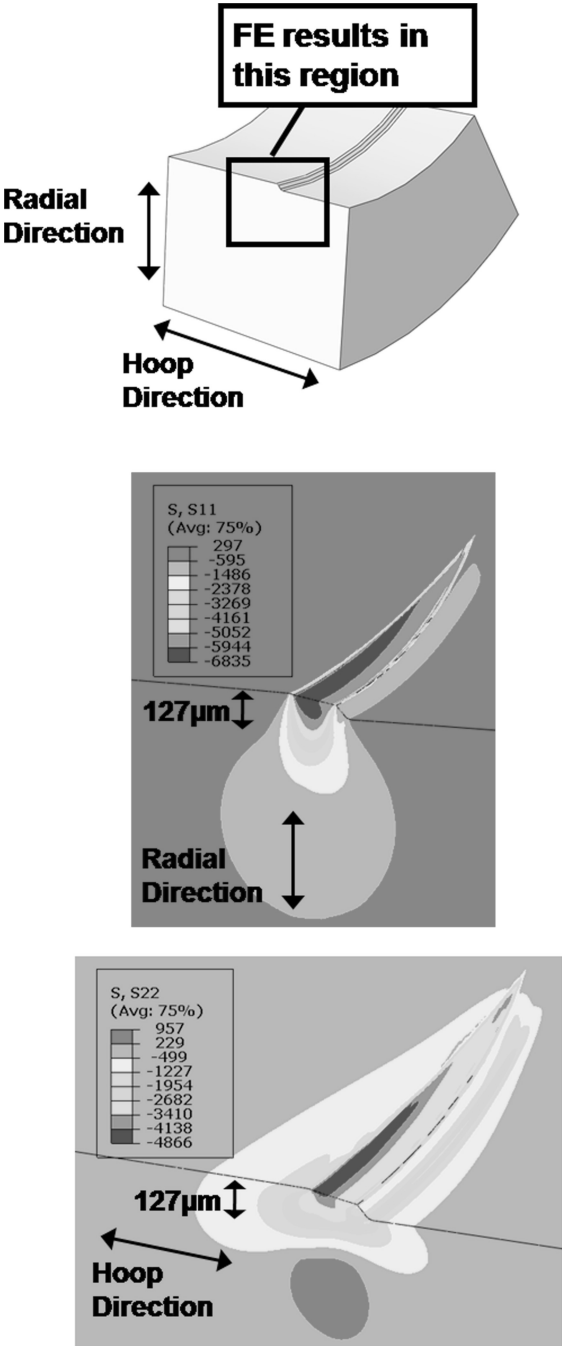


FIG. 9—Radial and hoop stresses during impact.

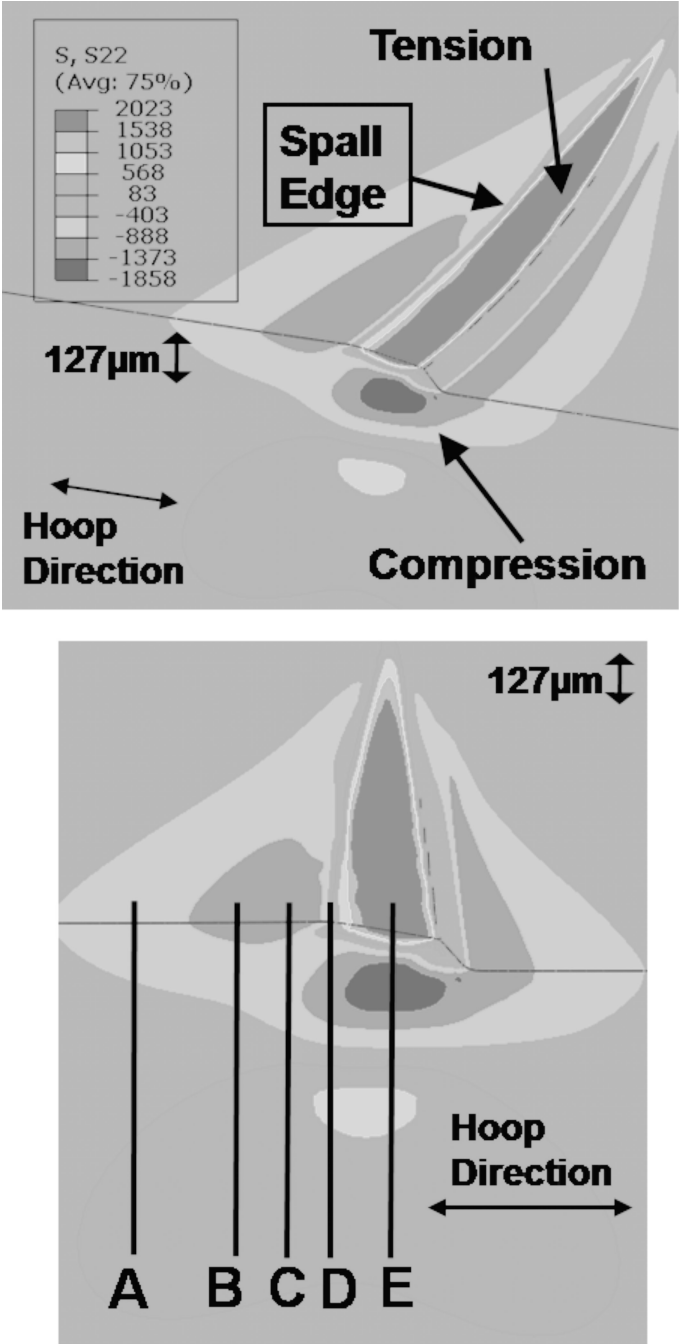


FIG. 10—(a) Residual stress profile in hoop direction after three impacts. (b) Locations of profiles in Fig. 11.

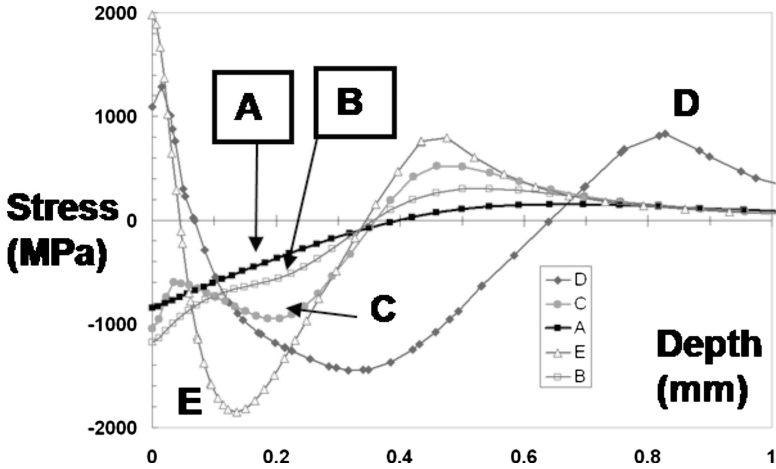


FIG. 11—Residual hoop stress profiles, finite element results.

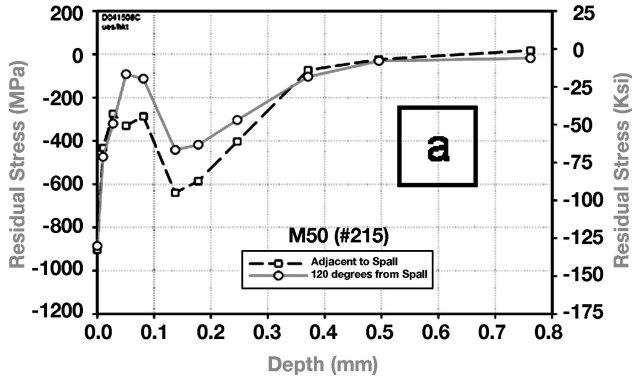
Many fatigue failure theories predict tensile *radial* residual stresses to develop in rolling contact [4], which lead to crack formation parallel to the surface. The finite element results show that tensile radial residual stresses exist within the edge of the spall after successive impacts (Fig. 13).

Radial tensile residual stresses encourage fatigue cracks to form and their location matches the spall depth (Figs. 13 and 2). Residual stress profiles are plotted (Fig. 14) at the same locations as the hoop direction plot (Fig. 10). Locations D and E capture the residual compressive stresses just below the spall's edge, while B and C show subsurface residual tensile radial stresses at the same depth as the depth of the spall. The location of residual tensile radial stress explains why the spall depth is consistent throughout spall growth.

The finite element results in Figs. 10 and 13 show compressive residual stresses in both the radial and hoop directions located just below the spall's edge. Compressive residual stresses below the spall's edge retard crack growth [18] and consequently keep material loss localized to the immediate surface. This is observed in experiments and shown in Figs. 2–4.

The location of residual tensile hydrostatic pressure also agrees with where cracks appear in spalled bearings (Figs. 15 and 3). ABAQUS defines tensile hydrostatic pressure as negative, which is why the tensile region is blue instead of red as before. Howell et al. [19] studied rolling contact over a surface indent and showed that cracks are likely to initiate in regions of tensile hydrostatic pressure.

- Bearing 215 – Life tested, suspended at
- Suspended $16,343 \times 10^6$ cycles at 3.01 GPa
- Propagation 31.0×10^6 cycles at 2.41 GPa



- Bearing 075 – Propagation initiated on new bearing from indents
- Initiation 19.6×10^6 cycles at GPa
- Propagation 187.3×10^6 cycles at 2.41 GPa

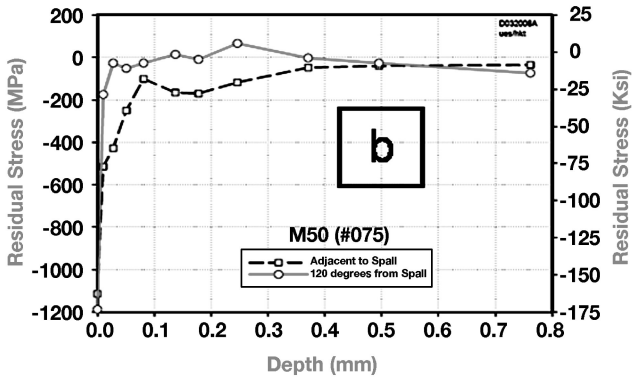


FIG. 12—XRD residual hoop stress profiles; M50 spalled bearings.

The strain-life approaches summarized in Ref 18, such as Coffin–Manson, have shown that fatigue cracks are likely to initiate in a region of high plastic strain over many load cycles. A detailed quantitative fatigue life analysis cannot be made here since the cyclic strain-life data for bearing steels is scarce, the plastic strains calculated by the FE model are for a spall that has only suffered three impacts, and the cyclic plastic strain amplitudes from the FE model are highly dependent on the cyclic hardening law, which is also limited for bearing steels. However, as a qualitative investigation it is worth comparing the distri-

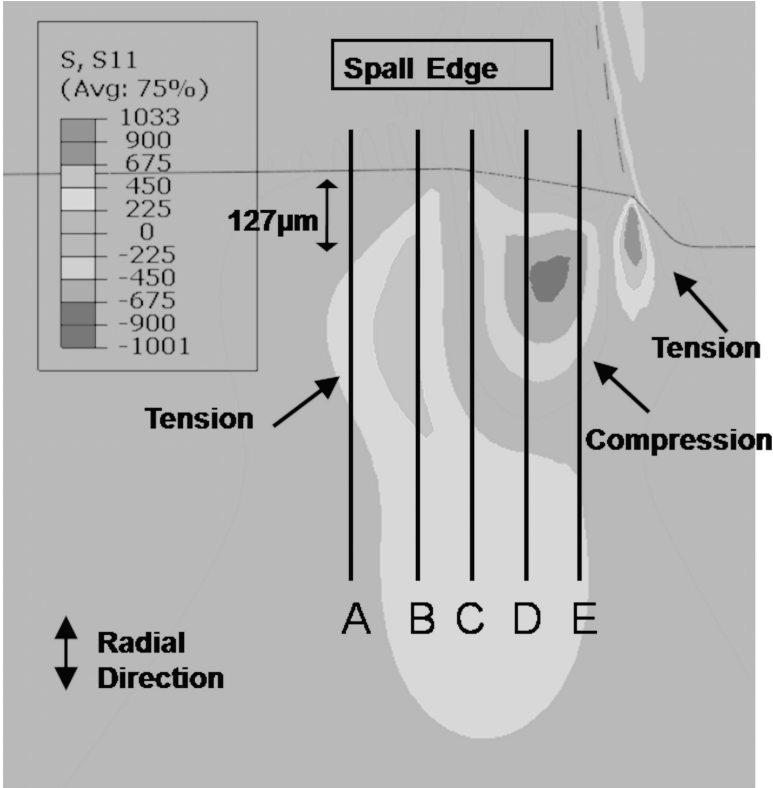


FIG. 13—Residual stress state in the radial direction and locations of stress profiles shown in Fig. 10.

bution of plastic strain within the spall edge with the location of cracks in the actual bearings to see if cracks form in the most damaged region as predicted by the FE model.

Large amounts of plastic strain in the hoop direction develop on the surface of the spall's edge (Fig. 16) and match closely to where cracks are observed in actual bearing spalls (Figs. 3 and 4). The maximum hoop plastic strain is below the surface and is a likely site of crack initiation.

The distribution of maximum principal plastic strain (Fig. 17) corresponds to the locations of residual tensile hoop stress (Fig. 10), residual radial tensile stress (Fig. 13), residual tensile hydrostatic pressure (Fig. 15), and hoop plastic strain (Fig. 16). Cracks are likely to follow this path of highly damaged material and also aided by the tensile and compressive residual stresses within the spall.

To verify the assumption that cracks follow the path of maximum principal plastic strain, it is shown that the distribution of maximum principal plastic

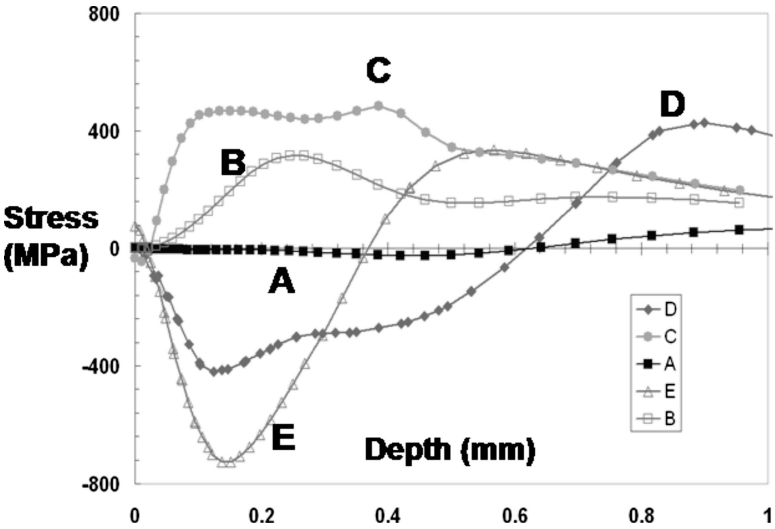


FIG. 14—Finite element residual stress profiles in radial direction.

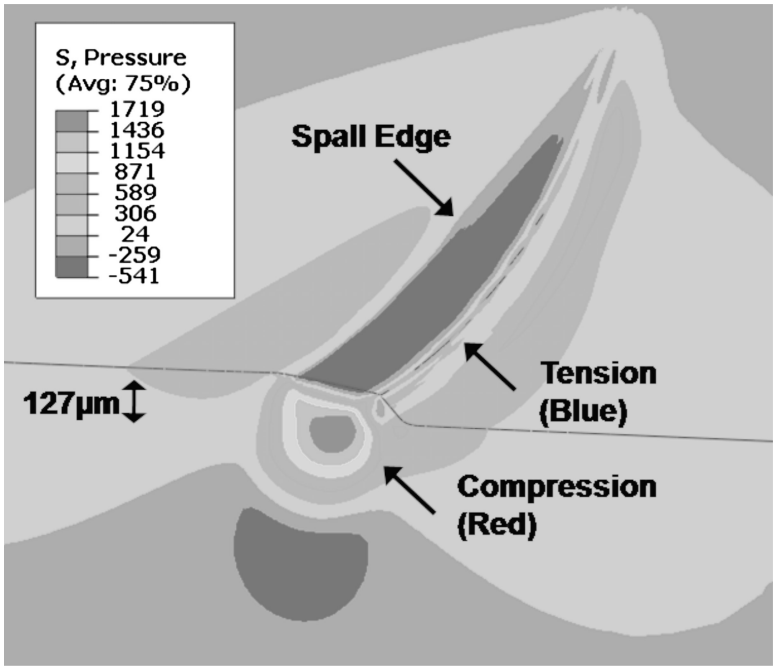


FIG. 15—Residual hydrostatic stresses.

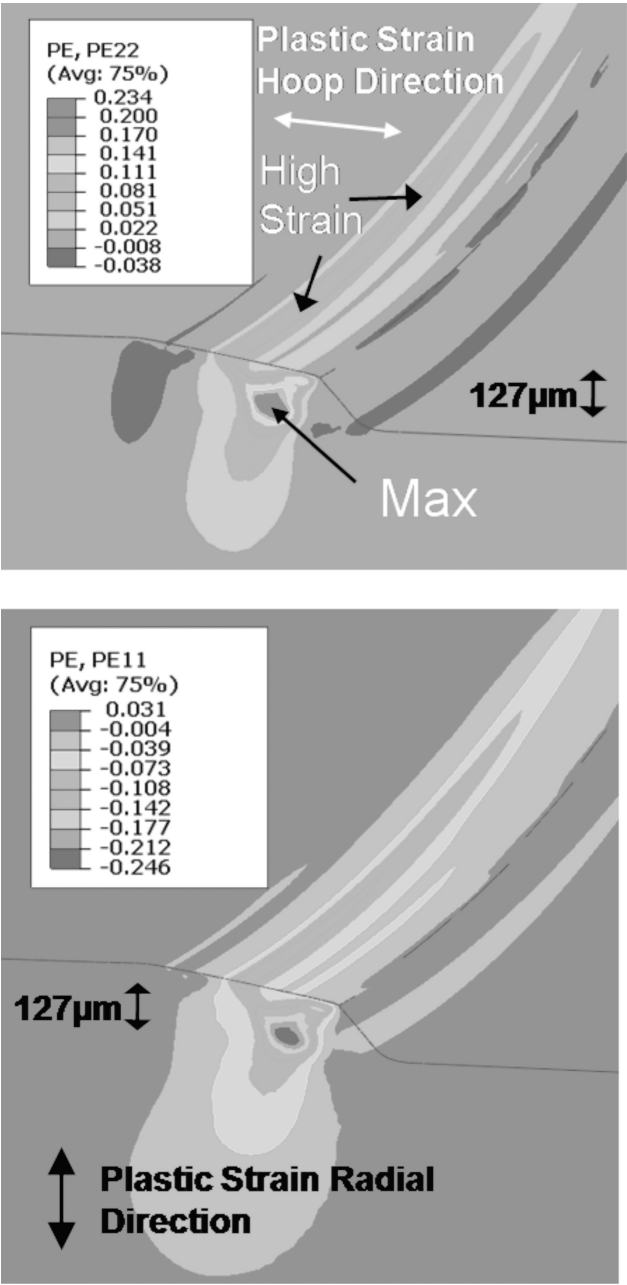


FIG. 16—Hoop and radial plastic strains. Radial plastic strain is mostly compressive around edge of spall. Locations of large tensile hoop plastic strain are close to where cracks appear.

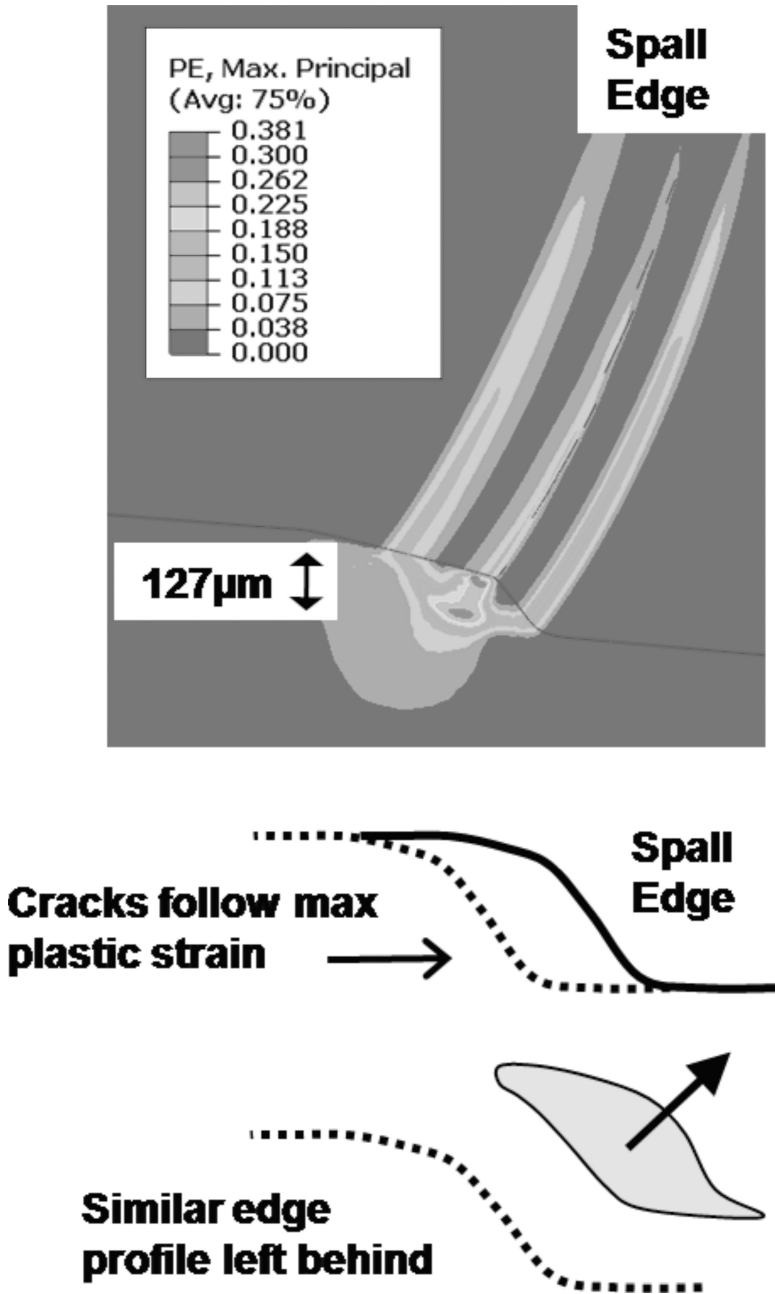


FIG. 17—Maximum principal plastic strain.

strain (Fig. 17) is also similar to the profilometer tracings of the spall edges (Fig. 6). After a fragment of material is liberated from the spall edge, the new spall edge profile left behind is a close match to the profilometer tracings *and* the distribution of maximum principal plastic strain. This process repeats itself and explains why the spall edge profile does not vary throughout spall propagation.

Spall edge geometry is expected to influence the calculation of stresses and strains in the finite element model; however it can be shown that similar residual stress gradients will develop regardless of spall edge sharpness (Figs. 18 and 19). Two different degrees of spall sharpness were modeled with the same ball velocity and material properties as before. Stress contours of the residual radial stresses within the spall's edge are shown along with a graph of the residual hoop stress profiles. As expected, the stresses around a blunt spall are not as severe as the sharper spall edge, but the fact that the stress gradients are similar helps show the dependency of spall shape in the finite element results.

Conclusions

This analysis represents a first attempt to capture the critical stresses and strains that develop in a spall edge during and after successive ball impacts. The validity of the finite element results is strengthened by the locations and magnitudes of the residual stress profiles calculated from XRD techniques and the locations of cracks on the impacted spall edges observed in spall propagation experiments.

It is well understood that residual tensile stresses decrease the fatigue life of a material [4]. The finite element model determines residual hoop, radial, and hydrostatic tensile stresses to occur within an impacted spall edge at the same locations where fatigue cracks are observed in experiments. The computed residual compressive stresses below the trailing edge of the spall retard crack growth and keep material loss localized to the immediate surface as seen in experiments.

The distribution of plastic strain within the spall edge provides a likely path of crack growth which leads to the liberation of material fragments during spall propagation. This is supported by the observation that the spall edge shape remains consistent throughout propagation and closely matches the distribution of maximum principal plastic strain. Qualitative strain-life methodologies predict cracks to initiate in regions of high plastic strain. Cracks appear on spall edges where the finite element model predicts large plastic strain.

Future research will investigate how the impact between the ball and spall edge is affected by the presence of the outer raceway and bearing cage. Changes in material properties due to microstructure decay are expected to occur in actual bearings, and this influence should be included. Future work will also explain why spall propagation rates differ between case and through-hardened bearing materials.

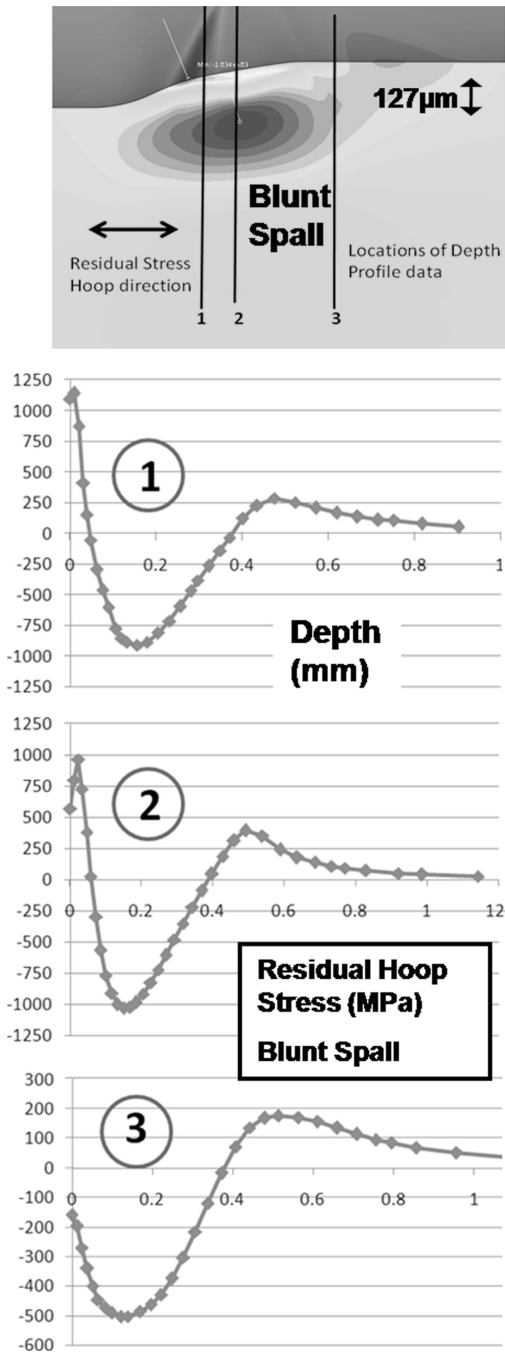


FIG. 18—Residual hoop stresses for blunt spall similar to that of sharp spall.

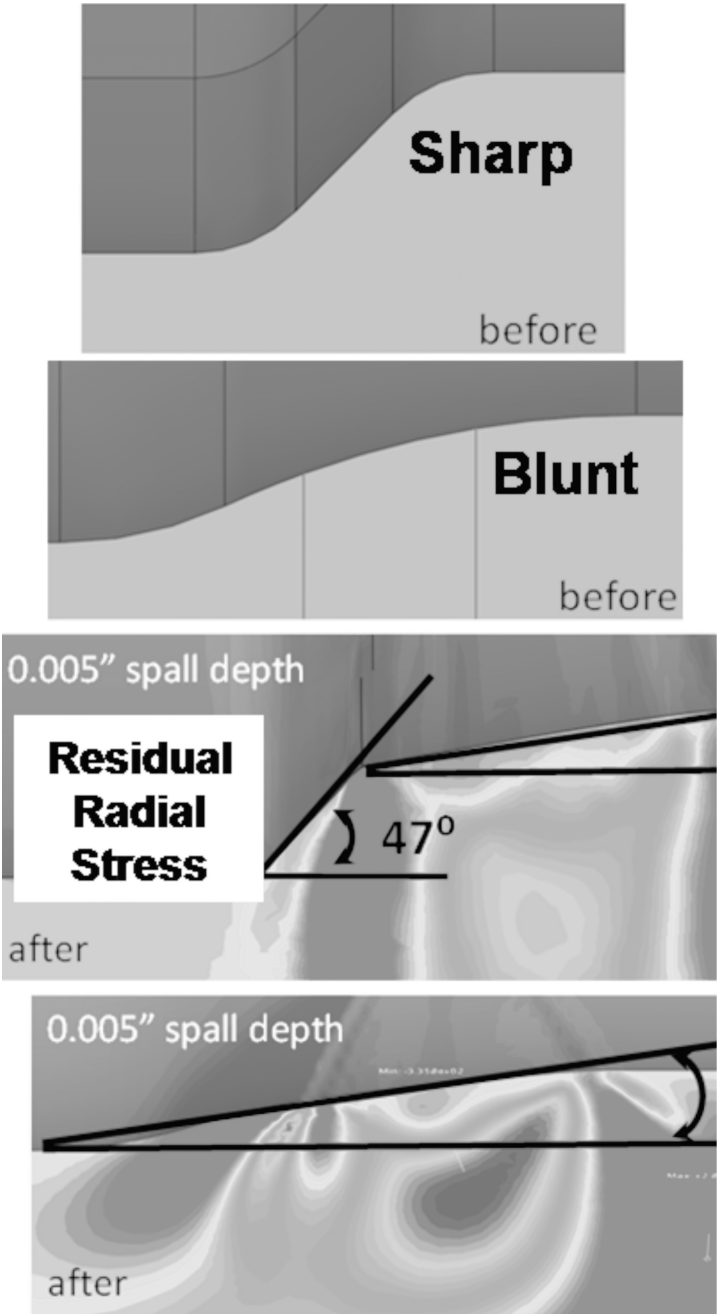


FIG. 19—Before and after images of sharp and blunt spalls. Similar radial residual stresses are obtained.

Acknowledgments

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Thore B. Lund¹

Sub-Surface Initiated Rolling Contact Fatigue—Influence of Non-Metallic Inclusions, Processing History, and Operating Conditions

ABSTRACT: A number of competing failure mechanisms are involved in bearing failure initiation. For well manufactured bearings operating under clean and well controlled running conditions, sub-surface initiated fatigue is the classical initiation form. Three mechanisms dominate the concept of sub-surface induced initiation and growth: (i) The well documented slow structural breakdown of the steel matrix due to accumulation of fatigue damage in a process superficially similar to tempering, (ii) stress induced generation of butterflies by a process enabling the growth of butterfly micro-cracks and accompanying wings at non-metallic inclusions, and (iii) surface induced hydrogen intrusion causing hydrogen-enhanced fatigue damage accumulation in the matrix. The development of butterflies as a function of contact stress, over-rolling, and non-metallic inclusion characteristics is presented, and the influence of metallurgical cleanliness and processing history on this progression is discussed. The results of laboratory conducted tests are compared to results from field applications where premature spallings have occurred. The progression from butterfly micro-cracks to extending cracks with non-etching borders has been studied. Special interest has been paid to the interaction between the non-metallic inclusion composition and morphology and their propensity to generate butterfly wing formations, as this may affect the way that inclusion harmfulness should be judged in rolling bearing steel quality assurance efforts. Complex oxy-sulfides are the main butterfly initiators in today's bearing steels.

KEYWORDS: bearing steel, steel making, hot forming reduction, non-metallic inclusions, fatigue

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Introduction

Bearing life theory still to a high extent is based on the Lundberg–Palmgren [1] developed carrying capacity to applied load relationships augmented with the Miner accumulating damage hypothesis [2]. These relationships were derived based on the assumption that fatigue gradually develops at sub-surface imperfections and that this development leads to spalling of bearings under application load conditions where the running conditions do not supersede the development of the sub-surface initiated “classic” fatigue development.

It seems well proven that the natural way of bearings to fatigue under moderately high contact stress conditions is a gradual decay of the matrix, which with accumulating fatigue damages leads to a finite bearing life after long to very long service times [3]. This introduced the concept of an infinite life of bearings if stressed below the fatigue limit [4].

In certain bearing applications today, however, premature failures are experienced. Such early failures are application dependent and occur very rarely seen to the total bearing population in use. They develop in all bearing types and for all heat treatment processes used. They are generally associated with sub-surface structural transformations where micro-crack developments with non-etching borders are present.

In order to increase the understanding of these phenomena, a detailed study of the origin, morphology, and development of such micro-structural transformations has been undertaken.

Classic Sub-Surface Initiated Fatigue

Lundberg–Palmgren early realised that non-metallic inclusions initiate sub-surface damages that lead to crack growth and spalling. This did not only lead to the formulation of the carrying capacity to load relationships later standardised and still the fundament of bearing life calculations (Fig. 1), it also led to the development of the procedures and ratings of non-metallic inclusions in steel that still is the fundament for bearing steel purchasing specifications; the Jernkontoret rating chart [5] later developed into the refined and currently world-wide-used ASTM E-45-05e2 [6].

Structural Decay in Bearings under Moderately High Stresses

In the late 1960s and during the 1970s, focus was on the slowly developing structural decay occurring in bearings under moderate to high contact stresses generating a gradual development of what superficially resembles tempering processes. The decay was shown to be a stage-wise distortion of the matrix structure into dark-etching regions followed by the development of ferrite bands aligned in two different directions towards the surface (Fig. 2) [7].

The process involved is a gradual fatigue damage mechanism, quite unlike tempering, and in the final stages this leads to a weakening of the structure, which tends to give the bearings a very long but finite life.

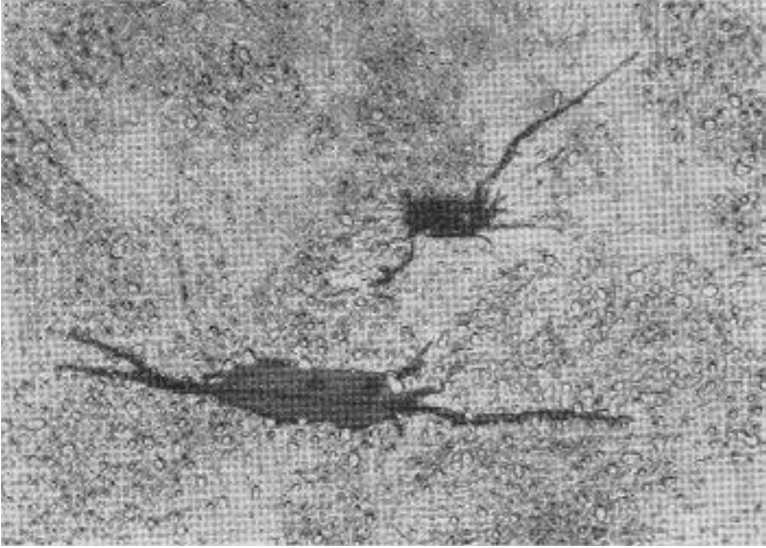


FIG. 1—*Bearing endurance tested and examined 1945.*

When bearings experience this fatigue process, few or no sub-surface structural transformations with non-etching crack borders have been observed.

Premature Failures in Specific Applications

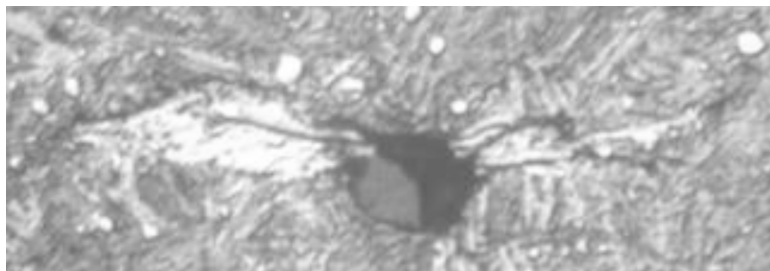
Under certain circumstances specific applications generate unpredicted early fatigue failures.

Such premature failures generally are associated with sub-surface micro-structural transformations of a completely different nature than the slow material decay.

Instead, non-metallic inclusions have initiated the growth of micro-cracks at non-metallic inclusions coupled with growth of non-etching areas at the micro-crack borders.



FIG. 2—*Material decay, dark etching region, and white etching bands.*

FIG. 3—*Butterfly*.

In everyday language we call the micro-crack development with associated non-etching constituents as “butterflies” (Fig. 3). Frequently, the growth of such formations is found in the form of propagating micro-cracks with non-etching borders, and on many occasions such micro-cracks start branching at pre-austenite grain boundaries into crack systems with non-etching borders.

The branching crack systems with non-etching borders, which can develop into considerable sizes, are denoted “white etching cracks.” This of course is a somewhat misleading denomination. The cracks are cracks, and cracks are just cleavages that of course do not etch. The branching crack systems develop borders that do not etch using the standard Nital etch in industrial use today in the same as the “wings” in butterfly formations. A more proper name for the crack systems would be “cracks with non-etching borders.” Or, if a distinction is desired between the cracks that develop as butterfly wings and then propagate into straight growing cracks and the large branching crack systems called WEC, then the WEC systems should be called “branching cracks with non-etching borders” (Fig. 4). Straight growing micro-crack formations with non-etching borders have not been given an accepted technical name and have seldom been discussed (Fig. 5). One reason for this is that the two phenomena were considered by many to be different species and that butterfly wing developments and WEC formations never were related to one another.

Characteristics of the Non-Etching Crack Borders

Using scanning electron microscope (SEM) and transmission electron microscope technology, the examination of the non-etching borders of initiating micro-crack formations associated with butterfly wings, developing large size butterfly wing formations, straight growing cracks with non-etching borders, and branching crack systems with non-etching borders (“WEC”) show that they are the same features in different stages of development.

They all are nano-sized ferrite cellular structures with the same morphology (Fig. 6). Due to their fine grained cellular structure, they have a very high hardness, and their hardness is surprisingly similar to that of an aluminum oxide inclusion, about 75 HRC (Fig. 7).

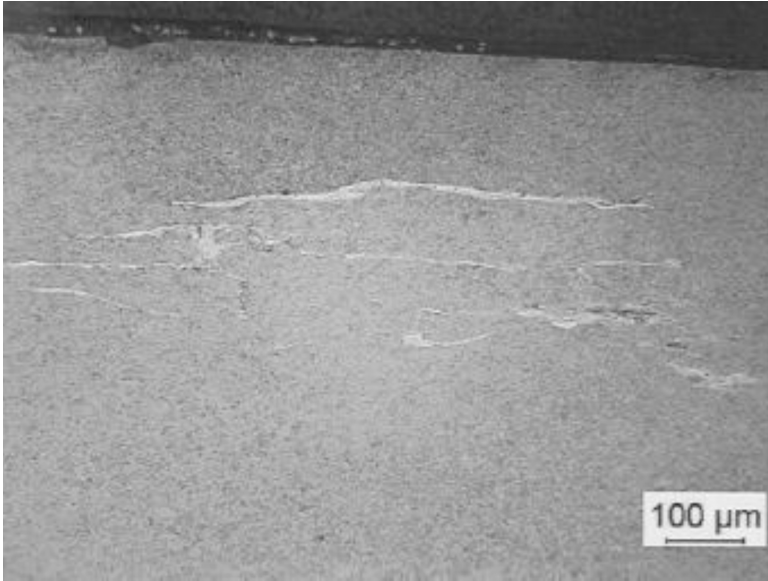


FIG. 4—WEC.

Micro-Crack Associated Butterfly Wing Development—Examination Procedure

In order to generate statistically significant data, two methods were used. A number of bearings having experienced early spalling damages in field applications were examined in order to assess the locations and the inclusion association of the micro-crack developments.

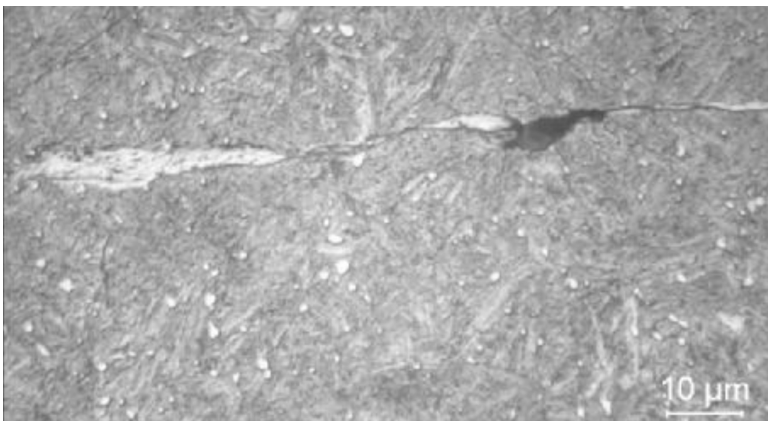


FIG. 5—Crack with non-etching borders.

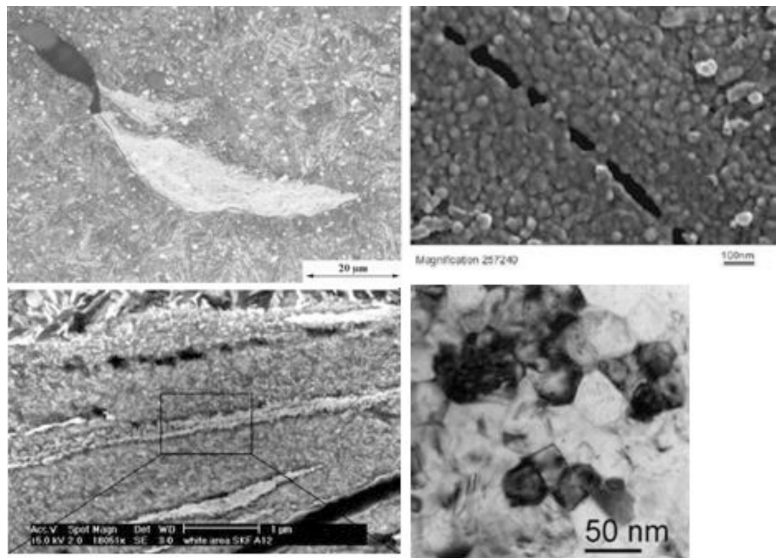


FIG. 6—*Butterflies, WEC in FEG-SEM.*

Additionally, a laboratory test procedure was developed where self-aligning ball bearings were used as the test vehicle in a way similar to the method used by Lundberg–Palmgren in developing the bearing capacity-load relationship and their attempts to relate this to sub-surface non-metallic inclusion initiation.

The tests were conducted on self-aligning spherical ball bearings using the outer ring as test vehicle. In doing this, a moderate load will generate contact stresses from 4.9 GPa to 0 in the outer ring of the bearing (Fig. 8).

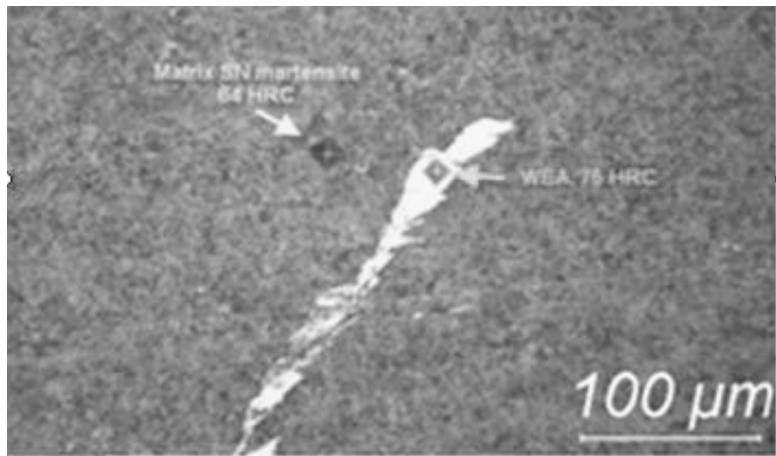


FIG. 7—*Butterfly wing and hardness.*

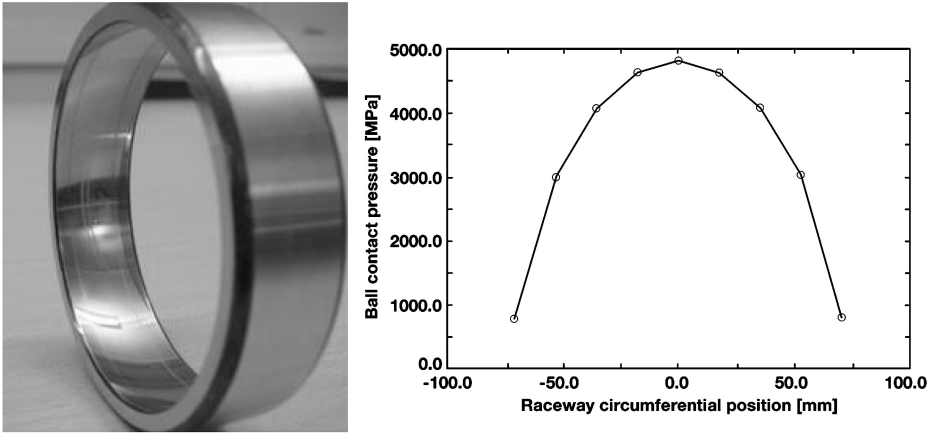


FIG. 8—SABB testing and stresses.

The tested bearings were sectioned using a circumferential cut through the raceway contact, which gives access to the sub-surface structural transformations developed for all contact stresses experienced in one sample (Fig. 9).

All tests were performed under well lubricated, clean, well lubricated, and non-aggressive conditions.

With the combination of the field bearing examinations and the laboratory tests, well over 1000 micro-crack associated wing formations were studied as regards to their location in relation to their position in the stress field and the inclusion association to their formation.

Micro-Crack Associated Butterfly Wing Growth: Micro-Inclusion Initiation

The micro-inclusion types associated with the generation of butterflies is of interest as it relates to the effectiveness of different inclusion types in generat-

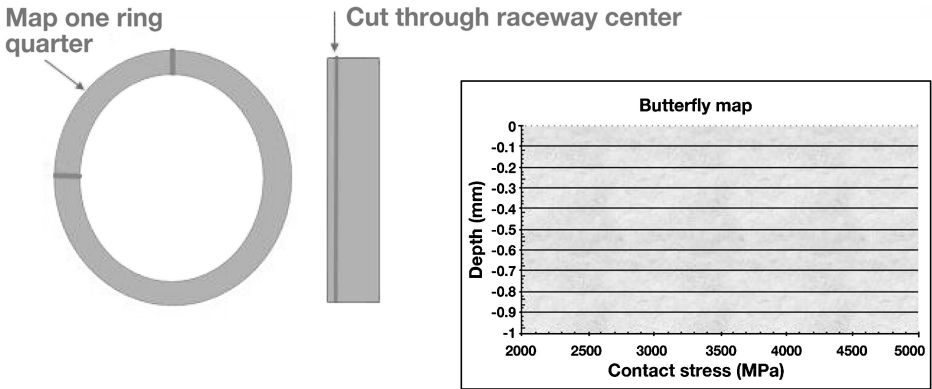


FIG. 9—Sampling of test rings.

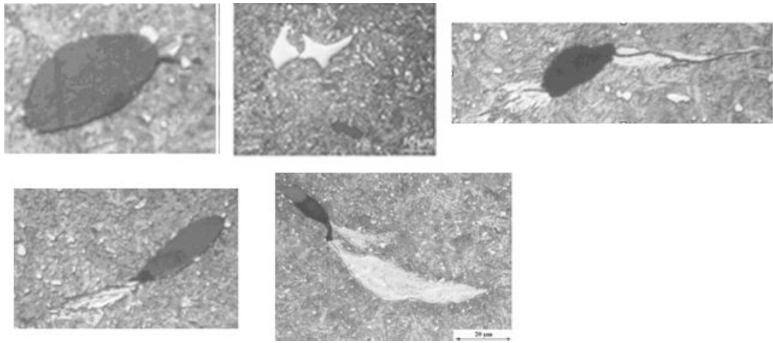


FIG. 10—Inclusions with and without butterfly formations.

ing micro-crack formation and growth. This potency should affect the way different inclusion types are rated in the micro-inclusions standards employed in the evaluation of the fitness for use of the steels used for bearing applications.

From the evaluations performed, it is evident that complex oxy-sulfide inclusions where the sulfide inclusions have encapsulated oxides are the main source for non-metallic associated micro-crack connected butterfly wing development.

This, however, only happens on the condition that the oxide inclusions entrapped in the sulfides are located in positions; here they have matrix contact and thus are visible to the stress field applied. Such complex inclusions frequently generate only one-sided wing formations due to the encapsulation location, while pure oxides generally generate two-sided wing formations. Titanium carbonitrides of the same size as oxides and present at positions experiencing the same stress conditions generally do not generate butterflies (Fig. 10).

In one field application, two bearings having experienced very similar application conditions could be compared (Fig. 11).

The difference between the bearings was that one had been produced to a

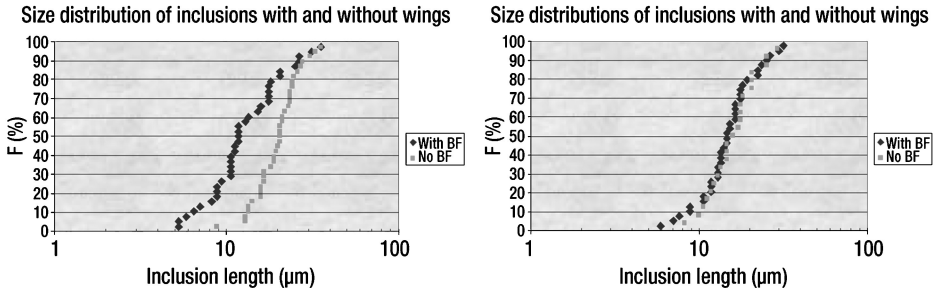


FIG. 11—Butterfly wing generating inclusions.

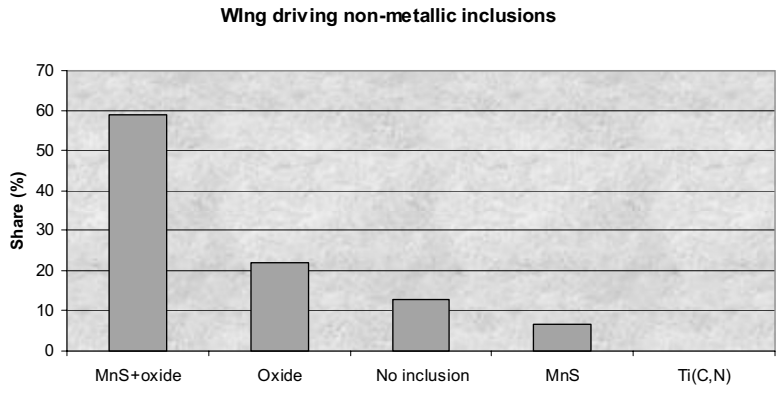


FIG. 12—*Inclusion and butterfly forming inclusion statistics.*

fairly low level of sulfur, while the other had been produced from a heat at the high end of the sulfur range.

While both bearings evidenced a similar number and size distribution of micro-crack associated wing developments, the bearing high in sulfur content had a significant number of sulfide inclusions present that did not.

The “surplus” sulfide inclusions having no oxide encapsulations thus did not generate any butterflies.

The main micro-crack associated wing former in modern bearing steel is thus sulfide inclusions with oxide encapsulations having matrix contact in positions open to the stress field.

Such inclusions also are the vast majority of the inclusions present in modern bearing steel.

These inclusions are also by far the largest present in bearing steel (Fig. 12).

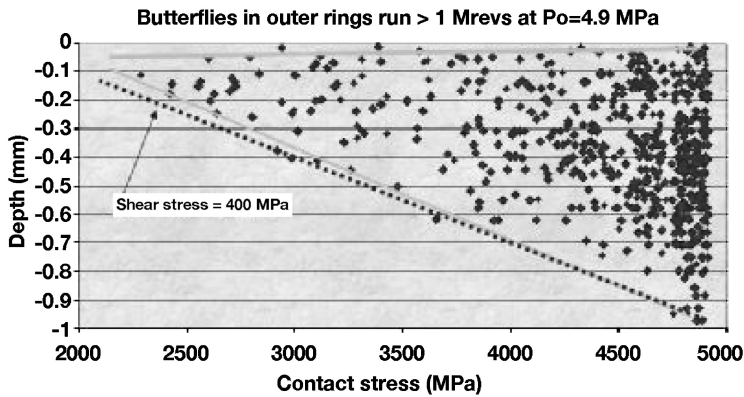


FIG. 13—*1309 butterfly map.*

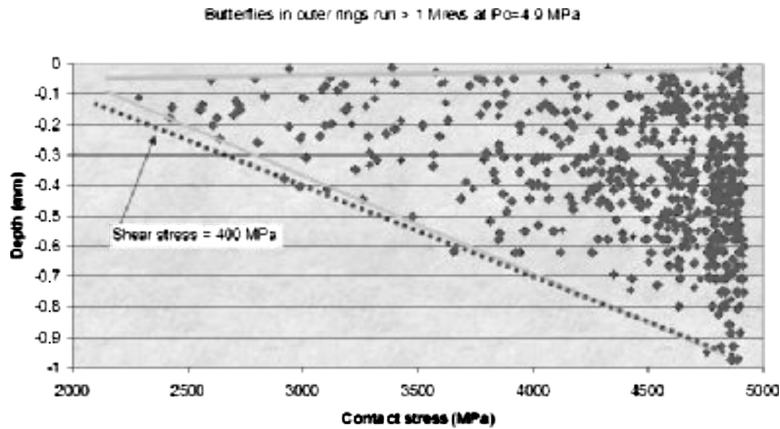


FIG. 14—1309 butterfly map and delimiting shear stress.

Micro-Crack Associated Butterfly Wing Development: Contact Stress Influence

By using the self-aligning ball bearing and the circumferential evaluation technique, the formation of micro-crack associated wing growth could be recorded. Associating the position of the wing formations to the contact stress experienced, a delimiting graph for their formation could be developed (Fig. 13).

The contact stress is of course not the factor deciding if a micro-crack will develop at a non-metallic inclusion; it is the local shear stress conditions prevailing at that inclusion, its orientation in the stress field and the composition and its morphology. The development of the butterfly is also dependent on the number of over-rollings, and thus there can be no way of establishing a direct

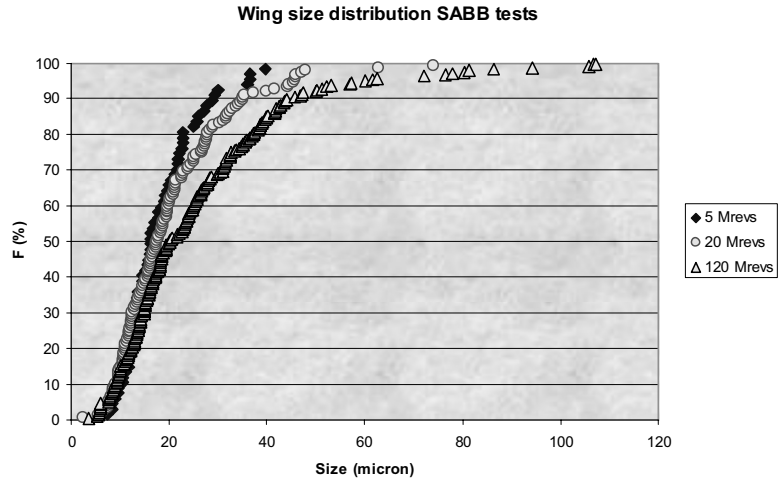


FIG. 15—Butterfly micro-crack growth.

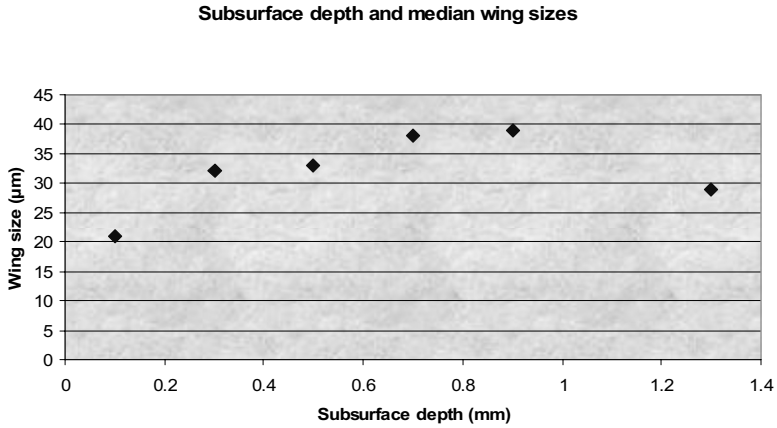


FIG. 16—Wing size and stress conditions.

relationship between contact stress, number of load cycles, and butterfly wing connected micro-crack development. However, a delimiting shear stress for the formation of butterflies is present and gives an indication of the boundary conditions for butterfly formation for the variant under test.

In this case, the lower bound for butterfly formation agrees well with a shear stress of 400 MPa (Fig. 14).

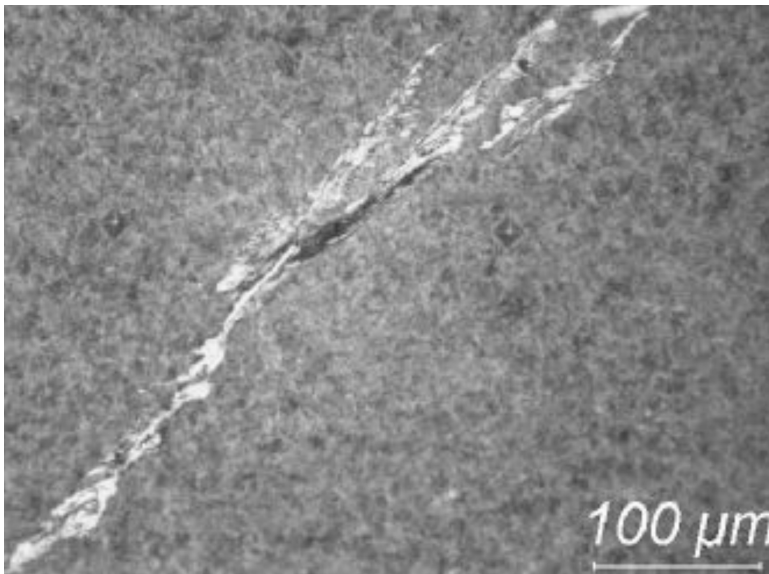


FIG. 17—XL wings.

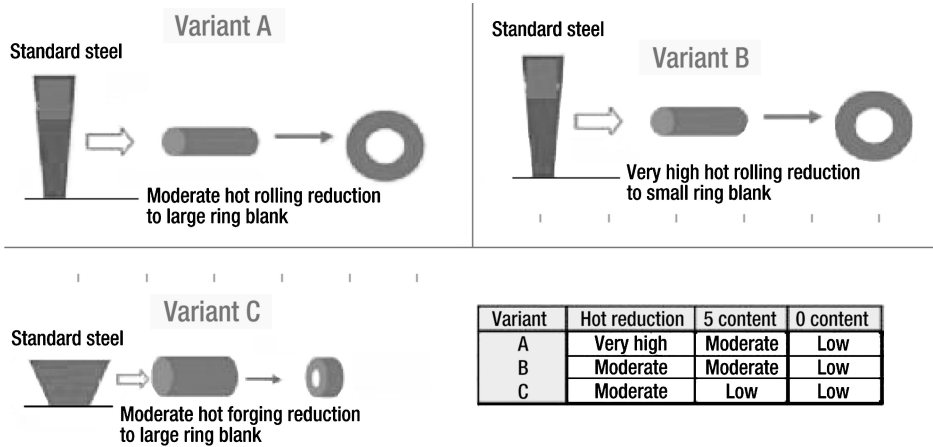


FIG. 18—Tested variants.

Micro-Crack Associated Butterfly Wing Development: Growth

The development of the micro-crack butterfly associated wing formations is stress and time dependent.

With prolonged stressing, under constant loading conditions, the frequency of the wing formations that do grow increases.

This is evidenced by the ratio of micro-crack formations that deviate from the normal size distribution seen in bearing rings at early running stages (Fig. 15).

There is also a relation between the developing micro-crack formations and the relative stress they have experienced (Fig. 16).

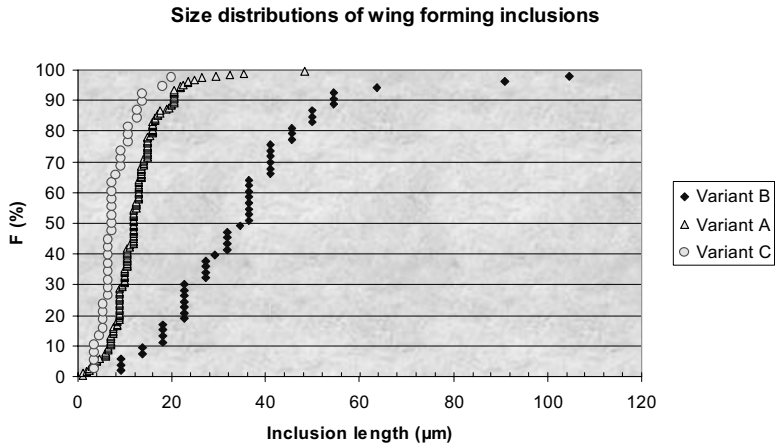


FIG. 19—Butterfly generation in tested variants.

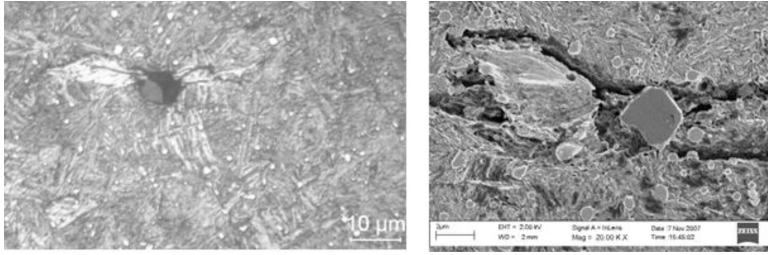


FIG. 20—*Butterfly appearance in tested variants—Variant A.*

On continued stressing, the micro-cracks developed in association with the wing formations can grow to considerable sizes. In the self-aligning bearing tests under clean lubricating conditions, micro-crack formations with non-etching borders of sizes approaching millimetres have been noted (Fig. 17).

Micro-Crack Associated Butterfly Wing Development: Steel and Heat Treatment Impact

Delaying micro-crack development and the associated wing growth could affect the development of early bearing damages. In order to evaluate the impact of steel non-metallic inclusions characteristics and heat treatment, a number of tests using the self-aligning ball bearing procedure were used.

In the steel testing program, three variants were used. The standard bearing for this application, a high hardness, extremely well, hot forming reduced product, was compared to a high cleanliness variant where virtually all oxy-sulfide inclusions had been removed, which had been given a high degree of hot forming reduction and a modern standard steel variant with a relatively low degree of hot forming reduction.

The test results show that the variant where the oxy-sulfide inclusions had been removed not only generated smaller butterfly wing associated wing formations, but there also was less tendency to micro-crack growth (Fig. 18).

The little reduced standard steel variant significantly deviates from the other two variants in both respects (Fig. 19).

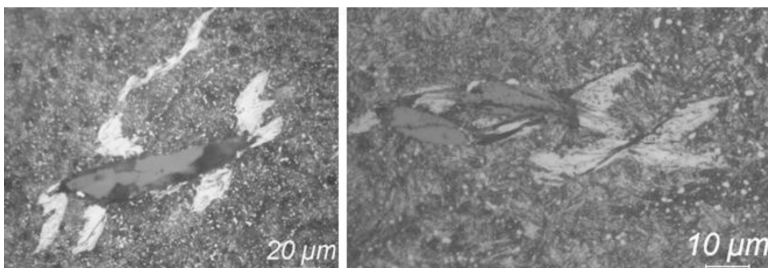


FIG. 21—*Butterfly appearance in tested variants—Variant B.*

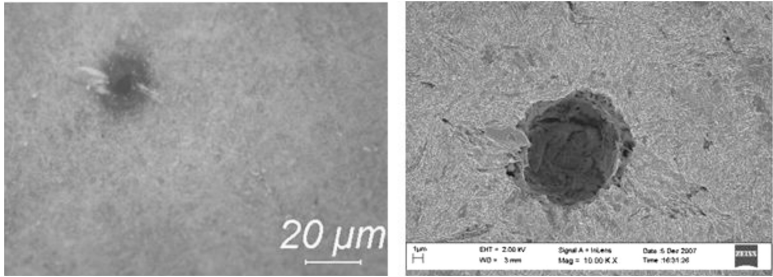


FIG. 22—Butterfly appearance in tested variants—Variant C.

The butterfly wing formations also differ significantly in appearance (Figs. 20–22).

Taking the steel matrix conditions to an extreme, a very lightly reduced billet rolled from a small as-cast dimension was used to produce test rings.

The total reduction from as-cast to tested product was only 2.2 times.

Under the same testing conditions as above, extremely short lives were attained, and significant development of WEC formations developed (Fig. 23).

At least for through hardened bearings, the development of butterflies is marginally influenced by the heat treatment condition used. No significant difference can be seen between bainite hardened and martensite hardened components at different hardness levels (Fig. 24).

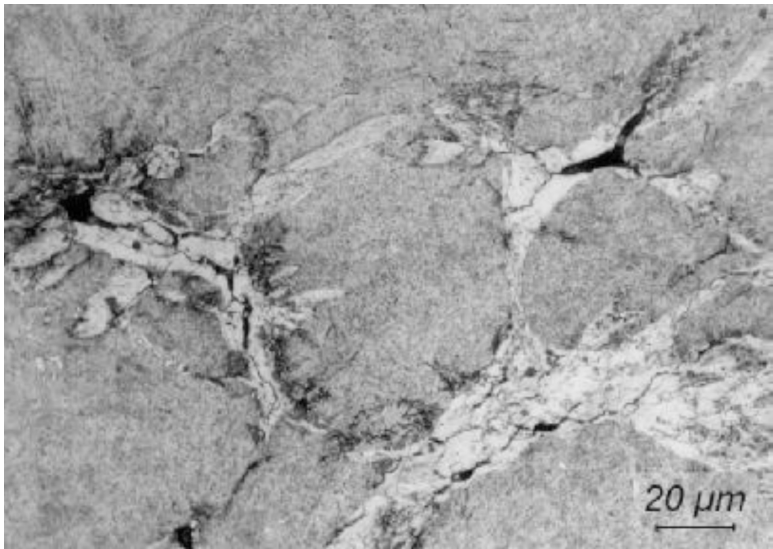


FIG. 23—WEC formation under clean lubrication and constant loading conditions

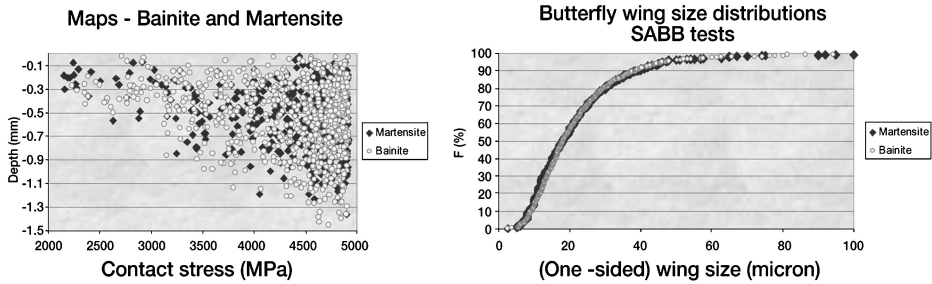


FIG. 24—1309 butterfly maps for bainite and martensite.

Micro-Crack Associated Butterfly Wing Development: Environmental Impact

It has been shown that hydrogen infusion can affect the development rate of micro-structural transformations significantly [8]. The development rate of micro-inclusion connected wing formations is thus not only related to the local shear stress conditions experienced, size, type, and composition of the non-metallic inclusion in the stress field but can also be influenced by the environmental conditions experienced by the rolling bearing.

Conclusions

Butterfly wings, straight growing cracks with non-etching borders, and branching cracks systems with non-etching borders have the same origin, have the same morphology, and are different development stages of the same phenomenon.

The micro-crack development associated with butterfly wing development can grow to substantial sizes and has mechanical properties that are similar to those of alumina oxide inclusions.

Micro-cracks associated with butterfly formations pre-dominantly form at complex sulfide inclusions with oxide encapsulations located in positions with matrix contact and do not generally form at plain sulfide inclusions or at titanium carbonitrides.

No simple relationship between contact stress and micro-crack formation can be established as this is related to applied stress, micro-inclusion alignment in the stress field, local matrix conditions, and non-metallic inclusion composition and shape.

Heat treatment conditions do not significantly affect the development of butterfly wing formations in through hardened bearing components. The micro-inclusion size distribution and the hot forming reduction do however, and the most favourable inclusion shape to delay micro-crack development is well dispersed, small spherical inclusions in a well hot reduced matrix.

Growth from micro-cracks at non-metallic inclusions into extending crack systems with non-etching borders occur, and with extending number of stress

cycles, the share of the micro-crack formations at non-metallic inclusions that do extend increases.

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Initiation Behavior of Crack Originated from Non-Metallic Inclusion in Rolling Contact Fatigue

ABSTRACT: Recently, it has been generally accepted that flaking life is dependent on the size of non-metallic inclusion under proper condition in bearing use. Statistics of extreme values to predict the maximum non-metallic inclusion size and ultrasonic test to assess large non-metallic inclusions in a large volume are widely used as practical methods for the evaluation of bearing steel cleanliness. *Murakami's formula* is well known, which describes the relationship between non-metallic inclusion size and fatigue strength. However, the formula cannot be directly applied to rolling contact fatigue because of the difference in applied stress mode. While the equation was invented to predict the fatigue limit in principal stress mode, rolling contact fatigue is mainly caused by shear stress. It is expected that the condition of bearing use will be more critical due to the downsizing tendency in various industrial or automotive machine units. Thus the research on the flaking mechanism becomes more important from the viewpoint of global ecology because that is beneficial to the improvement in bearing life and methods for cleanliness evaluation. Due to the difficulty in experimental observation, however, the details of the flaking mechanism in rolling contact fatigue caused by non-metallic inclusion has not yet been clarified. Focusing on the relationship between non-metallic inclusion and initiation of the crack, the flaking mechanism is proposed in this paper.

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KEYWORDS: bearing steel, rolling contact fatigue life, crack initiation, crack propagation, non-metallic inclusions, oxides, hot isostatic pressing

Introduction

Recently, various machine units tend to be downsized for the solution of the global warming issue. Thus the improvement of life and reliability has been demanded by bearings. Under a situation like this, the improvement of the fatigue life and cleanliness evaluation has been more important than before in bearing steels. Although the rolling contact fatigue (RCF) processes should be clarified to realize these improvements, they are not yet clear in detail. This is due to the difficulties of observations of crack initiation and propagation because the fracture surfaces suffer from wear by shear stress in a compressive field in RCF.

On the other hand, it is well known that the fatigue limit of steel can be estimated from the non-metallic inclusion size by Murakami's formula [1]. The formula was built on the basis of principal stress and fracture mechanics, where the size of non-metallic inclusion is assumed to be the crack length and the fatigue limits derived from the threshold stress intensity to the crack propagation rate. As the shear stress dominates in RCF and the concerned property is the fatigue life rather than the fatigue limit in RCF, it is not appropriate to apply the formula for RCF directly.

Nagao et al. [2] studied the relationship between the maximum non-metallic inclusion size predicted by statistics of extreme value and RCF life, L_{10} . They reported that L_{10} decreased as the predicted maximum inclusion size increased, where the life varied over one order of magnitude even at the same inclusion size. It was supposed that the RCF life is affected not only by the inclusion size but also by other factors.

In the present paper, flaking processes in RCF are attempted to be clarified from the view-point of initiation and propagation behavior of cracks originated from non-metallic inclusions. The first half of this paper will review the flaking processes through previous studies, and the latter half will verify the processes by experiments.

Review of Previous Studies

Previous studies on the crack behavior in RCF are summarized as follows.

Furumura et al. [3] reported on the stress cycles spent for butterfly generation in their attempt to find the fatigue limit of RCF. In this experiment, Japanese Industrial Standards (JIS) SUJ2 was chosen for the material of the specimens, and thrust-type testers were applied to the RCF test. The maximum contact stress $P_{\max}$ was 3.43 GPa. It is reported that a butterfly generated at approximately 1×10^5 cycles, which was about 1/1000 of L_{10} . The report suggests that the crack initiates at an extremely early stage of the life period, as the butterfly's wings are accompanied by cracks around the non-metallic inclusions [4].

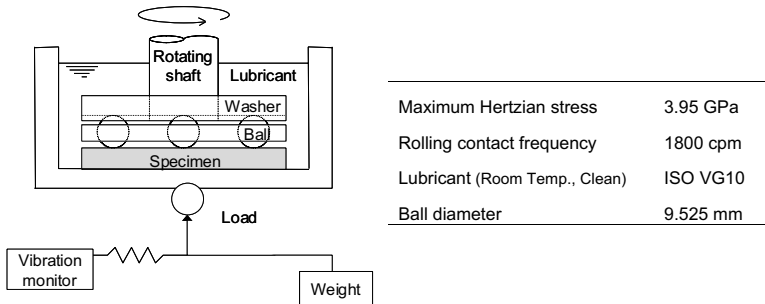


FIG. 1—Conditions and schematic diagram of thrust-type RCF tester.

Nélias et al. [5] investigated the behavior of crack initiation and propagation through a high frequency ultrasonic method. The testing material was American Iron and Steel Institute (AISI) M50, which contained many primary carbides since the carbides could be origins in RCF failure; the observation was expected to be easier. A high-speed twin-disk machine at $P_{\max} = 3.5$ GPa was applied to the RCF test. In their report, the echographic analysis showed that cracks were generated by 5×10^5 cycles and their very slow propagation followed.

Udagawa et al. [6] reported the crack initiation and propagation behavior around non-metallic inclusions by ultrasonic testing. JIS SUJ2 was chosen for this experiment, and thrust-type testers were applied to the RCF at $P_{\max} = 5.3$ GPa. They identified that the crack initiated at approximately 1×10^4 cycles by the growth of a defect echo image in the specimen flaked at 2.3×10^6 cycles. They made sure that such growth of a defect echo image corresponded to the crack initiation by a serial sectioning method in the previous study [7].

From the three papers above, it is strongly suggested that the RCF life caused by non-metallic inclusion consists of an extremely early crack-initiation stage and very long propagation stage. That is to say, the crack propagation substantially controls the RCF life.

We [8] reported the experimental results on the crack initiation behavior. In the paper, crack initiation behavior was estimated through the observation of a lot of cracks. The specimens contained artificial pores with a $25 \mu\text{m}$ or smaller diameter instead of scarce non-metallic inclusions as crack initiation sites. They were prepared by the powder metallurgy of hot isostatic pressing (HIP). JIS SUJ2 was chosen for the materials, and thrust-type testers in Fig. 1 were applied to the RCF at $P_{\max} = 3.95$ GPa. Each specimen was cut along the tangential line of the track and observed mainly by scanning electron microscope (SEM). Figure 2 shows the SEM images of cracks initiated from pores at 5×10^4 cycles. It was found that the cracks initiated from pores at a very early stage of RCF life as reported in the papers above [3,5,6], which used specimens bearing natural defects. Figure 3 shows the relationship between distance from the track surface and frequency of occurrence of cracks around pores. 60 % or more of the pores were accompanied by cracks regardless of pore size in the

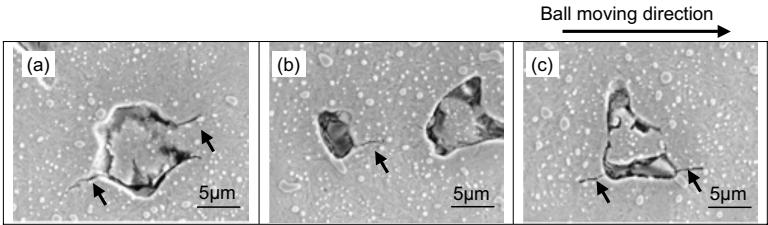


FIG. 2—Secondary electron images of typical pores with cracks after the thrust-type RCF of 5×10^4 cycles.

region from surface to 0.2 mm in depth. These observations showed that the crack initiation could not be described by the distribution of shear stress, which is known as a dominant factor of crack initiation in RCF [9,10], and that the cracks initiated or did not from pores at the same depth. It is suggested that the behavior of cracks that originated from non-metallic inclusions in RCF were not always controlled by the original defect size especially in the crack initiation stage.

In addition, the stress working on the crack initiation from pores was identified by computational simulation in the paper. Figure 4 shows the simulated maximum principal stress around pores with 20 μm in diameter at a different depth from the surface. The depth of the high tensile principal stress area corresponded well to that of the crack initiating area in Fig. 3. Moreover, the observed crack angles with the rolling direction in Fig. 2 agreed to the simula-

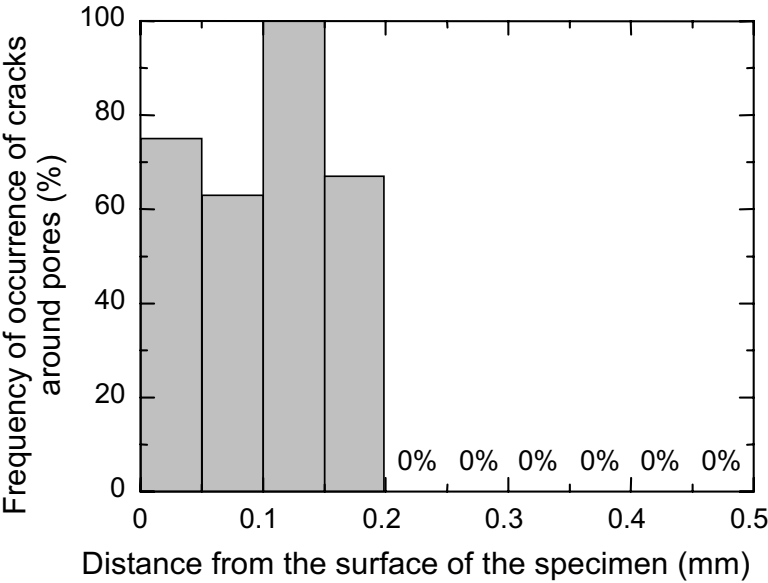


FIG. 3—Frequency of occurrence of cracks around pores after the thrust-type RCF of 5×10^4 cycles.

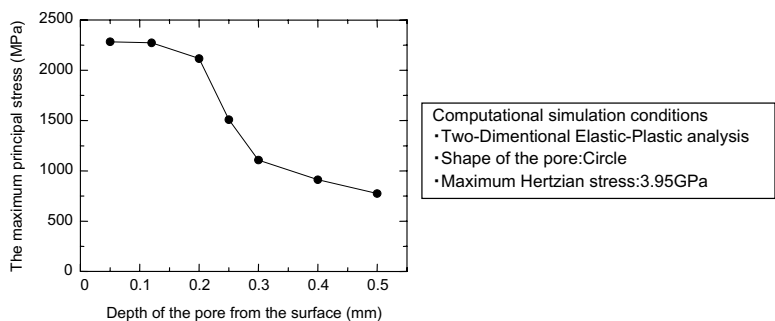


FIG. 4—The maximum principal stress around the pore 20 μm in diameter at different depths from the surface.

tion results shown in Table 1 shows the physical and mechanical properties of the matrix, non-metallic inclusions, and pore. As the Young’s modulus of non-metallic inclusion decreased, the tensile principal stress was likely to increase. It was found that the tensile principal stress around non-metallic inclusions depended on their constituents.

The inclusion-matrix interface was presumed to be bonded in this simulation. The tensile principal stress around the inclusion was quite lower than that around the pore. It is too low to initiate the mode I cracks especially from oxide inclusions with a high Young’s modulus. However it is known that there is a possibility of debonding between some types of non-metallic inclusions and the matrix [11]. In that case, tensile principal stress around the non-metallic inclusions is expected to increase close to that of the pores, which is detrimental to RCF life.

Previous studies were reviewed and the mechanisms of flaking processes in RCF are summarized as follows.

- (1) The crack initiates at an extremely early stage of life period, approximately 1/1000 of RCF life, and the crack propagation period substantially dominates the RCF life.
- (2) The direct life control factor in RCF is the initial crack size “2a” in Fig. 6 because the crack propagation rate “da/dN” is controlled by the crack size in fracture mechanics.

TABLE 1—The results of stress analysis around the pore.

Depth of the Pore (mm)	The Maximum Tensile Stress (MPa)	The Position and the Direction of the Maximum Tensile Stress (Ball Moving Direction: Left-to-Right)	The Direction of the Maximum Tensile Stress Plane
0.05	2286		14.1°
0.12	2270		17.6°
0.20	2122		29.2°

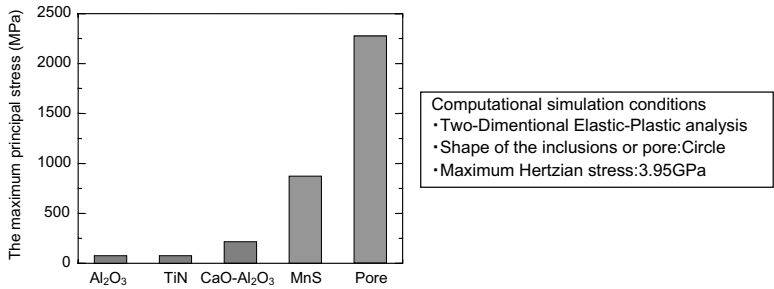


FIG. 5—The maximum principal stress at the inclusion—matrix interface or surface of the pore. Inclusions and the pore are 20 μm in diameter and 0.12 mm deep from the surface.

- (3) The initial crack size is affected not only by inclusion size but also the composition and interface condition of non-metallic inclusions through the change of stress concentration around the non-metallic inclusions.

In order to prove the hypothetical mechanisms, the following original samples were RCF tested, where crack initiation behavior was expected to vary depending on different stress conditions around non-metallic inclusions.

Experimental

RCF tests were carried out on samples containing different size, chemical composition, and interface condition of non-metallic inclusions to vary their initial crack size.

Oxide Modification

Experimental Procedure—The non-metallic inclusions as crack initiation sites in RCF are generally Al₂O₃ in aluminum deoxidized steels. Three heats of JIS SUJ2 were prepared by different deoxidation methods in 100 kg vacuum induction melting (VIM). The chemical compositions of specimens are shown

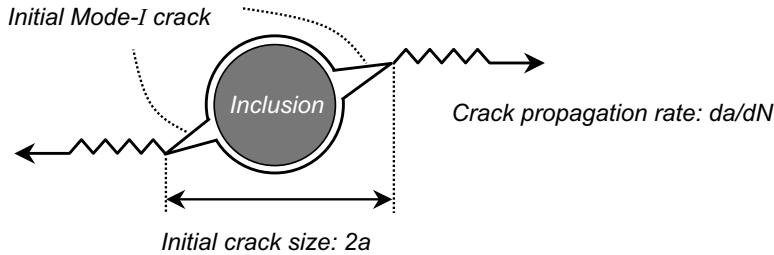


FIG. 6—Model of crack initiation and propagation originated from non-metallic inclusion.

TABLE 2—Physical and mechanical properties of the matrix, inclusions, and pore.

	Matrix	Al ₂ O ₃	TiN	CaO–Al ₂ O ₃	MnS	Pore
Young's modulus (GPa)	206	387	316	113	137	0
Tangent modulus (MPa)	85.8	...	...	...	...	...
Poisson ratio	0.30	0.25	0.192	0.25	0.25	0
Hardness (HV)	700	~1900	~2400	~2200	150	0
Yield strength (MPa)	1960(=σ _y)	3σ _y	3σ _y	3σ _y	σ ₅ /5	0

in Table 3. Sulfur content is decreased to minimize the effect of manganese sulfide inclusions to RCF. Steels A and B were deoxidized by aluminum to introduce Al₂O₃. The oxygen contents of steel A and B are 11 and 30 ppm, respectively. Steel C was not deoxidized by aluminum to provide SiO₂·Al₂O₃ type inclusions. The oxygen content of steel C is 28 ppm, which is close to that of steel B.

Ingot of the three heats were forged to 65 mm diameter bars at 1423 K, normalized at 1143 K for 3.6 ks, air cooled to room temperature, and spheroidized annealed at 1073 K maximum for 54 ks. After the heat treatment, they were machined to the shape for a RCF test specimen, i.e., 60 mm in outer diameter, 20 mm in inner diameter, and 5.8 mm in thickness. These specimens were austenitized at 1108 K for 1.8 ks, quenched into oil at 323 K, and tempered at 443 K for 5.4 ks with the resultant Rockwell hardness test (scale C) of 62 HRC.

The non-metallic inclusion size was evaluated by statistics of extreme value in accordance with the in-house standard [12]. A longitudinal section of forged bars was observed by microscopy to determine the each largest inclusion size of each oxide in a control area of 100 mm². The measurement was repeated for 30 fields, and the resultant extreme values were statistically treated to predict the maximum inclusion size, (AREA_{max})^{1/2}, in a reference area of 30 000 mm². The compositions of non-metallic inclusions were analyzed by SEM-energy dispersive X-ray spectrometer, and the inclusion-matrix interfaces of quenched and tempered specimens were observed by field emission-SEM after polishing by ion milling. Thrust-type testers were used to run the RCF tests, as shown in Fig. 1. The calculated maximum contact stress under Hertzian elastic theory, $P_{\max}$, was 5.3 GPa. Although the 5.3 GPa results in local plastic deformation, it is empirically acceptable for the evaluation of the RCF life associated with inclusions.

TABLE 3—Chemical compositions of the specimen, mass %.

Steel	C	Si	Mn	P	S	Ni	Cr	Mo	Cu	Al	O
A	0.98	0.20	0.39	0.015	0.001	0.10	1.39	0.03	0.10	0.013	0.0011
B	0.98	0.21	0.41	0.014	0.002	0.10	1.40	0.03	0.11	0.013	0.0030
C	0.99	0.22	0.41	0.017	0.002	0.11	1.41	0.03	0.11	0.001	0.0028

TABLE 4—Oxides type and predicted size by E.V.A., mass %.

Steel	Oxides Type	Predicted $(\text{AREA}_{\text{max}})^{1/2}$ (μm)
A	Al_2O_3	26
B	Al_2O_3	51
C	$\text{Al}_2\text{O}_3 \cdot \text{SiO}_2$	52

Experimental Results—Table 4 shows the representative type and $(\text{AREA}_{\text{max}})^{1/2}$ of the oxides in each heat. The basic chemical compositions of oxides were Al_2O_3 in steels A and B and $\text{SiO}_2 \cdot \text{Al}_2\text{O}_3$ in steel C as designed. The $(\text{AREA}_{\text{max}})^{1/2}$ were 26, 51, and 52 μm in steels A, B, and C, respectively. In this experiment, the oxide size tends to be larger when the oxygen content is higher regardless of deoxidation methods. Figure 7 shows the SEM micrographs of representative oxides. It was observed that there were apparent cavities around Al_2O_3 , while the $\text{SiO}_2 \cdot \text{Al}_2\text{O}_3$ -matrix interface was seemed to be closed.

Figure 8 and Table 5 show the results of the RCF test. As we expected, the life of steel A was longer than that of steel B because the $(\text{AREA}_{\text{max}})^{1/2}$ of steel A was smaller than that of steel B. Both of the heats contained the same Al_2O_3 type inclusions. On the other hand, the life of steel C was longer than that of steel A, although steel C exhibited a similar oxygen content and large $(\text{AREA}_{\text{max}})^{1/2}$ to those of steel B.

RCF life can be improved by reducing the inclusions size when the chemical compositions of the non-metallic inclusions are the same. In addition, it is suggested that the life can be improved by changing the composition of non-metallic inclusions by the reducing the cavities between the non-metallic inclusions and the matrix.

Hot Isostatic Pressing Treatment

Experimental Procedure—The results in the Oxide Modification section included the effects of the differences in physical properties and morphology of

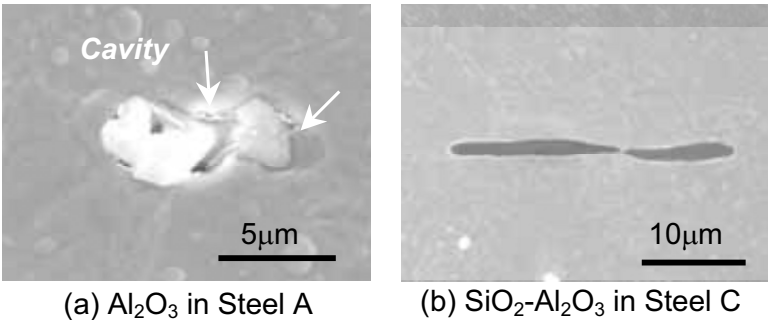


FIG. 7—Secondary electron images of typical oxides in the specimens after quenching and tempering.

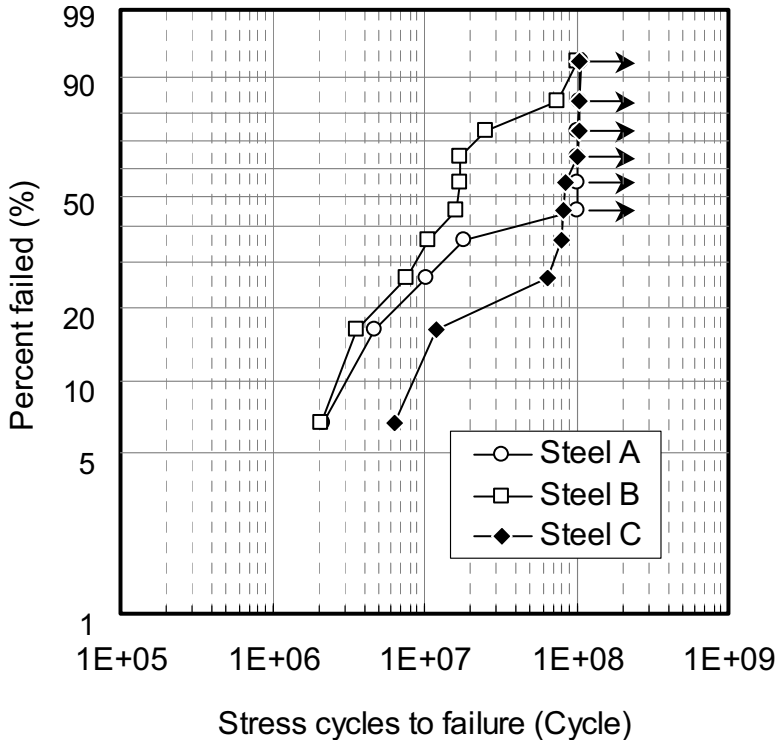


FIG. 8—The results of the thrust-type RCF test with specimens including different oxide inclusions ($P_{\max}=5.3$ GPa).

the oxide inclusions. In order to extract only the effect of the inclusion-matrix interface condition on the RCF life, HIPed and non-HIPed steels were RCF tested. The aim of the HIP treatment was to close the interface between the Al_2O_3 inclusion and the matrix. The 65 mm diameter forged bars of steels A and B in Table 2 were HIPed at 1423 K for 18 ks in 147 MPa. The specimens for thrust-type RCF were prepared by the same process as in the Experimental Procedure subsection of Oxide Modification. The hardness was 62 HRC. The observation procedure of the inclusion-matrix interface and test condition of RCF are also the same as in the Experimental Procedure subsection of Oxide Modification.

TABLE 5— L_{10} and L_{50} of thrust-type RCF test with specimens including different oxide inclusions.

Steel	L_{10} (cycles)	L_{50} (cycles)
A	3.13×10^6	27.49×10^6
B	2.59×10^6	16.68×10^6
C	10.48×10^6	101.58×10^6

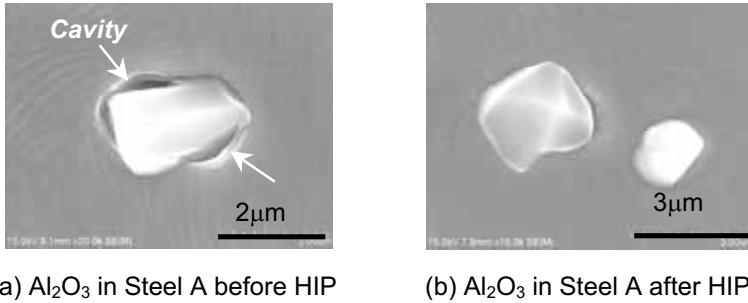
(a) Al_2O_3 in Steel A before HIP(b) Al_2O_3 in Steel A after HIP

FIG. 9—Secondary electron images of typical oxides in the specimens before and after HIP treatment.

Experimental Results—The main chemical compositions of oxides were Al_2O_3 in steels A and B. Some cavities were observed around the oxides in the specimens without HIP treatment, while it was confirmed that the cavities were closed in the HIPed specimens, as shown in Fig. 9.

Figure 10 and Table 6 show the results of RCF test. The life of steel A with low oxygen content and small $(\text{AREA}_{\text{max}})^{1/2}$ was improved by HIP treatment in the short life area in the Weibull distribution. The life of steel B with high oxygen content and large $(\text{AREA}_{\text{max}})^{1/2}$ was improved by HIP treatment as a whole. It has become clear that the life can be improved by simply closing the gap between the inclusion-matrix.

Discussion

The experimental results above proved the mechanisms of crack initiation and propagation originated from non-metallic inclusions in RCF, which were presumed from the review of previous studies including ours. Figure 11 shows the model of the clarified mechanisms of RCF in the present paper.

The crack initiates from non-metallic inclusions at the early stage around 1/1000 of the RCF life period. It is supposed that the plastic deformation area is already formed by RCF around the non-metallic inclusions prior to the crack initiation stage [5,13,14]. Then the crack propagation stage follows as the substantial RCF life. In fracture mechanics, the crack propagation rate “ da/dN ” depends on the stress intensity factor range (ΔK_{II}), which is derived from crack size and stress. The RCF life is controlled by the initial crack size “ $2a$ ” in Fig. 6. Since the initial crack size is affected not only by inclusion size but also by the stress condition around it, the interface condition between non-metallic inclusions and matrix is one of the critical factors affecting RCF life. Although the conditions of the bonded/debonded interface were not fully verified in this paper, it was observed that closing the gap identified by SEM at the interface resulted in longer RCF life.

From the viewpoint of this mechanism, it may be said that a rational way to improve the RCF life is to decrease the size of non-metallic inclusions in a

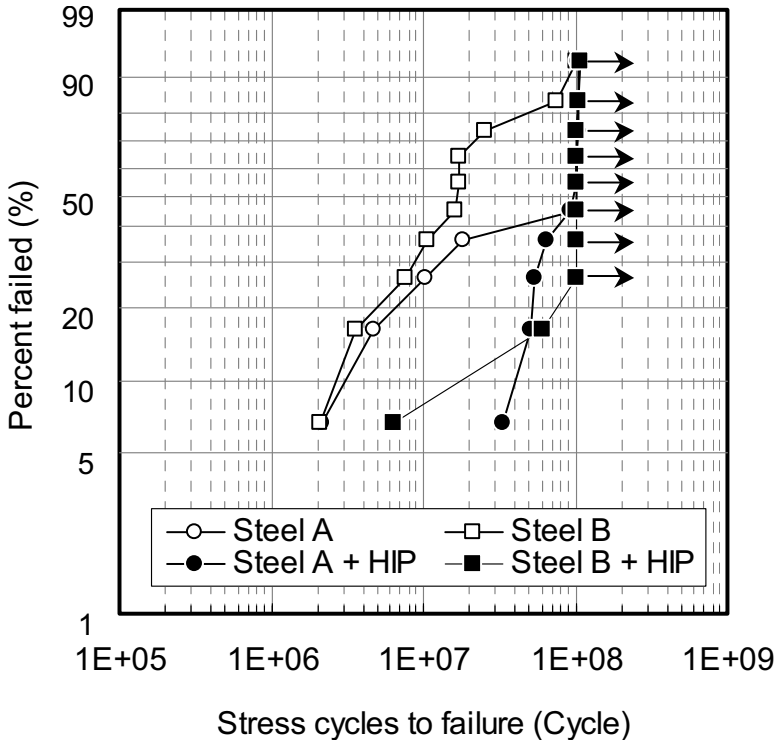


FIG. 10—The results of the thrust-type RCF test with specimens treated by HIP ($P_{\max} = 5.3$ GPa).

conventional manner [15]. In addition, the reduction of stress concentration around inclusions to minimize the initial crack size by HIP treatment, for example, may be valuable for further improvement of the RCF life. It is an important subject to develop more advanced cleanliness evaluation methods including analysis of the inclusion's composition to realize precise prediction of RCF life.

Conclusions

From the review of previous studies, it was presumed that the size of an initial crack originated from non-metallic inclusions at the very early stage of RCF life was the critical factor of life since it influenced the crack propagation rate.

TABLE 6— L_{10} and L_{50} of thrust-type RCF test with specimens treated by HIP.

Steel	L_{10} (cycles)	L_{50} (cycles)
A (+HIP)	38.53×10^6	88.21×10^6
B (+HIP)	17.28×10^6	161.82×10^6

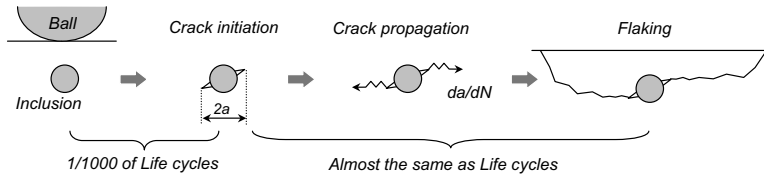


FIG. 11—Model of flaking process caused by non-metallic inclusion in RCF.

Furthermore it was suggested that the size of the initial crack changed depending not only on the size of the inclusion but also on the interface condition between the inclusions and the matrix because the presence of the cavity affects the stress field around them. In this report, this theory was supported through the two experiments below.

- (1) RCF life was improved by oxide modification from Al_2O_3 type inclusions to $\text{SiO}_2\text{--Al}_2\text{O}_3$ type inclusions. The cavities were identified between Al_2O_3 type inclusions and the matrix by SEM observation. They were probably introduced by hot forging due to the difference in deformability between them. On the other hand, a cavity was not observed around $\text{SiO}_2\text{--Al}_2\text{O}_3$ type inclusions. Therefore, it is suggested that the RCF life is improved by closing the gap between the inclusions and the matrix through reducing the initial crack size.
- (2) To clarify the effect of the interface condition between non-metallic inclusions and matrix on RCF life, the aforementioned steels with Al_2O_3 type inclusions were HIP treated to close the gap between the inclusions and the matrix. The HIP treatment resulted in remarkable improvement in the RCF life.

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Initiation and Propagation Behavior of Crack Originated from Non-Metallic Inclusion in Rolling Contact Fatigue, Dr. Norimasam Tsunekage, Mr. Kazuya Hashimoto, Dr. Takeshi Fujimatsu, and Dr. Kazuhiko Hiraoka; Journal of ASTM International Selected Technical Papers, STP1524, Bearing Steel Technologies: Developments in Rolling Bearing Steels and Testing, 8th Volume.

Page 3, Paragraph 2, Sentences 5 and 6 were omitted. The paragraph should read:

“...shown in Table 1. *It is strongly suggested that the tensile principal stress is concerned with the crack initiation. Since it was unsure that natural non-metallic inclusions might be treated as pores from the viewpoint of stress concentration, the maximum principal stress around non-metallic inclusion was computationally simulated with the results in Fig.5, where the inclusion-matrix interface was assumed to be bonded.*

Table 2... shows the physical and mechanical properties of matrix, non-metallic inclusions and pore.”

Erick S. Alley,¹ Krich Sawamiphakdi,² Patrick I. Anderson,² and Richard W. Neu³

Modeling the Influence of Microstructure in Rolling Contact Fatigue

ABSTRACT: Crack nucleation, first spall generation, and spall growth in rolling contact fatigue (RCF) are known to be highly sensitive to the heterogeneity of the microstructure. Yet the current state-of-the-art in the design of high performance bearing materials and microstructures is highly empirical requiring substantial lengthy experimental testing to validate the reliability and performance of these new materials and processes. We have laid the groundwork necessary to determine the influence of microstructure in RCF to aid in the development and processing of bearing steels. Microstructure attributes that may control the fatigue behavior are explicitly modeled in a 41xxx steel. The methodology is demonstrated by studying the role of an aluminum oxide inclusion embedded in a matrix of tempered martensite and retained austenite. The matrix is represented by crystal plasticity, which provides more realistic accumulations of localized plastic strains with cycling compare to homogenized J_2 plasticity. As a demonstration of the approach, the relative influence of the volume fraction of retained austenite on RCF is evaluated.

KEYWORDS: bearing steel, rolling contact fatigue, crystal plasticity, non-metallic inclusion, martensite, retained austenite, transformation

Introduction

Crack nucleation, first spall generation, and spall growth in rolling contact fatigue (RCF) are known to be highly sensitive to the heterogeneity of the mi-

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microstructure at the microscopic level. Features in the microstructure include non-metallic inclusions, either hard (e.g., Al_2O_3 and TiN) or soft (e.g., MnS), the size, morphology, and volume fraction of the different phases (e.g., martensite, retained austenite, bainite patches), and the crystallographic orientation of each of the phases relative to each other [1]. Furthermore, the elastic and plastic properties at the crystalline scale are highly anisotropic. The typical assumption of homogeneous, isotropic material, even with consideration of embedded inclusions, is often highly idealistic. Because of the difficulty in predicting the influence of microstructural heterogeneity on RCF life, the current state-of-the-art in the design of high performance bearing materials and microstructures continues to be highly empirical requiring substantial lengthy bench testing to validate the reliability and performance of new materials and processes.

It is well known that the empirical parameters correlated to bench tests are sensitive to the microstructural attributes of the material [2]. Based on fracture mechanics arguments, the fatigue life is known to be inversely proportional to the square-root of the size of the non-metallic inclusions [3]. Many other attributes of the microstructure can influence RCF. These include inclusion composition, size, orientation, their distribution and volume fraction, interfacial/particle strength, grain size, morphology and distributions, microtexture, compositional variations (e.g., banding), microhardness, percentage of retained austenite, and residual stress. Experimental work to evaluate the influence of all of these attributes on bearing fatigue can be costly and can take long amounts of time to complete, and thus it is desirable to have a numerical model for predicting the influence of different microstructural attributes on RCF. With a numerical model, many different microstructural realizations can potentially be evaluated more efficiently to help reduce the number of costly and time-consuming bench tests and hence to more quickly hone in on optimum processing conditions.

In RCF, the crack often initiates at a subsurface location where the cyclic shear stresses are greatest. Hence, the heterogeneity of the microstructure plays a more important role compared to typical fatigue scenarios where cracks form on the surface. There are key microstructural attributes known to be associated with RCF. For example, the formation of “butterfly wings,” typically around hard Al_2O_3 inclusions, as seen in Fig. 1, is driven by the accumulated deformation accentuated by the difference in the mechanical properties of the inclusion, matrix, and interface [1,4]. The location and orientation of the altered microstructure adjacent to the inclusion is not random but is linked to the directions of the local cyclic deformation in the vicinity of the inclusion [1,5,6]. Inclusions can appear singly, as shown in Fig. 1, or grouped together in long chains known as stringers. When grouped in close proximity, these are often treated as a single, elongated inclusion. The orientation of such stringers can have a significant effect on fatigue performance.

The amount of retained austenite (RA) affects the fatigue performance of steel based on the loading conditions, i.e., differently for RCF as opposed to bending fatigue. For bending fatigue, higher amounts of RA are detrimental to high cycle fatigue (HCF), while RA can be beneficial for low cycle, high strain fatigue [7], which suggests that RA should be detrimental for RCF. However, for

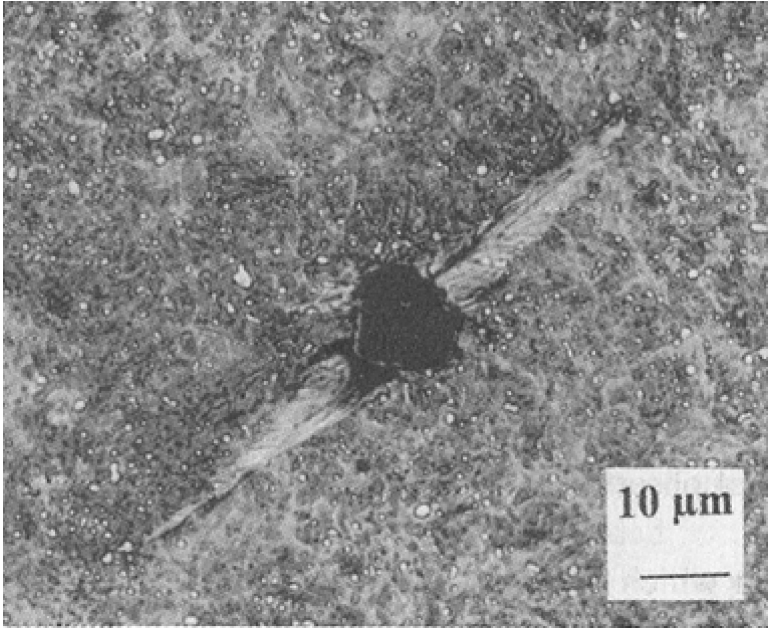


FIG. 1—White etching near Al_2O_3 inclusion [5].

some RCF conditions, the fatigue life has been found to increase as RA increases [8]. One beneficial role of RA in RCF is preventing fatigue cracking from debris dents, most likely due to the compressive residual stress field that forms around the debris dents due to transformation of austenite to martensite [1]. Since increased amounts of RA reduce the surface hardness, it is important to balance the appropriate amount of RA with the anticipated operating conditions. This balance is critical, as the benefit of increased life in debris containing environments can be negated by transformation-induced distortion when RA transforms to untempered martensite during RCF conditions [1]. When RA transforms to martensite, there is a volumetric transformation strain that needs to be modeled since it clearly alters the local residual stresses, which in turn affect the fatigue drivers.

Under long-term RCF conditions, changes in the microstructure are often observed in the subsurface regions undergoing the greatest loading. One change is the formation of microtexture, which is a preferred orientation of the crystallographic grains [9,10]. Generation of microtexture depends on loading, temperature, and number of cycles. Bearings that run cool at extreme loads tend to form deformation textures, whereas bearings that run warm at nominal loads tend to promote recrystallization textures. Texture has been shown to influence the character of the spall that forms under RCF [10].

Bearing steels comprise a complex hierarchical heterogeneous microstructure of mostly lath martensite in austenite. Plastic deformation is driven by the magnitude of resolved shear stress on individual slip systems. The resolved shear stress depends on the crystal structure and orientation with respect to the

loading. In polycrystalline and multiphase systems, the resolved shear stress also depends on the intergranular constraints and anisotropic properties of the matrix and inclusions. Hence, a crystal plasticity formulation is needed to capture the highly non-uniform cyclic strain field developed due to the heterogeneous microstructure. In HCF problems, crystal plasticity has been shown to be a much more realistic model to predict the localized cyclic deformation [11–13]. In fact, when using a conventional J_2 plasticity formulation to model HCF and RCF, the analysis almost always predicts the response remains elastic, yet it is well-understood that fatigue is driven by the accumulation of plastic strains locally in the microstructure [14]. Crystal plasticity captures these local plastic strains under conditions when HCF life is known to be finite. Because of the fine microstructure of steels that is comprised of multiple phases and structures, they are more challenging to model using crystal plasticity than other classes of metals. To reduce the complexity, polycrystal plasticity models typically assume that steel can be described by a single phase (i.e., all of the crystalline grains have the same crystal structure and properties). Single-phase crystal plasticity have been used to model low carbon steels [15], HSLA steels [16], pressure vessel steels [17], austenitic stainless steel [15], and martensitic gear steels [18]. Simulations using crystal plasticity have been used to study the influence of texture on fatigue of a HSLA steel [19] and the influence of intergranular constraints in both low carbon steel and austenitic stainless steel [15]. Both studies clearly showed the importance of the local ratcheting as a driver for fatigue damage.

In this paper, a new two-phase crystal plasticity model for martensitic bearing steels containing RA is described. Prior crystal plasticity models of steels only considered a single phase. This material model is coded as a User MATERIAL (UMAT) subroutine for the general-purpose finite element code ABAQUS [20]. Using a volume element approach described in a companion paper [14], a fully three-dimensional finite element simulation of RCF is performed to capture end effects of elongated aluminum oxide inclusions not possible with two-dimensional plane-strain models, providing for a more realistic assessment of inclusion morphology and arbitrary orientations. The scaling of the finite element model is optimized to capture the cyclic microplasticity around the inclusion accurately and efficiently. The links between the microstructural attributes and RCF are evaluated using fatigue indicator parameters. Using the new material model, the influence of the volume fraction of RA on RCF is examined.

Representative Bearing Steel Microstructure and Properties

The 41XXX series steel was chosen as a representative bearing steel for this study. The microstructure consists of lath martensite formed in a prior austenite grain. Pockets of RA can remain between the laths depending on heat treatment. The response behavior of these steels can vary greatly by the initial volume fraction of RA, heat treatment, and distribution of primary carbides.

Two variations with different carbon content, 4145 and 41100, were used to calibrate the material model. The heat treatment of 41100 steel was varied to control the initial volume fraction of RA, as illustrated in Fig. 2. Three different

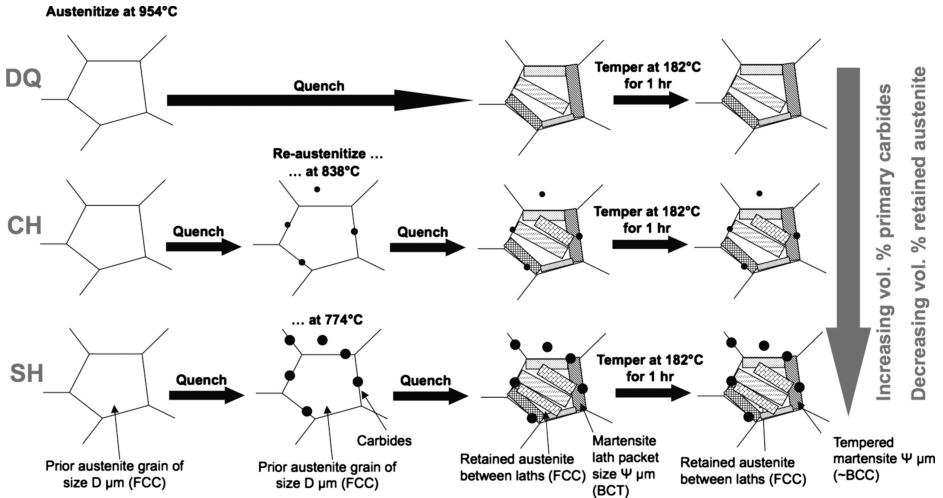


FIG. 2—Heat treatments for 41100 steel.

heat treatments were used. The direct quench resulted in the greatest amount of RA. It was 35 %, as measured by x-ray diffraction. A conventional heat treatment (CH) resulted in 22 % RA and a special heat treatment resulted in 13 % RA. The lower levels of RA were primarily due to the increasing amount of primary carbides that form during the re-austenitization step. The 4145 steel was processed using the CH.

Optical images of the microstructures are shown in Fig. 3. The 4145 material is nearly fully martensitic, as indicated by the densely packed martensitic laths in Fig. 3(a). Thus it can be reasonably treated as a single-phase martensite component and is used to calibrate the martensitic plasticity portion of the material model. The 41100 steel with 13 % RA, shown in Fig. 3(b). As the volume fraction of RA increases to 22 % [Fig. 3(c)], visible portions of RA (lighter shade areas) are detectable and when RA is 35 % [Fig. 3(d)], significant RA is visible. As seen in Fig. 3(c), the RA distribution is also often inhomogeneous, with some prior austenite grain regions containing more RA than others. This can be accounted for and studied using the newly developed model described herein by assigning individual elements or element regions different initial volume fractions of RA.

Uniaxial mechanical tests were conducted on cylindrical dog-bone specimens with a gage diameter of 6.35 mm (0.25 in.) and gage length of 12.7 mm (0.50 in.). Both the axial and diametral strains were measured simultaneously using two extensometers. The experiments were conducted at room temperature and a strain rate of $1.0 \times 10^{-4} \text{ s}^{-1}$. The tensile response of the 4145 steel is shown in Fig. 4 and the responses of the three 41100 steels are shown in Fig. 5. Note that the response of the fully martensitic 4145 steel is similar to the 41100 steel with lowest amount of RA. With increasing RA, the apparent yield strength decreases due primarily to the transformation of RA to martensite.

Since there is a volume change associated with the transformation, the

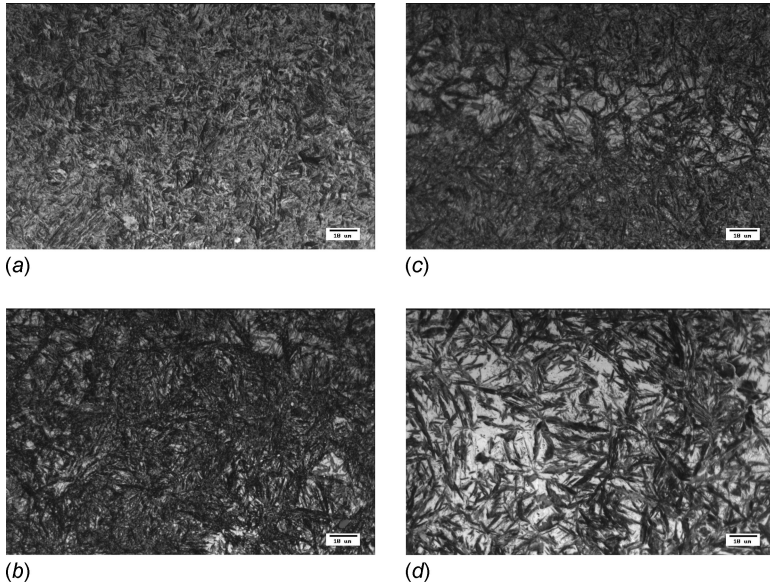


FIG. 3—Optical images of (a) 4145 steel, (b) 41100 steel with 13 % RA, (c) 41100 steel with 22 % RA, and (d) 41100 steel with 35 % RA.

amount and evolution of the transformation can be determined by tracking the volumetric transformation strain, as shown in Fig. 6. The volumetric transformation strain was determined from the strain readings after subtracting the elastic strain, assuming plastic strains do not cause any volume change, following the procedure described in Neu and Sehitoglu [21]. The results shown in Fig. 6 confirm that the apparent yielding is due to transformation of RA to martensite.

The uniaxial compression response is shown in Fig. 7. The amount of transformation is reduced in compression compared to tension for the same amount of applied axial strain. This suggests that the transformation is primarily controlled by a stress-assisted mechanism and that hydrostatic stress, which is smaller under uniaxial compression than under uniaxial tension, plays a role in controlling the transformation. In addition, cyclic deformation experiments were conducted to study the evolution of transformation with cycling. The rate of measurable transformation diminishes to negligible levels per cycle after the first couple of cycles under constant amplitude cycling.

Hybrid Transformation and Crystal Plasticity Model

To explore bearing performance in multi-phase alloys, the effects of RA must be considered. The austenite can promote ductility and can also induce residual stress fields due to volume expansion during the phase change. Thus, the two-phase model was developed to account for both stress-assisted transformation

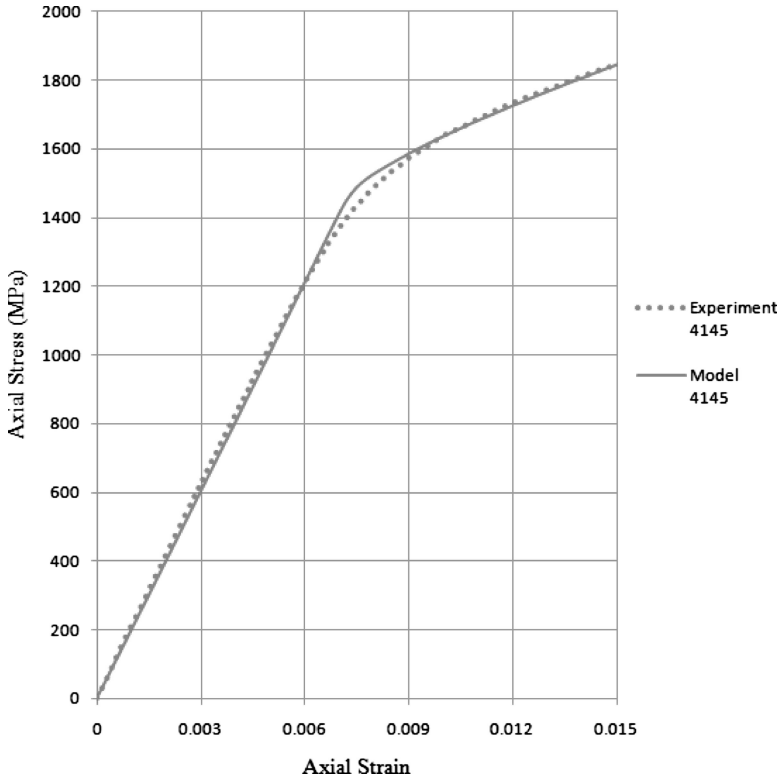


FIG. 4—Uniaxial tensile experimental response and calibration of the two-phase crystal plasticity model for 4145 steel.

and plasticity. The transformation model tracks the evolution of the volume fraction of RA that is transformed, as well as captures the volumetric strains produced by such activity.

Austenite-to-martensite transformation deformation parameters are formulated in a manner described by Gall and Sehitoglu [22], with transformation directions and habit plane normals relative to orientation of a single crystal depicted in Fig. 8. Based on the mechanical tests, the stress-assisted transformation model proposed by Suiker and Turteltaub [23] was identified as the best choice for implementing transformation. Their formulation was chosen over other formulations for steels such as that proposed by Karaman et al. [24] because a key feature of the model is its multiplicative decomposition, consistent with formulations of crystal plasticity models. A rate-dependent crystal plasticity model formulated by Asaro [25] and McGinty [26] was chosen to describe the dislocation slip behavior.

The new two-phase hybrid model is built around a triple multiplicative decomposition of the deformation gradient $\underline{F}$, given as

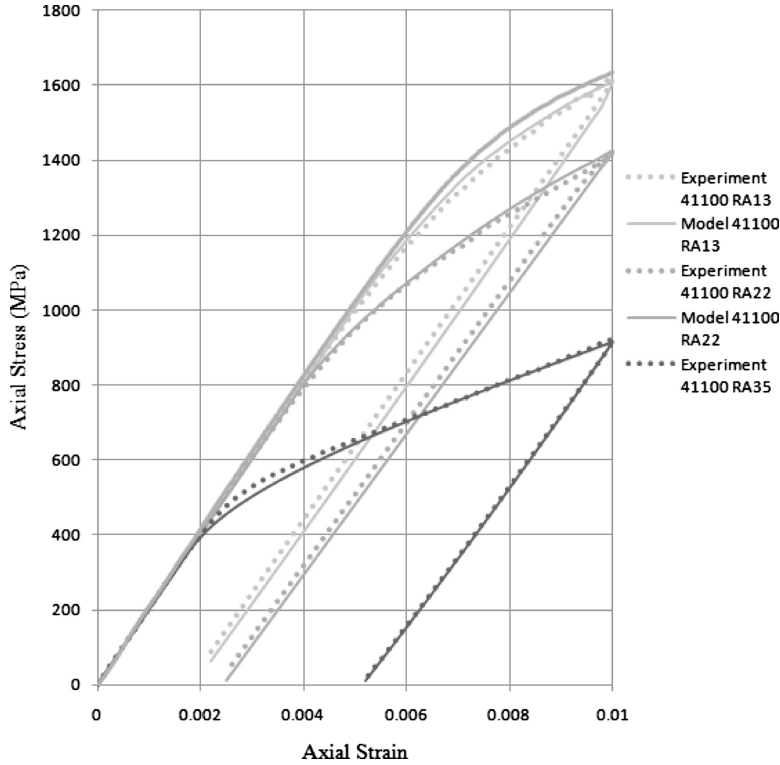


FIG. 5—Uniaxial tensile experimental response and calibration of the two-phase crystal plasticity model for 41100 steel.

$$\tilde{F} = \tilde{F}^e \cdot \tilde{F}^p \cdot \tilde{F}^{tr} \quad (1)$$

where $\tilde{F}^{tr}$ accounts for the volumetric strain produced by the austenite-to-martensite phase transformation, $\tilde{F}^p$ accounts for polycrystalline plasticity, and $\tilde{F}^e$ accounts for elastic deformations and rigid body translations. Plasticity and transformation occur along slip and transformation systems associated with the lattice structures of the martensite and austenite. The evolution of the crystal plasticity follows the formulation work of Asaro [25]. The model uses the 48 body-centered-cubic (bcc) slip systems to represent the behavior of the tempered martensite. This is a reasonable assumption since during the early stages of tempering, the body-centered-tetragonal (bct) lattice of virgin martensite quickly relaxes to obtain a bcc symmetry. The transformation is controlled by the formulation based on Suiker and Turteltaub [23]. This is calculated on the 24 face-centered-cubic (fcc) transform systems for austenite-to-martensite transformation.

For the transformation model, the unit vectors $\hat{b}_0^{(\lambda)}$ and $\eta_0^{(\lambda)}$ correspond to the transformation direction, also referred to as the shape strain direction, and the habit plane normal, respectively, for the λ^{th} transformation system in the

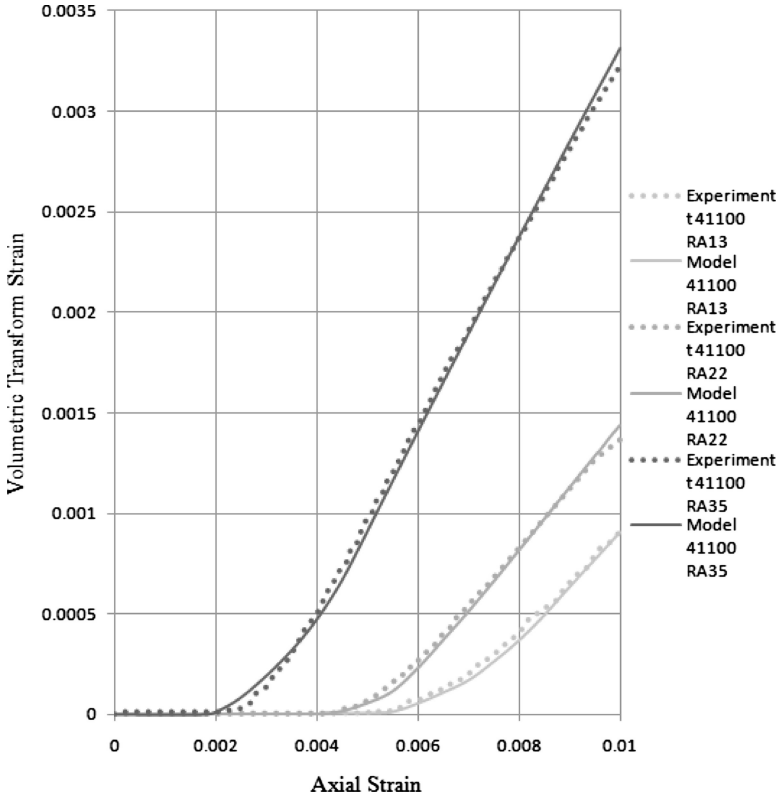


FIG. 6—Measured volumetric transformation strain under uniaxial tensile loading and correlation to the two-phase crystal plasticity model for 41100 steel with different initial volume fractions of RA.

intermediate configuration. Transformation is said to occur when a critical driving stress on a given transformation system is reached. The driving stress $f_{tr}^{(\lambda)}$ on a transformation system λ is related to the transformation and habit vectors, and the Cauchy stress by

$$f_{tr}^{(\lambda)} = \sigma : (\gamma_T \cdot \hat{b}^{(\lambda)} \otimes \hat{n}^{(\lambda)}) \quad (2)$$

where the transformation direction and habit plane normal are rotated into the current configuration by

$$\hat{b}^{(\lambda)} = F^e \cdot \hat{b}_0^{(\lambda)} \quad (3)$$

and

$$\hat{n}^{(\lambda)} = (F^e)^{-T} \cdot \hat{n}_0^{(\lambda)} \quad (4)$$

and γ_T is the shape strain magnitude, a parameter which is uniform for all transformation systems. When the driving stress given by Eq 2 exceeds the

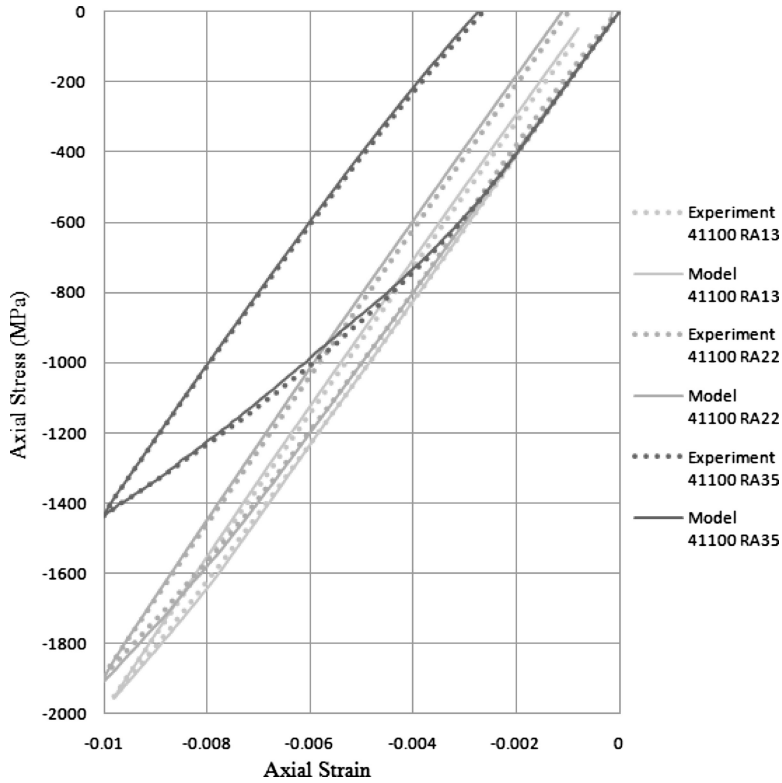


FIG. 7—Uniaxial compression experimental response and prediction of the two-phase crystal plasticity model for 41100 steel.

critical driving stress $f_{cr}^{(\lambda)}$ for the transformation system, the rate of the volume fraction of transformation $\dot{\xi}^{(\lambda)}$ on that system is given by

$$\dot{\xi}^{(\lambda)} = \dot{\xi}_{\max} \cdot \tanh\left(\frac{1}{\nu_{tr}} \cdot \frac{\langle f_{tr}^{(\lambda)} - f_{cr}^{(\lambda)} \rangle}{f_{cr}^{(\lambda)}}\right) \tag{5}$$

where:

$\dot{\xi}_{\max}$ = the maximum rate of transformation, as the hyperbolic tangent reaches a maximum value of one.

The transform viscosity parameter ν_{tr} controls the amount of rate dependence in the transformation. When all of the transformation systems are summed together,

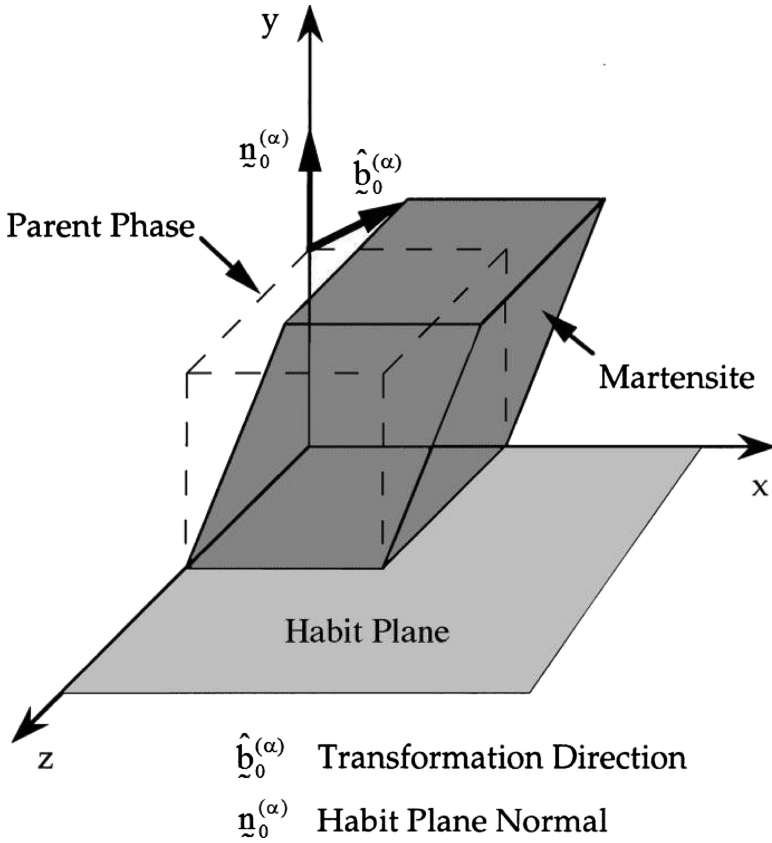


FIG. 8—Transformation direction and habit plane normal [22].

$$\dot{V}_{\text{trans}} = \sum_{\lambda=1}^{N_{\text{trans}}} \dot{\xi}^{(\lambda)} \quad (6)$$

they give the rate of change of the volume fraction transformed from austenite to martensite, which is then multiplied by the time step, Δt , to obtain the total volume fraction transformed,

$$V_{\text{trans}} = \dot{V}_{\text{trans}} \cdot \Delta t \quad (7)$$

Limits placed on the system transform amounts $\xi^{(\lambda)}$ are based on the initial volume fraction of RA in the material, designated as $\text{RA}_{\text{initial}}$. The total transformation, as well as on any given system, cannot exceed the initial amount of austenite present, as governed by

$$0 \leq V_{\text{trans}} \leq \text{RA}_{\text{initial}} \quad (8)$$

$$0 \leq \xi^{(\lambda)} \leq \text{RA}_{\text{initial}} \quad (9)$$

When activated, the transform rates $\dot{\xi}^{(\lambda)}$ on the transformation systems drive the rate of change of the transform deformation gradient according to

$$\dot{\tilde{F}}^{\text{tr}} = \sum_{\lambda=1}^{N_{\text{trans}}} \dot{\xi}^{(\lambda)} \cdot (\gamma_T \cdot \hat{b}_0^{(\lambda)} \otimes n_0^{(\lambda)}) \quad (10)$$

Additionally, the transform rates control the evolution of the critical driving force $f_{\text{cr}}^{(\lambda)}$ via

$$\dot{f}_{\text{cr}}^{(\lambda)} = \sum_{\eta=1}^{N_{\text{trans}}} Q \cdot |\dot{\xi}^{(\eta)}| \quad (11)$$

where Q is the direct hardening parameter of transformation. This accounts for the increased resistance to transformation as more of the RA becomes surrounded by transformed martensite. The increase in resistance to transform on any system is assumed to be the same.

For the crystal plasticity portion of the material model, unit vectors $\underline{s}_0^{(a)}$ and $\underline{m}_0^{(a)}$ denote the slip direction vector and the slip plane normal direction vector, respectively, for the α^{th} slip system in the intermediate configuration. The values for these vectors for bcc martensite crystal plasticity are given in Alley [27]. The resolved shear stress on each slip system is related to the Cauchy stress tensor according to

$$\tau^{(a)} = \underline{\sigma} : (\underline{s}^{(a)} \otimes \underline{m}^{(a)}) \quad (12)$$

where the slip direction and plane normal are rotated into the current configuration by

$$\underline{s}^{(a)} = \underline{F}^e \cdot \underline{s}_0^{(a)} \quad (13)$$

and

$$\underline{m}^{(a)} = (\underline{F}^e)^{-T} \cdot \underline{m}_0^{(a)} \quad (14)$$

Under the application of resolved shear stress, the shearing strain rates on the each slip system, $\dot{\gamma}^{(a)}$, due to dislocation slip depend on the resolved shear stress on the slip system according to the flow rule,

$$\dot{\gamma}^{(a)} = \dot{\gamma}_0 \cdot \left| \frac{\tau^{(a)} - \chi^{(a)}}{g^{(a)}} \right|^m \cdot \text{sign}(\tau^{(a)} - \chi^{(a)}) \quad (15)$$

where:

$\dot{\gamma}_0$ = the shearing rate coefficient.

In this rate-dependent formation, the slip systems are always active, and rate dependence is governed by m . High values of m can effectively make the model nearly rate independent. The drag stress, $g^{(a)}$, and the back stress, $\chi^{(a)}$, on the α^{th} slip system evolve according to

$$\dot{g}^{(\alpha)} = \sum_{\beta=1}^{N_{\text{slip}}} H_{\text{dir}} \cdot |\dot{\gamma}^{(\beta)}| - g^{(\alpha)} \cdot \sum_{\beta=1}^{N_{\text{slip}}} H_{\text{dyn}} \cdot |\dot{\gamma}^{(\beta)}| \quad (16)$$

and

$$\dot{\chi}^{(\alpha)} = A_{\text{dir}} \cdot \dot{\gamma}^{(\alpha)} - \chi^{(\alpha)} \cdot A_{\text{dyn}} \cdot |\dot{\gamma}^{(\alpha)}| \quad (17)$$

where:

H_{dir} = the direct isotropic hardening coefficient,

H_{dyn} = the dynamic recovery coefficient for drag stress,

A_{dir} = kinematic hardening coefficient, and

A_{dyn} = the dynamic recovery coefficient for the back stress.

The plastic velocity gradient and its relationship to the plastic deformation gradient in the intermediate configuration depends on the shearing strain rates on all of the slip systems, $\dot{\gamma}^{(\alpha)}$, according to

$$\underline{L}_0^P = \dot{\underline{F}}^P \cdot (\underline{F}^P)^{-1} = \sum_{\alpha=1}^{N_{\text{slip}}} \dot{\gamma}^{(\alpha)} \cdot (\underline{s}_0^{(\alpha)} \otimes \underline{m}_0^{(\alpha)}) \quad (18)$$

The two-phase formulation is implemented via an implicit integration algorithm. This was chosen because an explicit algorithm at the low rate-dependence cases investigated would require extremely small steps and thus not be practical. The implicit martensitic crystal plasticity portion of the algorithm follows the method laid out by McGinty [26,28] using a Newton–Raphson iterative technique and scaled error norm to judge iteration convergence. The formulation is coded as an ABAQUS UMAT [20]. Details on the implementation of this material algorithm can be found in Refs. [27,29].

Some of the material parameters were chosen from published values in literature, while others require correlation to experimental data. Experimental calibration involves correlating a virtual test simulation with the experimental response. For comparison to experimental measurements based on the aggregate response of the gage section volume, it is necessary to model the averaged response of many grains. To facilitate this, a finite element model was constructed with 125 regions of $2 \times 2 \times 2$ elements, shown in Fig. 9. Each region is assigned a random set of Euler angles, which define the crystal orientation, as schematically represented in Fig. 9(b). Past work [30] has shown that 125 grains is sufficient to represent the average response. Increasing the number of grains does not significantly change the response. To allow for localized deformations and measure the overall response of the grains collectively, special boundary conditions are applied to the model, as described by Zhang [31]. This entails creating a reference node in the model, which is linked to surface nodes on the desired surface via constraint equations. Rather than requiring all nodes to displace a prescribed amount on the surface, the collective movement of all tied nodes must resolve with the displacement of the reference node. The virtual uniaxial test is conducted by prescribing the displacement of the reference node and recording its reaction force. This is then translated into stress-strain data for the model using the dimensions of the cube. Reaction force is divided by the

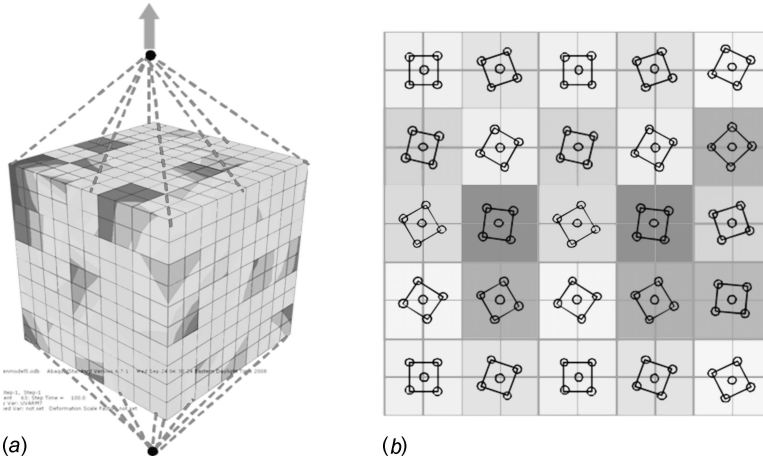


FIG. 9—(a) Mesh and loading of the volume element used to calibrate the crystal plasticity model showing the plastic strains, and (b) representation of the individual grains and random orientation scheme.

area of the face to return average uniaxial stress, and displacement of the reference node is divided by the cube side length to yield the axial strain. Similarly, transverse strain is measured by taking the averaged displacements of lateral faces relative to one another and dividing by cube side length. The incremental displacement of the reference node replicates a displacement-controlled experimental uniaxial test.

The two-phase model requires constants to be fitted to both the crystal plasticity and stress-assisted transformation model. Since these deformation mechanisms may occur simultaneously in both the experiment and simulation, noting that both mechanisms are driven by stress, the calibration process was performed in two steps. First, the 4145 experimental response was used to calibrate the fully martensitic crystal plasticity model. Using the ABAQUS UMAT routine, with martensite only (i.e., RA_{initial} is set to zero), virtual tests in displacement-control were conducted to match the experimental stress-strain curve as closely as possible, as shown in Fig. 4. This involves taking initial guess values from parameters of similar steels, and using some trial-and-error to gauge the effect of varying each parameter on the curve's appearance. Then the transformation parameters are calibrated using the experimental response of three 41100 steels containing different amounts of RA, while the crystal plasticity parameters for the martensitic phase are held constant. In addition to the stress-strain data, the evolution of the volumetric transform strain data was used to determine the parameters of the rate equation of the stress-assisted transformation model. The calibrated responses compared to the experimental responses are shown in Figs. 5 and 6. Using the calibrated parameters, the volume fractions of RA predicted to have transformed at 1 % tensile strain are 2 %, 3.5 %, and 8.5 % in specimens with initial RA of 13 %, 22 %, and 35 %, respectively [27]. X-ray diffraction measurements of RA on these tensile

TABLE 1—Parameters for the hybrid crystal plasticity and transformation model.

Parameter	4145	41100 13 % RA	41100 22 % RA	41100 35 % RA
C_{11}^0	269 230 MPa	269 230 MPa	269 230 MPa	269 230 MPa
C_{12}^0	115 384 MPa	115 384 MPa	115 384 MPa	115 384 MPa
C_{44}^0	76 923 MPa	76 923 MPa	76 923 MPa	76 923 MPa
$\dot{\gamma}_0$	0.001 1/s	0.001 1/s	0.001 1/s	0.001 1/s
m	50	50	50	50
$g_0^{(\alpha)}$	770 MPa	770 MPa	770 MPa	770 MPa
H_{dir}	8500 MPa	8500 MPa	8500 MPa	8500 MPa
H_{dyn}	0	0	0	0
$\chi_0^{(\alpha)}$	0 MPa	0 MPa	0 MPa	0 MPa
A_{dir}	0 MPa	0 MPa	0 MPa	0 MPa
A_{dyn}	0	0	0	0
RA_{initial}	0	0.13	0.22	0.35
γ_T	N/A	0.1809	0.1809	0.1809
ν_{tr}	N/A	0.17	0.17	0.17
$\dot{\epsilon}_{\text{max}}$	N/A	0.003 1/s	0.003 1/s	0.003 1/s
$(f_{\text{cr}}^{(\lambda)})_0$	N/A	120 MPa	87.4 MPa	40 MPa
Q	N/A	650 MPa	450 MPa	150 MPa

specimens after straining to 1 % and then unloading were 12 %, 18 %, and 26 %, respectively, which is in good agreement with the model predictions.

Virtual compression tests were conducted using the parameters correlated to the tensile response and compared to the experimental response, shown in Fig. 7. Although all of the parameters were calibrated to the tensile response data, the excellent agreement with the experimental compression response provide further validation that the hybrid stress-assisted transformation and crystal plasticity model correctly captures the physics of the process. A summary of the parameters are given in Table 1. Additional details on determining the material parameters are given in Refs. [27,29].

Fatigue Indicator Parameters

We are interested in assessing the drivers for local fatigue damage accumulation, assuming that it is related to the accumulation of plastic strains associated with dislocation slip locally in the material. Since there is no one accepted driver, we choose to evaluate several potential fatigue indicator parameters (FIPs) that capture the expected mechanism of crack formation.

Critical plane based FIPs are often used to represent HCF crack formation [13], noting the orientation of cracks that form are related to the cyclic loading. Here, we use the cyclic shear-based Fatemi–Socie parameter [32],

$$P_{\text{FS}} = \frac{\Delta \gamma_{\text{max}}^p}{2} \cdot \left(1 + k \cdot \frac{\sigma_n^{\text{max}}}{\sigma_y} \right) \quad (19)$$

where:

- $\Delta\gamma_{\max}^p/2$ = the maximum plastic shear strain amplitude for the cycle,
 k = a parameter found by fitting uniaxial test data against torsion test data (here set to 0.5),
 $\sigma_n^{\max}$ = the maximum normal stress acting on the $\Delta\gamma_{\max}^p$ plane, and
 σ_y = the yield stress.

The Fatemi–Socie parameter is an effective measure of the driving force for fatigue crack nucleation in scenarios where cracks grow in planes of high shear [33], which is often the scenario in RCF.

As an alternative mechanism for crack formation in RCF, we also considered a ratcheting-based FIP, using the definition given by Ambrico and Begley [34]. The ratcheting plastic strain increment accumulated over one pass of the roller is given as

$$(\Delta\varepsilon_{ij}^p)_{\text{ratch}} = (\varepsilon_{ij}^p)|_{\text{end of cycle}} - (\varepsilon_{ij}^p)|_{\text{beginning of cycle}} \quad (20)$$

Then the effective ratcheting plastic strain increment, hereon referred to as ERPSI, is defined as

$$\Delta\varepsilon_{\text{ratch,eff}}^p = \sqrt{\frac{2}{3}(\Delta\varepsilon_{ij}^p)_{\text{ratch}}(\Delta\varepsilon_{ij}^p)_{\text{ratch}}} \quad (21)$$

As will be shown later, both of these FIPs give similar trends when examining the relative RCF behavior. Other FIPs could also be considered. A couple others include the effective cyclic plastic strain range [34] and a modified ratcheting version of the Fatemi–Socie parameter proposed by Zhang et al. [35]. These other FIPs give similar trends as those exhibited by the ERPSI and FS parameters [27].

RCF Simulations

An investigation of the effect of initial volume fraction of RA on the fatigue performance was conducted. An elongated aluminum oxide inclusion, assumed isotropic elastic with $E=300$ GPa and $\nu=0.2$, was embedded in a small-scale volume with sides $150 \times 150 \times 150 \mu\text{m}^3$, as shown in Fig. 10. The dimensions of the inclusion were $10 \times 10 \times 50 \mu\text{m}^3$ with its long axis oriented in the z -direction, which is also the axis of the roller, as depicted in Fig. 10. The volume surrounding the inclusion ($110 \times 70 \times 70 \mu\text{m}^3$) was represented by the hybrid transformation-crystal plasticity model. The remaining elements were isotropic elastic. The element size throughout the volume was $5 \times 5 \times 5 \mu\text{m}^3$. The interface between the inclusion and matrix was assumed to be a perfect bond.

The small-scale volume was located at the depth of the maximum Tresca stress for an applied Hertzian pressure of 2.026 GPa and contact half-width of $260 \mu\text{m}$. This loading was selected as representative of a typical bearing bench test that results in finite life. It is of interest to note that a conventional J_2 plasticity model does not capture any localized plastic strains for this loading condition because the localized response adjacent to the inclusion remains

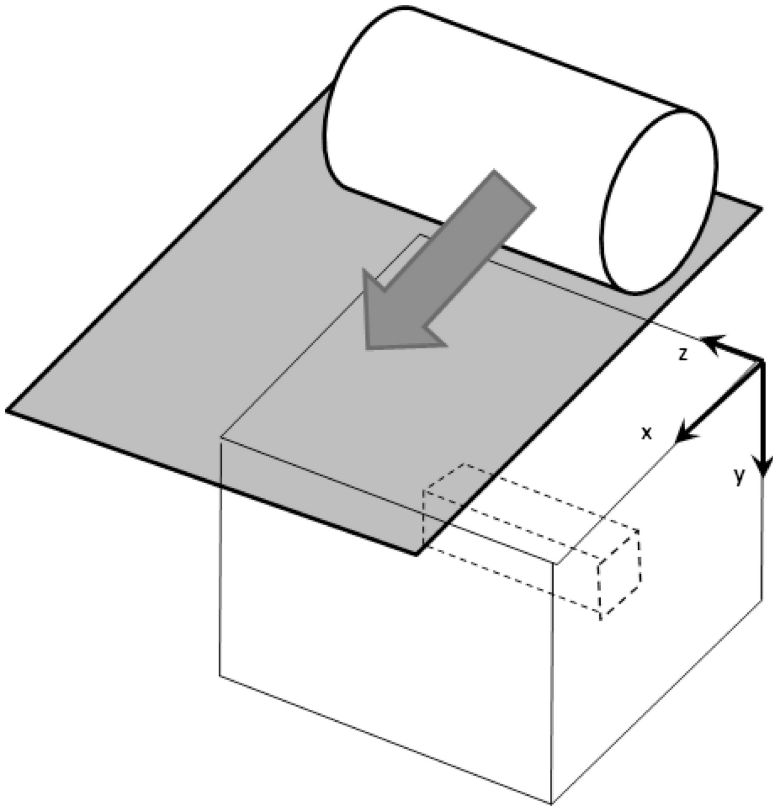


FIG. 10—Schematic of volume element containing embedded inclusion showing coordinate system and rolling direction.

below the macroscopic yield strength [27]. This in itself demonstrates the utility of using the crystal plasticity model in capturing localized plastic strains for RCF analyses, particularly when the desire is to have a physics-based approach to assess the role of microstructure on life. All of the simulations utilize the same mesh with same initial random orientation of the grains (i.e., random texture).

Three roller passes were simulated by prescribing the time-varying tractions to the sides of the small-scale volume. For each roller pass, nine load steps were used [14]. The tractions were obtained from an elastic Hertzian analysis at the desired depth [14,27]. Here, the volume was assumed to be sufficiently small to be able to prescribe uniform tractions. The inhomogeneity of the microstructure in the volume element controls the inelastic response. Each element within the volume was assumed to have a fixed crystallographic orientation, which essentially sets the size of the prior austenite grain. Several simulations were conducted starting with a microstructure seeded with different initial volume fractions of RA between 0 % and 22 %, the latter typically the largest percentage found in practical bearing manufacturing. The initial critical

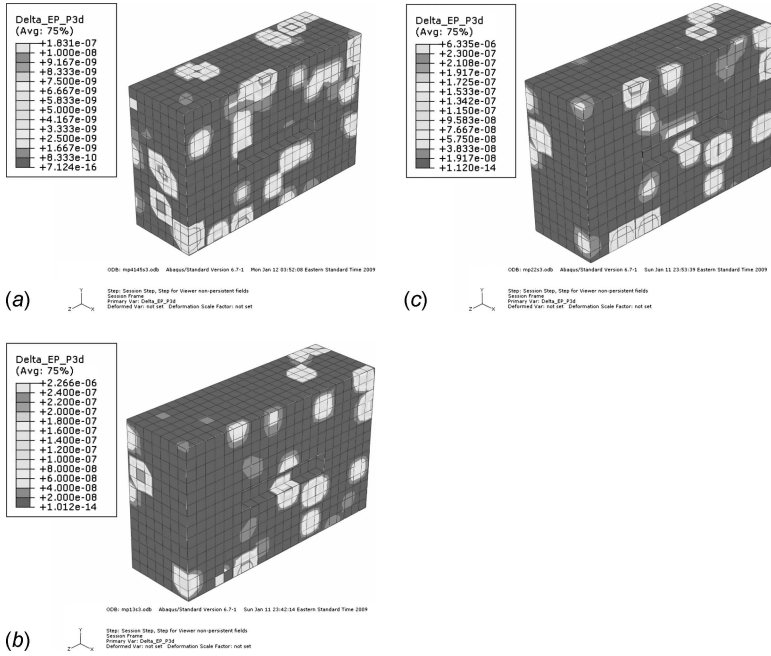


FIG. 11—ERPSI after three passes of the roller for (a) 0 %, (b) 13 %, and (c) 22 % initial volume fraction of retained austenite.

driving stress, $(f_{cr}^{(\lambda)})_0$, and the direct hardening parameter of transformation, Q , both depend on the initial RA by the following relations:

$$(f_{cr}^{(\lambda)})_0 = (167.38 - 363.76 \cdot RA_{initial}) \text{ MPa} \quad (22)$$

and

$$Q = (947.55 - 2275.2 \cdot RA_{initial}) \text{ MPa} \quad (23)$$

Results

Contour plots of the ERPSI after the third pass of the roller are shown in Fig. 11 for 0 %, 13 %, and 22 % initial RA. The figures show the view of the region surrounding the inclusion along the y - z plane, based on the coordinate system shown in Fig. 10. Additionally, only the region in the mesh that contained elements described by the hybrid transformation-crystal plasticity model are shown for clarity. Due to the differences in the response of the isotropic elastic elements surrounding the hybrid model region, some edge effect exists within one element of this boundary. Hence, there are artificial stress concentrations appearing along the outer edges of the microplastic zone. For this reason, the results collected for graph comparisons do not include these outermost crystal plasticity elements. The effect of initial volume fraction of RA on ERPSI is

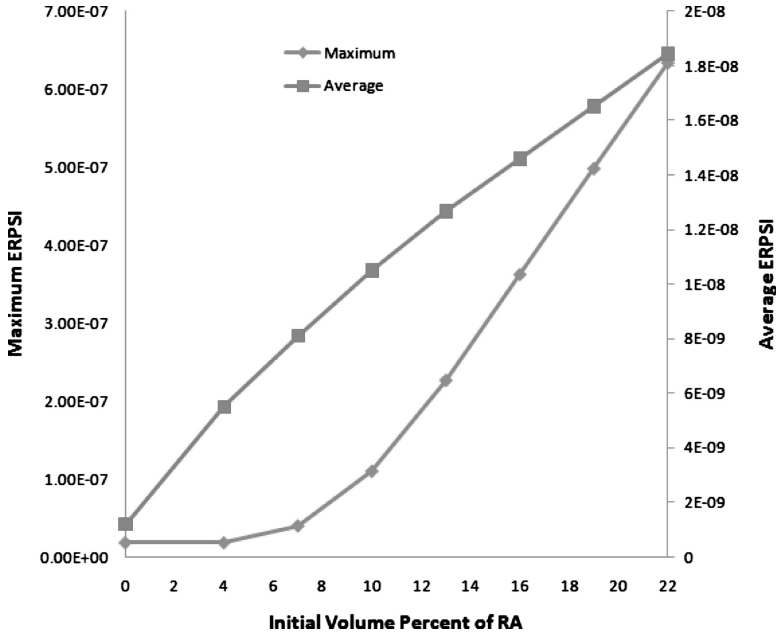


FIG. 12—ERPSI after three passes of the roller vs initial volume fraction of RA.

shown Fig. 12. Two values are shown: the maximum ERPSI found anywhere in the crystal plasticity region (excluding elements on the outermost boundary), and the average value of this FIP in this region (also excluding the elements on the outermost boundary).

To understand the distribution of plastic strain accumulation, it is helpful to examine the amount and distribution of transformation, expressed as a percentage of the whole volume, after the three passes, shown in Fig. 13. Of interest to note is that while some localized regions in both cases exhibit volume fractions of transformation near 4 %, there is much more comprehensive transformation occurring in the higher RA case. This is apparent in the maximum and average percentages of volume fraction transformed for different initial RA levels shown in Fig. 14. As expected, there is an increase in transformation with increasing initial RA, as there is a decrease in the threshold for transformation with increasing austenite. The left-hand side of both graphs is anchored at the origin, as no transformation can occur in the 0 % initial RA case. While the maximum values appear to reach a saturation value between 3.5 % to 4 % when the initial RA is greater than 10 %, the average value steadily increases nearly linearly with increasing initial volume fraction of RA.

The residual stress field is also of interest to examine. Here we examined both the hydrostatic and von Mises effective stresses. First, the residual hydrostatic stresses are shown in Fig. 15 for the 0 %, 13 %, and 22 % initial RA cases. The hydrostatic stress field in the 0 % initial RA case exhibits a distinct pattern where the inclusion is the central feature controlling the residual stress field. The 13 % and 22 % RA cases exhibit a more inhomogeneous response, with the

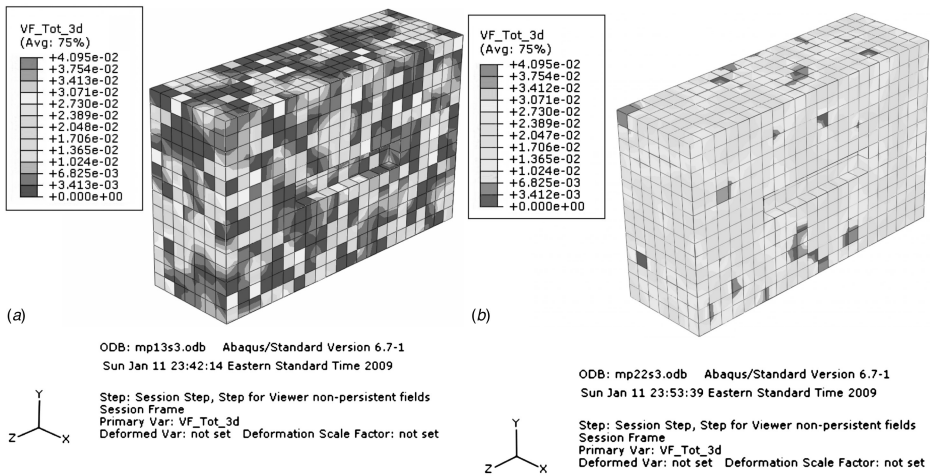


FIG. 13—Volume fraction transformed for (a) 13 % and (b) 22 % initial volume fraction of retained austenite after three passes of the roller.

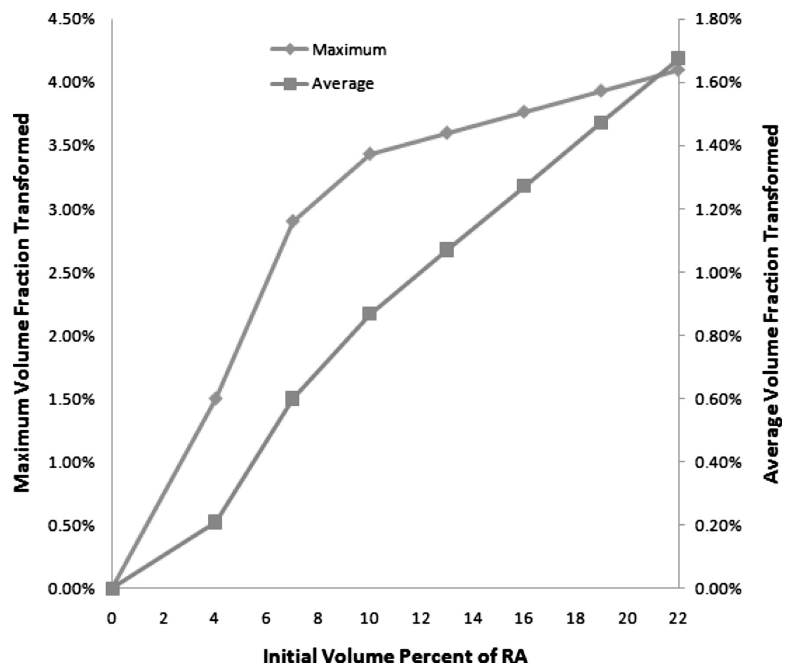


FIG. 14—Volume fraction transformed after three passes of the roller vs initial volume fraction of RA.

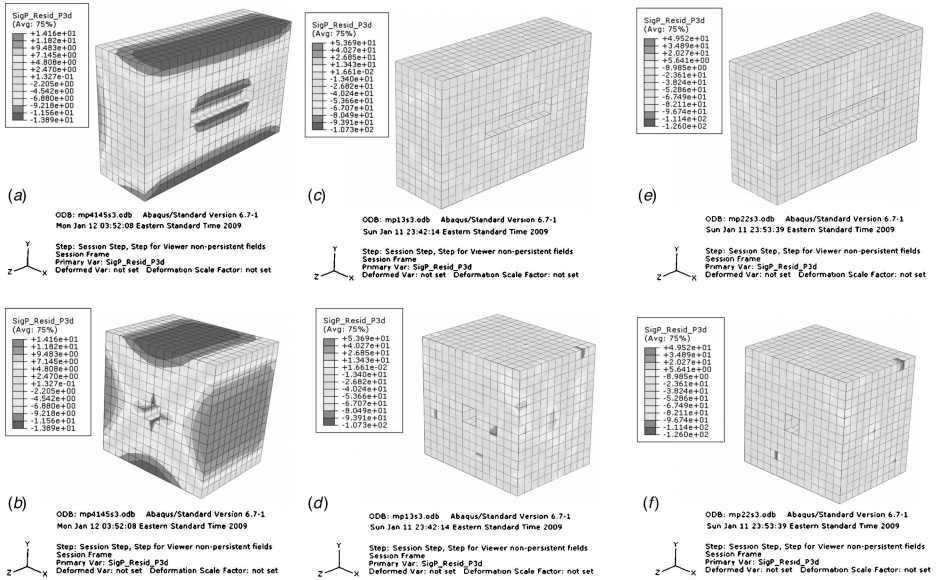


FIG. 15—Residual hydrostatic stress after three passes of the roller for (a) 0 % RA showing x-plane normal cut view, (b) 0 % RA showing z-plane normal cut view, (c) 13 % RA showing x-plane normal cut view, (d) 13 % RA showing z-plane normal cut view, (e) 22 % RA showing x-plane normal cut view, and (f) 22 % RA showing z-plane normal cut view.

transformation strains playing a more significant role than the inclusion in controlling the residual stresses. Note that the “hot spots” for the residual hydrostatic stresses tend to occur in areas other than immediately adjacent to the inclusion in these cases. The maximum positive, maximum negative, and average values of the residual hydrostatic stresses are shown in Fig. 16 for different initial RA. The hydrostatic state of stress at 0 % initial RA as depicted in Fig. 15(a) and 15(b) shows the compressive stress state above and below, and tensile stress state to the sides of the inclusion. The symmetry of the positive and negative contours nets an average residual hydrostatic stress of approximately 0 for the entire local region as the maximum positive and negative values are nearly equal. This trend continues until about 5 % RA when the average becomes negative. It was previously shown that significant portions of the austenite become susceptible to transformation for initial RA greater than about 5 %. For initial volume fractions above 5 %, there is a significant shift in the magnitude difference between the positive and negative maximum values, with the negative (compressive) maximums becoming much larger. The average value also becomes compressive, because the transformation of RA causes a volume expansion, thereby compressing grains adjacent to regions with higher levels of transformation.

The contours of the residual von Mises effective stresses for 0 %, 13 %, and 22 % initial RA after the third roller pass are given in Fig. 17. The maximum

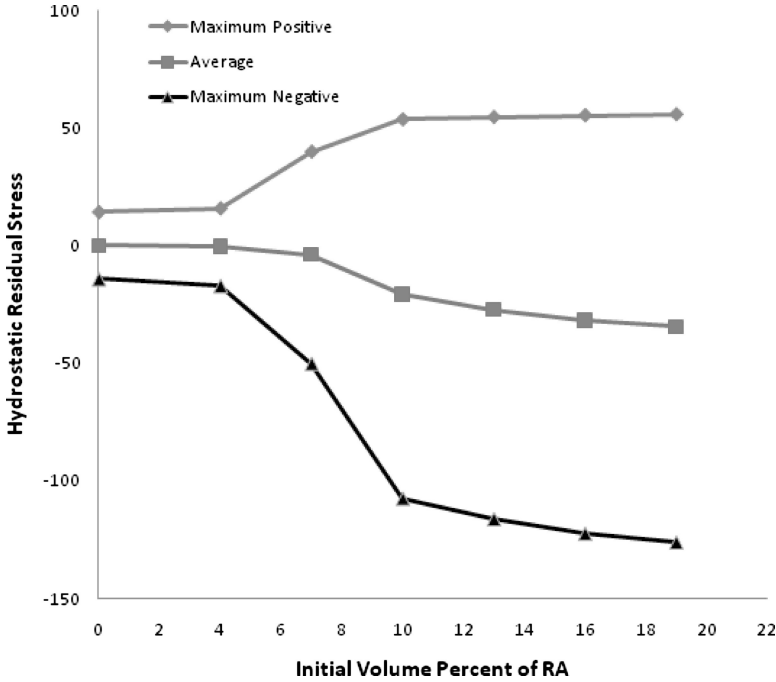


FIG. 16—Residual hydrostatic stress after three passes of the roller vs initial volume fraction of RA.

and average value of the residual von Mises stress are shown in Fig. 18 for different levels of initial RA. It can be seen that for the case of no initial RA, the von Mises effective residual stress is greatly reduced surrounding the inclusion, due to the high state of hydrostatic stress that the inclusion creates. With increasing amounts of RA, the residual von Mises stress increases, and the distribution of the residual von Mises stress is inhomogeneous similar to the hydrostatic stress.

Discussion

Of importance in this study is that the polycrystalline structure leads to localized plastic strain accumulation in regions that are not necessarily immediately adjacent to the inclusion. When there is no RA, the inclusion is the primary stress riser as indicated by the residual stress field shown in Fig. 15(a) and 15(b). But for steels containing metastable RA, transformation of the RA increases the heterogeneity of the residual stress field. This tends to blunt the effect of the inclusion's ability to induce plastic strain accumulation which should be beneficial to RCF. However, the distortions and high residual stresses induced by the local transformation strains do nevertheless raise the magnitudes of the FIPs when there is increasing amounts of RA that can transform. The local distribution of residual stress becomes much more inhomogeneous,

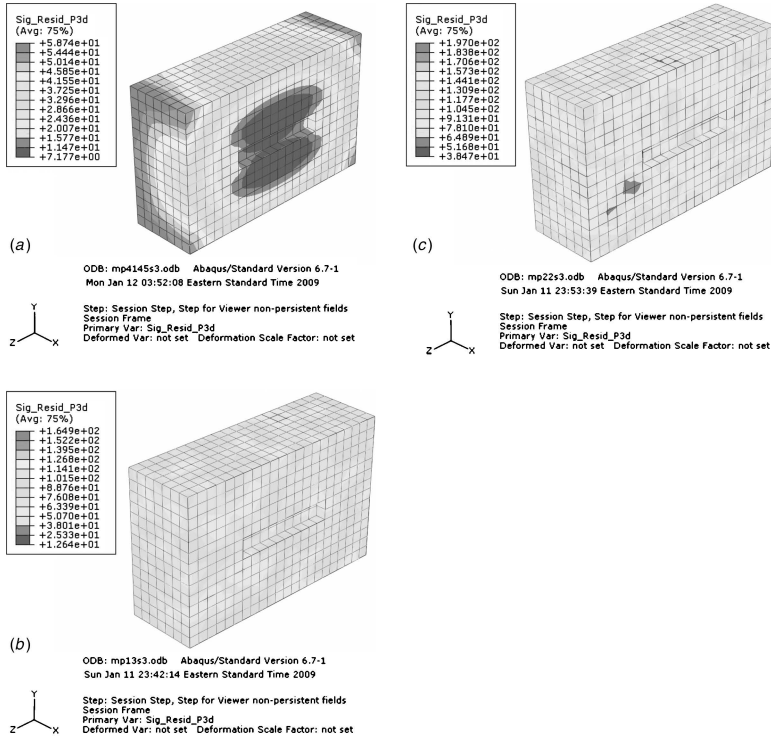


FIG. 17—Residual von Mises effective stress after three passes of the roller for (a) 0 %, (b) 13 %, and (c) 22 % initial volume fraction of retained austenite.

which in turn leads to stress concentrations resulting in regions in the microstructure where the contribution to the driver for dislocation slip is increased. As a consequence, the local accumulation of plastic strains due to dislocation slip also increases with increasing amount of transformation.

The demonstration case showed that increasing amounts of RA lead to higher FIP values and thus lower relative fatigue life. Even though a compressive residual stress should be beneficial, the distortion caused by the transformation strains actually contributes to the driver for dislocation slip plasticity, resulting in enhanced accumulation of plastic strain under repeated roller passes. This seems to be an alternative explanation for the reduction in life with increased RA, consistent with the known influence of RA on HCF life [7] or bearing life in clean, well-lubricated environments [1]. Other explanations have concentrated on the cyclic plastic strain accumulation within the austenite itself, which may be plausible if (i) the transformation is strain-induced or (ii) the transformation has saturated and the remaining RA, now having a much reduced driving force for transformation coupled with an increase in residual stress and distortion, deforms by a dislocation slip mechanism. But here, the focus is primarily on the early stages of cycling for which experiments suggest that a stress-assisted transformation mechanism is dominant. Hence, these al-

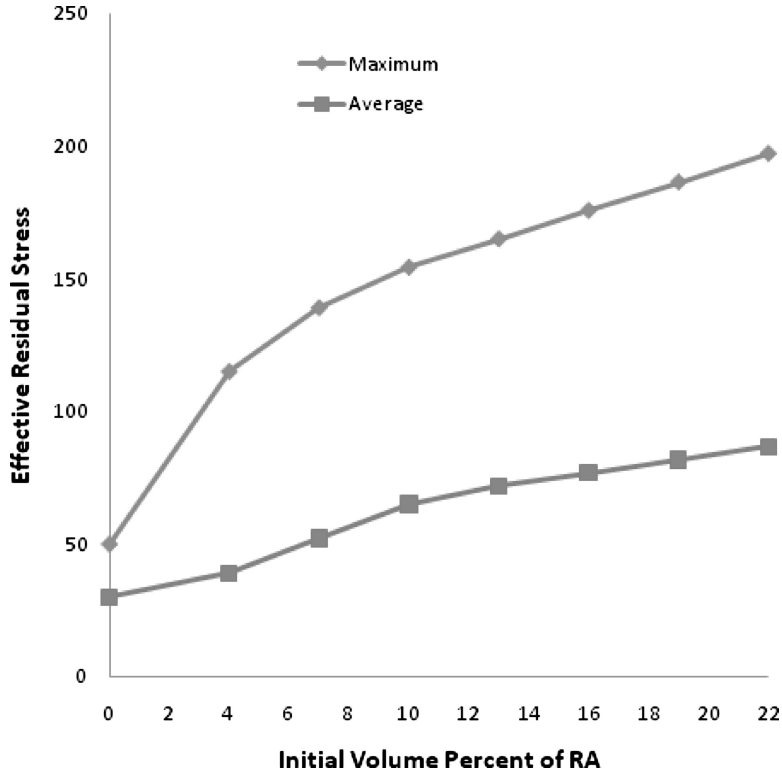


FIG. 18—*Residual von Mises effective stress after three passes of the roller vs initial volume fraction of RA.*

ternative mechanisms were assumed to be negligible in this study.

The goal was to develop a set of tools that can be used to investigate microstructural parameters and relate them to performance, thus reducing the amount of expensive physical bearing testing required. The work presented herein was just an initial attempt to capture key microstructural attributes on RCF performance. Admittedly, even though the present model represents an advancement over the use of conventional homogeneous plasticity models, the present material model does not capture dislocation slip in austenite, partially due to the lack of information on relative influence of transformation and dislocation slip deformation mechanisms within the austenite. Presently, the model assumes that the nonlinear behavior of the austenite is completely controlled by stress-assisted transformation. Certainly, there is some dislocation slip to help maintain compatibility accompanying the transformation. Nonetheless, the correlations of the model to the experimental deformation response were excellent for this particular bearing steel. This suggests that the magnitude of the plastic strain in the austenite is negligible in comparison to the transformation strain and that the influence of these plastic strains is effectively included in the fit of the parameters of the stress-assisted transformation

model. This is a reasonable assumption for the behavior during the first couple of roller passes, but long-term evolution of the transformation and local deformation may also involve considerable dislocation slip in the austenite. In that case, a more complex coupled transformation and dislocation slip model of the austenite phase would be needed.

Other enhancements of the approach are also desirable and left for future work. In particular, the effect of varying composition through the depth, for example due to case carburization, should be considered. In addition, the initial residual stresses could be prescribed based on processing calculations. Certainly, the presence of a compressive residual stress at the location of the maximum cyclic shear should reduce the FIP values, as one might expect. One could also prescribe crystallographic texture instead of assuming random texture in the critical volumes. Studies [9,10] have clearly shown that texture can form in the highly loaded volumes during operation of a bearing and that this texture influences the RCF behavior, manifested by the spall characteristics. Crystal plasticity is an ideal tool to study and understand this behavior and potentially could be implemented in spall growth analyses.

The interfaces between phases and grains in the microstructure were assumed to be ideal perfect bonds. In reality, this is not always a good assumption particularly with aluminum oxide inclusions. Debonded inclusions have been shown to be worse than perfect bonded in related gear steel analyses [36,37]. In fact, the effect of the interface condition has been shown to be more significant than inclusion morphology [36]. However, making the interface weak is a more complex computational problem due to the instabilities that can arise, which is why in this paper, we limited our scope to the simpler perfect bond condition. Clearly, future work should focus on the interface condition between the non-metallic inclusions and crystal plastic matrix.

Summary and Conclusions

An approach is presented to determine the relative RCF performance as a function of microstructural attributes. Several challenges were addressed. First, fully three-dimensional finite element modeling allows for end effects to be captured that were not previously possible with two-dimensional plane-strain models. This additional capability provides for a more realistic assessment of inclusion morphology and arbitrary orientations. In addition, the new hybrid two-phase transformation and crystal plasticity model captures the intense local plastic strain accumulation that can develop in pockets within the polycrystalline microstructure of bearing steels.

By accounting for the volume fraction evolution of the RA, the residual stress state due to the inelastic transform strains are accounted for and their effect on subsequent accumulation of the cyclic plastic strains is determined. Thus, important links between the initial volume fraction of RA and FIPs (and thus relative fatigue performance) considering initial phase distributions can be drawn with this approach. It was observed through this study that initial volume fractions of RA above about 5 % increased the magnitude of the fatigue indicator parameters which corresponds to a reduction in the relative fatigue

life. This appears to be linked to the high residual stresses and local distortions caused by the extensive transformation in the higher initial RA cases.

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FATIGUE LIFE PREDICTION METHODOLOGIES

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Pascal Daguiet,⁴ Ferhat Merchi,⁵ and Alain Vincent⁶

A New Methodology for Predicting Fatigue Properties of Bearing Steels: From X-Ray Micro-Tomography and Ultrasonic Measurements to the Bearing Lives Distribution

ABSTRACT: This work aims at developing a new methodology for predicting the distribution of fatigue lives of bearings versus steel microstructure, namely, concentration, morphology, and properties of residual inclusions in steel. On the experimental side, X-ray micro-tomography and high frequency ultrasonic testing are used to provide the required inclusion characteristics. A physically based model is used to compute the number of cycles to crack nucleation and crack propagation up to the surface. For the statistics predictions, the inclusions and/or stringers are distributed randomly in the bearing steel according to the concentration of stringers provided by ultrasonic data.

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The distributions of the fatigue lives predicted by the model are compared successfully to the experimental distributions determined by fatigue tests performed on flat washer machines. Finally, it is shown that the model is able to predict the influence of the orientation of stringers on the fatigue lives.

KEYWORDS: rolling contact fatigue, bearing steel, inclusions, stringers, distribution of fatigue lives

Introduction

Non-metallic inclusions are known to have a detrimental effect on the fatigue performance of high strength steels, but producing steels with perfect cleanliness is not possible for mass productions at competitive costs. However, international standards, such as ISO 281 [1], do not include explicitly inclusion characteristics in the methods of calculating the basic rating life. Hence, predicting the influence of inclusions on bearing lives remains a major challenge for steel producers, bearing manufacturers, and machine industries. In previous works [2–4], a physically based model for predicting the distribution of bearing lives was developed on the basis of spherical inclusions distributed randomly in the steel. In fact, due to the elaboration process of bearing steels, real inclusions have complex shapes and are often lined up, thus forming so-called stringers. To improve the relevance of such a physically based model, these configurations should be taken into account, especially to be able to describe the influence of the forming process on fatigue properties. This is the topic of the present paper.

For that purpose, a new methodology has been developed. It is based on (i) the determination of the concentrations of detrimental inclusions in the steel by high frequency ultrasonic testing; (ii) the study of the morphologies of single inclusions or stringers by X-ray micro-tomography, which now enables one to obtain three-dimensional (3D) images of inclusions, thanks to the high flux beams of X rays provided by synchrotrons of third generation; (iii) the characterization of the mechanical properties, in terms of elasticity and plasticity, of the inclusions by nano-indentation; and (iv) the random distribution of inclusions and/or stringers in the bearing steel according to the concentrations provided by ultrasonic data and the computation of the resulting fatigue lives by using a physically based model for crack nucleation and crack propagation up to the surface.

The experimental conditions are described in the first part of the paper. In the second part, some experimental results are illustrated, especially for micro-tomography observations. Then, the modeling conditions are described, and the distributions of fatigue lives predicted by the model are compared to the experimental data derived from fatigue tests performed on flat washer machines. Finally, the influence of the orientation of stringers on the fatigue lives is studied.

Experimental Conditions

Material

The work has been carried out on various heats of 100Cr6 steel, heat treated according to standard conditions: 20 min austenitization at 850°C followed by isothermal quenching at 220°C and 1 h tempering at 170°C. The resulting 0.2 % yield strength is 2360 MPa at room temperature. Additional details about the mechanical properties of the studied material can be found in Ref 2. It should be noted that the studied heats were selected with poor cleanliness according to the purpose of this work.

Ultrasonic Testing

Ultrasonic testing is commonly used to detect and estimate the concentration of non-metallic inclusions in steel. The main advantage of this technique arises from its ability to scan a large volume of material [5]. Details about the technique can be found in dedicated literature [6]. In the present work, this technique has been used for the purposes of (i) estimating the inclusion population in the various heats that have been tested in fatigue and (ii) detecting and locating typical inclusions prior to the X-ray micro-tomography observations. An immersion C-scan Epsilon[®] device from Metalscan,⁷ equipped with an 80 MHz focused transducer, has been used. With this equipment, inclusions of 10 μm in diameter located in the focal zone can be detected. However, the spatial resolution of the technique does not enable one to reveal the detailed shape of inclusions. Such a goal can be reached by using X-ray micro-tomography, as described in the following section.

X-Ray Micro-Tomography

X-ray micro-tomography is a 3D imaging technique that is increasingly used in materials science [7–10]. It allows one to visualize the inner constituents (pores, inclusions, phases, etc.) of a material, provided they exhibit an X-ray absorption coefficient different from that of the matrix. Although several papers dealing with constituent particles in aluminum alloys can be found in the literature (see, for example, Ref 11), to the best of the authors' knowledge, the detailed characterization of inclusions embedded in a high strength steel matrix has not been reported yet.

The examples of observations presented in this work have been obtained on beam line ID19 at the European Synchrotron Radiation Facility (ESRF). This beam line provides an X-ray beam with high brightness and high spatial coherence, allowing one to perform micro-tomography with a spatial resolution in the micrometer range. Figure 1 shows the experimental micro-tomography set-up available at beam line ID19 [12]. It includes a monochromator that allows one to select photons with a suitable energy, a scintillator converting X-rays into visible light, and a digital camera to record the radiographs ob-

⁷Metalscan, Industrial Echography: www.metalscan.fr.

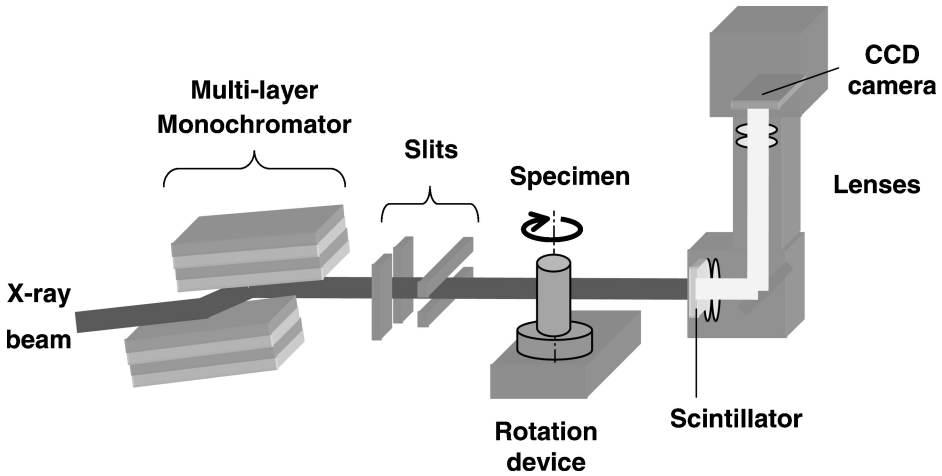


FIG. 1—Experimental set-up available at line ID19 for micro-tomography experiments.

tained for successive angular positions of the specimen mounted on a rotating stage. To observe inclusions with the high spatial resolution required for this work, the size of the samples has to be limited. This limitation arises mainly from two factors. First, the size of the specimen affects the choice of the energy of the photons. As a rule of thumb, a minimum of 10 % of the intensity of the incident beam should remain transmitted throughout the specimen to get good quality images, which plays in favor of an increasing energy of the photons with increasing the size of the specimen. But, in contrast, a lower energy tends to be required to keep a good contrast of absorption between inclusions and matrix as well as a good efficiency of the scintillator used to detect X-rays. For the present investigations, these conflicting requirements were fulfilled by setting the energy of the incoming photons to 35 keV. Second, the image of the specimen must remain within the field of the charge-coupled device detector of the camera. This depends on the size of the specimen, the optical magnification, and the desired voxel size in the reconstructed image. With the optical magnification chosen, the effective voxel size in the reconstructed 3D image is $0.7 \times 0.7 \times 0.7 \mu\text{m}^3$. Altogether, these conditions inferred a cross section of $0.4 \times 0.4 \text{ mm}^2$ for our needle-like specimens. Once reconstructed by the software available at ESRF (classical filtered back projection method), the 3D images have been analyzed using ImageJ⁸ and Amira⁹ softwares. 3D renditions of the inclusions have been obtained by the region growing segmentation method [13].

⁸ImageJ: <http://rsb.info.nih.gov/ij/>.

⁹Amira: www.amira.com.

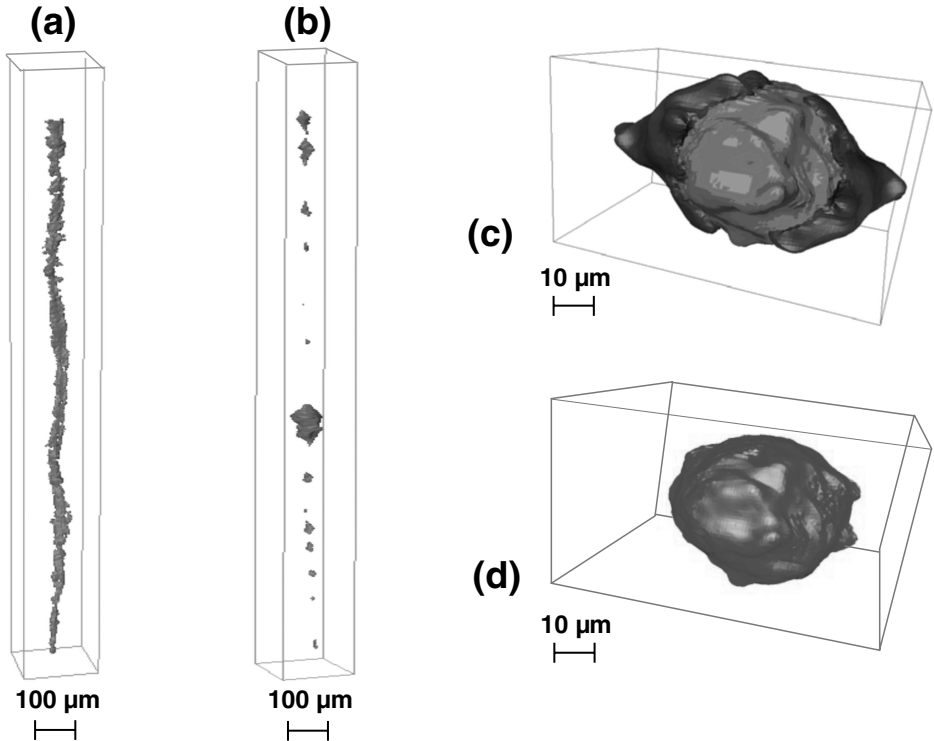


FIG. 2—3D renditions of typical inclusions observed by X-ray micro-tomography in bearing steels: (a) Non-fragmented elongated inclusion; (b) fragmented elongated inclusion; (c) globular inclusion with adjacent conical ends; and (d) same inclusion as (c) but the conical ends have been removed by image processing.

Experimental Results

Inclusion Observations through X-Ray Micro-Tomography

The various types of non-metallic inclusions most often met in bearing steels have been observed successfully by X-ray micro-tomography: isolated inclusions with a globular shape, stringers constituted of lined up inclusions, and continuous elongated inclusions. Examples of such observations are shown in Fig. 2 for a few inclusions. The 3D renditions are presented after image processing, which enables one to make the matrix transparent and visualize only the inclusions in a virtual parallelepiped box showing the limits of the volume analyzed around of the inclusion.

Figure 2(a) shows an example of elongated inclusion that appears as a long continuous ribbon (length > 1 mm), with its main direction almost parallel to the axis of the bar from which the specimen was cut. Its section is approximately constant throughout the ribbon, typically $8 \times 20 \mu\text{m}^2$. Furthermore, the inclusion is not perfectly straight but rather undulating around its main direc-

tion. It should be emphasized that due to this undulating character, the continuity of such type of inclusion could not be easily evidenced by standard two-dimensional micrographic observations. By means of destructive energy dispersive X-ray analysis, such inclusions have been identified as manganese sulfides in the batches of steels investigated in this work.

Figure 2(b) shows an example of a stringer with lined up inclusions. For a given stringer, the size of the constituent inclusions is widely scattered (from a few microns to 70 μm in the case presented in Fig. 2(b)). The distances between neighboring inclusions are also widely scattered and were found to vary from a few microns to a few tens of microns for the investigated samples. It should be pointed out that such detailed characteristics of the stringers could not be obtained by ultrasonic echography. Besides, in such stringers, destructive analyses have shown that the constitutive inclusions are alumina inclusions and calcium aluminate inclusions, sometimes with some coating of calcium enriched sulfide.

Figure 2(c) shows an example of an isolated globular inclusion. It should be mentioned that they often include two conical ends, as the one selected for the figure. The axis going through the two tips of the cones is nearly a revolution axis of the whole defect and is parallel to the rolling direction of the bar. The gray level for these conical ends, darker than those of the steel matrix and the inclusion core, suggests that they are cavities. This contrast can be used to remove the conical ends by image processing (Fig. 2(d)). Such globular isolated inclusions have been identified either to alumina or calcium aluminate inclusions.

The flexibility of the tomography representation is illustrated in Fig. 3. It shows three different virtual tomography cuts of the inclusion shown in Fig. 2(c) in orthogonal median planes. The inclusion is globally spherical but exhibits some surface irregularities. The aforementioned cones are also visible in Fig. 3(b) and 3(c) (darker zone).

Finally, Fig. 4 shows a comparison between a virtual tomography cut and a classical micrograph on a polished section of the same inclusion. This comparison confirms the ability of tomography to depict all the details of inclusions with a spatial resolution close to that of optical microscopy.

Quantitative Characterization of Stringers from X-Ray Micro-Tomography Images

As it will be detailed in the section devoted to modeling, the detrimental character of a stringer is linked to the interaction between constitutive inclusions. This interaction is linked to the distance between neighboring inclusions, which is widely scattered, as mentioned above. Moreover, the size of the interacting inclusions plays also an important role. Hence, the edge-to-edge distance d between two neighboring inclusions has been normalized by the average of the radii of these inclusions $R_m = (R_i + R_j)/2$ (see Fig. 5). In order to characterize this parameter, several 3D images of stringer were analyzed to establish statistical data related to the tested heats. The histogram of the parameter d/R_m thus obtained is given in Fig. 6. It shows that the maximum of the distribution of the

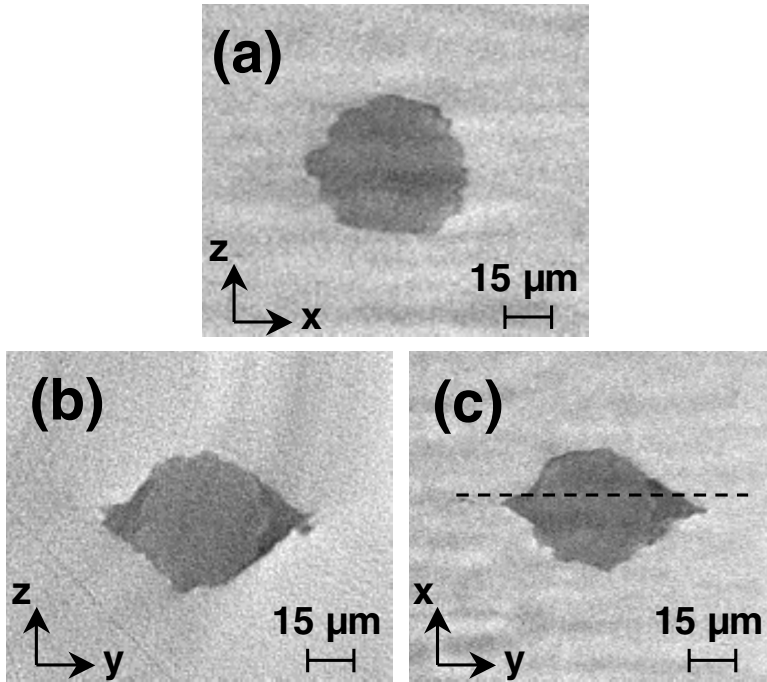


FIG. 3—Reconstructed tomography slices of the globular inclusion shown in Fig. 2(c) observed in three perpendicular median planes. y axis is parallel to the rolling axis of the bar. Two cones appear in darker gray in (b) and (c) on each side of the inclusion. The dotted line in picture (c) corresponds to the observation plane in Fig. 4.

normalized distance occurs for d/R_m values close to one. For further use of these data, it can be seen that the distribution can be reasonably fitted by a log-normal distribution (continuous curve in Fig. 6).

Stringer Concentration Obtained through High Frequency Ultrasonic Testing

The stringers can be easily detected by using the ultrasonic equipment referenced in the previous section. For that purpose, parallelepiped samples were taken from the laminated bars, machined, and then scanned with the point focused 80 MHz probe. The scanning plane was parallel to the bar axis. Figure 7 shows an example of a C-scan image thus obtained. As pointed out previously, the resolution of the technique does not enable one to distinguish the details of the stringer. Only its total length L (see Fig. 5) can be estimated. Thus, the ultrasonic testing and the X-ray micro-tomography technique appear quite complementary. Furthermore, for each heat tested in this work, a total volume of about 4 cm³ has been investigated. The stringers have been classified according to their length. An example of histogram of the length of the stringers is shown in Fig. 8. Most of the stringers have a length shorter than 200 μm, but a few were found to belong within categories of longer length.

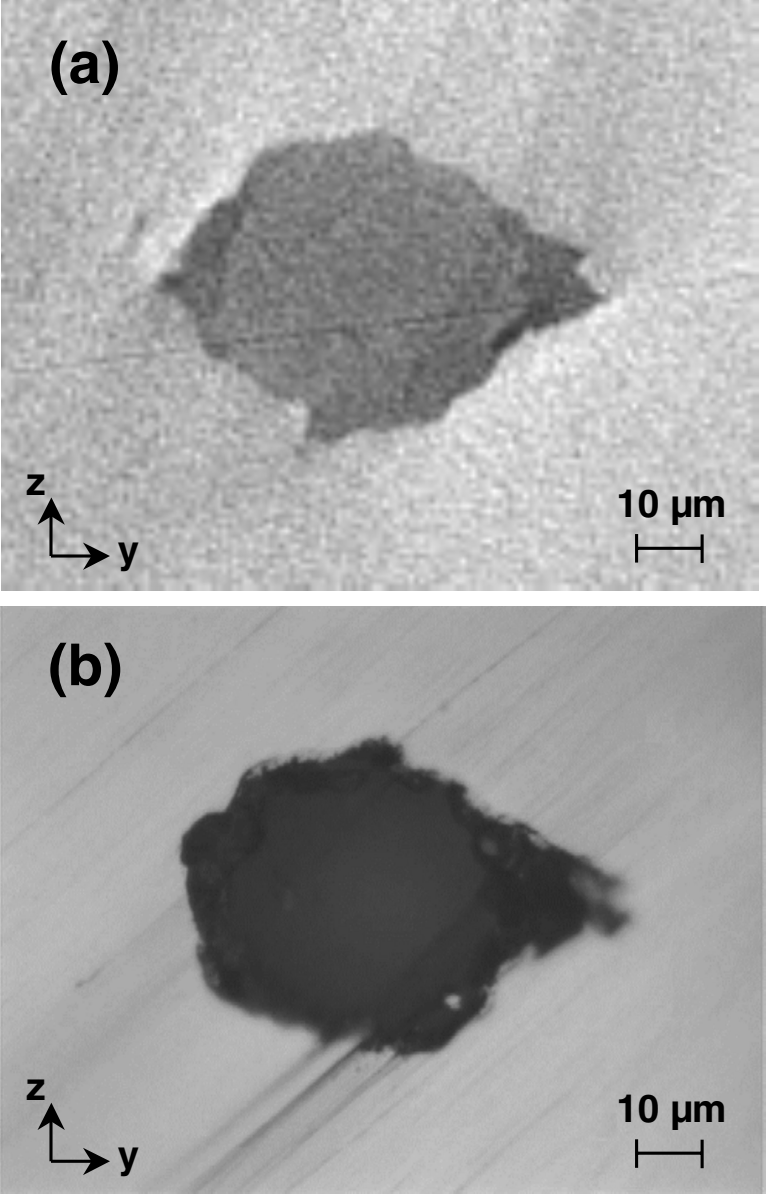


FIG. 4—Comparison between (a) X-ray micro-tomography and (b) optical micrograph of the globular inclusion shown in Figs. 2(c) and 3.

Finally, it should be mentioned that, up to now, the ultrasonic testing was not able to recognize the nature of inclusions in the detected stringers. Hence, the nature of the inclusions, or at least the majority of them, has to be known using another technique.

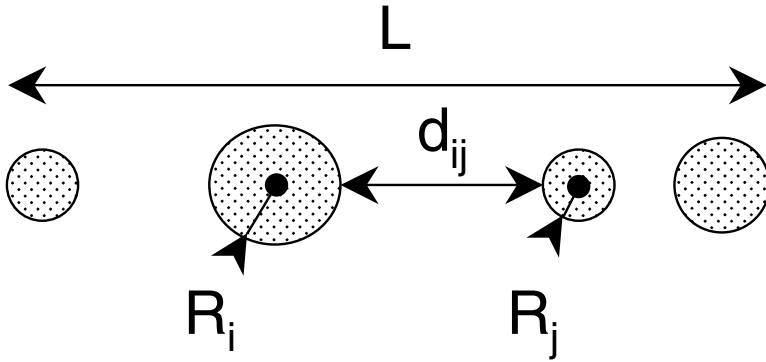


FIG. 5—Schematic of a stringer: Definition of parameters.

Computational Modeling of the Distribution of the Fatigue Lives

The approach used for predicting the distribution of fatigue lives in the present work consists in introducing the effects of interaction of neighboring inclusions in stringers in a model initially designed for isolated spherical inclusions, which has been presented in previous works [2–4]. This is made possible due to both the internal characteristics of the stringers provided by X-ray microtomography and the statistical data provided by ultrasonic testing.

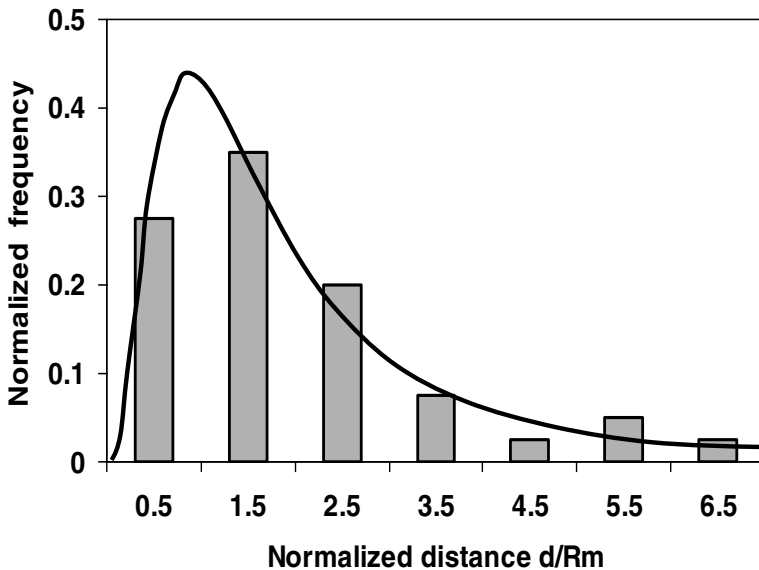


FIG. 6—Histogram of the normalized distances between inclusions in the stringers (for the normalized distance axis, numbers refer to the medium value of each class whose width is 1); the corresponding distribution is fitted by a log-normal function (continuous curve).

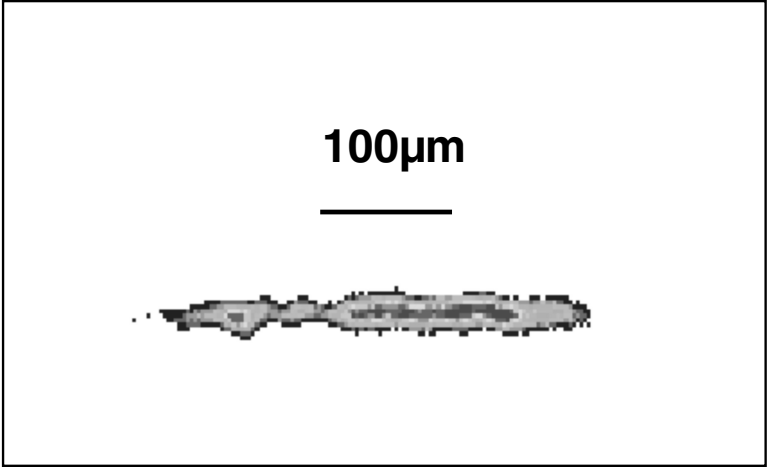


FIG. 7—C-scan image of a stringer detected by high frequency ultrasonic echography.

The computational modeling of the bearing lives distribution includes two parts. The first one is of determinist nature: It aims at calculating the number of cycles to failure induced by a given inclusion. This problem is solved by considering the crack nucleation and the crack growth stages successively. The second part deals with the statistical aspect of the problem: The inclusions are statistically distributed in the bearing steel (position, size, etc.).

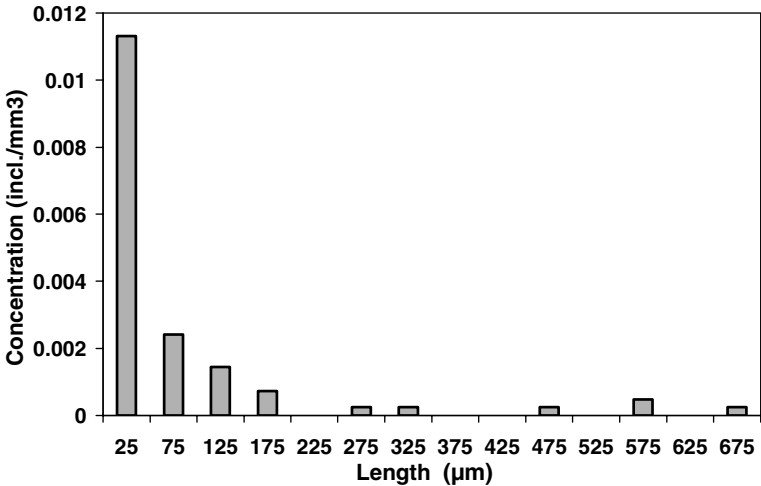


FIG. 8—Histogram of the lengths of the stringers in the heat D (for the length axis, numbers refer to the medium value of each class whose width is 50 µm).

Determinist Model for Crack Nucleation and Growth [2]

The material is loaded cyclically under the repeated contact pressure produced by the rolling bodies. Crack nucleation originating from inclusions arises from the stress concentration around the inclusions. This induces cyclic micro-plasticity leading to the accumulation of dislocations. Finally, crack nucleation occurs once a critical density of dislocations is reached in this area.

In the present work, the contact stress field is that produced by circular balls (9.992 mm in diameter) on a plane surface. The resulting contact stress, 4.2 GPa, is that of the fatigue tests on flat washers described in the last section of the paper. The stress concentration around the inclusions has been computed by means of the Abaqus finite element software, as described in Ref 14. Moreover, in this reference, it has been shown that the real geometry of the inclusions provided by the 3D images from micro-tomography can be reasonably modeled by simplified shapes in order to derive by finite element a realistic stress field around the inclusions (for instance, the inclusion shown in Figs. 2(c) and 3 can be valuably modeled by a spherical core associated with two conical empty sides [14]). The elasto-plastic behavior of the 100Cr6 matrix and the inclusions are also given in this reference. The accumulation of dislocations in the vicinity of inclusions is characterized by the damage accumulation factor f^* defined in Ref 2, and the criterion for crack nucleation can also be found in Ref 2. The resulting cycle number for crack nucleation is N_n . The orientation of the micro-crack is given by the direction of the maximal shear stress for dislocation glide. Its initial length is taken equal to that of the zone where localized dislocation movements occur, i.e., the stress concentration zone around the inclusion [2].

Crack propagation is calculated by considering a physically based model as in Ref 3. The crack is assumed to grow in mode II with a circular shape. The amplitude of the stress intensity factor for crack growth is calculated by considering an effective shear stress that accounts for friction of the crack lips due to the compression stress normal to the crack plane. To account for the limited extension of the Hertzian loading, the width extension of the crack is bounded to the contact width of the rolling ball. The current crack length, a , is calculated numerically by integrating the local crack growth rate da/dN , which is updated (in relation to the present crack length and the crack position in the loading stress field) after each numerical step. The number of cycles to bearing failure, N_p , is considered necessary for the crack to reach the surface. This means that the number of cycles from the surface breaking stage of the crack to complete spalling is neglected.

Finally the total cycle number to failure, N_R , is given by the sum $N_R = N_n + N_p$.

Influence of the Interaction between Inclusions

In this work, the excess of harmfulness of stringers has been assigned to the interaction between neighboring inclusions, which reduces the number of cycles to crack nucleation N_n . Indeed, the superposition of the stress concentration zones around the neighboring inclusions enhances locally the stress

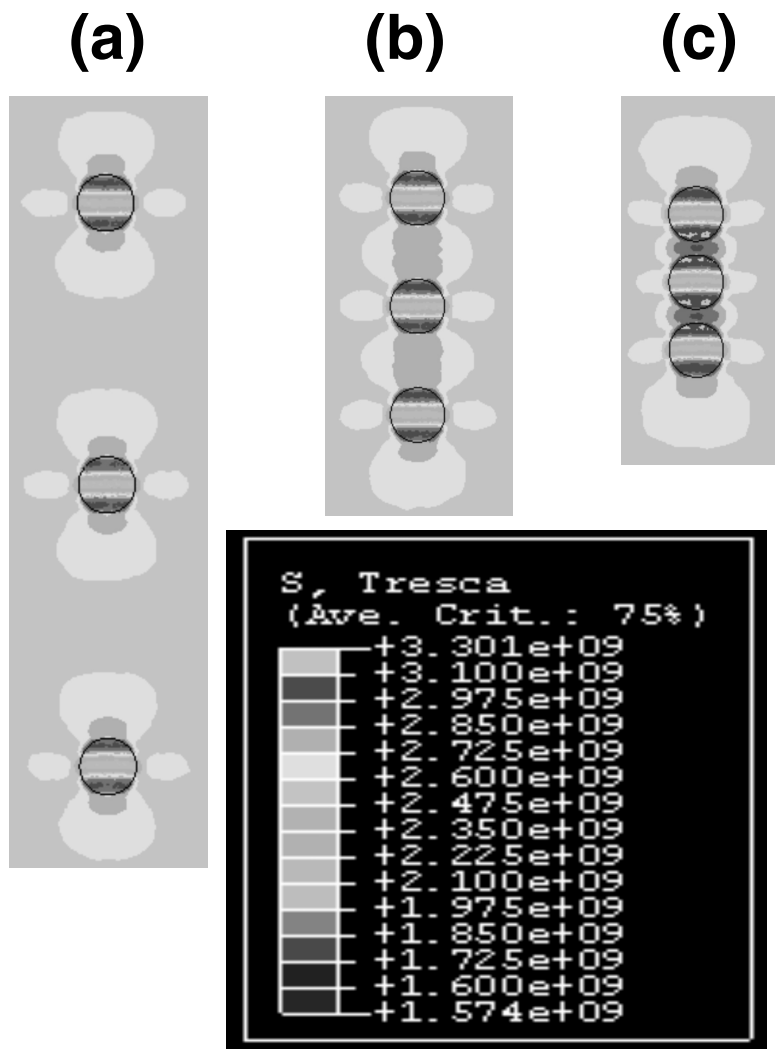


FIG. 9—Tresca stress maps for a stringer oriented perpendicularly to the contact surface for various distances between constitutive inclusions: (a) $d/R_m=8$; (b) $d/R_m=2$; (c) $d/R_m=0.5$; case of alumina inclusions; and applied stress taken at the center of the Hertz field for a Hertz pressure of 4.2 GPa.

concentration. This mechanism is illustrated in Fig. 9 for a simple stringer made of three spherical inclusions of equal diameter, with the axis of the stringer oriented perpendicularly to the contact surface. Figure 9 shows the Tresca stress maps in a plane passing through the axis of the stringer for different normalized distances d/R_m between the inclusions. It appears clearly that the local stress increases markedly when d/R_m is decreased down to 0.5. It should be pointed out that any local stress increase generally induces a stronger

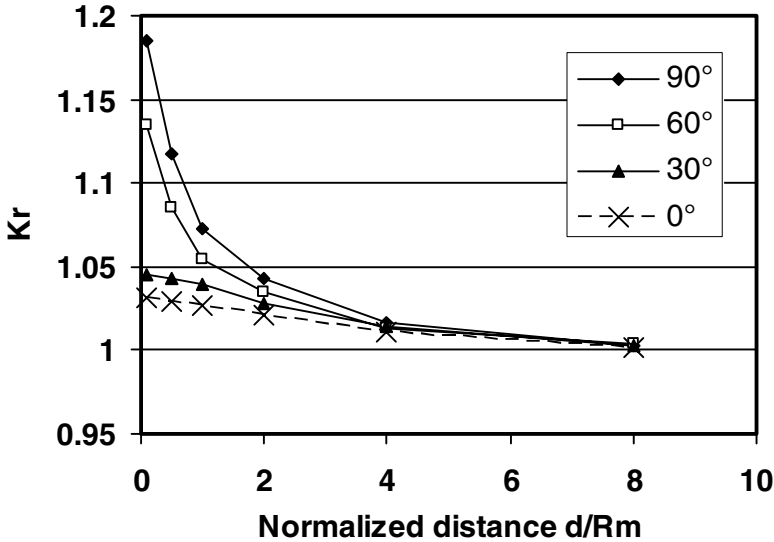


FIG. 10—Stress enhancement factor, K_r , for alumina stringers as a function of the normalized distance between neighboring inclusions for various orientations of the stringer (0° =parallel to the contact surface; 90° =perpendicular to the contact surface).

effect on the local cyclic micro-plasticity that governs the damage accumulation. Moreover, it has been shown elsewhere [15] that the number of inclusions in the stringer does not influence significantly the maximal value of the local stress. In other words, for a given inclusion the stress concentration in the surrounding matrix is mainly influenced by the next neighboring inclusion. This fact enables one to characterize the interaction between constitutive inclusions by studying simple stringers made of three inclusions.

Furthermore, a stress enhancement factor has been defined for the stringers by

$$K_r = \frac{[\sigma_{Tr}^{\max}]_{\text{stringer}}}{[\sigma_{Tr}^{\max}]_{\text{single I}}}$$

where:

$[\sigma_{Tr}^{\max}]_{\text{stringer}}$ and $[\sigma_{Tr}^{\max}]_{\text{single I}}$ =maximal values of the Tresca stress in the steel matrix in presence of interaction between neighboring inclusions and for a single inclusion of the same type, respectively.

The stress enhancement factor, K_r , for alumina stringers has been plotted versus the normalized distance between neighboring inclusions in Fig. 10 for various orientations of the stringer with respect to the contact surface. As already mentioned, the stress enhancement increases steeply as the normalized distance decreases. Moreover, Fig. 10 shows that K_r is also markedly influenced by the orientation of the stringer, the orientation perpendicular to the contact surface (orientation of 90°) being the more detrimental. It is worth mentioning that this stress enhancement factor has to be calculated prior to the statistical

calculations described in the following paragraph. Thus, the stress concentrations do not need to be computed again for every particular stringer, but they can be simply derived from a single constitutive inclusion. This method drastically reduces the computation time for the statistical predictions.

Statistical Prediction of the Distribution of Fatigue Lives

The statistical implementation of the model for predicting the distribution of fatigue lives has been carried out in the following way.

- For each considered steel heat, a series of virtual flat washers is built using a Monte Carlo method for drawing successively the number of stringers of a given type in each flat washer, the position of each stringer center (uniform distribution), its total length, and the corresponding d/R_m parameter. It should be specified that the number and the total length of each stringer are drawn according to the experimental distribution determined by ultrasonic echography (see example in Fig. 8), while the d/R_m parameter is drawn according to the distribution determined by X-ray micro-tomography (see Fig. 6).
- Every flat washer is subjected to a virtual fatigue test, and the number of cycles N_F to fatigue failure is computed by applying the models for crack nucleation and growth described above for every stringer in every flat washer; according to the weakest link rule, the fatigue life N_F of a given washer is defined as the minimum value among all the values corresponding to the various stringers; once they are analyzed, all these minimum values form the distribution of lives related both to the heat and the conditions of fatigue testing.

Application and Comparison with Experimental Data

The assessment of the methodology for predicting the distribution of fatigue lives has to be carried out by comparing the computed fatigue lives to experimental ones. This has been achieved by investigating eight different heats chosen with different cleanliness levels. For that purpose, on the one hand, rolling contact fatigue tests have been carried out on flat washer specimens using SNR Roulements type FB2 test rigs [16]. On the other hand, the cleanliness of every heat has been characterized by ultrasonic testing, as explained previously (see Fig. 8).

The experimental tests were conducted according to the “sudden death testing method” [16]. Thus, 50 washers were tested by groups of five units under a nominal Hertzian pressure of 4.2 GPa. For time reasons, fatigue tests without failure after running 1000 h or so were interrupted.

On the simulation side, for a given heat, first a series of 10 000 virtual washers has been built up and the corresponding lives have been calculated. From experience feedback, the constitutive inclusions in the distributed stringers were assumed to be alumina inclusions in the presented computations. This procedure provides the overall rolling fatigue properties of the heat, which can be presented as the cumulative probability of failure versus life, in a classical logarithmic diagram. Figure 11 shows an example of such a representation for

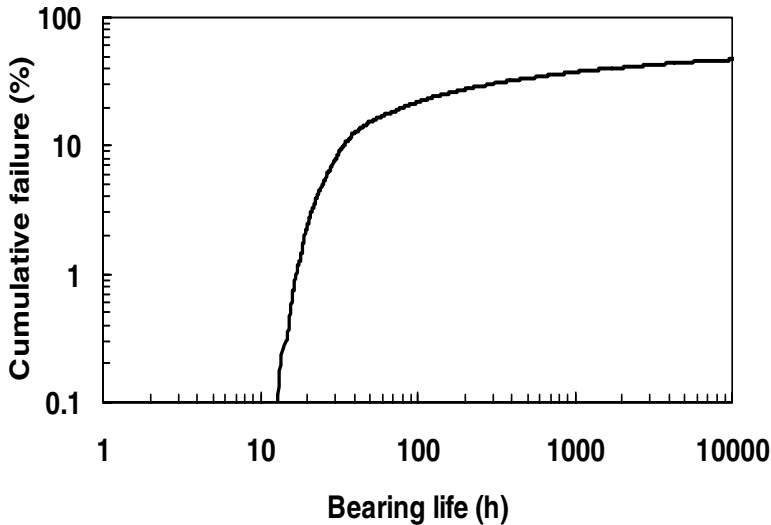


FIG. 11—Simulated cumulative probability of failure versus life for 10 000 flat washers plotted in a log-log diagram for heat D; $1\text{ h} \equiv 675\,000$ cycles.

heat D characterized in Fig. 8. As shown in earlier works [3], the overall cumulative distribution of fatigue lives exhibits a typical Γ shape. The tendency of the cumulated failures to saturate well below 100 % is linked with the significant probability to have washers with no detrimental stringers in the core of the Hertzian stress field.

Finally, in order to compare simulations with experimental data, the simulated data have to be analyzed according to the “sudden death” method. Moreover, since a single experimental test exhibits a random character, the scattering range of the results has to be determined. For that purpose, 100 virtual sudden death tests of 50 washers were built up from the 10 000 virtual tests. The cumulative distribution function of each number of failure (Nos. 1–5) can be derived from these results. Then, the scattering range of the results with a confidence interval of 90 % was obtained from these distribution functions. Finally, the five experimental data points and the computed scattering range were plotted in the same Weibull diagram for every heat. Four examples are reported in Fig. 12.

First, it can be seen that the computed scattering ranges are generally wide. As a matter of fact, in Fig. 12 the scattering range appears to be narrower for heat D that is less clean than the other heats. Such a relationship between the scattering width and the inclusion concentration can be explained easily. For a clean heat the probability to find a harmful inclusion in the stressed region below the raceway is low, and hence fatigue failure is a quite random process. In contrast, for a less clean heat the probability to find harmful inclusions in the same region is close to one, thus reducing the randomness of the fatigue process. Second, the experimental data fall well within the computed scattering ranges, except for some interrupted tests (open symbols in Fig. 12). In fact,

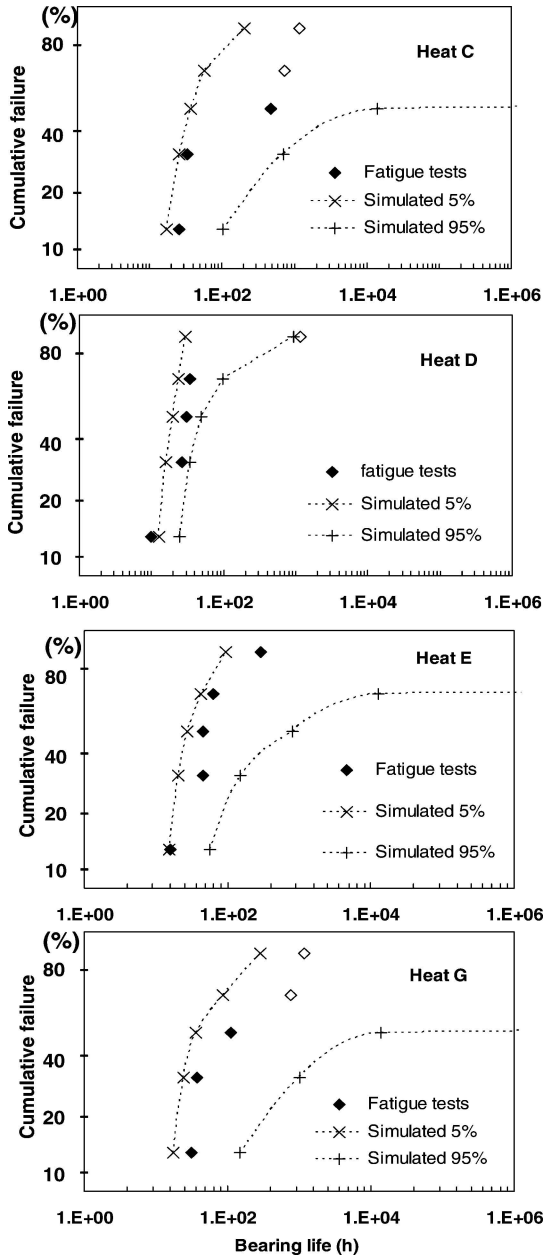


FIG. 12—Comparison between the experimental cumulative probability of failure versus life for sudden death tests plotted in a Weibull diagram and the computed scattering ranges for a confidence interval of 90 %: Examples of heats labeled C, D, E, and G. The open symbols indicate that the corresponding tests were interrupted.

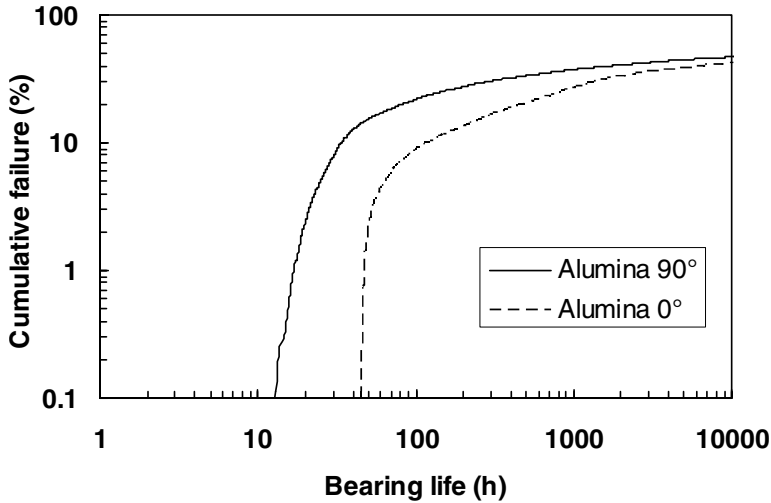


FIG. 13—Comparison between the predicted cumulative probability of failure versus life for stringers oriented parallel or perpendicular to the contact surface (inclusion concentrations are those of heat D).

these results remain consistent with the predicted ranges because an interrupted test potentially refers to a real life longer than the number of cycles to interruption. A similar agreement was obtained for all the investigated heats. All these good agreements show that the new model presented in this section is able to predict reasonably well the statistical distribution of fatigue lives of bearings in relation with the steel cleanliness characterized by high frequency ultrasonic testing and X-ray micro-tomography. It is worth pointing out that estimating the scattering ranges by means of the experimental fatigue tests would take about 4 years for one heat with seven rigs running continuously. This shows the great advantage of the modeling route.

Finally, regarding the influence of the stringer orientations, Fig. 13 shows the comparison between the distributions of fatigue lives for stringers oriented parallel and perpendicular to the contact surface. It appears that for the perpendicular orientation (90°), the first failures occur significantly earlier than for the parallel orientation (0°). This is mainly due to the higher stress concentration for the perpendicular orientation, which has been shown in Fig. 10. In fact, such a larger stress significantly reduces the number of cycles to crack nucleation, but it does not influence the tendency to saturate as much. This is not surprising because, as mentioned above, the saturation level of cumulated failures is mainly linked to the concentration of stringers, which was the same for both simulations. It should be mentioned that this influence of the orientation is in good agreement with the trend observed experimentally, which exhibits a L_{10} fatigue life about five times longer for washers machined from a bar to obtain the “parallel orientation” of the stringers than for those machined from the same bar to obtain the “perpendicular one.” Unfortunately, for these results,

the high frequency ultrasonic characterization of the heat was not available, and hence, the corresponding simulation was not possible.

Conclusion

A new methodology has been developed for predicting the distribution of fatigue lives of bearings from appropriate characteristics of the inclusion population in the steel.

This work has shown that high frequency ultrasonic testing provides reliable data on concentrations of harmful stringers in bearing steels. X-ray microtomography has been used for the first time to visualize the 3D shape of inclusions and stringers. This new technique appears to be quite complementary of the high frequency ultrasonic testing. Indeed, it is able to provide the internal details of stringers, but it is limited to small volumes, while in contrast ultrasonic testing enables one to investigate larger volumes of material but without distinguishing internal details of stringers.

The physically based model presented here makes a clear link between the cleanliness of the steel and its fatigue performances. It has been validated by comparing the predicted distribution of lives with a set of fatigue test data carried out according to the sudden death method.

Moreover, this model is able to (i) account for the influence of the stringers orientation with respect to the contact surface, (ii) make predictions of distribution of fatigue lives for large batches of bearings representative of populations of bearings used in industry, and (iii) estimate the scattering of results for fatigue tests performed on reduced numbers of parts.

Furthermore, it should be pointed out that, in its present form, the applicability of the model is limited to operating conditions for which inclusion initiated damage is the controlling process for bearings failure. For instance, this excludes applications with extremely high contact pressures involving bulk metal fatigue or with very clean steels and surfaces indented by pollution of the lubricant.

In the future, this methodology could be used to optimize the forming process of bearing steels. For instance, the reduction ratio of bars or tubes used for manufacturing bearings governs the characteristics of inclusions and particularly of stringers. Using the presented methodology would enable one to predict the fatigue properties of the final product versus the parameters of the forming process.

Acknowledgments

The ASCOMETAL company is gratefully acknowledged for its continuous support. The writers also wish to thank Dr. G. Baudry and Dr. E. Henault for fruitful discussions and encouragements.

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Claude Bathias¹

Gigacycle Fatigue Properties of Bearing Steels

ABSTRACT: The concept of a fatigue limit at 10^6 cycles that was introduced by Wohler in 1850 seems nowadays to be the wrong magnitude when considering the fatigue life of modern industrial applications such as automotive engines that run typically billions of cycles. The necessity to increase performances in terms of lifetime and safety in steel fabricated products remains an issue, although testing is becoming more time consuming and costly with current design products. It is the aim of our research to develop better understanding of fatigue properties of steels in the gigacycle domain. One way to approach this is to use a piezoelectric fatigue testing system working at ultrasonic frequency. This paper will describe the research done on several 52100 steels coming from Europe (NF100C6) and Japan (SUJ2). Different heat treatments have been applied to obtain different microstructures: Martensite and bainite with and without residual austenite. The fatigue tests, up to 10^{11} cycles, have been carried out using piezoelectric machines working at 20 and 30 kHz under tension-compression, tension-tension, and torsion only. A set of specimens has been tested also with a circular notch to observe the effect of stress concentration. Finally the thermal dissipation during testing has been determined using an infrared camera. It is confirmed that the influence of inclusions is a key problem.

KEYWORDS: gigacycle fatigue, SN curve, bearing curves, fish eye, flaw effect

Introduction

According to the technical applications, a modern approach of the fatigue of components should be divided into three domains.

- Low cycle fatigue (LFC) when the fatigue life is less than 10^5 cycles (bulk-plasticity)

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- Megacycle fatigue (MCF) between 10^5 and 10^7 cycles (surface plasticity)
- Gigacycle fatigue (GCF) beyond 10^7 cycles (local-plasticity)

It is interesting to point out that most of the structural components are working between 10^4 and 10^{11} cycles. The fatigue life of a car engine is ranging beyond 10^8 cycles; the large diesel engines for ships work up to 10^9 cycles. It should be realized that, currently, the technical limit for fatigue life is about 10^{10} cycles.

It is well known that LCF is predicted using a Coffin–Manson law, and MCF is related to the Wohler curve concept. But there is no general model for GCF.

From a historical point of view, it was said for the first time (1984–1997) by Japanese researchers [1–3] that metals can fail between 10^7 and 10^8 cycles. However, the gigacycle fatigue failure in many alloys was experimentally proven very recently (1990s) up to 10^{10} cycles by Bathias and co-workers [4,5]. Up to now, the SN curve (stress-number of cycles) is still assumed to be a hyperbolic function, but in reality the asymptote is not horizontal. This means that fatigue initiation mechanisms from 10^6 – 10^9 cycles is a new topic of great interest for advanced technologies, and SN curves that are not asymptotic must be determined in order to guarantee the real fatigue strength in high cycle regime.

Also it should not be forgotten that the fatigue strength is strongly dependant on the number of cycles chosen to stop the test.

This study was focused on investigating the gigacycle fatigue properties of bearing steel 52100, using three different frequencies, 35 Hz and 20 and 30 kHz, under push-pull loading, tension-tension loading, and torsion loading. A comparison with the Japanese results using a rotating bending fatigue test and the results from research using axial loading demonstrates a difference between the obtained SN curves.

Experimental Procedure

Testing Methods

The test methods for determining gigacycle fatigue properties have been developed and applied by many researchers around the world. In 1950, Mason and Wood built the first 20 kHz piezoelectric machine. Thanks to the ultrasonic fatigue technology, we can establish the typical life of a great number of components that achieve 10^9 – 10^{10} cycles during its life in service [6–14].

For some 20 years, our laboratory has carried out diverse fatigue life investigations on many materials with the use of a piezoelectric fatigue testing machine at 20 kHz, which has approved capability for producing fatigue property data in gigacycle regime.

A new piezoelectric fatigue machine with 30 kHz was used to explore the fatigue properties of bearing steel AISI-SAE 52100 in the domain of the 10^9 – 10^{11} cycles. Here a comparison of fatigue testing with different frequencies was made. Using a 20 kHz piezoelectric fatigue machine, it takes us around 14 h to obtain 10^9 cycles, 6 days for 10^{10} cycles, and 58 days for 10^{11} cycles.

The stress excitement principle is the same for both at 30 kHz as 20 kHz piezoelectric fatigue machines, where the vibration of the specimen is induced by a piezo-ceramic converter, which generates acoustical waves in the specimen through a power concentrator (horn) in order to obtain a desired material displacement and an amplification of the stress [10,11]. The resonant specimen dimension and stress concentration factor were calculated by FEM for 20 and 30 kHz.

Fatigue tests were performed in an open environment at 20 and 30 kHz with the piezoelectric fatigue machine using a symmetrical cyclic stress ($R = -1$) and cooling with compressed air at 20°C to keep the temperature of the specimen below 30°C.

The dynamic displacement amplitude of the specimen end face is controlled by an optical sensor and computer control in order to keep the stress constant during the test.

The test is automatically stopped when the frequency reduced to 19.5 kHz that is attributed to cracks development prior to fracture. Both the specimen and machine are described in other papers [10].

In order to test the torsion fatigue limit of the alloy up to 10^{10} cycles, an ultrasonic torsion fatigue system was designed. Attached to the transducer are two horns, one serve to amplify the longitudinal mechanical displacement and the other is to amplify the torsion angular displacement. A torsion fatigue specimen designed to run in resonance with the system is then attached to the horn. The specimens were designed so that the maximum strain is located in the minimum gage section.

Testing Material and Specimens

Three different kinds of specimens made in steel type 52100 were used in this research. (Fig. 1).

The smooth specimens to be tested at 20 kHz come from three different manufacturers (NF 100C6*, NF 100C6**, and JIS SUJ2). These specimens have an hourglass-shape with a minimum diameter of 3 mm and a radius of curvature at the reduced section (notch radius) of 31 mm. The fatigue results obtained from every group of specimens were compared in order to investigate the effect of the used processing method.

The smooth specimen to be tested at 30 kHz was machined by only one manufacturer (NF 100C6*), an axisymmetric specimen with a minimum diameter of 3 mm and having a hourglass-shape radius of 16 mm.

Finally, the notch specimen from only one manufacturer (NF 100C6*) with a minimum diameter of 6.4 mm and 60° notch with a round notch radius of 2 mm was used to evaluate the intrinsic resistance of the microstructure and the thermal effect on vibratory fatigue.

The three steel-makers have manufactured the specimens independently although they have used the steels with similar chemical compositions and applied similar heat treatments and machining operations. Tables 1 and 2 show the chemical compositions and some measured mechanical properties for the different the bearing steels.

The surface roughness (Ra) was measured to be $3.21 \mu\text{m}$ maximum in the

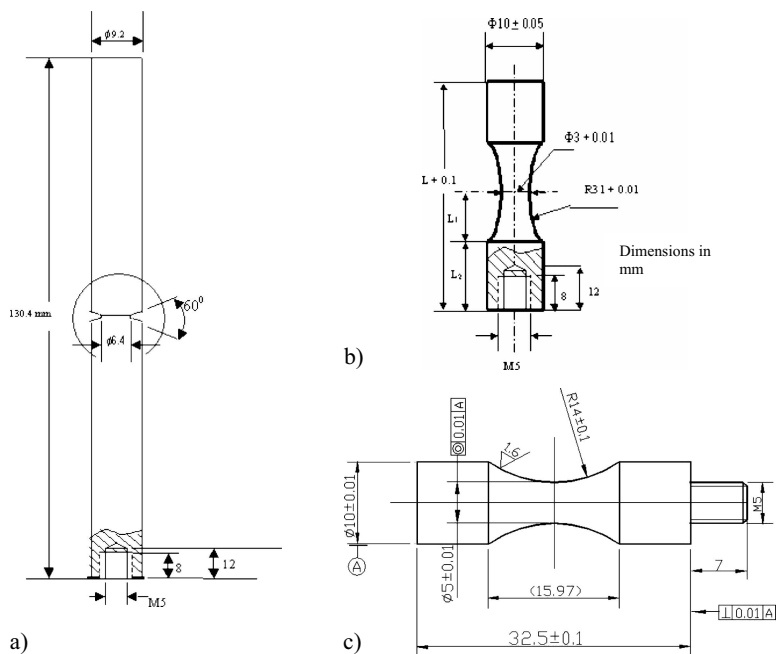


FIG. 1—Specimens used in gigacycle testing for bearing steels. Tension of 20 kHz with notch (a) and without notch (b). Torsion (c) at 20 kHz.

longitudinal direction on all specimens. Some temperature measurements using an infrared camera were carried out to evaluate the thermal effect on vibratory fatigue at 20 kHz.

Gigacycle Behavior of Bearing Steels

The SN curves obtained from all the fatigue tests are shown in Figs. 2–8. Regarding material (NF 100C6), it is very important to obtain its fatigue property at a very high number of cycles. It is essential to point out that this kind of bearing steel can fail up to 2×10^{10} cycles in tension-compression. The main purpose of Fig. 3 is to demonstrate that the shape of the SN curve is continuously decreasing from the mega- to the gigacycle regime. No step is observed in the SN curve when the load condition is tension-compression.

The results obtained using different kind of specimen at different frequency levels show that fatigue rupture can occur beyond 10^9 cycles, and the differ-

TABLE 1—Chemical composition (wt %) of tested bearing steels.

	C	Si	Mn	P	S	Cr	Mo	Ni
NF 100C6 martensitic and NF 100C6 bainitic (wt %)	1.0	0.2	0.3	0.001	0.008	1.4	0.03	0.14

TABLE 2—Mechanical properties of tested bearing steels.

	E_s (GPa)	σ_y (MPa)	UTS (MPa)	P (kg·m ⁻³)	HV30	HRc
NF 100C6 martensitic	210	1158	2316	7860	778	63.2
NF 100C6 bainitic	210	1179	2358	7860	703.9	60.3

ence of fatigue resistance can decrease by 150 or even 200 MPa between 10^6 and 10^{10} cycles under these conditions. The fatigue limit of bearing steels defined with a statistical analysis between 10^6 and 10^7 cycles cannot guarantee a safe design.

As to the notch specimen, fatigue resistance is far lower than for the smooth specimen, which is due to the existence of a larger stress concentration at notch surface. Again, the SN curve for the notched specimens is slowly decreasing after 10^6 cycles if compared with the smooth specimens curve. It means that the difference between the two SN curves at 10^9 cycles is smaller than that at 10^6 cycles (about 150 and 300 MPa), but the fatigue strength for notched specimens is ranging at about 600 MPa. It seems that the effect of stress concentration at the tip of a notch is similar in the megacycle than in the gigacycle regime (Fig. 7).

In order to obtain the experimental fatigue limit at 2×10^6 and 10^{10} cycles, the staircase method was used for calculating the fatigue resistance. The NF 100C6* smooth specimens have a fatigue strength of 809.82 MPa at 10^{10} cycles and a fatigue strength of 939.1 MPa at 2×10^6 cycles. On the other hand, to compare the fatigue resistance results according to the ASTM standard [12,13], a new calculation was made using another statistical method in which it is possible to avoid the use of the staircase method to find out the fatigue resistance [14].

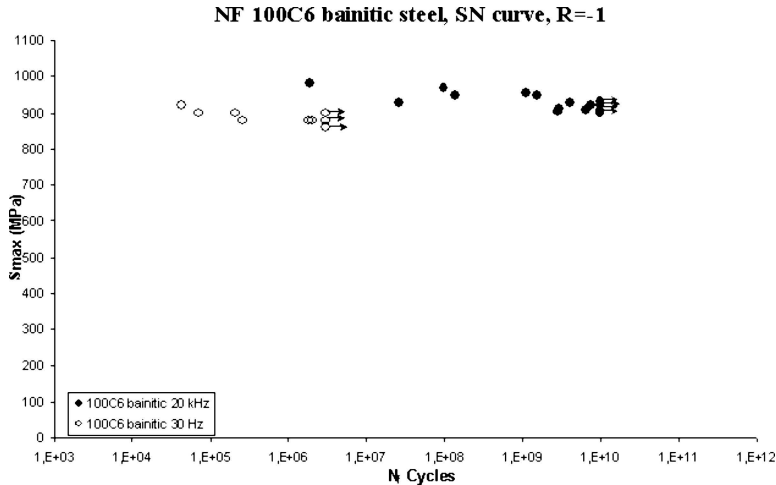


FIG. 2—Fatigue SN curve NF 100C6 bainitic, R=-1, 20 kHz, and 30 Hz.

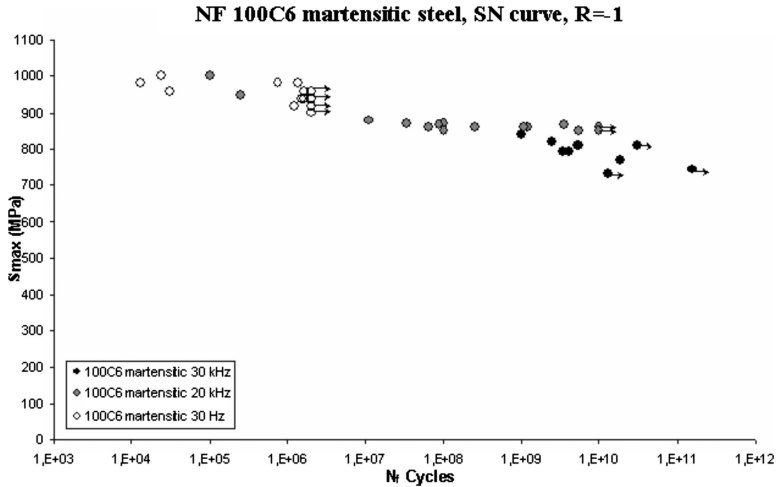


FIG. 3—SN curve, martensitic steel NF100Cr6, $R=-1$, 30 Hz, 20 kHz, and 30 kHz.

Roughly speaking, it is said that $\sigma_f - 3s$ gives a probability of fracture close to 0. In our testing, the standard deviation (s) is close to 10 MPa; the true infinite fatigue limit should be $\sigma_f - 30$ MPa. Thus, we can get the fatigue limit at 10^9 cycles from the fatigue resistance at 10^6 cycles, which is calculated to be 905 MPa using the statistical method. The experimental fatigue limit at 10^{10} cycles is 809.8 MPa. This differs by 100 MPa if compared with the calcu-

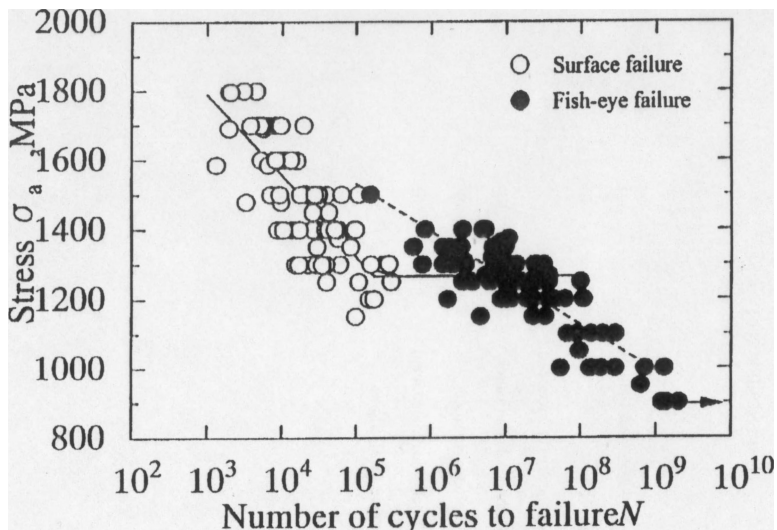


FIG. 4—Typical SN curve for SUJ2 in rotating-bending loading. Results from Sakai *et al.*

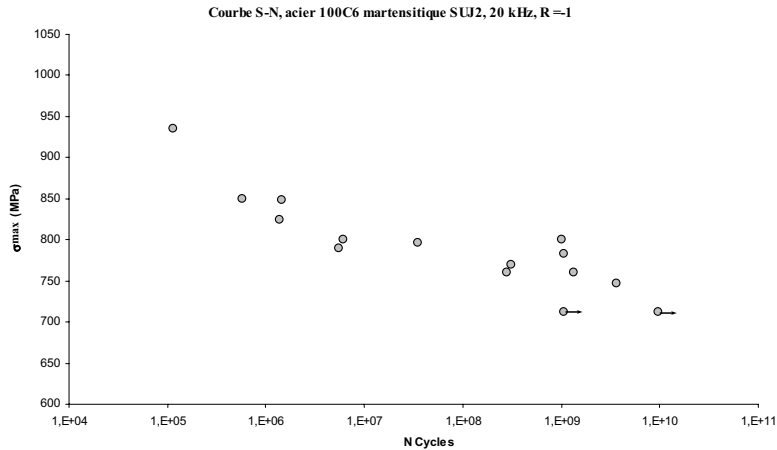


FIG. 5—Typical SN curve for SUJ2 in tension-compression. $R=-1$ at 20 kHz.

lated results. Consequently, it seems that the high frequency fatigue test is the most effective way to obtain the materials fatigue strength in the high cycle regimes.

Figure 8 shows the SN curve of 100C6 steel for torsion fatigue test results at 20 kHz. Fatigue lifetime increases as the shear stress amplitude decreases in the life range of $10^6 \sim 10^{10}$ cycles such that the life is between 380 and 390 MPa in the very high cycle regime.

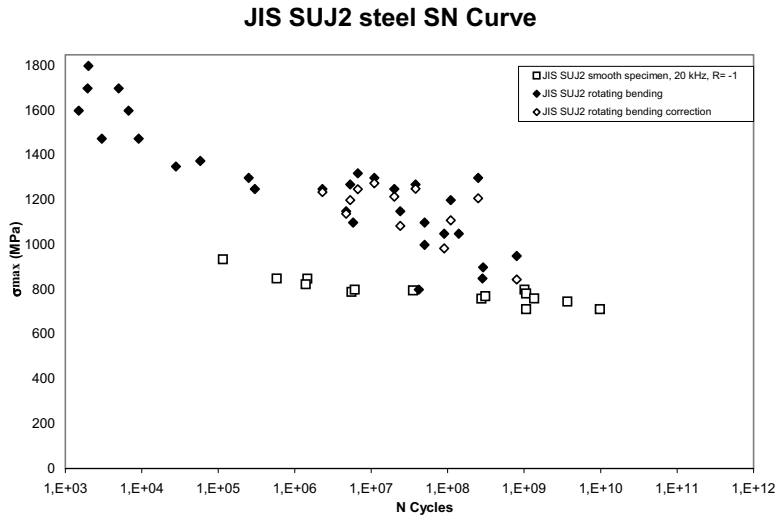


FIG. 6—Comparison of tension-compression and rotating bending results for SUJ2 (original and after correction).

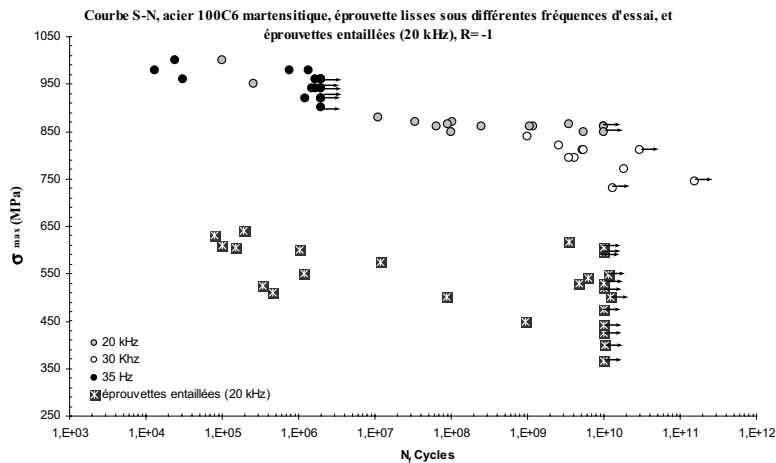


FIG. 7—SN curves of NF100C6 steel without notch (35 Hz, 20 kHz, and 30 kHz) and with notch (20 kHz) in tension-compression ($R=-1$).

Fractography

Scanning electron microscopy observations show that the fatigue crack initiation site for the NF100C6* bearing steel appeared in two typical rupture modes, i.e., at the surface between 10^5 and 10^6 cycles, and within the specimen for above 10^7 cycles where the initiation site is principally associated with nonmetallic inclusions like Al_2O_3 , CaO, and SiO_2 .

As regards the NF100C6** bearing steel, failure was always initiated within the specimen, principally located at nonmetallic inclusion like Al_2O_3 , SiO_2 , CaO, and MgO under tension-compression.

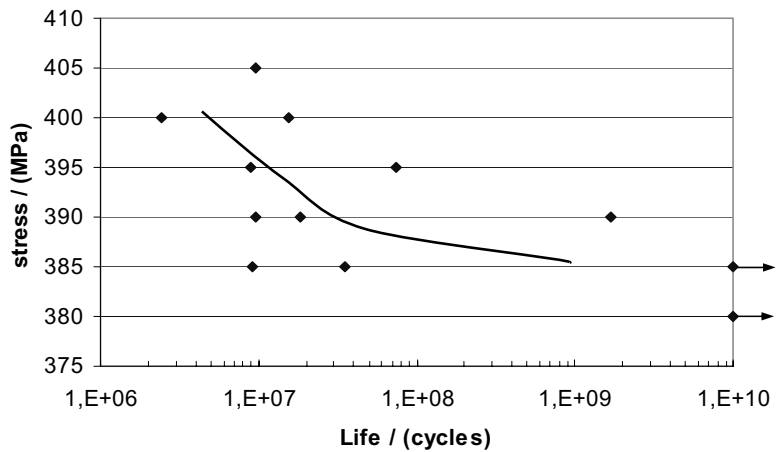


FIG. 8—Gigacycle SN curve in torsion for 100C6 steel (20 kHz).

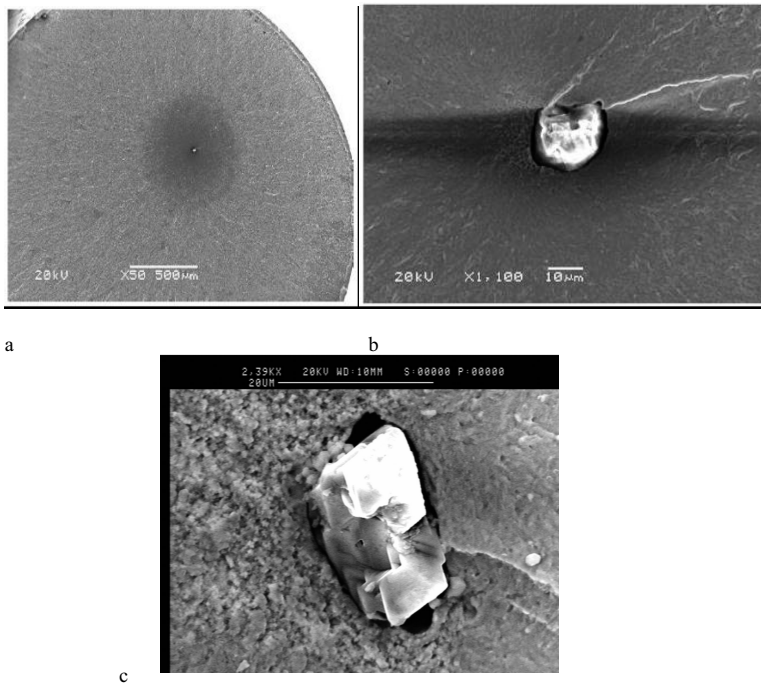


FIG. 9—NF 100C6 martensitic steel. [(a) and (b)] Internal rupture mode fish eye at 716 μm of the surface, $\sigma_{\text{max}}=865$ MPa, $N_f=3.5429 \times 10^9$ cycles, $R=-1$, and $f=20$ kHz. Inclusion (8.9 % MgO+13 % Al_2O_3 +28.1 % CaO+50.1 % SiO₂) NF 100C6 bainitic steel. (c) Internal rupture mode $\sigma_{\text{max}}=929$ MPa, $N_f=2.6608 \times 10^7$ cycles, $R=-1$, and $f=20$ kHz. In tension-compression the critical inclusion is an oxide.

For JIS SUJ2 bearing steel, there are also two typical rupture modes, with initiation sites both at the surface and within the specimen. The fracture initiation site of the notched specimen NF100C6 is always located at the surface of the notch. The remaining machining marks of the notched specimen accelerate the fatigue rupture.

For the smooth specimen tested under the lower cyclic stress, the internal stress concentration coincides with the internal rupture mode. Figure 9 shows the rupture mode of a smooth specimen made from NF 100C6* steel that was tested under 865 MPa cyclic stress. The fatigue life is up to 3×10^9 cycles, where the initiation site is below the surface and associated with an Al_2O_3 inclusion. In Fig. 9(a), we find the optical dark area found by other researches [5,15]. Figure 9 shows the internal inclusion rupture mode of a smooth specimen made from NF 100C6** steel tested under 965 MPa cyclic stress with a resultant fatigue life up to 2.6×10^7 cycles. X-ray microprobe analysis (Energy Dispersive System) revealed that the inclusion consists of 54 % Al_2O_3 , 9.95 % O₂, 25.7 % CaO, and 10.4 % MgO.

As a result, the inclusions within the specimens are seen as the main rupture mode when tested in the very high cycle regime. We can see for the smooth

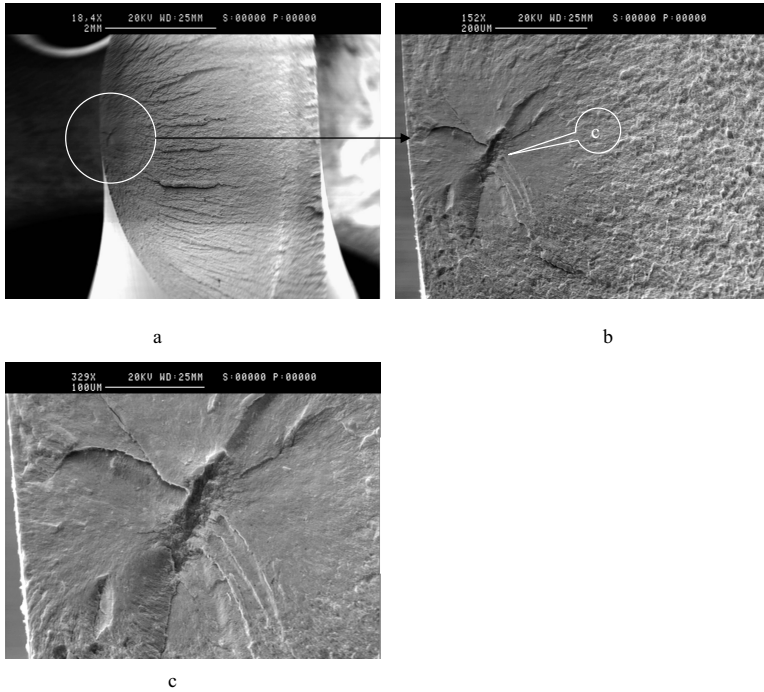


FIG. 10—Fatigue crack initiation from the subsurface in torsion. In torsion the critical inclusion is a sulfur. (a) Fatigue fracture surface, (b) fish eye, and (c) MnS inclusion ($\tau_{max}=370$ MPa; $N_f=3.06 \times 10^7$ cycles)

specimens that fatigue failure continues to occur between 10^7 and 10^{10} cycles.

The first failure after 10^{10} cycles has occurred using 30 kHz vibratory fatigue system in NF 100C6* steel under a maximal stress of 750 MPa.

For all the fatigue tests that were conducted using a cyclic shear stress just above the fatigue limit, a uniform cracking behavior in the failed specimens has been observed. A small crack is nucleated on one of the planes of maximum shear that is able to propagate up to a length of several hundred microns. Thereafter crack branching occurs with propagation on one of the planes corresponding to the maximum normal stress that grows and eventually leads to failure.

As expected under torsion, the fatigue crack initiation starts at the surface at higher stress amplitudes, but there is a transition where crack initiation changes from surface to subsurface mode. The fatigue crack initiation under torsion for most the specimens is due to a subsurface inclusion as shown in Fig. 10 (near to a dark area region) The EDS analysis shows that the composition in the fatigue initiation zone corresponds more often with a MnS type inclusion rather than a metallic oxide inclusion.

The subsurface crack initiation was observed in most fatigue tested samples with the fatigue life between 10^7 and 10^9 cycles. The fatigue fracture surface has a typical fish eye appearance; however, it is different from the fa-

tigue fractures in specimens subjected to axial cyclic loading, where the fatigue crack always initiates at subsurface or near to the specimen surface and not from internal inclusions, although it has very high cycle fatigue life. Also, the dark zone around the inclusion is not circular but sometimes more elliptical and smaller in area than that observed on the fatigue fracture surfaces subjected to the axial cyclic loading [1–4].

Rotating Bending Internal Crack Stress Correction

At this point, it is interesting to compare the fatigue curves in rotating bending and in tension-compression [16]. In Japanese literature, a lot of results had been reported for JIS SUJ2 by Sakai [3], Murakami [2,10,17], and others. It has been found by Japanese researchers that the internal failure initiation that appears after a plateau is more important in rotating loads than in tension-compression. The same kinds of curves have been produced by Nishijima [1] in other steels tested with rotating bending loads.

Whereas the SUJ2 is not exactly the same as NF100C6 steel, it seems that the stepwise SN curve is common for both and therefore more related to the rotating bending behavior.

The SUJ2 has been tested in our laboratory under tension-compression mode at 20 kHz, and no step in the SN curve could be derived. Thus, it is assumed that this step depends on the calculation of the nominal stress, which is decreasing from the surface to the center to get zero for a given loading (Figs. 4–6).

Since the maximum stress in rotating bending is located at the surface of the specimen, a correction of the stress should be considered for the specimens with internal crack mode in function of the distance to the surface (Fig. 6).

Figure 4 shows the curve originally obtained by professor Sakai, where we can observe the formation of two straight lines, the first corresponding to a short fatigue lifetime at a high stress level followed by a step. A correction of the stress is realized by considering the distance of the inclusion to the surface for finally obtaining a continuously decreasing curve. The comparison of both of them is shown at where we can obtain a correction value of 110 MPa considering the bending effect.

Finally, the corrected values could be used to develop a part of the SN curve under tension-compression for the same steel (Fig. 6).

This correction is aimed to give differences between the curves to obtain a relationship that allows us to identify and predict the lifetime in the future. By using an experimental curve and the derive relationship, we can develop a rotating bending curve without making experiments and vice versa.

Thermal Dissipation

Investigations were made to study thermal effects on the occurrence of fatigue rupture. This was done by producing an infrared thermograph that allows us to reveal a heating at the tip of the notched specimens, during high frequency testing. The intention being to find an explanation on why the fatigue strength

of the smooth specimens is higher than that of the notch specimens.

These tests were possible using an infrared camera, which converts the infrared radiation into numeric signals. These signals show the thermal image of the considered object. The temperature, which we will consider later, is the maximum temperature for a pixel of the resonance zone. In order to increase the emissivity and also to avoid different reflection problems, the specimen was darkened with the "soot." The resultant emissivity was accordingly kept at 0.9.

During the test, we observed a rise in the surface temperature of the specimen. The most important heating is slightly localized below the notch, where the stress is the highest. This confirms the results obtained by FEM (ANSYS), where the maximum stress is not located at the notch bottom as during classic tests but at a distance of 0.4 mm above and below the notch.

Without cooling the specimen, the temperature at the tip of the notch increases quickly. However, it is strongly dependent on the cyclic stress and the fatigue life. It is important to notice that the maximum temperature is less than 50°C for a cyclic stress around 600 MPa in the gigacycle regime. Without cooling the temperature of the notched specimens loaded under tension-compression, at 618 MPa and 20 kHz, stays constant at 45°C between 5×10^7 to 10^9 cycles. Even in the megacycle regime the temperature is not exceeding 65°C at 820 MPa.

Those results confirm that the heating dissipation is quite low in high strength steels loaded at 20 kHz. The measured temperature shows that it is also not a problem for bearing steels if the design of the specimens is correct and the control of the piezoelectric machine is accurate. The heating dissipation is also a good way to detect the fish eye initiation in bearing steels during a fatigue test using an infrared camera, which detects the temperature increase.

Murakami Model

The Murakami model allows the prediction of the material fatigue limit for materials containing three-dimensional defects like nonmetallic inclusions or porosities. The necessary input data are Vickers hardness of material (HV), the defect dimension $\sqrt{\text{area}}$ in μm , and the loading criterion R . Murakami based his model mainly on fatigue steel tests under rotational bending, but he affirms that it is applicable to other metals.

Murakami does not specify the number of cycles since the model considers that the lifetime is unlimited. Previously, it has shown that the "fatigue limit" can vary appreciably between 10^7 and 10^{10} cycles. Results from Murakami and other literature do not originate generally from high frequency fatigue tests, and we can suppose that the maximum number of cycles is lower than 10^8 cycles. Therefore, the necessity to predict a fatigue limit at long lifetime (gigacyclic regime) and verify the Murakami model efficiency into this regime (Eq 1) is possible using NF 100C6 bainitic data

$$\sigma_w = \frac{1.56(HV + 120)}{(\sqrt{\text{area}})^{1/6}} \left[\frac{1 - R}{2} \right]^\alpha \quad (1)$$

where:

TABLE 3—Application of the Murakami model for the gigacycle fatigue of bearing steel ($R=-1$).

Experimental Fatigue Strength (MPa)	$\sqrt{\text{Area}}$	Murakami Model Fatigue Strength σ_w (MPa)	Murakami Error (%)	Murakami Modified Model σ_w (MPa)	Error Percent After Correction
912	16.7	803.9	13.4	943.1	3.3
907	21	773.8	17.2	907.7	0.1
949	21.6	770.2	23.2	903.5	5.0
929	21.1	773.2	20.2	907.0	2.4
956	21.7	769.6	24.2	902.8	5.9
903	20.4	777.5	16.1	912.1	1.0
980	14	827.9	18.4	971.2	0.9
929	13.2	836.1	11.1	980.8	5.3
967	17.8	795.4	21.6	933.1	3.6

$C = 1.56$ for the internal defects,
 σ_w = fatigue resistance (MPa), and
 R = loading ratio

$$\sigma = 0.226 + HV \times 10^{-4}$$

Because loading ratio is $R = -1$, Eq 1, will be changed as follows:

$$\sigma_w = \frac{1.56(HV + 120)}{(\sqrt{\text{area}})^{1/6}} \tag{2}$$

Table 3 shows the results between our experimental results and those from Murakami using NF 100C6 bainitic data representing long lifetime, where the error percent is between 11.1 % and 24.2 %. However by changing the internal defect constant, the error can be between 0.1 % and 5.9 % using Eq 3.

Figure 11 shows the modified SN curve data using NF 100C6 bainitic steel, where the difference between Eqs 2 and 3 can be illustrated

$$\sigma_w = \frac{1.83(HV + 120)}{(\sqrt{\text{area}})^{1/6}} \tag{3}$$

It shows that the Murakami model can be useful into gigacyclic regime to predict the fatigue resistance, taking only care of inclusion size at nucleation crack.

Fish Eye Formation

The Murakami model is a good empirical approach to predict the fatigue strength of bearing steels depending on the defect size. However, the fatigue life that covers both crack initiation and crack growth is not evaluated. Assuming that the gigacycle fatigue is related to the fish eye formation, it is reasonable to

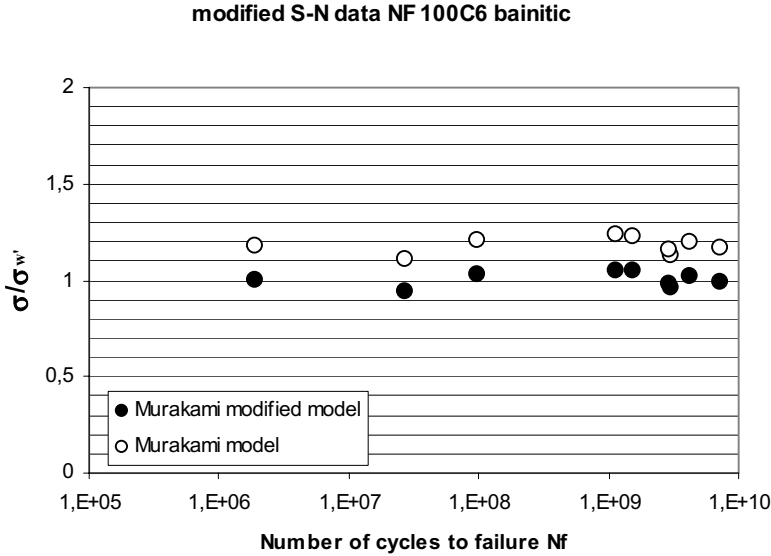


FIG. 11—Application of the Murakami model for NF100C6 steel.

compute the number of cycles for propagation by the integration of the Paris–Hertzberg law applied to a circular short crack. In gigacycle fatigue regime, the geometry of the fish eye initiation is a circle, which collapses arriving at the surface of the specimen.

In order to do this, one should refer to the general behavior pattern of the crack growth rate curve as illustrated by the equations in Fig. 12. It is assumed that small cracks such as those growing from small inclusions do not exhibit crack closure, so these equations in terms of ΔK_{eff} apply fairly well. They form an upper limit on crack growth rates for the small cracks in the “fish eye” range for which crack closure is minimal.

Estimating the life for a crack of this type beginning just above threshold, it is then appropriate to consider the growth law as

$$\frac{da}{dN} = b \left(\frac{\Delta K_{\text{eff}}}{E\sqrt{b}} \right)^3 \quad (4)$$

where for the circular crack growing in a fish eye the stress intensity factor formula is

$$\Delta K = \frac{2}{\pi} \Delta \sigma \sqrt{\pi a} \quad (5)$$

The integration to determine the crack growth life will begin here with the crack growth rate corner (as indicated by the star in Fig. 12), which we shall denote as ΔK_0 corresponding to an initial circular crack of radius, a_0 . It is found that

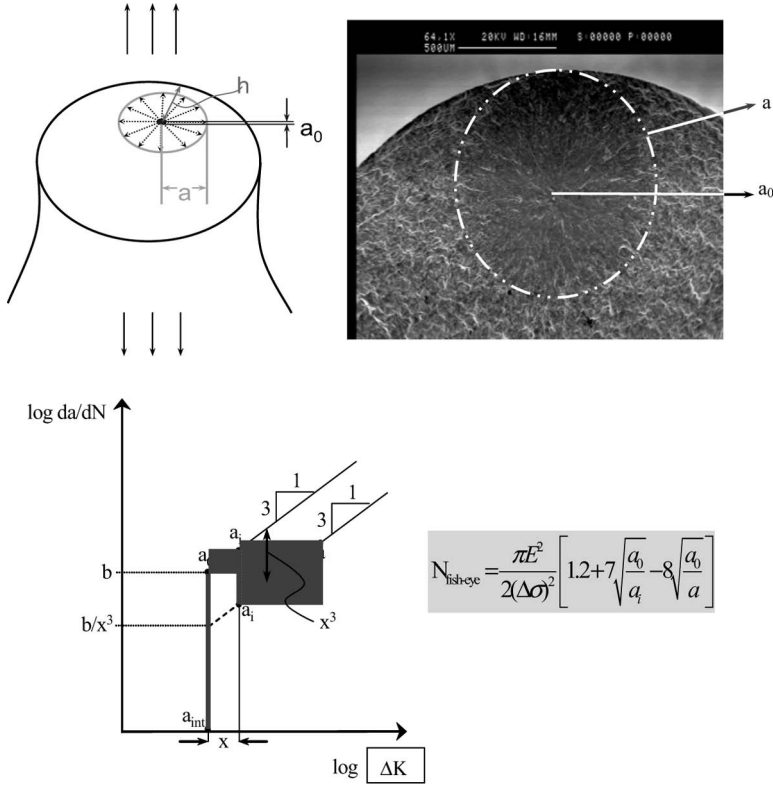


FIG. 12—Modelling of the fish eye formation from the Paris-Hertzberg law.

$$N_f = \frac{\pi E^2}{2(\Delta\sigma)^2} \tag{6}$$

This will result in the approximate number of cycles from the threshold point to eventual failure for an initial crack size a_0 as expressed above. For the completeness of these results, let us consider the number of cycles, which can be considered crack growth below the threshold point. The relation becomes

$$N_f \cong \frac{\pi}{12} \left(\frac{E}{\Delta\sigma} \right)^2 \tag{7}$$

Therefore it is shown that crack growth before the threshold point is of little consequence in gigacycle fatigue.

In the results presented in Tables 4 and 5, one may observe the numerical data and the measurements from fractographic evidence that verify reasonably the integration of crack growth laws for bearing steels. The number of cycles of crack growth, N_f , is about three orders of magnitude smaller or much less than

TABLE 4—Computation of the number of cycles of propagation in the fish eye for different assumptions in gigacycle fatigue (100C6 steel, 20 kHz).

$\Delta\sigma$ (MPa)	N_{total}	a_0 (μm)	a (μm)	$R \approx 0, b/27, \alpha = 100, \text{ and } a_{\text{int}} = 0.94a_0$						$R \approx 0.4, b/8, \alpha = 100, \text{ and } a_{\text{int}} = 0.94a_0$					
				$a_i = 100 \mu\text{m}$		$a_i = 125 \mu\text{m}$		$a_i = 200 \mu\text{m}$		$a_i = 100 \mu\text{m}$		$a_i = 125 \mu\text{m}$		$a_i = 200 \mu\text{m}$	
				$N_{\text{fish eye}}$	$N_{\text{fish eye}}$	$N_{\text{fish eye}}$	$N_{\text{fish eye}}$	$N_{\text{fish eye}}$	$N_{\text{fish eye}}$	$N_{\text{fish eye}}$	$N_{\text{fish eye}}$	$N_{\text{fish eye}}$	$N_{\text{fish eye}}$	$N_{\text{fish eye}}$	$N_{\text{fish eye}}$
860	2.97×10^9	12	413	5.25×10^5	4.36×10^5	2.78×10^5	2.12×10^5	1.88×10^5	1.45×10^5	2.12×10^5	1.88×10^5	1.45×10^5	1.45×10^5	1.45×10^5	1.45×10^5

TABLE 5—Computation of fish eye propagation for several failed specimens in the gigacycle regime and different bearing steels. $N_{fish\ eye}$ is of the order of 10^5 cycles or less.

Material-Specimen Number (Elastic Modulus GPa)	$\Delta\sigma$ (MPa)	N_{total} (cycles)	N_f (cycles)	a_o (μm)	a (μm)	h (μm)
SUJ2 steel (210) (UTS 2316 MPa)	-2	2.77×10^8	1.19×10^5	10	325	327
	-4	1.01×10^9	1.08×10^5	6	273	325
	-7	6.1×10^8	1.08×10^5	10	263	491
	-8	3.11×10^8	1.16×10^5	3.3	664	943
	-14	1.06×10^9	1.14×10^5	8	406	615
	-15	1.34×10^9	1.19×10^5	7	682	1150
100C6 steel (210) (martensitic steel) (UTS 2358 MPa)	-1	1.18×10^7	7.84×10^4	8	463	756
	-2	1.49×10^7	7.67×10^4	17	216	243
	-4	2.40×10^7	7.52×10^4	6	641	1441
	-5	4.61×10^7	7.52×10^4	17	614	535
	-6	1.08×10^8	7.52×10^4	10	304	371
	-7	2.97×10^9	9.37×10^4	12	413	1088
100C6 steel (210) (bainitic steel) (UTS 2358 MPa)	-10	8.71×10^7	7.68×10^4	13	420	448
	-3	2.93×10^9	8.36×10^4	4	300	1070
	-8	1.53×10^9	7.67×10^4	10.5	217	927
	-9	1.53×10^9	8.18×10^4	11	329	1070
	-10	4.15×10^9	8.01×10^4	10.5	217	235
	-11	1.11×10^9	7.60×10^4	11	313	636
	-12	2.86×10^9	8.46×10^4	10.2	305	420
	-14	2.66×10^7	8.01×10^4	6.8	324	858

1 % of the total life. It means that the key problem to improve the gigacycle fatigue of bearing steels is to reduce the internal crack initiation around inclusions starting from persistent slip bands.

Conclusion

The conclusions from this research are the following.

- The bearing steels can fail up to 10^{10} cycles in axial loading or in shear.
- The initiation site in the gigacycle regime is always related to nonmetallic inclusions even under shear conditions.
- The effect of microstructure flaws is more important in gigacycle fatigue than the surface roughness quality used in this study.
- The initiation sites produced under the gigacycle fatigue testing conditions are mainly within the steel (core) under axial loading and close to the surface under shear loading.
- There is a relation between the size of the inclusion and the fatigue strength.
- Generally, no significant effect of the stress cycle frequency from 35 Hz to 30 kHz could be observed both under axial loading and under shear loading conditions.
- Comparing our results with those of others, a difference between axial loading and rotating bending is found in the gigacycle regime.
- A model accounting for fish eye fracture formation is given in this paper.

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Takumi Fujita¹

Rolling Contact Fatigue Life Test Design and Result Interpretation Methods Maintaining Compatibility of Efficiency and Reliability

ABSTRACT: In this report, several methods for both rolling contact fatigue (RCF) life test design and result interpretation are introduced. These methods generate results using random numbers followed by Weibull distribution (i.e., Weibull random number). The first method illustrates a relationship between the minimum number of test specimens and the suspension time in a fixed time test required at L_{10} and L_{50} lives with an arbitrary reliability. This relationship is useful to maintain the qualitative reliability and avoid excessive quantitative testing. The second method can clarify a relationship between the given number of test specimens and resultant significant differences at L_{10} and L_{50} lives in an accelerated test with an arbitrary reliability. This relationship is also useful to estimate the appropriate number of test specimens based on statistical logic. Of note, calculations employing the Weibull random number can apply to not only RCF life test design but also estimations of the test results. The third method enables the determination of a range of L_{10} and L_{50} lives with an arbitrary reliability even if the number of test specimens is too small to estimate L_{10} or L_{50} lives from the Weibull plots. The fourth method can determine significant differences of L_{10} and L_{50} lives between any two given lots and allow a quantitative estimation of the minimum difference between their lives from data obtained by experiments. These methods provide techniques that are easier to understand as compared to the recent mathematical model, and they show enough flexibility to apply to almost all types of testing. These systems will therefore eliminate the need for qualified experiences related to the statistical design and result interpretation for RCF life testing.

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Introduction

Rolling contact fatigue (RCF) life testing is one of the essentials prerequisites for bearing performance estimation. Generally, the probability density distributions of RCF lives follow a Weibull distribution [1,2], and wide variations in life data are common. Therefore, conservative estimation of RCF life is extremely important in order to avoid potential quality problems regarding RCF life in practical use.

RCF life testing is classified into two main groups. The first is where conditions are similar to practical use. The second is where conditions are severe beyond practical use. The first group is called a fixed time test to estimate RCF life from little or few failure data. In the fixed time test, a determination whether to accept the required RCF life is derived from the suspension time without failure because the testing rarely causes failure in a short duration. The second group is called an accelerated test, and the test conditions are more severe than that of the fixed time test. In the accelerated test, differences in RCF life can be established as compared with L_{10} or L_{50} lives between any two given lots because the severe test condition causes several failures enough to estimate the accurate L_{10} and L_{50} lives through Weibull plots.

In a fixed time test, it is not easy to clearly illustrate a statistical and quantitative standard between the number of test specimens and the suspension time, although it is basically understood that a better RCF life is established by increasing both the number of test specimens and the suspension time without failure. For this problem, several mathematical solutions based on statistics have been introduced [3–5]. However, the methods used in the mathematical solutions are too confusing to interpret the content of formulas and to apply to multifactor RCF life test results.

In an accelerated test, it has been a complex problem to determine the statistically quantitative number of test specimens, although it can be expected that the greater the life difference, the greater the number of test specimens ensures a significant life difference with a high reliability. For this problem, Johnson proposed a series of diagrams for varying several conditions to estimate a significant difference between two lives [6]. However, this method was rarely employed in actual examination because the diagrams were not well established for all conditions.

In this report, atypical methods for RCF life test design and result interpretation are introduced. These methods use random numbers followed by Weibull distribution (i.e., Weibull random number). These techniques are utilized in order to ensure that equations are based on universally accepted mathematical logic. However, it is a rare case to apply calculation results obtained from random numbers to the practical testing strategies, especially to RCF life testing.

TABLE 1—*Calculable contents by this method.*

Test	Classification	Calculable Contents
Fixed time test	<u>Test design number 1</u>	How to determine the number of test specimens in order to finish testing before a required time limit
	Test design number 2	How to determine a suspension time when either the number of test specimens or the number of test equipments is fixed
	Test design number 3	How to renew a suspension time when failures of several specimens occur before designated suspension time without failure
	Test design number 4	How to determine whether the testing should be stopped or not when several specimens are failed at an extremely brief time
	<u>Result interpretation number 1</u>	How to estimate L_{10} and L_{50} lives from a suspension time without failure
	Result interpretation number 2	How to estimate the ranges of L_{10} and L_{50} lives from a few failure data
Accelerated test		How to determine the number of test specimens in order to determine significant difference at L_{10} and L_{50} lives between any two given lots with an arbitrary reliability, or how to determine required the minimum difference at L_{10} and L_{50} lives with an arbitrary reliability when either the number of test specimens or the number of test equipment is limited
	<u>Test design number 5</u>	
	<u>Result interpretation number 3</u>	How to estimate the significant difference at L_{10} and L_{50} lives between any two given lots with an arbitrary reliability from a set of L_{10} and L_{50} lives
	<u>Result interpretation number 4</u>	How to estimate the minimum difference at L_{10} and L_{50} lives between any two given lots with an arbitrary reliability from a set of L_{10} or L_{50} lives

Note: Calculations for the classifications with underlines are carried out in this paper.

Calculation Methods and Weibull Random Number Generation Method

These methods produce solutions for all contents in Table 1 by using Weibull random numbers. The basic idea to produce the solutions is the same in all contents of Table 1 and parallels applications of RCF life testing simulation employing Weibull random numbers to the RCF life test design and the result interpretation.

When generating Weibull random numbers, three parameters in the Weibull distribution must be defined. The probability density function of Weibull distribution is shown as follows:

$$f(x) = \frac{e}{\alpha} \left(\frac{x - \gamma}{\alpha} \right)^{e-1} \exp \left[- \left(\frac{x - \gamma}{\alpha} \right)^e \right] \quad (1)$$

where:

e = shape parameter (Weibull slope),

α = scale parameter, and

γ = location parameter. Thus, Weibull distribution has the three parameters as shown above: e , α , and γ . Therefore, three conditions related to the parameters are required in order to define a Weibull distribution function. The conditions are explained below.

- 1) The Weibull slope “ e ” approximated to 10/9 in ball bearings and 9/8 in roller bearings was established from the many past testing data [1,2,7].
- 2) A relationship between the location parameter “ γ ” and the L_{10} life was established [8] from the many past testing data [9–12] as follows:

$$\gamma = 0.05 \cdot L_{10} \quad (2)$$

- 3) The scale parameter “ α ” is calculated as a relationship of $\alpha(L_{10})$ Eq 3. The $\alpha(L_{10})$ equation is obtained by substituting the relationship of Eq 2 into a Weibull cumulative distribution function of Eq 4. The scale parameter α can then be evaluated from the resulting Eq 3 because a Weibull slope of 10/9 or 9/8 is already fixed and a L_{10} life is assumed for the purposes of simulations on all testing

$$\alpha(L_{10}) = \frac{0.95 \cdot L_{10}}{\{\ln(10/9)\}^{1/e}} \quad (3)$$

$$F(x) = 1 - \exp \left[- \left(\frac{x - \gamma}{\alpha} \right)^e \right] \quad (4)$$

Sample calculation with the Weibull slope value approximated at 10/9 and with the L_{10} life set at 100 h, the probability density function of Weibull distribution can be determined as follows:

$$f(x) = \frac{10/9}{720} \left(\frac{x - 5}{720} \right)^{1/9} \exp \left[- \left(\frac{x - 5}{720} \right)^{10/9} \right] \quad (5)$$

Thus, the probability density function of Weibull distribution is prepared in order to generate Weibull random numbers. Once a probability density function of Weibull distribution is identified, Weibull random numbers can be obtained from uniform random numbers by one of two well-known approaches: (1) Moro’s algorithm [13] or (2) the rejection sampling [13]. At this juncture, the uniform random numbers are essential to maintain high quality in terms of uniformity and periodicity within the specified subject range. Accordingly, Mersene twister method is adopted to generate the required high-quality random numbers up to $2^{19,937} - 1$ random numbers [14]. Of note, the calculation results in this paper contain a statistical error of approximately 5 % due to the use of random numbers. This paper presents only basic contents in Table 1, specified by classifications with underlines.

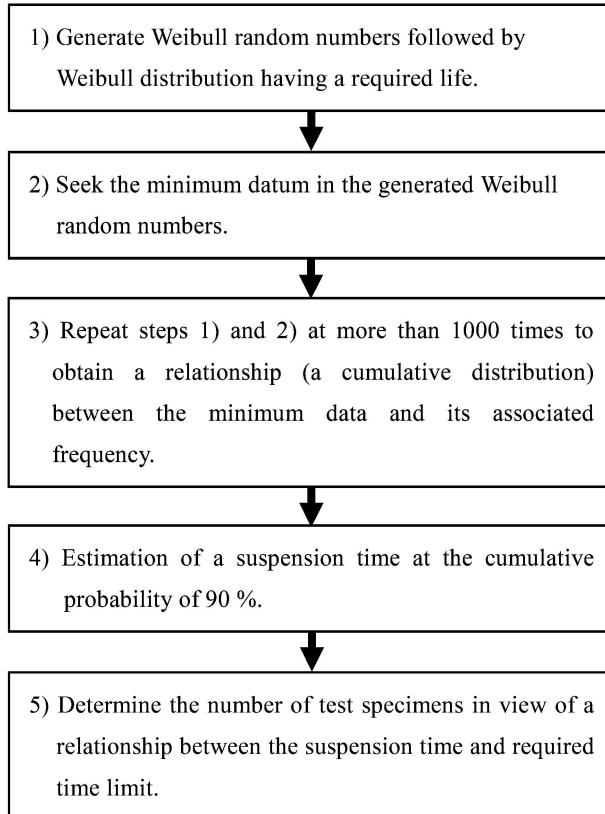


FIG. 1—Flowchart to design the number of test specimens in fixed time test.

Fixed Time Test Design and the Result Interpretation

Design of the Minimum Required Number of Test Specimens and a Suspension Time

Figure 1 is a flowchart to show the design process for establishing the number of test specimens in a fixed time test. In order to establish a flowchart proof of concept, a case study in practical RCF life testing was carried out. For purpose of evaluation, bearings having a L_{10} life of greater than 500 h were considered, and the testing time limit was set at 3 months (about 2100 h). Following these criteria, we need to design the number of test specimens to complete an actual RCF life testing up to 2100 h. It is a common belief that having a greater number of test specimens and a longer testing time without failure will ensure a longer RCF life. Therefore, it is logical to prepare both a sufficient number of test specimens and RCF life test equipments in order to achieve a L_{10} life of greater than 500 h in the limited time scenario. In the following, we consider whether or not the testing of the realistically selected six test specimens is

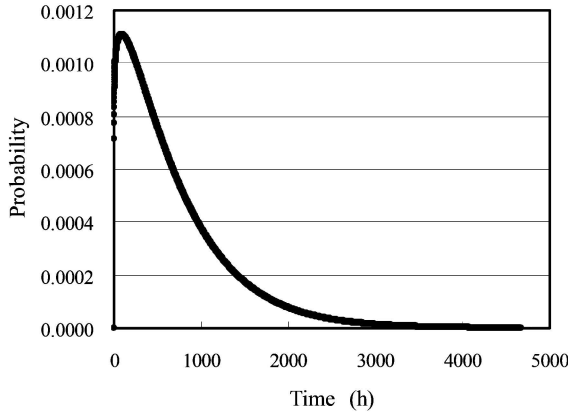


FIG. 2—Weibull probability density distribution for $L_{10}=500$ h and $e=10/9$.

adequate to achieve a L_{10} life of greater than 500 h up to the limited time of 2100 h. We also reconsider the rationale for establishing the appropriate number of test specimens, assuming that the six test specimens do not achieve the L_{10} life of greater than 500 h up to the limited time of 2100 h.

Figure 2 shows the Weibull probability density distribution with a L_{10} life of 500 h and a Weibull slope e value of 10/9. The Weibull slope e value of 10/9 in the distribution is a common value applied to the probability density distribution of RCF life in ball bearings. This procedure allows for the use of Weibull random numbers followed by the above probability density distribution. First, six Weibull random numbers are generated by a computer, which correspond to a simulation of RCF life testing for the six test specimens with a L_{10} life of 500 h. Subsequently, the minimum datum is obtained from the generated six random numbers. Both the generating of the six Weibull random numbers and the obtaining of the minimum datum are repeated at a minimum of 1000 times, and consequently we can obtain a relationship between the minimum data from six test specimens and its associated frequency. Of note, increasing the repetition results in the improvement of the statistical error. This relationship is represented as a cumulative distribution in Fig. 3.

Figure 3 shows that at least one of the generated six Weibull random numbers is less than 1499 h at the probability of 90 %, i.e., a suspension time of 1499 h obtained in this case results in ensuring the L_{10} life of greater than 500 h at the probability of 90 %. Therefore, the suspension time for six test specimens becomes 1499 h in order to ensure the L_{10} life of greater than 500 h. This suspension time of 1499 h is consequently acceptable because it is within the time limit of 2100 h. The procedure for the case study can be applied for an arbitrary reliability and an arbitrary probability of failure such as L_1 , L_{10} , or L_{50} life. In this paper, only the easiest calculation was presented in the case of non-failure. This approach can be applied for the others such as the calculations of the test design No.3 in Table 1. For example, when observing one

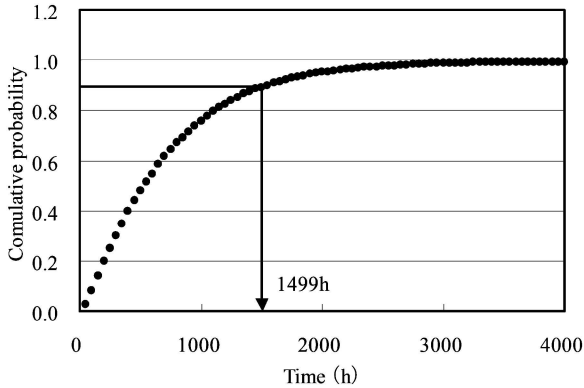


FIG. 3—Cumulative distribution to estimate a suspension time prior to failure (required L_{10} life=500 h, $e=10/9$, and number of test specimens=6).

failure before the end of the test, a second suspension time can be obtained from the relationship between the second minimum data and its associated frequencies.

Figure 4 shows the suspension time prior to failure depending on the number of test specimens and values of Weibull slope. The suspension time decreases exponentially with an increase in the number of test specimens. This goes without saying that the greater the number of test specimens can remarkably shorten the testing time. Additionally, the suspension time decreases exponentially with an increase in the value of Weibull slope because Weibull probability density distribution with the same fixed L_{10} life is distributed toward a shorter life with an increase in the value of Weibull slope. For purpose of efficiency, increasing the number of test specimens and/or selecting such conditions, which allow for an increase in the value of the Weibull slope, is important to shorten the testing time.

Estimation of Rolling Contact Fatigue Life from Fixed Time Test Results

Figure 5 is a flowchart showing the steps required in order to estimate L_{10} life from a suspension time prior to failure. As explained in the previous section, a case study in practical RCF life testing was carried out. In the previous section, we arrived at the conclusion that the suspension time for six test specimens became 1499 h in order to ensure the L_{10} life of greater than 500 h. In this section, we consider the estimation of a L_{10} life when the suspension time without failure attains 3000 h, exceeding the designed suspension time of 1499 h in the previous section. This will go without saying that these data are required to be the L_{10} life of greater than 500 h pre-calculated in the previous section.

First, six Weibull random numbers followed by a Weibull distribution with a L_{10} life of less than 3000 h are generated by computer. At the points, the Weibull slope value of $10/9$ is used, and the L_{10} life is set at less than $1/10$ of the suspension time of 3000 h. Subsequently, the minimum datum is obtained from

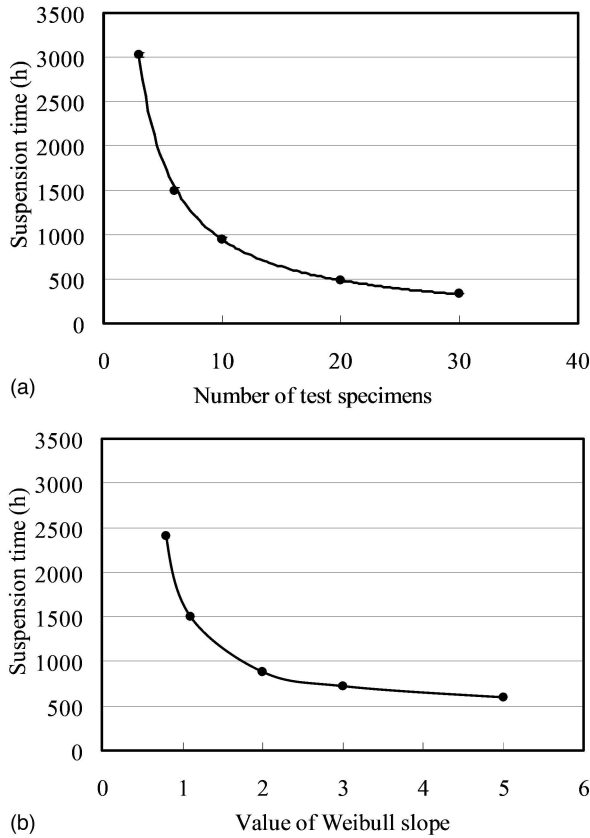


FIG. 4—Suspension time prior to failure depending on the number of test specimens and value of Weibull slope. (a) Dependence of suspension time on the number of test specimens (required L_{10} life=500 h and $e=10/9$). (b) Dependence of suspension time on the value of Weibull slope (required L_{10} life=500 h and number of test specimens=6).

the generated six Weibull random numbers. Both generating of the six Weibull random numbers and obtaining the minimum datum are repeated at a minimum of 1000 times. Consequently we can obtain a probability that the minimum datum in the six Weibull random numbers exceeds 3000 h. In the other words, this simulation confirms the probability that all the data obtained from six test specimens with L_{10} life of 300 h exceed 3000 h.

Second, the above process is applied over a gradually increasing L_{10} life span. With the results, we can obtain the probability of exceeding the suspension time of 3000 h in all six test specimens for each additional L_{10} lives in their respective Weibull distributions. This relationship in Fig. 6 is represented to estimate a L_{10} life from the suspension time prior to failure. Figure 6 shows that the cumulative distribution produces a clear L_{10} life of 975 h at which the probability exceeding 3000 h in the all six test specimens is representative of

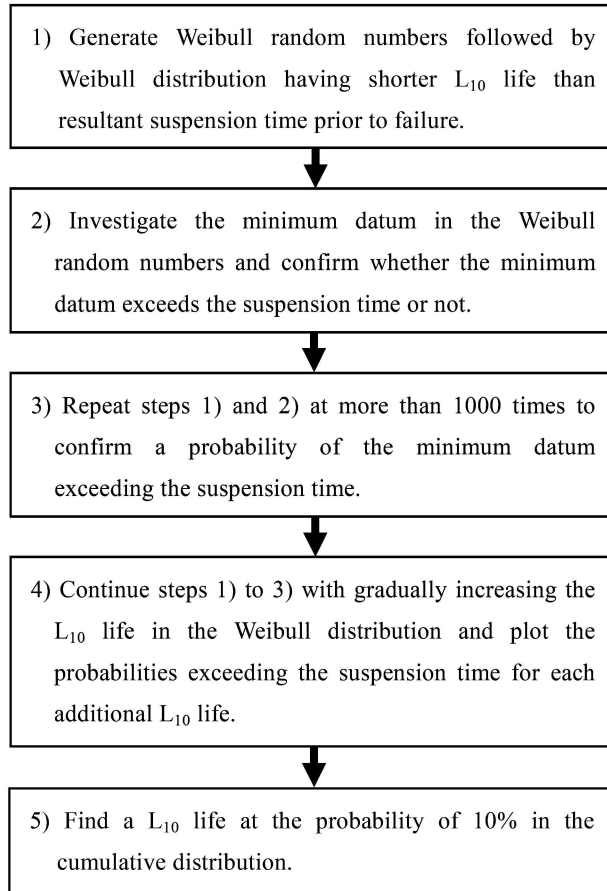


FIG. 5—Flowchart to estimate L_{10} life from suspension time prior to failure.

only 10 %. Therefore, we conclude that the suspension time exceeding 3000 h in the all six test specimens ensures a L_{10} life of greater than 975 h. This approach can be applied for the case with a few failures, such as the calculations of the result interpretation No. 2 in Table 1.

Accelerated Test Design and Result Interpretation

Design of the Minimum Required Number of Test Specimens in Accelerated Testing

Figure 7 is a flowchart to explain the rationale for the design process required to set the number of test specimens in an accelerated test. As explained in the previous sections, a case study in practical RCF life testing is carried out. In the actual accelerated testing, the life differences among several lots are discussed.

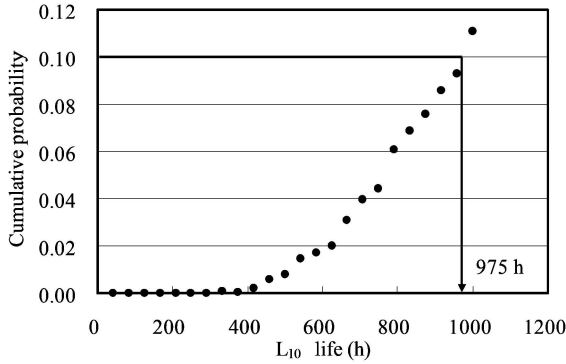


FIG. 6—Cumulative distribution to estimate L_{10} life from the suspension time prior to failure ($e=10/9$ and number of test specimens=6).

In this section, we consider first how to clarify the minimum required life difference to ensure significant statistical difference between any two given lots having three test specimens. A Weibull slope value of 10/9 is employed in below calculation.

First, a set of three Weibull random numbers followed by a Weibull distribution with an arbitrary L_{10} life is generated by a computer. Subsequently, a set of L_{10} lives is obtained by Weibull plots for each of a set of three Weibull random numbers. In a series of Weibull plots, median ranks are set by Johnson's method [6], and the estimation of regression curve is carried out by the least-squares method with regard to location parameter in the Weibull distribution [15]. Following this, a ratio of L_{10} lives is then calculated from the set of L_{10} lives.

Second, the process as stated above, to find a L_{10} life ratio, is repeated at least 1000 times, and a resulting cumulative distribution of L_{10} life ratio is plotted as in Fig. 8. From Fig. 8, a wide variation of the L_{10} life ratio from 0.042 to 21.8 is demonstrated with a reliability of 90 %, even though the estimated L_{10} lives are obtained from Weibull random numbers with the same L_{10} life. It is easily understood that testing containing a small number of test specimens, such as three specimens, produces results with a wide variation of L_{10} life ratio from 0.042 to 21.8 with a reliability of 90 %. In other words, either the life difference of greater than 21.8 times or that of less than 0.042 times is required in order to determine reliable significant statistical difference between any two given lots when limited to only three test specimens.

Third, variations in the L_{10} and L_{50} life ratios are calculated with an increase in the number of test specimens. Figure 9 shows a relationship between the number of test specimens and the variations of L_{10} and L_{50} life ratios with a reliability of 90 %. The variation illustrates an exponentially decrease with an increase in the number of test specimens. From these results, it can be concluded that 45 and 15 test specimens are required in order to create a rational

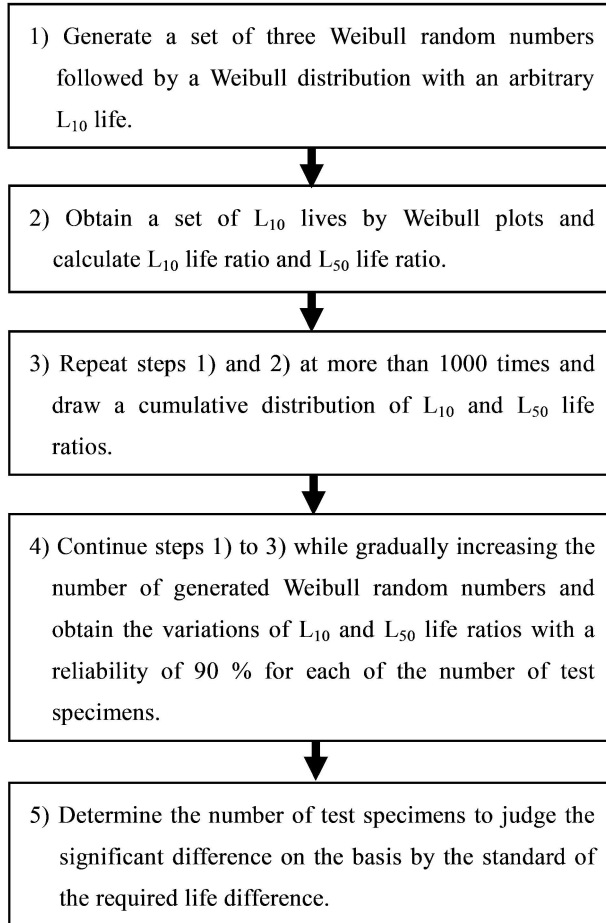


FIG. 7—Flowchart to determine the number of test specimens for an accelerated test.

comparison for respective L_{10} life and L_{50} life when the significant difference between any two given lots is judged by the standard of the twice the life difference.

Figure 10 shows the relationships for each of the Weibull slope value $e = 0.8, 10/9$, and 3 between the number of test specimens and their variations in the L_{10} life ratio. As explained before, the essential points in accelerated testing are as follows.

- 1) The L_{10} life difference between any two given lots decreases with an increase in the number of test specimens in order to establish a significant difference.
- 2) The number of test specimens can be reduced by comparing results at L_{50} life rather than that at L_{10} life.

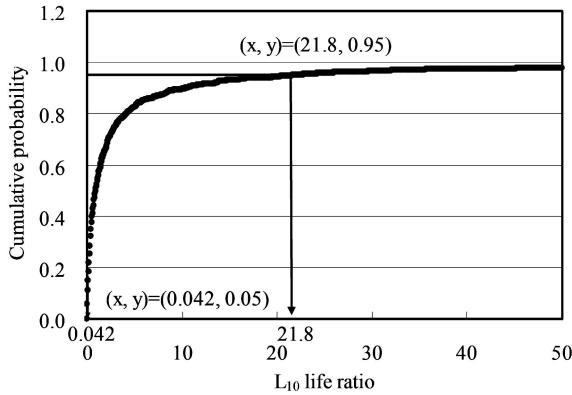


FIG. 8—Cumulative distribution of L_{10} life ratio ($e=10/9$ and number of test specimens=3).

- 3) The number of test specimens can be reduced when the test conditions allow for a higher Weibull slope value.

Estimation of Significant Difference and the Minimum Life Difference in Accelerated Testing

Figure 9 reveals the required life difference to establish a significant difference in accelerated testing. For example, when preparing ten test specimens, we can easily estimate the required life difference of five times according to the simulated data plots at the L_{10} life in Fig. 9. However, Fig. 9 does not allow for an

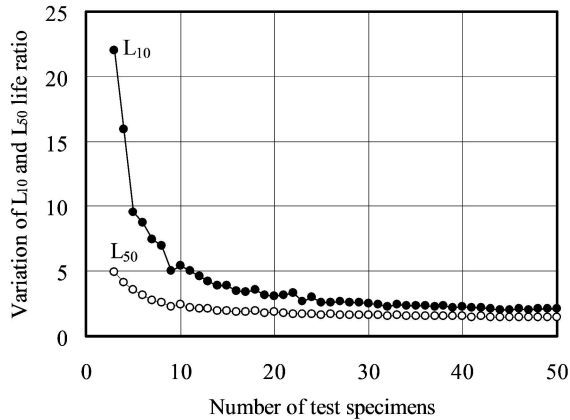


FIG. 9—Relationship between the number of test specimens and the variation of L_{10} and L_{50} life ratios with a reliability of 90 % ($e=10/9$).

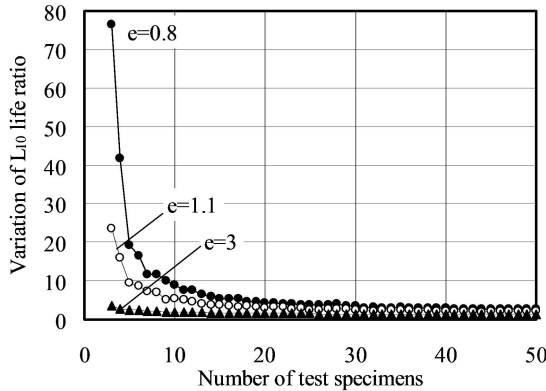


FIG. 10—Relationship between the number of test specimens and the variation of L_{10} life ratios with a reliability of 90 % ($e=0.8$, $10/9$, and 3).

estimation of the minimum life difference between any two given lots. In this section, we consider how to clarify the minimum life difference after confirming a significant difference between two specific lots.

Figure 11 is a flowchart showing the steps involved in estimating the minimum life difference of any two given lots with respect to accelerated test results. As explained in the previous sections, a case study in practical RCF life testing was carried out to form the basis of the flowchart. Initially, we suppose that estimating a ten times life difference between L_{10} lives can be confirmed by preparing ten test specimens. At this juncture, it is easy to ascertain the significant difference between two specific lots from Fig. 9.

The initial step to estimate the minimum life difference between any two given lots is the same as that of obtaining Fig. 8 in the above section, with the exception of generating ten Weibull random numbers. Figure 12 shows a cumulative distribution of the L_{10} life ratios as applied to ten test specimens. Figure 12 makes it clear that a variation of L_{10} life ratio produces a decrease from 21.8 to 4.90 at cumulative probability of 0.95 as a result of the greater test specimens, as contrasted with Fig. 8. Subsequently, in the same steps as obtained in Fig. 12, a cumulative distribution of L_{10} life ratios is calculated for each set of the Weibull random numbers with respective 1.1, 1.2, ..., 3 times difference between L_{10} lives. Here, the same cumulative distribution plots can be obtained to apply to L_{50} life. Accordingly, the variations of L_{10} and L_{50} life ratios with a reliability of 90 % are obtained for the respective given sets of the 1.1, 1.2, ..., 3 times different L_{10} lives, as shown in Fig. 13. Figure 13 shows diagrams for L_{10} and L_{50} lives between the variations in horizontal axis and their respective given sets of the 1.1, 1.2, ..., 3 times different L_{10} lives in vertical axis. Therefore, from this diagram, it is pointed out that a ten times life difference at L_{10} life is a rare result when testing a set of ten test specimens with less than a 2.1 times life difference. According to conservative estimate with a reliability of 90 %, a greater than 2.1 times life difference can be concluded from a result of ten times life difference between L_{10} lives.

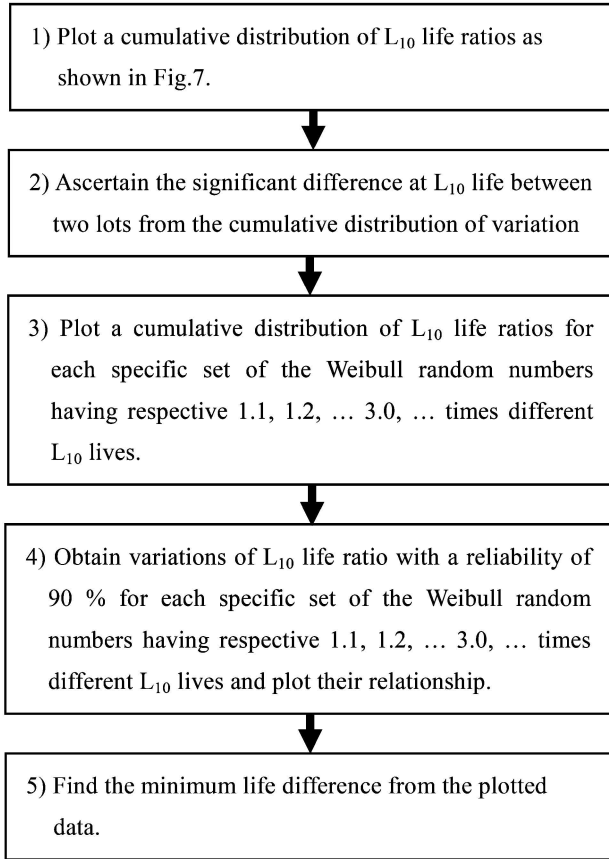


FIG. 11—Flowchart to estimate the minimum life difference of any two given lots with respect to accelerated test results.

Figures 14 and 15 are the data plots to estimate the minimum life differences calculated by varying the Weibull slope value and the number of test specimens, respectively. These results allow us to conclude that the greater the number of test specimen and the greater the value of Weibull slope acquire a greater life difference.

Discussion

All methods in this paper require the three assumptions listed below, which are based on the many past testing data [1,2,9–12].

- 1) RCF life follows the Weibull distribution.
- 2) The location parameter can be obtained from Eq 2.
- 3) The Weibull slope is fixed at 10/9 in ball bearings and 9/8 in roller bearings.

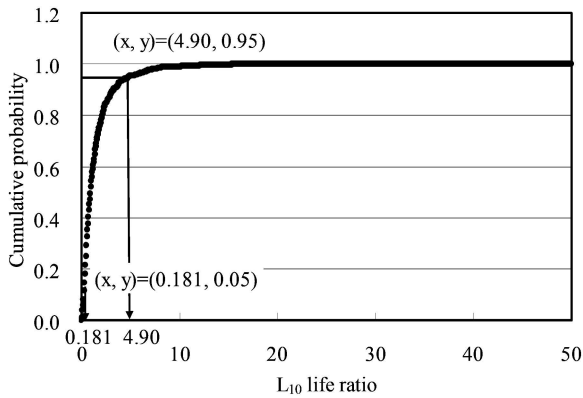


FIG. 12—Cumulative distribution of L_{10} life ratios ($e=10/9$ and number of test specimens=10).

The validity in terms of these assumptions is being investigated by Shimizu et al. [16]. If the universally accepted RCF life distribution is clarified on the basis of new experimental result or theorem, then the calculation processes as shown in this paper allow for modification. In such cases, the random number calculation can be easily applied for the other distributions, although the modulus of the distribution will still need to be determined on the basis of new experimental result or theorem.

Supposing, as presented in this paper, that the RCF life follows the Weibull distribution, the effect of location parameter and Weibull slope on the suspen-

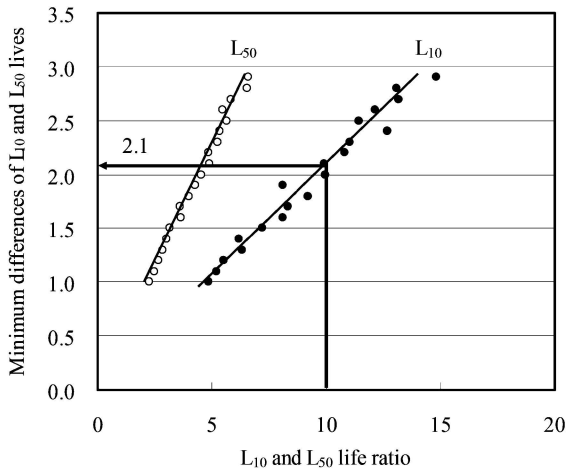


FIG. 13—Diagram plots to estimate the minimum differences of L_{10} and L_{50} lives with respect to accelerated test results ($e=10/9$, number of test specimens=10, and reliability=90 %).

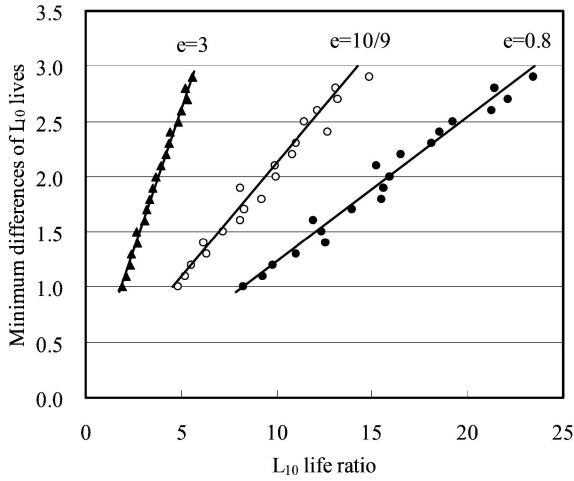


FIG. 14—Diagram plots to estimate the minimum L_{10} life differences with respect to accelerated test results ($e=0.8$, $10/9$, and 3 , number of test specimens= 10 , and reliability= 90%).

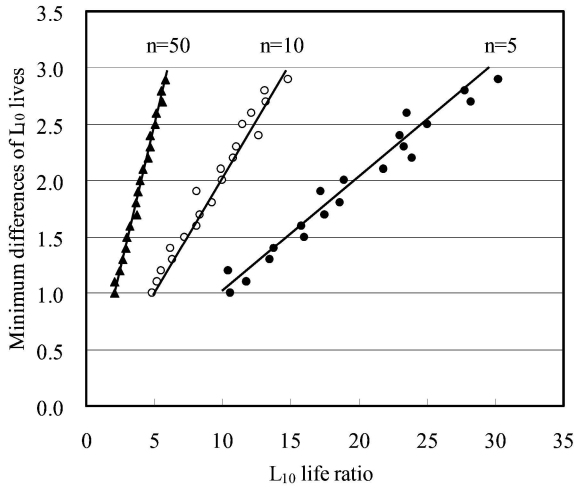


FIG. 15—Diagram plots to estimate the minimum L_{10} life differences with respect to accelerated test results of two lots (number of test specimens= 5 , 10 , and 50 , $e=10/9$, and reliability= 90%).

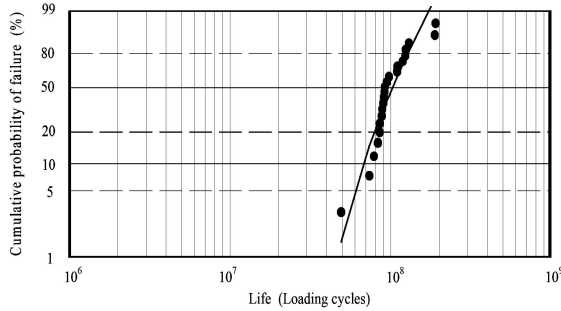


FIG. 16—An example of RCF life test result (point contact $P_{\max}=5.88$ GPa).

sion time calculation can be considered by the well-known equation [4] as follows:

$$T = (L_p - \gamma) \cdot \left(\frac{\ln(1 - 0.01R)}{n \ln(1 - 0.01p)} \right)^{1/e} + \gamma \quad (6)$$

where:

- T =suspension time with no failure,
- L_p =required Life at cumulative probability p ,
- p =cumulative probability (%),
- n =number of specimen,
- γ =location parameter, and
- R =reliability (%).

A relationship between the location parameter γ and the L_{10} life was established [8] from the many past RCF testing data [9–12] as shown in Eq 2. The effect of the location parameter on the calculation results produce relatively small. For example, the suspension times in the case of $\gamma=0$ and $\gamma=0.05L_{10}$ are 1600 and 1545 h, respectively, in order to ensure the L_{10} life of greater than 500 h at the probability of 90 % for the six specimens, assuming a Weibull slope of 10/9.

On the other hand, Eq 6 shows that the calculation result is affected exponentially by the uncertainty of the estimated Weibull slope. Therefore, in all methods for RCF life test design and result interpretation, it is critical to employ, as close as possible, to an actual Weibull slope as determined from real testing. Herein, we discuss only a practical rational process needed in order to obtain an appropriate value of Weibull slope from actual test results by using random number calculations.

In general, the value of Weibull slope in RCF life distribution changes depending on the state of the contact stress and/or the mode of failure, i.e., the value of Weibull slope changes depending on the test conditions. Figure 16 represents an example of an RCF life test result where the contact and lubrication conditions are point contact at $P_{\max}=5.88$ GPa and oil film thickness parameter of 5 when considering Hertzian contact condition, respectively. The value of Weibull slope in Fig. 16 is estimated to be 2.88, which is greater than the generally accepted values of 10/9 and 9/8 in their respective ball bearings

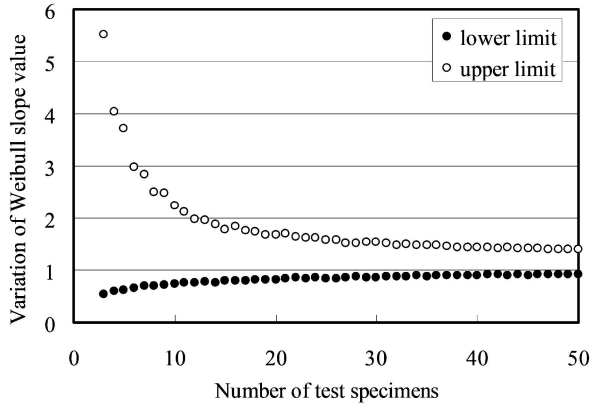


FIG. 17—Relationships between the number of test specimens and variations of Weibull slope values ($e=10/9$).

and roller bearings. Thus, it is obvious that the RCF life test conditions practically affect the value of Weibull slope in RCF life distribution. Accordingly, the value of Weibull slope used in all calculations should be adequately determined for each RCF life testing condition. Figure 17 illustrates a relationship with a reliability of 90 % between the number of test specimens and variations of Weibull slope value. The plots are obtained by 1000 times simulations of Weibull plotting of Weibull random numbers at a preset Weibull slope value of $10/9$. The variations of the Weibull slope value in Fig. 17 are defined at both lower confidence limit of 5 % and upper confidence limit of 95 %. The data plots of both variations clearly show asymptotic relationships to $10/9$. Additionally, the asymptotic behavior for both variations is pronounced until the number of test specimens increases to about ten test specimens. Figure 18 shows the normalized differential values of both the variations of the Weibull slope values depending on the number of test specimens. The normalized differential values demonstrate a decrease to about 0.2 at ten test specimens and reach about 0.1 at 20 test specimens regardless of the Weibull slope values. Therefore, the rational for Weibull slope value used in the calculations is established from the resultant value of the Weibull slope at ten test specimens, and the accuracy of calculation results can be further enhanced with 20 test specimens. If the values of the Weibull slope cannot be obtained from more than ten test specimens, it is rational to employ the estimated values of $10/9$ for ball bearings and $9/8$ for roller bearings in hypothetical testing because the use of the smaller Weibull slope results in a conservative estimation of life and numbers of specimens.

Summary

This paper explains RCF life test design and result interpretation methods while maintaining both efficiency and reliability. The methods provide techniques that are easier to understand than currently accepted mathematical

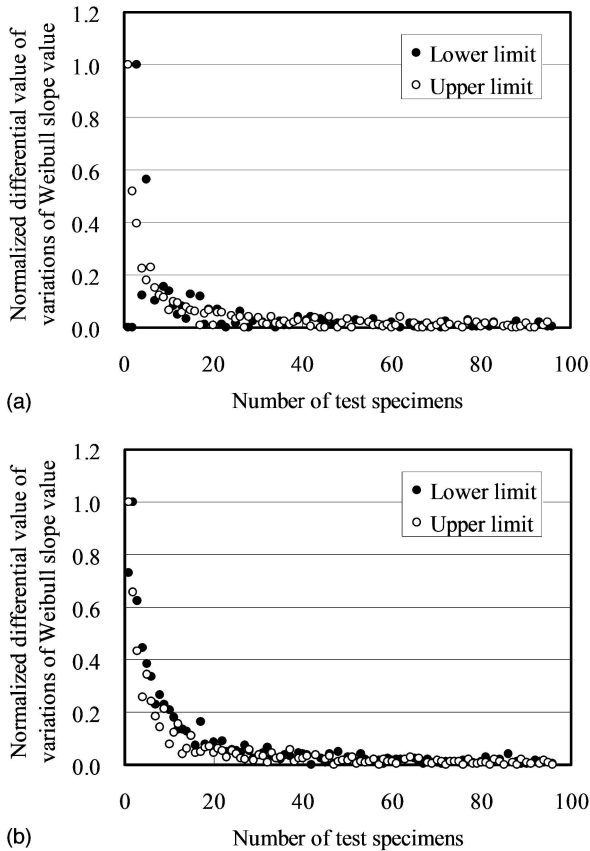


FIG. 18—Normalized differential values of variations of Weibull slope values depending on the number of test specimens: (a) $e=10/9$; (b) $e=3$.

models and allow for application to almost all aspects of fatigue life testing. This system will therefore eliminate the need for qualified experiences as related to the statistical design and result interpretation in RCF life testing.

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**CORROSION RESISTANT STEEL
AND HYDROGEN EFFECTS IN
BEARING STEELS**

R. H. Vegter¹ and J. T. Slycke¹

The Role of Hydrogen on Rolling Contact Fatigue Response of Rolling Element Bearings

ABSTRACT: It is generally accepted that the rolling contact fatigue (RCF) life of bearing components is strongly reduced when the hardened steel matrix contains hydrogen. Although frequently reported in the literature, a physically sound explanation to this phenomenon has not yet been presented. In recent work on building understanding around the nature of the RCF damage process, we have reported on the development of a physically based fatigue damage model. Here we discuss an attempt to interpret the role of hydrogen on the micro-plastic fatigue damage mechanism on the basis of this concept. In this context, the role of hydrogen on the RCF response has been studied in order to find evidence for and to determine its weakening effect on hardened and low-temperature tempered (bearing) steels. In order to perform valid tests, effort was put in controlling the hydrogen content in bearing components before testing using an electrochemical hydrogen charging process. A number of bearing tests were performed clearly showing differences in fatigue response with different amounts of hydrogen present in the micro-structure. The results of testing of bearings with hydrogen-containing steel matrices are given. It was found that increased hydrogen content in bearing steel at 5 ppm (by weight) will significantly promote bearing spalling failure and enhance the formation and growth of so-called white-etching crack systems. This experimental information is interpreted in the light of the physically based fatigue damage model.

KEYWORDS: hydrogen, bearing steel, fatigue life

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Introduction

This paper describes a number of rolling contact fatigue tests of bearings that contain increased contents of hydrogen. The effect of hydrogen on the strength of (bearing) materials has been reported in a many of publications [1,2]. The documented cases where hydrogen plays a role in decreasing the fatigue strength of materials include wide ranges of steel grades, including bearing steels. There are many indications in the literature that hydrogen does not only induce embrittlement of materials leading to delayed brittle fracture under static tensile stress but also enhance local plasticity phenomena [3–10], which precede the failure. This enhanced micro-plastic effect may also play a role in the response to stress also under conditions when the load is not excessively high, such as during exposure to structural or rolling contact fatigue under moderate load [11], leading to strongly reduced fatigue lives. In this contribution we suggest an atomistic description for the role of hydrogen on the high-cycle fatigue mechanism and aim at demonstrating the effect of hydrogen on the rolling contact fatigue damage mechanism in hardened bearing steels, leading to rapid localization of the fatigue damage. This tends to promote early butterfly crack initiation and accelerated growth of fatigue cracks, resulting in early bearing failures. Rolling contact fatigue can manifest itself in many ways, but the emphasis here will be on the initiation and growth of characteristic, so-called “white-etching crack systems,” leading to the development of extensive networks of often branching cracks in the Hertzian shear stress zone under the raceway surface. Similar crack systems are often observed in bearing failures in applications where hydrogen is suspected to be accumulated in the bearing steel matrix. The formation of such crack systems has earlier been described [12], demonstrating a link between the pick-up of atomic hydrogen in the bearing steel and the bearing operating environment. This contribution focuses on observation of the formation of white-etching crack systems in rolling contact fatigue testing under running conditions resembling those of bearings in hydrogen-active applications. In order to investigate the effect of hydrogen in a controlled way, an electrochemical process was used to introduce a given elevated hydrogen content in bearing steel test elements. With this method, bearings containing hydrogen-charged components can be prepared prior to testing, for comparison with standard bearings.

On the basis of the observations made during these tests and in the light of the suggested mechanism for the hydrogen-enhanced fatigue damage mechanism, conclusions on the effect of hydrogen on rolling contact fatigue damage can be drawn.

Theoretical Basis

In a recent work [13] we have reviewed the theoretical basis and experimental evidence for that two micro-plastic damage mechanisms are active during high-cycle rolling contact fatigue in hardened (bearing) steels. In the early stage of the fatigue process, small-scale (classic) obstacle-controlled dislocation glide, or “micro-yielding,” dominates the damage accumulation. The yielding mechanism leads to generation of large quantities of crystal defects like dislocations

and vacancies [14–17] and a range of vacancy-carbon clusters [18,19]. This leads to a gradual work hardening such that the yielding mechanism is retarded and eventually comes to a standstill. This stage of the bearing life cycle is called the “shakedown” stage. In rolling contact fatigue of bearings, this typically occurs before 10^4 – 10^5 inner ring revolutions, corresponding to less than 0.1–1 % of the total bearing life.

After the shakedown stage, the steady-state fatigue damage stage commences. The rate by which the micro-plastic damage accumulates during this stage is several orders of magnitude slower than during the shakedown stage, and it therefore only comes to expression during prolonged exposure to fatigue. The presumption presented in our earlier work [13] is that this slow damage rate is controlled by climb-controlled dislocation glide, that is, by a cyclic creep-like damage mechanism. This damage mechanism entails that a gliding dislocation, which has become trapped by an obstacle, only can continue its glide motion after that it is freed from the obstacle by stepwise climb of the dislocation segment around the obstacle. This climbing process requires local rearrangement of the matrix atoms around the dislocation core, which means that the climb process will be rate-controlled by iron self-diffusion [20]. Self-diffusion in iron and steel is normally not possible at low temperatures, that is, at around room temperature and at typical bearing operating temperatures, below approximately 150°C. The reason to this is that diffusion, that is, a successful atom jump from one lattice position to the next, requires two sequential atomistic events: The formation of an empty lattice position (that is, a vacancy) and the subsequent jump of the vacancy to a neighboring iron atom lattice position (or, in other words, that the iron atom concerned moves in the opposite direction, into the earlier vacant lattice position), the well known atom-vacancy exchange mechanism [21]. The thermal energy available at low temperatures is not high enough to realize both these steps [22], thereby preventing iron self-diffusion.

Hardened steels, with their non-equilibrium microstructure, already in their as-heat treated state contain high presence of various crystal defects. The numbers of dislocations and vacancies are further strongly increased during plastic deformation during the shakedown stage and the subsequent fatigue damage process. The fatigue-exposed bearing steel matrix therefore already contains high concentrations of vacancies and other point defect clusters. This means that for self-diffusion to occur; it is only necessary to thermal activate the migration step of a vacancy to an iron atom lattice position. The thermal energy available at low temperature is sufficient for this to occur, but still, the number of successful jumps remains low at low temperatures [22,23]. This means that the overall dislocation climb mechanism and the fatigue damage mechanism will be quite slow under these conditions, corresponding to the good fatigue properties we are used to see in hardened bearing steels.

One critical aspect is here the availability of mobile crystal point defects. When vacancies are generated during dislocation glide or in interaction between moving dislocations [14–17], they tend to cluster in different configurations, not the least when interstitially dissolved elements like carbon [18,19] and hydrogen [24–28] are present. In hardened steel matrices most vacancies will be bound to carbon atoms, making the resulting carbon-containing point

defect clusters practically immobile [22]. Only a limited fraction of the total number of vacancies remains as free single or double vacancies that are mobile at low temperature [22,23]. The reason is that the carbon-containing clusters become immobile is that if the vacancy tends to jump, it has to drag along its accompanying carbon atom(s) or to dissociate from its companion(s) to successfully complete its jump. This would require additional thermal energy, which is not available at low temperatures. The presence of carbon therefore tends to reduce the number of mobile point defects and therefore to enhance the fatigue resistance of (bearing) steels. This is the likely atomistic explanation to that the composition of bearing steels over the years has evolved towards their current high carbon contents (around 1 %).

When atomic hydrogen is present in iron, it will, just like carbon, interact with the iron lattice, tending to strongly increase the equilibrium concentration of vacancies, more so the lower the temperature is [25,26]. In carbon-containing steels, hydrogen will tend to further increase the total concentration of vacancies. Further, the point defect spectrum will broaden to also include vacancy-hydrogen, vacancy-carbon-hydrogen clusters, etc. This means that with increasing contents of atomic hydrogen, the number of mobile point defects will increase. While vacancy-carbon pairs are essentially immobile [22] at low temperatures, vacancy-hydrogen pairs are, due to the small size and low weight of the hydrogen atom, likely to retain some degree of mobility and may therefore contribute to the overall self-diffusion. In conclusion, the presence of atomic hydrogen will tend to strongly promote the iron self-diffusion process and thereby to acceleration of the fatigue damage process.

In another contribution [29], a detailed description of the formation of so-called butterfly features at oxide inclusions in hardened bearing steels will be presented. In brief, the mechanism involves accumulation of highly localized damage in the steel matrix in contact with the debonded oxide inclusion. The damage is induced by nano-scale rubbing action across the inclusion/matrix interface, leading to a sequential (diffusion) bonding and debonding cycle during the fatigue stress cycle. During the rubbing action between the debonded inclusion and the steel matrix, material is transferred from the steel matrix and deposited on the inclusion surface. With continued rubbing damage and material transfer continues, almost by an atom-by-atom mechanism, across the evolving crack-like gap between the unchanged matrix and the deposited material at the inclusion surface. This constitutes the early development of the butterfly feature with its (in the optical microscope) white appearing "wing" (the deposited material) and its associated crack (separating the wing from the unchanged steel matrix). This sequence clearly demonstrates the cooperative nature of the growth of the butterfly wing and its crack, as it is also observed in the white-etching crack systems that will be shown in this paper. The deposited white butterfly wing areas undergo (dynamic) recrystallization during their formation, resulting in nanometre sized and essentially equiaxed ferrite grains, with grain sizes in the range of 10–30 nm. This structure is identical to the microstructure as formed along the white-etching cracks. The fact that recrystallization occurs is a strong direct indication that iron self-diffusion is active during the butterfly formation process [29]. The butterfly cracks do not grow in an irreversible manner like classic cracks but migrate instead sideways

into the steel matrix during their development through material transfer across the crack faces. This leads to the deposition of white-etching material on the receiving side of the (butterfly) crack. Butterflies (rolling contact fatigue), just like fish-eye cracks (structural fatigue), only develop in hardened and low-temperature tempered steels and grow (if at all formed, depending on loading conditions, steel cleanliness, etc.) only quite slowly during more than 99 % of the total fatigue life [30], preceding the final stable crack growth to unstable spalling failure.

In the presence of atomic hydrogen in the steel matrix, the butterfly formation process is strongly accelerated. This leads to that butterflies form at lower bearing load levels than otherwise needed to generate these damage features. Although the hydrogen diffusivity is considerably lower in the hardened microstructure than in soft iron or steel, the diffusivity in hardened bearing steel is still sufficient to allow hydrogen transport over millimetres or even centimetres during the use of a bearing (weeks, days, or years) at typical operating conditions. This readily allows diffusion of hydrogen, for example, from the raceway surface to the fatigue-exposed Hertzian subsurface zone. This also means that even if the average hydrogen content in the steel matrix is low (ppm range), it can locally be considerably higher by redistribution within the matrix. Since hydrogen strongly binds to various crystal defects [5–8,25–28], atomic hydrogen will be attracted to areas of local plastic deformation where the densities of crystal defects (dislocations and vacancies) are high. As mentioned, hydrogen itself tends to further promote the formation of yet more vacancies, which again stimulates the dislocation climb-controlled damage process. This leads to a self-generating damage localisation process. This is particularly apparent in the steel matrix adjacent to oxide inclusions, which is exposed to the above discussed rubbing action between debonded inclusion and the steel matrix. This is the likely mechanism that governs the hydrogen-enhanced butterfly formation and subsequent further cooperative growth of the butterfly cracks and their white-etching features. Eventually these cracks develop into longer and often branching (white-etching) crack systems, leading to early bearing failures in hydrogen-active operating environments. The white-etching decoration of butterfly cracks and longer cracks in presence of hydrogen is the result of rubbing actions across the crack faces. This leads to transfer of material from one side of the crack and deposition of this material (at atomistic rates) on the receiving side of the crack during course of the continued fatigue exposure. This leads to that the crack “meanders” back and forth through the microstructure, a behavior that is highly characteristic for crack growth in hydrogen-containing bearing components. This is signalled by the presence of more or less white-etching decoration along the fatigue cracks and by frequent branching of the cracks.

Experimental

Bearing Type

The bearing used in this investigation was a Deep Groove Ball Bearing (DGBB) of type 6309 with a plastic snap-cage. The bearing was made of ASTM 52100

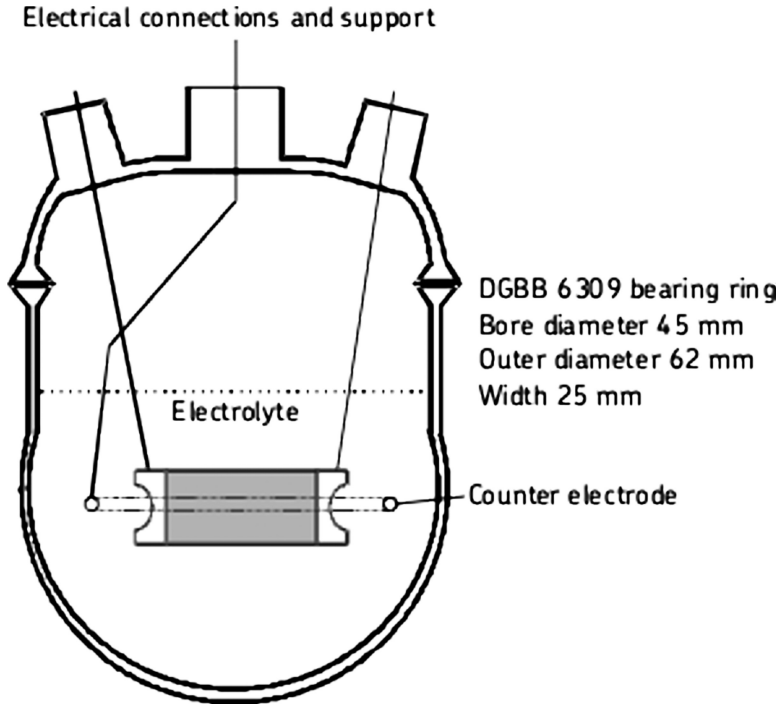


FIG. 1—Schematic set-up for hydrogen charging of inner rings of a DGBB.

[31] bearing steel, which was martensitically hardened and tempered (165°C) to obtain a hardness of 63 ± 1 Hardness Rockwell C. The bearing inner ring was used as the test element.

Hydrogen Charging

The bearings were charged with hydrogen using an electrochemical process. The set-up used in the charging process consists of a glass container filled with electrolyte. A schematic diagram is shown in Fig. 1. The work piece, in this case the DGBB inner ring, and a counter-electrode were positioned in the electrolyte and connected to a current source. For this charging experiment, a specially designed counter-electrode was used to conform to the shape of the DGBB inner ring sample. The electrical current between the work piece and the counter-electrode was supplied by a Keithley 2400 stabilized current source. This unit maintains a stable current even if the impedance of the charging cell changes, for example, due to the changing chemical composition of the electrolyte. The work piece is cathodically charged with hydrogen using a current density of approximately 10 mA/cm^2 . The electrolyte used for hydrogen charging was a dilute alkaline solution based on NaOH and with additions of thio-urea.

Before the inner ring test elements were charged for testing, a series of

TABLE 1—*Hydrogen content distribution in a hydrogen-charged DGBB inner ring.*

Sample	H-Content Surface, ppm	H-Content Core, ppm
1	4.58	3.20
2	5.96	5.66
3	4.30	3.21
4	4.72	4.80
5	4.66	5.17
6	4.64	4.95
7	4.75	4.95
Average content and standard deviation	4.80 (0.53)	4.56 (0.97)

charging pre-trials was carried out on identical rings to determine the distribution of hydrogen around the circumference of the ring and in the centre or the ring wall after charging at the selected conditions, in this case made at a temperature of 80°C with a charging time of 24 h. The results from these pre-trials are given in Table 1. On the basis of these results, it was decided to charge the bearing inner ring test elements using these conditions. With reference to the average hydrogen contents in the raceway-near material in the rings in the pre-trial, we can conclude that the hydrogen content in the surface zone of the inner ring at the start of the test was 4.8 ± 0.5 ppm. Information on the exact development of the hydrogen content during the test is not available.

Equipment

The microstructures of the tested bearing inner rings were investigated after metallographic sectioning, polishing, and chemical etching (Nital etchant) using standard optical microscopy and scanning electron microscopy (SEM). The total hydrogen content was measured using the ELTRA ONH-2000 hydrogen extraction analyser. Calibration of this instrument was made using standard samples from Alpha Resources, having a hydrogen content of 5.8 ± 0.3 ppm. The hydrogen desorption rate as function of temperature was measured using a temperature-programmed induction furnace equipped with an Ametek Proline mass-spectrometer for hydrogen detection.

Testing

The assembled test bearings were mounted on a double-sided standard (R2) test rig and exposed to the selected test conditions during pre-determined running times. In each test run one standard bearing (without hydrogen-charged inner ring) and one bearing with a hydrogen-charged inner ring were run in parallel. The running conditions used are given in Table 2. The temperature of the outer rings was measured on the outside of the ring. The actual inner ring temperature was not measured. It is expected that this is higher. Therefore, the given temperature is an indication.

TABLE 2—*Testing conditions for the bearing tests.*

Applied load	25 kN radial load
Contact pressure	3.2 GPa
Test temperature outer ring	83°C outer ring
Speed	6000 rpm
Lubricant	Shell TT100
Viscosity at operating temperature 88°C (calculated)	14.8
Kappa	2.42

Results

Hydrogen Desorption Analysis

A hydrogen desorption analysis was carried out to verify that the hydrogen dissolved in hardened bearing steel remains essentially mobile at low temperatures, so-called “diffusible” hydrogen. In this case, a sample of a bainite hardened ASTM 52100 bearing steel was charged with hydrogen using the same conditions as mentioned above. The sample was put into a temperature-programmed vacuum induction furnace, and during the linear heating process, the rate of hydrogen desorption was recorded using a mass-spectrometer. This analysis was repeated for a hydrogen standard sample with known hydrogen content (5.8 ppm). The applied heating rate was 25°C/min, and the results are shown in Fig. 2. It can be seen here that the hydrogen desorbs from the bearing steel sample already at temperatures lower than 200°C, while temperatures of around 400°C are required for the hydrogen standard sample. The analyses

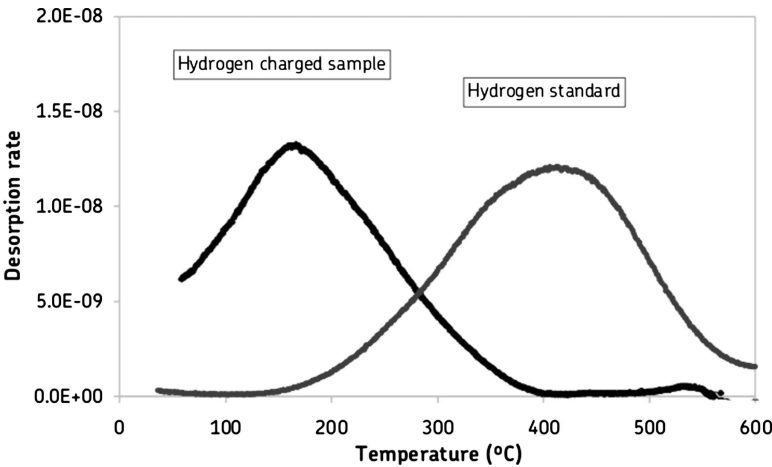


FIG. 2—*Hydrogen desorption as function of temperature. Experiments were carried out with cylindrical sample of 4 mm diameter and 7 mm length.*

TABLE 3—Test set-up and results of the hydrogen-charged (a) and standard bearings (b).

Test Number	Planned Number of Revolutions	Actual Number of Revolutions (Suspended/Failure)
1a	2.6×10^4	2.6×10^4 (suspended)
1b	2.6×10^4	2.6×10^4 (suspended)
2a	2.6×10^5	2.6×10^5 (suspended)
2b	2.6×10^5	2.6×10^5 (suspended)
3a	2.6×10^6	1.9×10^6 (failure)
3b	2.6×10^6	2.6×10^6 (suspended)
4a	2.6×10^7	4.0×10^6 (failure)
4b	2.6×10^7	2.6×10^7 (suspended)
5a	2.6×10^8	1.5×10^6 (failure)
5b	2.6×10^8	1.6×10^7 (failure)

therefore clearly show that hydrogen in the charged bearing steel sample is only loosely bound to weak hydrogen traps, such as dislocations, crystal point defects [5–9], etc., while hydrogen in the standard sample is, as expected, much less mobile and likely bound to strong hydrogen traps or even as hydrides. The total hydrogen contents are represented in Fig. 2 by the area under the curves. It can be visually seen that the areas for the two samples are similar; verifying that the total hydrogen content in the charged bearing steel sample is around 5 ppm and that most of this hydrogen remains mobile and thereby available for interaction with the fatigue damage mechanism.

Rolling Contact Fatigue Tests

The chosen test strategy was to expose the bearings to the test conditions for a range of pre-determined numbers of revolutions (or running times). However, it was found that it was not possible in all cases to reach the pre-determined exposures due to early failures of the test bearings. In total ten bearings were tested in five parallel tests. In each test run, a standard bearing was mounted on one position of the double-sided test rig, and a bearing with a hydrogen-charged inner ring was mounted on the other position of the test rig. An overview of the test results is given in Table 3.

The results from the test series show that three of the bearings with hydrogen-charged inner rings did not reach the pre-determined number of revolutions. This also happened in the case of one standard bearing, test number 5b (cf. Table 3), which failed before the required number of revolutions was reached. In all cases, the failures occurred in the inner ring and had a subsurface initiation character.

The microstructural features developing in the Hertzian zone of the inner rings were investigated and selected observations will be discussed here. Figure 3 shows a cross-section of the inner ring of bearing no. 3a (cf. Table 3). This bearing was charged with hydrogen and failed after 5.3 h testing. The micrograph from a cross-section of this inner ring shows a significant development of a white-etching crack system, several millimetres in length and is, in this sec-

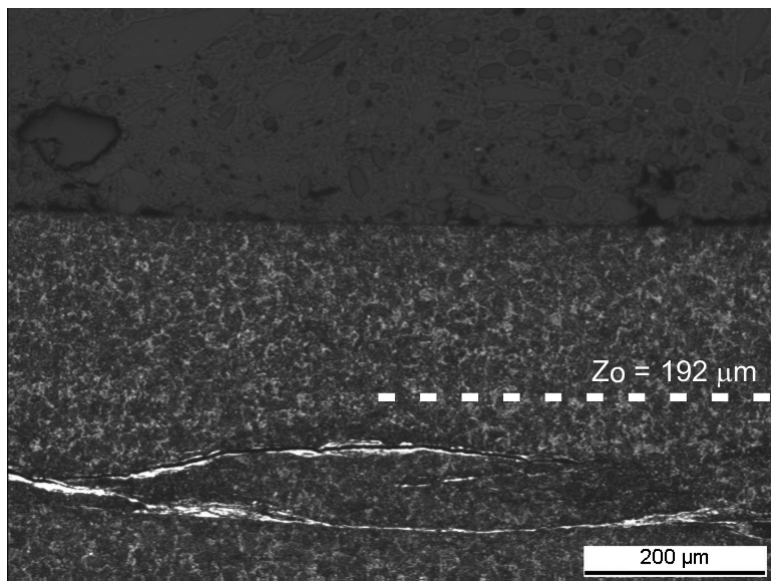


FIG. 3—Microstructure of the inner ring of bearing 3a, hydrogen charged, tested for 1.9×10^6 revolutions before failure (equivalent to 5.3 h), cross-section.

tion, oriented in parallel to the raceway surface. Similar observations were made on metallographic sections that were taken from the hydrogen-charged inner ring of bearing no. 4a (cf. Table 3), which had been running for 11.1 h. In this case, a parallel section was made through the raceway surface. The micrograph is displayed in Fig. 4 and shows that white-etching crack systems are present in the microstructure at more or less regular intervals below the raceway surface. They all have a typical morphology, with the upper part of the crack being inclined to the surface and the lower part being oriented in the radial direction. The microstructure of the hydrogen-charged inner ring from bearing 4a (cf. Table 3) is also shown in Fig. 5, where a cross-section is shown. This micrograph reveals several, more or less parallel, white-etching cracks below the raceway surface at depths corresponding to the crack systems shown in Fig. 4.

Discussion

In this test program, ten bearings, five of which had hydrogen-charged inner rings, were tested for a range of pre-determined numbers of revolutions. It was observed that several failures occurred during the test before the pre-determined numbers of revolutions were reached. Three of these failures occurred in hydrogen-charged inner rings and one failure occurred in the inner ring of a standard bearing. Investigations of the microstructure of the bearing inner rings show that subsurface cracks, often decorated by white-etching microstructural features, develop early in the hydrogen-charged inner rings, while

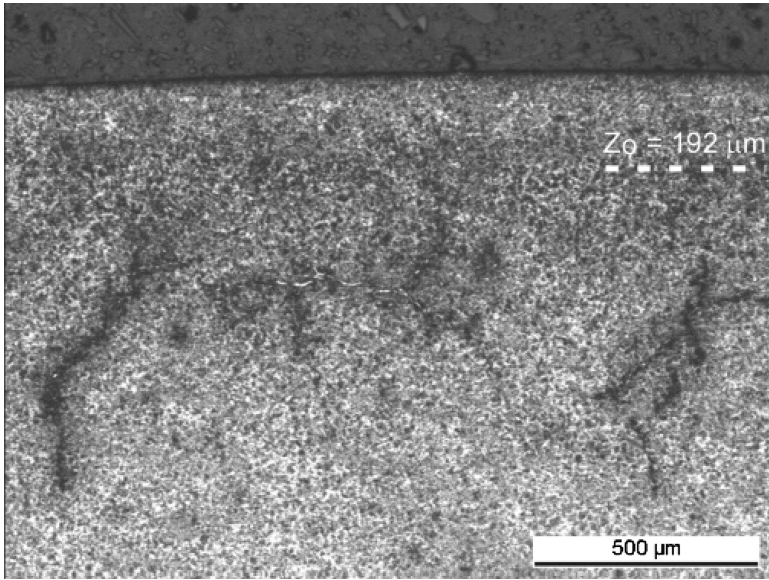


FIG. 4—Microstructure of the inner ring of bearing 4a, hydrogen charged, tested for 4.0×10^6 revolutions before failure (equivalent to 11.1 h). Inner ring circumferential section taken 180° opposite to the spalling failure.

these features are absent in the standard inner rings. All hydrogen-charged inner rings exhibit more or less well developed systems of white-etching cracks. The most advanced states of such crack systems were observed in bearing 3a and 4a (cf. Table 3), which were run for 5.3 and 11.1 h, respectively, and which failed prematurely. In both these cases, the cross-section of the inner rings (Figs. 3 and 5) show that the white-etching crack system, extending over several millimetres, have formed and grown in parallel to the raceway surface in the axial direction. The white-etching crack systems do not reach contact with the raceway surface until the point of spalling failure. From our observations, we conclude that the white-etching crack systems originate at depths below the raceway surface approximately corresponding to the z_0 -depth, that is, the depth where the alternating shear stress reaches its highest value. For the DGBB 6309 bearing geometry and for the applied loading conditions, the z_0 -depth is $192 \mu\text{m}$. Although not directly observed in this study, it can be assumed that (a part of) the white-etching cracks initiate early in the fatigue process at (small) non-metallic inclusions located around the z_0 -depth and initially develop as butterfly cracks [29]. While butterflies in standard bearings only grow quite slowly and remain as (almost) non-propagating cracks through the overwhelming part of the bearing life, the presence of hydrogen strongly promotes the further growth of the butterfly cracks. The presence of hydrogen will accelerate the crack growth; it will itself not change the appearance of the newly formed microstructure. During the development of the crack systems, they grow towards deeper depths below the raceway surface. The cracks can grow to con-

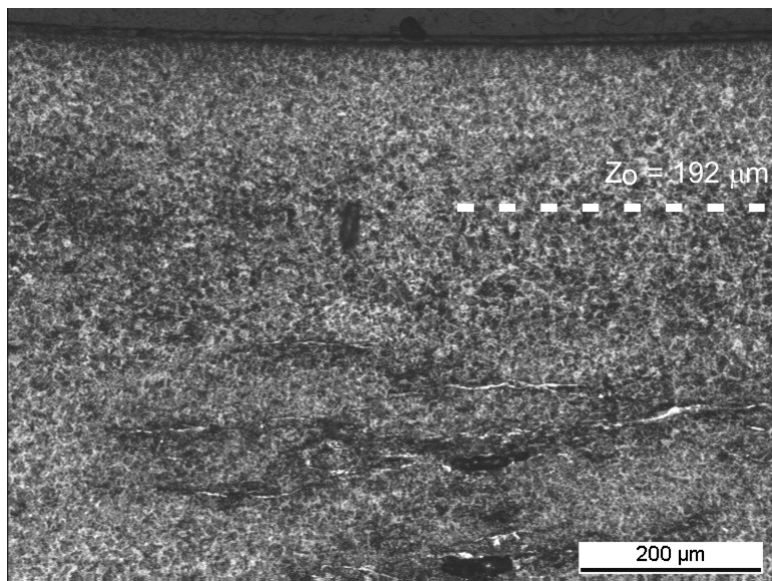


FIG. 5—Microstructure of the inner ring of bearing 4a, hydrogen charged, tested for 4.0×10^6 revolutions before failure (equivalent to 11.1 h). Inner ring cross-section taken 180° opposite to the spalling failure.

siderable sizes before they eventually find their ways to the raceway surface and trigger spalling failure. The location of the white-etching zone can also be seen in Fig. 5, which also shows that some degree of general microstructural change has developed in the overall Hertzian zone, the so-called dark-etching region (DER). The DER overlaps in part with the white-etching crack system.

For comparison with the hydrogen-charged rings, an inner ring that was not charged and that reached the pre-determined number of revolutions (2.6×10^6), bearing 3b (cf. Table 3), was also investigated. The micrograph in Fig. 6 shows that the microstructure does not exhibit any cracking, while a tendency to DER formation also can be seen in this inner ring. This observation emphasizes the quite strong effect of hydrogen in enhancing the crack initiation and growth processes, mechanisms that are orders of magnitude slower in absence of hydrogen.

Further investigations using SEM was performed on bearing inner ring 2a (cf. Table 3), which was hydrogen charged and had been running for 260 000 revolutions (43 min). SEM analysis was employed to this specimen since optical microscopy indicated that cracks were present in this inner ring, but that they were too small to allow proper documentation. It was observed in the SEM that very small cracks occur at the z_0 -depth and deeper. The detailed images of the cracks do not show formation of larger white-etching areas, only a few small patches scattered along the crack faces could be identified. This can be seen in the SEM micrograph in Fig. 7, showing an approximately $100 \mu\text{m}$ long crack with several instances of branching. This demonstrates that these are

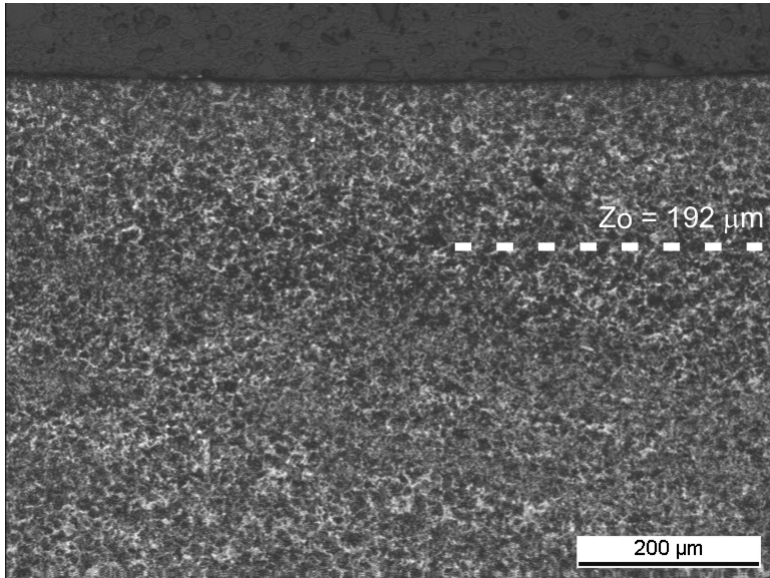


FIG. 6—Microstructure of the inner ring of bearing 3b, not hydrogen charged, tested for 2.6×10^6 revolutions (equivalent to 7.2 h). The inner ring cross-section was taken at random position. No failure was observed in the bearing.

cracks present already after short time of running and constitute the early stages of the large networks of white-etching cracks that are observed in bearings that have been running during longer times. The inner ring 5a has been running during 1.5×10^6 revolutions (equivalent to 4.2 h), that is, it has only obtained 5.8 times longer fatigue exposure than inner ring 2a. Still, the inner ring 5a exhibits a much more progressed white-etching crack system. This illustrates how rapidly the white-etching crack systems develop when hydrogen is present in the microstructure. It can therefore be concluded that the formation and growth of the white-etching crack system is very fast in presence of high hydrogen contents. This is further emphasized by the fact that the hydrogen-charged inner rings 3a and 4a (5.2 and 11.1 h running times) have an even more progressed state of the white-etching crack system, as compared to specimen 5a.

As hydrogen tends to remain mobile during the bearing operating conditions [11], it may also effuse out from the hydrogen-charged test element during prolonged running times. This makes it potentially more difficult to draw conclusions related to the progress of the crack systems. The measurement data as presented in Table 1 indicates a relatively homogeneous distribution of the hydrogen through the ring section and from ring to ring, representing the testing start conditions. However, the effect of changing hydrogen content due to the test conditions (temperature and microstructure changes) remains an open question. As indicated, hydrogen effusion may occur, but this process can potentially also be retarded by the generation of more crystal defects in the

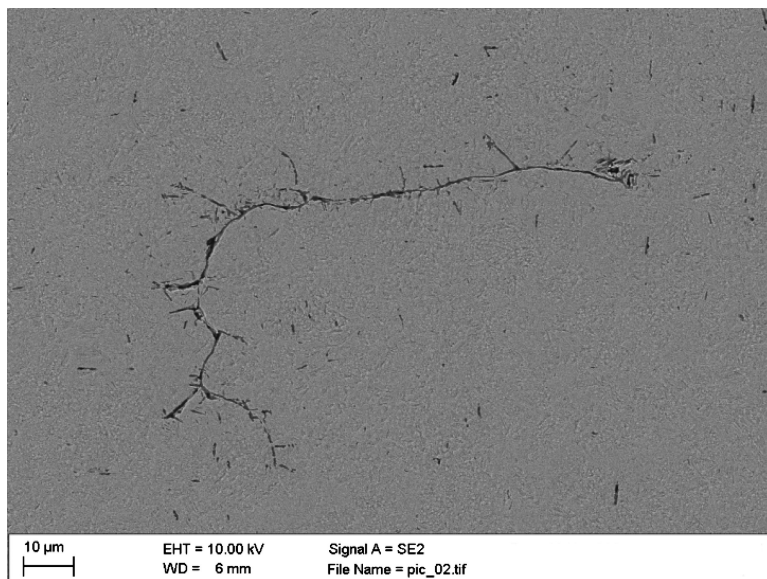


FIG. 7—Microstructure of the inner ring of bearing 2a, hydrogen charged, tested for 260 000 revolutions (equivalent to 43 min). Circumferential section, image taken with the SEM, approximately 200 μm below the raceway surface.

fatigue damage process zone, as the fatigue damage accumulation progresses, which tends to trap the hydrogen. So far, no systematic relation between the initial hydrogen content and the content after testing of the bearing has been established.

Although a quantitative relation between white-etching crack system development and the number of revolutions is difficult to establish at this stage, the qualitative relation is quite clear. It is obvious from our observations that the white-etching cracks gradually develop with increasing exposure to fatigue. The detailed growth process and its relation to the (varying) hydrogen contents should be further investigated.

During the test, five standard bearings were tested as a reference. Although one failure was observed in this set, there was no indication that a white-etching crack system was formed. The microstructure development showed the well known formation of DER in the Hertzian subsurface zone. It should be noted that the life of the failed standard bearing exceeded the calculated L_{10} life for this bearing type and for the running conditions used ($N_f = 16$ Mrevs $> L_{10} = 11$ Mrevs). It is clear that the crack that developed in this inner ring has a quite different appearance than the white-etching crack systems observed in presence of hydrogen and can therefore be characterized as a traditional subsurface initiated fatigue crack. In this project, no residual stresses were measured, but we assume that such measurements would show a gradual build-up of compressive residual stresses in the subsurface microstructure, in accordance with the findings of Voskamp [32]. The fact that no white-etching crack

systems were formed in the standard bearing inner rings strongly supports the hypothesis that the presence of hydrogen strongly accelerates the fatigue damage mechanism that is responsible for the initiation and growth of fatigue cracks under rolling contact fatigue conditions. It can therefore be concluded that the presence of hydrogen has a decisive effect on the occurrence of white-etching crack systems and the related early failures.

Conclusions

On basis of the results obtained in the executed bearing tests, it can be concluded that the presence of hydrogen in bearings steel has a quite strong effect on fatigue life. These findings, which are supported by many similar observations in the literature, is directly compatible with the here forwarded theoretical interpretation on the indirect role of hydrogen in promoting the number and the mobility of crystal point defects. This enhances the different stages of the crack initiation and crack growth processes, which presumably are governed by iron self-diffusion controlled mechanisms. When hydrogen is present in bearing steel, large systems of white-etching crack will be formed during bearing operation. This is not observed in bearings that were not charged with hydrogen and tested under identical conditions. Again, the role of hydrogen is here to indirectly promote the crack growth by stimulating the self-diffusion controlled rubbing damage across the crack faces, leading to transfer of material from one side of the crack to the other. This leads to a meandering crack, which accumulates white-etching microstructure on the receiving side of the crack. Furthermore it is observed that the white-etching crack systems formed during the tests, just like butterflies at non-metallic inclusions, preferentially originate at positions around or deeper than the z_0 -depth (192 μm). In addition to that, the white-etching crack systems formed under the testing conditions applied never reach the surface of the bearing until the point of failure.

Acknowledgments

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MICRO CLEANLINESS QUALITY ASSURANCE IN BEARING STEELS

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Quality Function Deployment Application on the Development of 100Cr6 Bearing Tubes

ABSTRACT: In 2001, a Brazilian bearing seamless tubes producer started a 5-year-long research and development (R&D) program to upgrade the global quality of its 100Cr6 bearing tubes, nowadays becoming in line with the requirements of first class producers. This work describes the quality function deployment methodology applied to understand the customer's needs and how the acquired knowledge was used to promote changes and developments in the production process, to improve quality control procedures on the key processes, and to prioritize investments along the whole manufacturing line (steel plant, tube rolling mill, and cold drawn plant). Present customer audits and quality reports demonstrate the great success of this R&D program, increasing customer satisfaction and enabling the company to sustain and enter new markets.

KEYWORDS: QFD, 100Cr6 steel grade, bearing tubes, quality improvement

Introduction

In 2001 the Brazilian bearing steel market demonstrated a high demand for quality improvement: the Brazilian bearing mills were gradually changing their specifications to get into agreement with their mother companies and to become global players, exporting part of their production to worldwide destinations. There was a lot of different requests at the same time. While one customer was asking for better microcleanliness figures, another one was requiring higher sulfur content in the steels to improve their productivity. A

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third customer signed with the interest to buy products with lower oxygen content in the steel, and another bearing company, which was becoming a new producer in Brazil, looked for a local supplier to be industrially qualified as its global supplier.

Under these circumstances, a Brazilian bearing seamless tubes producer decided to launch a comprehensive 5-year-long research and development (R&D) program in 2001 to upgrade the global quality of 100Cr6 bearing tubes in order to be able to become, in a short period of time, in line with the requirements of first class bearing steel producers.

During the planning and execution of this R&D program, the quality function deployment (QFD) methodology [1–3] was applied, which allowed the whole development team to understand well the customer's needs and prioritize them, to use the acquired knowledge obtained through the trials and experiments, to promote changes in the production process, to improve quality control procedures of the key process, and to establish the investment precedence along the whole manufacturing line (steel plant, tube rolling mill, and cold drawn plant).

QFD Methodology

Identification of Customer Needs and Definition of Quality Planning Matrix

During the first 3 months after the beginning of the project, a series of meetings between the customer's quality, production, and purchase managers and their correspondent managers from the producer was planned and executed.

On such work meetings, the managers discussed together carefully and very open-mindedly all quality and logistic aspects related to the product, enabling the producer to understand deeply the main necessities of each customer, to deploy them into two levels of specification, defining explicitly the correlated quality characteristics, to well define the customer targets to those necessities and to evaluate the producer performance against its best competitor. At the end, obtaining such information, it was possible to establish the producer's quality planning matrix for each customer (Fig. 1 exemplifies part of such matrix for customers 1 and 2) and to define the relative priority between all customers requirements.

Although the product itself, which was delivered to the market for many years, was essentially the same for all customers, it could be noted at the end of this learning process that besides some common requirements, many of the main needs of each customer were different from each other.

For example, while for customer 1 the most important point was related to increasing the steel machinability, for customer 2, the focus was to decrease the decarburization tolerances. Customer 3, on the other hand, asked for lower residual stresses and lower variability of mechanical properties, and customer 4 required very clean steel, with low level of micro- and macroinclusions.

Quality Planning Matrix			Necessity		[1]	[2]	[3]	[4]	[5] = [4] / [2]	[6]	[7] = [1][4] / [6]	[8] = [7] / Sum [7] * 100	Priorities	
			(1st level)	(2nd level)	Importance Level for Customer	Producer Performance	Best Competitor Performance	Planned Quality	Increasing Index	Sales Potential	Absolute weights	Relative Weights		
Planning Problem Identification Project or modify the product	CUSTOMER #1	reduction of overmetal	dimensional changes for tubes with Outside Diameter < 50 mm	5	2	4	4	2.0	1.0	10.0	11.3	4		
		lower frequency to replace pieces on turning machine	adequate length for the customer's turning machines	5	2	5	5	2.5	1.0	12.5	14.1	2		
				5	2	5	5	2.5	1.0	12.5	14.1	2		
		Steel with lower oxygen content	reduction of oxygen level in the steel	4	3	5	5	1.7	1.5	10.0	11.3	4		
				4	3	5	5	1.7	1.5	10.0	11.3	4		
		Increase machinability and surface quality	adjustment of tube hardness to optimize customer's machining operation	5	2	5	5	2.5	1.5	18.8	21.1	1		
			lower decarburization tolerances	5	2	4	4	2.0	1.2	12.0	13.5	3		
			reduction of residual stresses	3	3	3	3	1.0	1.0	3.0	3.4	5		
		TOTAL									88.8	100.0		
		CUSTOMER #2	Increase productivity	higher sulfur content in steel	5	2	4	4	2.0	1.5	15.0	14.4	3	
				3	5	5	5	1.0	1.0	3.0	2.9	8		
	lower decarburization tolerances			5	1	4	5	5.0	1.5	37.5	36.1	1		
	reduction of claims due to decarburization			5	2	5	5	2.5	1.5	18.8	18.0	2		
	reduction of residual stresses			3	3	3	3	1.0	1.0	3.0	2.9	8		
	adequate length for the customer's turning machines		5	2	5	5	2.5	1.0	12.5	12.0	4			
	Better intrinsic material quality		macrocleanliness control for each heat using US high frequency test	4	3	5	4	1.3	1.0	5.3	5.1	5		
			microcleanliness control	3	3	4	4	1.3	1.0	4.0	3.9	7		
		microstructure and segregation control	3	3	4	4	1.3	1.2	4.8	4.6	6			
	TOTAL									103.9	100.0			
					Importance Level for Customer		extremely important	5						
							very important	4						
							important	3						
							less important	2						
						not important	1							
				Producer Performance and Competitors		excellent	5							
						very good	4							
						good	3							
						bad	2							
						very bad	1							
						Sales Potential, taking into consideration the company performance and marketing strategy to emphasize any quality product characteristic		strong	1.5					
								middle	1.2					
								low	1.0					

FIG. 1—Part of quality planning table related to two customers.

The producer concluded that the strategy of market positioning and the different manufacturing processes of each customer at that time had a big influence on their product quality requirements, being necessary to act at different fields of development and quality improvement simultaneously in order to be able to satisfy the whole market and have a first class product at the end of the development process.

Quality Matrix						correlation between customer needs and product quality characteristics									
Step	Correlation Level		5 Strong correlation	3 There is some correlation	1 It is possible to have some correlation	NECESSITIES OF CUSTOMER #1									
	Who	What													
Steel Plant	Quality Manager of Steel Plant	Internal quality	Metallographic lab	Intermediate product	Type	Priorities									
						PRODUCT QUALITY CHARACTERISTICS					CORRELATION LEVEL				
Rolling Mill	Quality Manager of Rolling Mill	Structural quality	Metallographic lab	Intermediate product	Type										
Cold Drawn	Quality Manager of Cold Drawn	Structural quality	Mechanical and metallographic lab	Interm. prod.	final product										
Production Manager of Cold Drawn Plant	Physical defects	Dimensional aspects	NDT, visual and dimensional inspections	final product	Type										
						</									

FIG. 2—Part of 100Cr6 quality matrix related to customer 1.

New Product Concept—Quality Matrix

Taking into consideration the different correlation levels between the customers' needs, projected on producer's quality planning matrix (Fig. 1) with all intermediate and final product quality characteristics (chemical composition, micro- and macrocleanliness, decarburization, carbides (size, network, and banding), microstructure, grain sizes, mechanical properties, dimensional aspects and tolerances, superficial quality, etc.), it was possible to identify which product quality characteristics are the most representative ones concerning each customer's request (Fig. 2 exemplifies part of such matrix related to the case of customer 1).

New Product and Process Projects

As soon as the product quality characteristics, which must be improved or maintained, and their target values were defined, it was necessary to confirm the technical viability of the process to achieve such targets, planning in detail all new production and controlling steps.

The first step consisted in registering all previous operational standards and the statistical process capability to produce the bearing tubes and to correlate them to the quality characteristics of the intermediate and final products. This information was used afterwards as a comparison base to evaluate the performance efficiency of all process and control changes proposed along the development phase.

The changes along the process flow were previously discussed and agreed to between R&D and production engineers, registered on formal experiment plans, and distributed to all people involved in the production and quality inspection of the experiment batches. In some cases, the Tagushi design of experiment (DOE) methodology was used to identify all different alternatives to be tested, simplifying and reducing the total amount of experiments. In other cases, it was necessary to conduct cross-check experiments between different production mills to better understand the relative influence of the raw material on the surface quality of the intermediate and final products and the correspondent production yield.

Tables 1–3 show some examples of experiments, controlling actions, and investments realized along the whole producer process (steel plant, hot rolling mill, and cold drawn plant) between 2002 and 2005.

Many other classical quality planning and development tools were also used during the experiments to help on the evaluation of global or local process performance, for example SPC, control plan, and failure mode and effect analysis.

Using such tools during the planning and execution of the experiments, the producer's R&D team was not only able to analyze the influence of process changes on the product quality characteristics but also to check and establish, step by step, a new operational standard, enabling the production of bearing steel tubes with an improved process capability and low failure risks.

Nowadays such documents are part of the current material quality documentation and are used as reference to any other improvement plan.

Incremental Quality Improvement

Simultaneous with the trials and developments, the product quality was verified to be in accordance with customer requirements, and new samples were provided to them to be checked and homologated. This development method enables an incremental quality improvement in the production process without a large increase in stock and material losses.

Figures 3–6 show some examples of the producer's quality evolution during the development phase. Many quality characteristics were checked simultaneously for each production lot at different process steps, and all of these val-

TABLE 1—Experiments and analyses realized at a steel plant of a Brazilian bearing tubes producer between 2002 and 2005.

2002	2003	2004	2005
Global check of operation capacity and efficiency	SEM investigation on the nature and the chemical composition of inclusions	Use of a special ladle shroud during ladle changing	Special device to avoid turbulence into the tundish
Comparative analysis of O ₂ , Al, and Ti content measurements between producer's laboratory and external ones	present in the standard hollows and their relationship to the position inside the casting sequence	Introduction of new practice of deoxidization	Standardization of flux powder addition
Macrocleanliness DOE		Use of special intakes	Development of new synthetic slag to tundish
Definition of a standard hollow enable comparison of the experiments results		Application of an adequate slag for refining	Study of microinclusion profile along the transversal section of a cast bar
Evaluation of the influence of continuous casting parameters on macrosegregation and on the generation of internal defects	Microcleanliness DOE	Use of new synthetic tundish cover	Changes in the chemical composition to decrease macrosegregation
on rolled and cold drawn tubes	Use of special traceable intakes to check the inclusions origins		
	Surfboard analysis		

TABLE 2—Experiments and analyses realized at a tube rolling mill of a Brazilian bearing tubes producer between 2002 and 2005.

2002	2003	2004	2005
Definition and study of influence of cast structure in piercing operation	Statistical control of process parameters at the rolling mill		
Establishment of a standard for process parameters on hot rolling heating furnace	and analyses of their influence on the final decarburization of cold drawn tubes	Detailed visual and ultra-sonic inspections	Engineering analysis crossing the performance results of each heat at rolling mill with its particular steel making parameters and casting conditions

TABLE 3—Experiments, analyses, and investments realized at a cold drawing plant of a Brazilian bearing tubes producer between 2002 and 2005.

2002	2003	2004	2005
Initial development of high frequency ultra-sonic (U.S.) macrocleanliness measurements (cross-check experiments with other companies)	Definition of a common method for high frequency U.S. macrocleanliness measurement (Brazilian, French, and German research centers)	Quality defect catalogue for each product family	Continuous evaluation of linear density of macroinclusions inside cold drawn tubes
Initial production of tubes with higher sulfur content	Introduction of a new index to evaluate the regular U.S. defect indications: Number of indications per metre	Initial evaluation of linear density of macroinclusions inside cold drawn tubes	Investments at the soft annealing furnace
Cross-check analysis for macrocleanliness measurements	Definition of a quality defect catalogue for cold drawn tubes		
Cross-check experiments using steel from the Brazilian producer and from other French steel plants, hollows from the Brazilian producer and German rolling mills, and cold drawing tubes from Brazilian and French producers	Evaluation of quality impacts on the whole production chain due to producing heats with higher sulfur content Improved decarburization control on the soft annealing furnace		

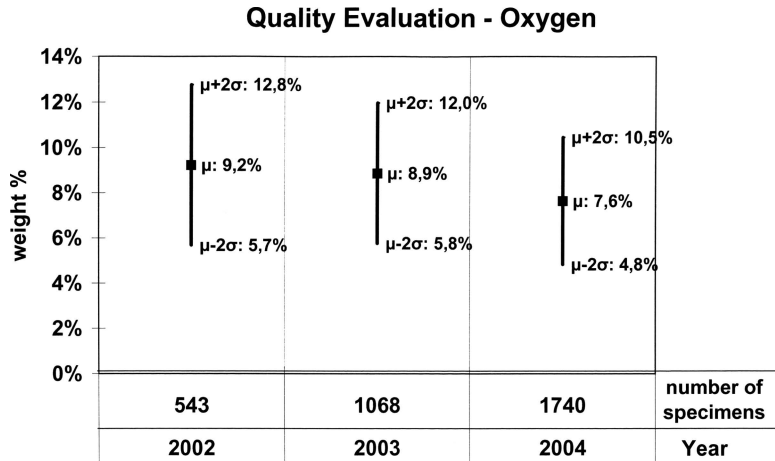


FIG. 3—Quality evolution between 2002 and 2004—oxygen content in steel.

ues were collected and statistically analyzed to evaluate the benefits or the negative consequences of any trial along the production lines. All practices, which provided global quality improvement, were afterwards incorporated into the producer's production standard for 100Cr6 steel.

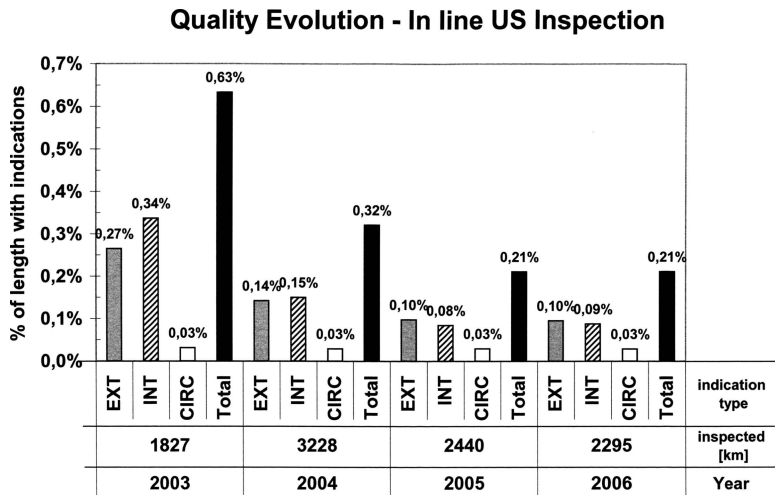


FIG. 4—Quality evolution between 2002 and 2006—results of in line ultra-sonic inspection (EXT and INT are the detection of external and internal superficial defects, respectively, corresponding to longitudinal and transversal cracks, risks, etc.; CIRC is the detection of defects like laps, roles, etc.).

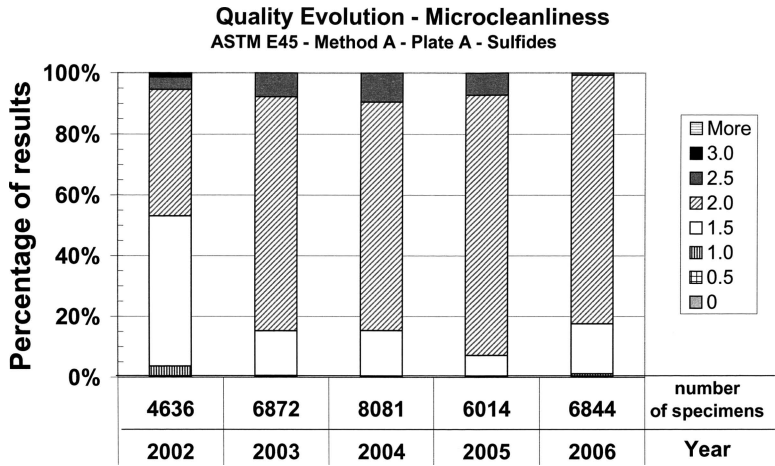


FIG. 5—Quality evolution between 2002 and 2006—microcleanliness analysis, sulfides.

Conclusions

Following a strong and focused R&D plan based on QFD methodology, a Brazilian bearing seamless tubes producer achieved a significant success concerning its strategy to be a premium 100Cr6 bearing tubes producer. Present customer audits and quality reports demonstrate the increase in customer satisfaction, enabling the company to sustain and sell with great competitiveness its products for first class markets worldwide.

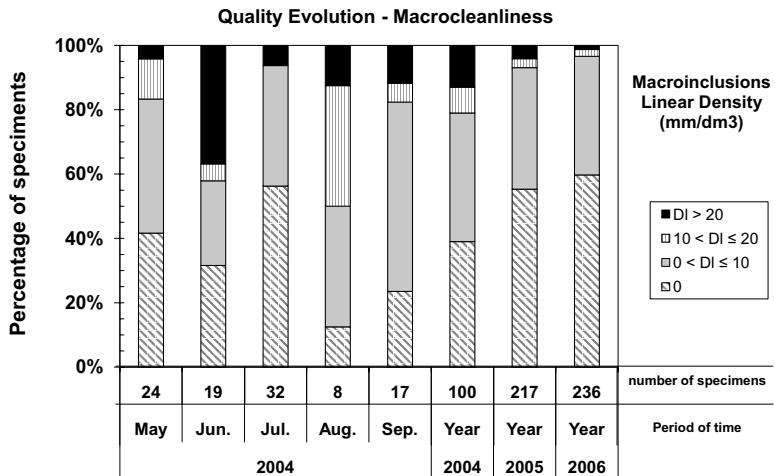


FIG. 6—Quality evolution between 2002 and 2006—macrocleanliness analysis.

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Comparison of Inclusion Assessment Rating Standards in Terms of Results and Reliability by Numerical Simulation

ABSTRACT: Today, the cleanliness assessment of bearing steel is usually performed by using standard metallographic methods such as ASTM E45, DIN 50602, Norme Internationale ISO 4967, ASTM 2283, etc. These methods are based on the estimation of indexes, and they use either reference images given by charts (Plate I-r for use with ASTM E45) or the principle of the extreme values. The obtained indexes contribute to the quality assessment of a heat. As all these methods do not give the same results, they must be compared to determine the following: First, what the most appropriate method for a given case is and, second, what the reliability of each obtained result is. It is nearly impossible to answer these questions on the basis of a set of experimental measurements coming from these methods. Indeed, this approach is inevitably time-consuming and does not offer any guarantee as to the conclusions. The main reason is that no standard sample exists with known cleanliness properties. To solve this problem, we have developed a simulation approach. In this case, the different methods of cleanliness assessment are simulated on virtual samples. The inclusion populations are perfectly known in this kind of sample (number of inclusions per mm³, sizes, positions, etc.). To create them, the characteristic parameters of these populations (size distribution, elongation distribution, etc.) must be precisely obtained through experiments. To carry this out, an automatic system of measurement has been developed using a scanning electronic microscope and an energy dispersive spectrometer system. The model in this simulation approach takes into account the experimental conditions (detection limit, observed area, etc.) and gives numerical results according to the typical chart taken from the standard methods. So, it is possible to compare and to evaluate the reliability of the results from the different methods or to quantify the

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effects of a parameter of a method on the results. Moreover, it provides the reliability of an experimental result to meet the requirement of a given customer in a more precise way.

KEYWORDS: inclusions, cleanliness, standard metallographic method, simulation, reliability

Introduction

Usually, the cleanliness assessment of bearing steel is performed by using standard metallographic methods such as ASTM E45 [1], DIN 50602 [2], Norme Internationale ISO 4967 [3], ASTM 2283 [4], etc.

These methods are based on the estimation of an index using reference images given by charts (Plate I-r to be used with ASTM E45 [1]) or on a statistical method of extreme values (ASTM 2283 [4]). In the latter method, the obtained index can be represented by a morphological parameter of the largest inclusion that is “probably” present in a volume superior to the examined volume and that can be estimated through the principle of the statistics of the extreme values [5,6].

These indexes contribute to the quality assessment of a heat.

As all the methods show different results, it is worth evaluating their reliability to define the most appropriate method for a given case [7].

It is nearly impossible to evaluate this through a set of experimental measurements coming from these methods. Indeed, this approach is always time-consuming and without any guarantee as to the conclusions. The main reason is that no standard sample exists with known cleanliness properties.

In this article, an approach that is based on simulations, and which can avoid these difficulties, is presented.

The Principle of the Simulation Method

The general principle of the simulation method is described in Figs. 1 and 2.

In the model:

- (1) A virtual sample is created, defined by its dimensions.
- (2) The inclusion populations are introduced in the volume by a Monte Carlo method: The position of each inclusion depends on the criteria of a random distribution. The diameters of the inclusions follow a perfectly defined histogram. A virtual sample is obtained for which you clearly know the cleanliness parameters.
- (3) A metallographic method to assess the inclusions is described.
- (4) By calculation, the distribution of the results obtained on a lot of random areas is defined.

Using this approach, the different methods can be tested by calculation on the virtual sample with a lot of measurements in a very short time.

The quality of a method can be assessed by its ability to classify different samples on the basis of the differences between the measured indexes.

The simulation method permits one to evaluate this property (Fig. 2).

Indeed, the thinner and the more separated the histograms are (with no

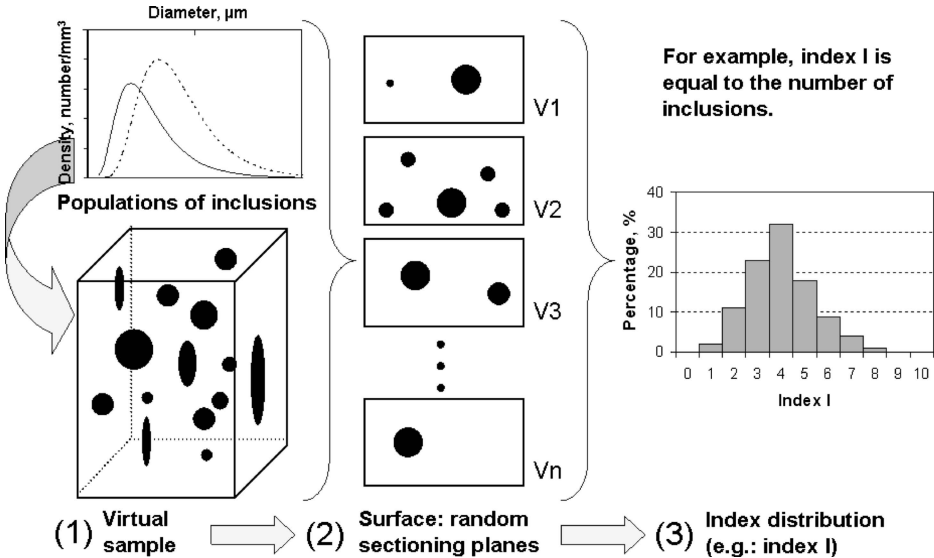


FIG. 1—Principle of the simulation method.

superimposing), the more efficient the method is to classify samples. Generally speaking, by using simulation, it is possible

- (1) To evaluate the three-dimensional (3D) density (number of inclusions per mm³) and the real size distribution of the inclusions according to the two-dimensional (2D) parameters,
- (2) To obtain the mean and the standard deviation of an inclusion index,
- (3) To compare the different rating methods, and

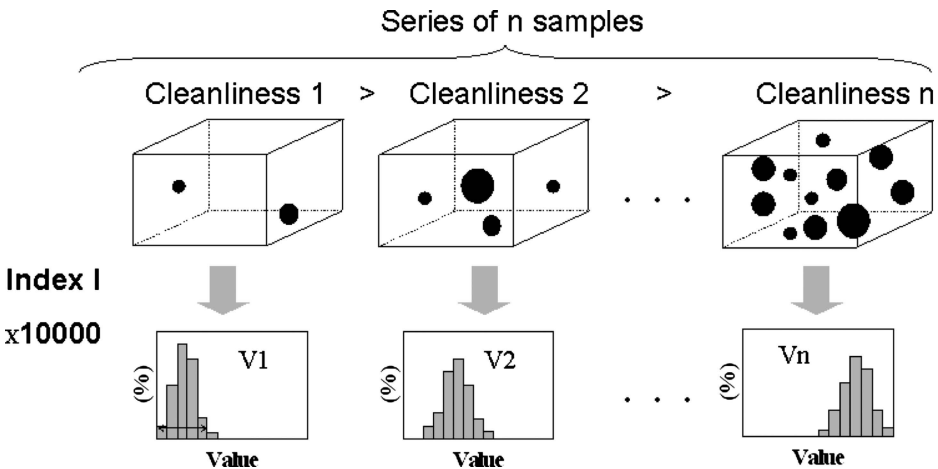


FIG. 2—Histograms of the distributions of the I parameter (e.g., 10 000 values, for the same method).

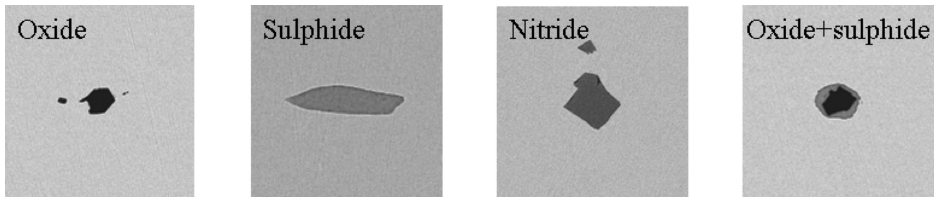


FIG. 3—Examples of inclusions in a bearing steel sample.

- (4) To define the smallest area that should be analyzed to obtain a precise result.

As is often the case the quality of the results of the simulation is going to depend on the quality of the data of entrances to the model. In our case, it is mainly about the description of the population of inclusions in samples. It is partially for this reason that we have developed a system for precise measurements.

The Characterization of Inclusion Parameters

In bearing steel, the main inclusion populations are composed of oxides, sulfides, and nitrides (and bi-phased inclusions). Inclusions are always present in steel (examples in Fig. 3) but their densities (number of inclusions per mm^3) or their sizes can vary greatly.

Regarding oxide inclusions, they appear during the steel-making and casting process. So, endogenous inclusions are formed during the de-oxidation process in liquid steel and exogeneous inclusions are produced by “incident.” For example, contact between the liquid steel and the refractory bricks can produce this second type of inclusion.

According to the probabilities, the inclusions observed in an area have generally equivalent diameters ranging from one or several micrometres to several dozens of micrometres. We can often describe these micro-inclusion populations by considering their sizes, which follow a log-normal distribution. For the inclusions that are likely to be deformed (sulfides), it is necessary to have the curves of elongation distributions to describe the populations.

To compare the rating methods, it is necessary to have accurate data about the different inclusion populations present in bearing steel heats. For that, it is better to use the system of measurement that best suits our needs.

So, an optical microscope is generally used to observe and characterize the inclusions on polished areas according to a rating method. An image analyzer system can sometimes be used to obtain results automatically [8]. But the main limits of the inclusion characterization by optical microscopy are caused by the following:

- A lack of discrimination among the different types of inclusion (the morphological criteria are often insufficient),
- Problems caused by the artifacts (dust, scars, etc.), and
- Problems caused by the poor depth of field.

To avoid these risks of errors, the features of an ideal system have to be the following:

- A signal stability to obtain comparable images,
- A stage control to be able to scan a sufficient area in order to obtain representative results,
- A magnification control to obtain accurate inclusion sizes whatever the size of the inclusion,
- A chemical contrast to discriminate among the various inclusion phases,
- A chemical composition to discriminate among the various inclusion populations,
- A resolution suited to detect inclusions smaller than 1 μm , and
- An improved depth of field to maintain image clarity.

A field-emission scanning electron microscope provides these specifications. In particular, its beam can operate for hours without any signal variation. The accuracy of the motorized stage motion allows one to detect and to observe the inclusions with various magnifications.

The functions necessary to develop the general application of the inclusion population's characterization have been defined. A general algorithm using these functions has been developed. Thus, an apparatus running automatically 24 h a day and 7 days a week has been set up.

So, more than 100 inclusions can be characterized per hour on a sample with a low density of inclusion (for example, it is necessary to measure 300 inclusions to get results with an error lower than 10 %). The morphological and analytical measurements of each inclusion are saved in a result file. So, this data can be post-treated according to the needs.

An Example of the Use of the Method of Measure

The sample is observed with a scanning magnification, which makes it possible to detect and measure inclusions larger than 3 μm in size (Fig. 4). This magnification is chosen so that enough area can be observed in a reasonable time. In the method used, this magnification is equal to 400, which corresponds to 11 images per mm^2 .

The defined thresholds permit one to segment the image. The sizes of the objects are measured on the obtained binary images.

The objects whose size is sufficient (small diameter superior to 3 μm) are analyzed automatically (Fig. 4). Under these conditions of observation, the measurement accuracy is not sufficient. So, each selected object is observed with a higher magnification. This magnification of analysis is higher than 2000 (defined during the setting in data). The stage moves so that each selected inclusion is placed at the center of the acquired image.

The morphological parameters of each inclusion and the various phase compositions are then measured. A threshold with two levels permits one to isolate oxide and sulfide phases. If the inclusion width is larger than 3 μm (the greatest magnification accuracy allows one to determine it), a chemical analysis of the different phases is made. The analysis time is equal to a few seconds. These measurements are made for each phase either in the center of the objects

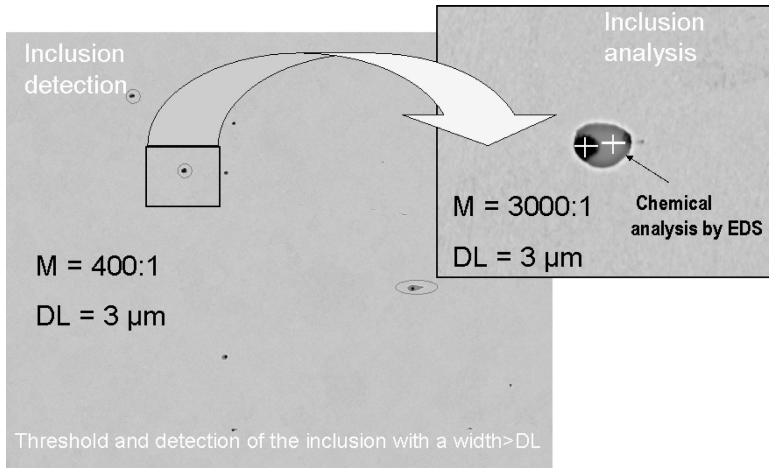


FIG. 4—Detection and analysis of an inclusion by an automatic system.

(obtained by ultimate erosion operator) or on the whole area by scanning it. The measurements are carried out on a sufficient area to reach the quality of results required from a statistical point of view.

The distributions of the morphological parameters of the different inclusion populations have been determined by this automatic system. These accurate data permit one to define the inclusion properties of bearing steel heats. They are used to continue the study and, in particular, to describe virtual samples.

Virtual Sample and Simulation

METIS is the software we developed to simulate the different rating methods. In a general way, some of its functions are used to evaluate a rating method:

- (1) The definition of the distribution of the real sizes of the inclusions,
- (2) The description of a virtual sample,
- (3) The modifications of the sample and of the inclusions (reduction ratio, not presented in this article),
- (4) The description of the rating methods,
- (5) The simulated measurements, and
- (6) The treatment of the data.

Evaluation of the Real Inclusion Size Distribution

The description of an inclusion population in a sample can be done by assessing the functions of distribution of the morphological parameters of the inclusions. Most of the time, it is a matter of defining the distribution of the real equivalent diameters and the distribution of the elongations. However, when

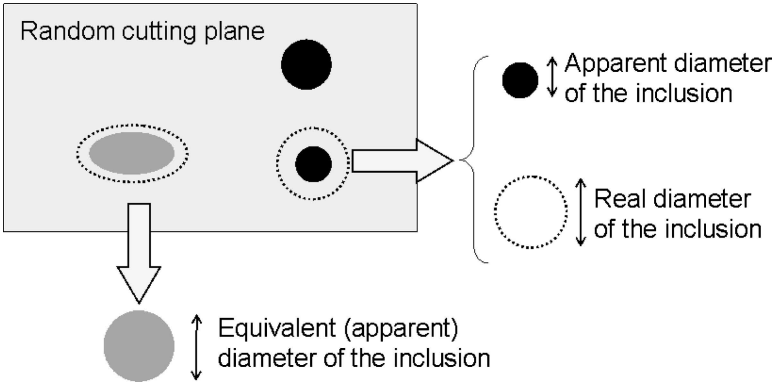


FIG. 5—3D/2D effect on diameters: definitions.

the size of an inclusion is estimated by observing a sectioning plan, the apparent diameter is lower than or equal to the true diameter of the inclusion (Fig. 5).

According to the principles shown in Fig. 6, calculation and simulation permit one to evaluate the distribution of the real diameters and the volume density of the inclusions (number per mm³) from data obtained by measurements using a scanning electronic microscope.

So, once these characteristics are known, it is possible to constitute virtual samples that represent well bearing steel heats.

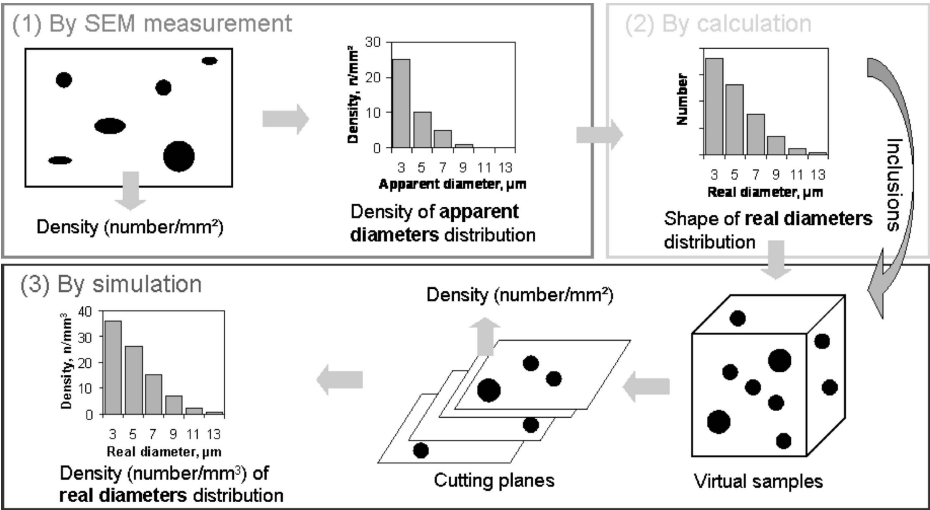


FIG. 6—Evaluation of the real size distribution.

Virtual Sample Creation

To create a virtual sample, it is necessary to describe at least the following data:

- (1) The geometry of the sample and
- (2) The number of inclusion populations.

And for each inclusion population:

- (1) The volume density of the inclusions (number of inclusions per mm^3),
- (2) The distribution of the real diameters, and
- (3) The distribution of the elongations (if the inclusions are deformable).

Through this method, all the samples you need for a study can be created. A sample can contain an unlimited number of populations. You can introduce stringers. You can simulate a reduction of the sample geometry taking into account the modifications of inclusion geometry (another function, not described in this article).

Description of a Rating Method

Another part of the software permits one to describe the different rating methods.

For example, it is possible to define the standard method Norme Internationale ISO 4967. In this case, the inclusions must be rated in a sectioning plan in the rolling direction. This method is based on the comparison of an observed field with different types of images. So, the parameters of the charts are contained in a file of the software (rating limits and inclusion thickness parameters). The other defined parameters are the detection limit (the thickness of the smallest inclusion detected) and the observed area.

All the methods you need for a study can be described.

For a given method, each parameter can be modified to study its effect (e.g., the analyzed area).

Simulated Measurements

The principle of the simulated measurements is to be as similar as possible to the experimental measurements. So, in the case of the Norme Internationale ISO 4967 standard method, an area is determined. The characteristics of the inclusions contained in this area are calculated (positions, apparent sizes, and inclusion types). The elementary images are defined to be compared with standard charts (Fig. 7). The indexes of each elementary field corresponding to an image of the diagram charts are obtained by calculation.

All the parameters of the rating method are saved in a data file. Thus, the distribution of the values of an index can be obtained.

Whatever the data, the time you need for calculation does not exceed a few minutes.

Data Treatments

The data treatments consist mainly of obtaining the distribution of the values of all the indexes. In Fig. 8, you can see the distributions obtained for two parameters described in the Norme Internationale ISO 4967 method:

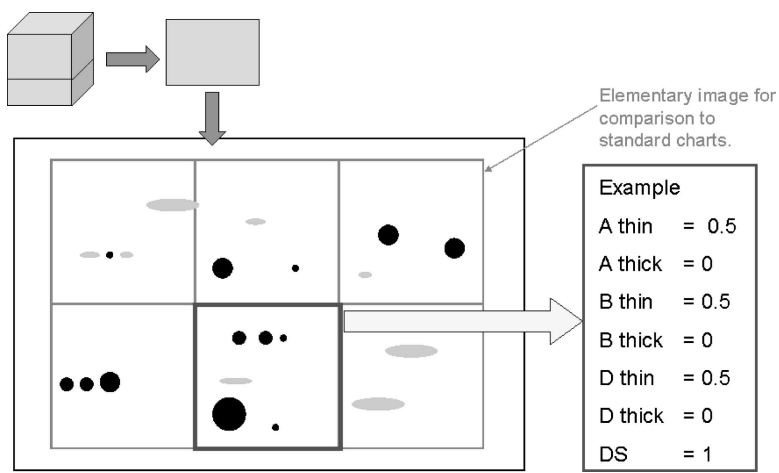


FIG. 7—Example of a simulated measurement.

- The cleanliness index (I73) and
- The index of the worst field for the B thick type inclusion (I77).

To highlight some possibilities of the simulation software, two examples of study are presented in the following paragraphs:

- (1) The comparison of ASTM E45, DIN 50602, and Norme Internationale ISO 4967 indexes and
- (2) The influence of the elementary area on the extreme value statistical method (ASTM E2283).

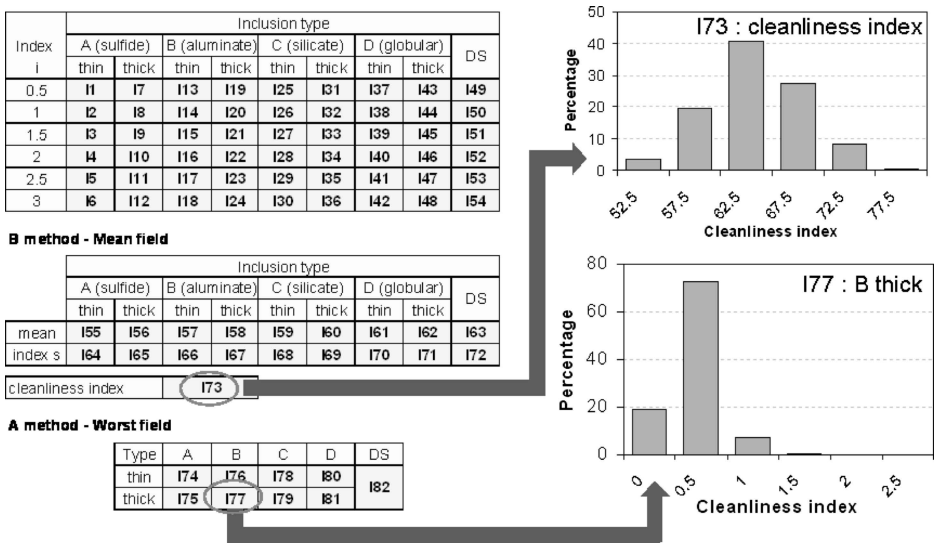


FIG. 8—Example of a data treatment.

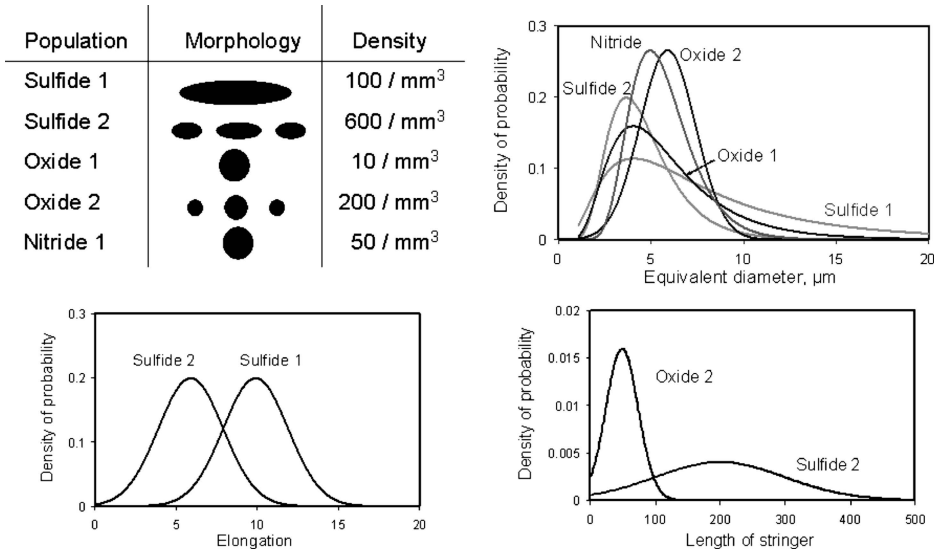


FIG. 9—Description of the five inclusion populations: experimental data.

Examples of Results of Simulation

For the two examples, only one virtual sample has been created containing five different types of inclusions. The inclusion populations in this sample are typical of those observed in a bearing steel (Fig. 9).

Example 1: Comparison of ASTM E45, DIN 50602, and Norme Internationale ISO 4967 Indexes

In Table 1, the statistical data (worst field method for Norme Internationale ISO 4967 and ASTM E45 methods) are obtained through 1000 simulated measurements (5 min simulation versus 1 man-year experimental measurement).

For each index average, standard-deviation, maximal value, minimal value, and median value are evaluated.

You can observe that the data obtained through the two standard methods, Norme Internationale ISO 4967, and ASTM E45, are similar.

These results can be used to determine the uncertainty related to each index.

For example (Fig. 10), in the same sample, according to the random sectioning surface, $D_{thick}=1(80\%)$ or $D_{thick}=1.5(20\%)$ can be obtained.

So, according to these data, it is possible to determine the best method to classify the produced heats. Presently, this aspect is studied according to specific grades and in-service applications.

Example 2: The Influence of the Elementary Area on the Results of the Extreme Value Statistical Method

The extreme value statistical methods are more and more commonly used. A standard, ASTM E2283, has been in use since 2003. Through this method, we observe a bigger area than for the other standard methods.

TABLE 1—Results obtained by simulation with Norme Internationale ISO 4967, ASTM E45, and DIN 50602 methods.

Norme Internationale ISO 4967							
	A Thin	A Thick	B Thin	B Thick	D Thin	D Thick	DS
Average	0.64	1.22	0.66	0.30	1.09	1.12	0.23
Standard deviation	0.14	0.30	0.16	0.28	0.19	0.23	0.30
Min	0.60	1.00	0.60	0.00	1.00	1.00	0.00
Max	1.00	2.00	1.00	1.00	1.60	2.00	1.00
Median	0.60	1.00	0.60	0.60	1.00	1.00	0.00
ASTM E45							
	A Thin	A Thick	B Thin	B Thick	D Thin	D Thick	
Average	0.68	1.07	0.72	0.46	1.04	1.00	
Standard deviation	0.16	0.17	0.26	0.26	0.14	0.19	
Min	0.60	1.00	0.60	0.00	1.00	0.60	
Max	1.00	1.60	1.00	1.00	1.60	1.60	
Median	0.60	1.00	0.60	0.60	1.00	1.00	
DIN 50602							
	K0 Sulfide	K0 Oxide	K0 Total	K4 Sulfide	K4 Oxide	K4 Total	
Average	36.89	23.52	90.41	16.42	0.15	16.58	
Standard deviation	13.36	3.98	13.96	9.19	0.87	9.24	
Min	30.90	16.10	48.00	0.00	0.00	0.00	
Max	101.30	33.70	131.40	35.70	5.10	40.80	
Median	36.55	23.10	90.55	15.30	0.00	15.30	

Its main disadvantage is the time needed to obtain a result. So, it is interesting to verify if it is possible to change the size of the elementary area or the number of these elementary areas without changing significantly the quality of the results.

Through the simulation approach, it is possible to study the influence of the different parameters of the method.

In this case, the calculations and graphical representation of extreme value data analysis are obtained for the following elementary areas equal to

- 25 mm²,
- 150 mm² (standard ASTM E2283 requirement), and
- 900 mm².

The other parameters are set as follows:

- The inclusions=all types,
- The detection limit=2 μm,
- The size parameter=equivalent diameter,

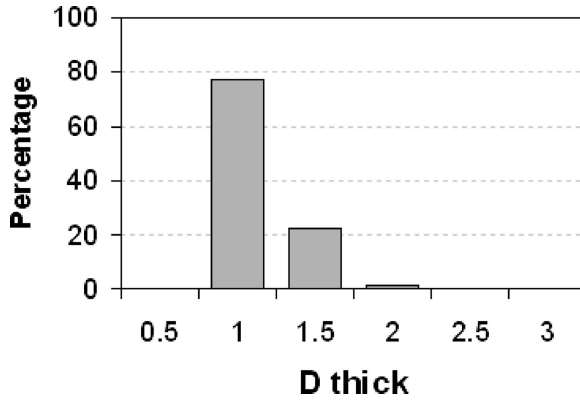


FIG. 10—*D thick*: results obtained from the same samples (1000 measurements).

- 24 elementary areas are analyzed,
- The maximum likelihood analyses are used to represent the best-fit line for the data, and
- Deq_{max} is calculated for a $A_{ref.} = 36\,000\text{ mm}^2$.

These statistical data are obtained by 100 simulated measurements (10 min simulation versus several man-months experimental measurements).

In Fig. 11, an example of SEV curves for the three conditions of measurement is presented. Each point represents the biggest inclusion observed on each elementary area.

You can see the impact of the elementary area on the obtained results. The biggest inclusions are not the same according to the elementary area. The observed part of the distribution of the size of the inclusion is not the same.

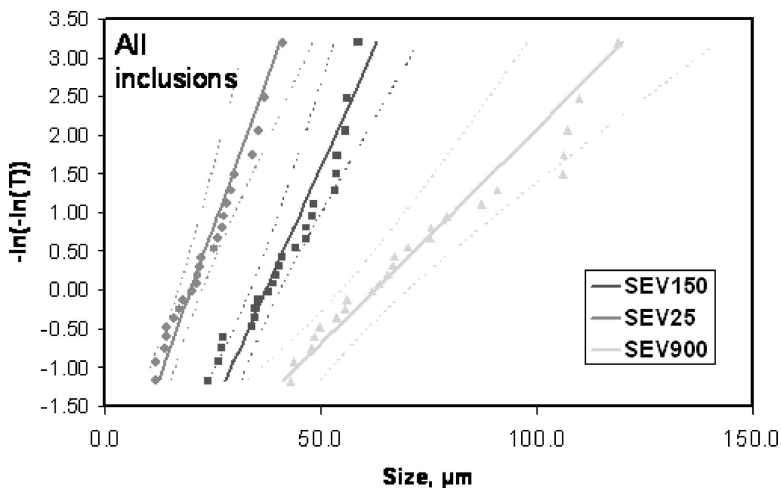


FIG. 11—*SEV curves*.

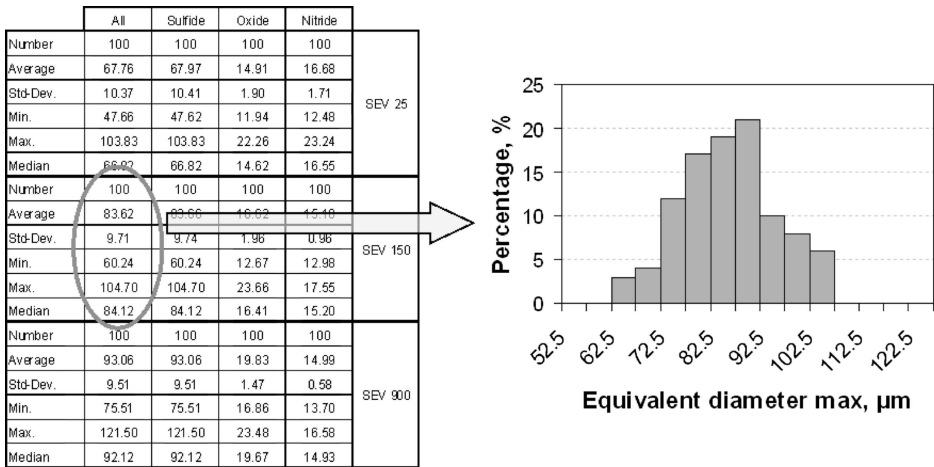


FIG. 12—Comparison of *Deq(max)* distributions according to the elementary area observed.

Comparisons of *Deq(max)* distributions can be observed according to the values of the elementary area. *Deq(max)* has been calculated for an area of 36 000 mm² (extrapolation of the lines).

In Fig. 12, the influence of the elementary area on the results for *Deq(max)* is visible. So, to compare different results, it is necessary to verify if they have been obtained with the same measurement conditions.

Conclusions and Perspectives

A methodology based on simulation has been developed to study the reliability of the different methods that permit one to assess the cleanliness in bearing steel.

Virtual samples are thus defined according to experimental data obtained by observing different areas.

The use of a scanning electronic microscope permits one to obtain more accurate data about inclusion populations for a heat.

The different inclusion populations are described thanks to the distribution of equivalent diameters, the distribution of elongations (for sulfide or silicates), and the description of the stringers (length, number of particles, and sizes of particles).

By means of calculation and simulation, it is possible to obtain 3D data on the basis of 2D data.

Each virtual sample can be characterized by numerous simulations according to the principle of the different standard methods that are based on comparisons between chart diagrams or on the estimation of an extreme value, and all this can be done in a very short time (minutes versus months or years).

The results of the standard methods (Norme Internationale ISO 4967 [3], ASTM E45 [1], and DIN 50602 [2]) can be compared on different samples ac-

cording to the distribution of the corresponding indexes obtained by simulation. Thus, it will be possible to define both the best method to classify heats and the limits of use of the standard methods.

Furthermore, we have presented an example of the results obtained on different elementary areas according to the principle of the ASTM 2283 [4] method on a sample. We have shown the important influence of the elementary area on the results.

This methodology will be used to define the uncertainty of each index.

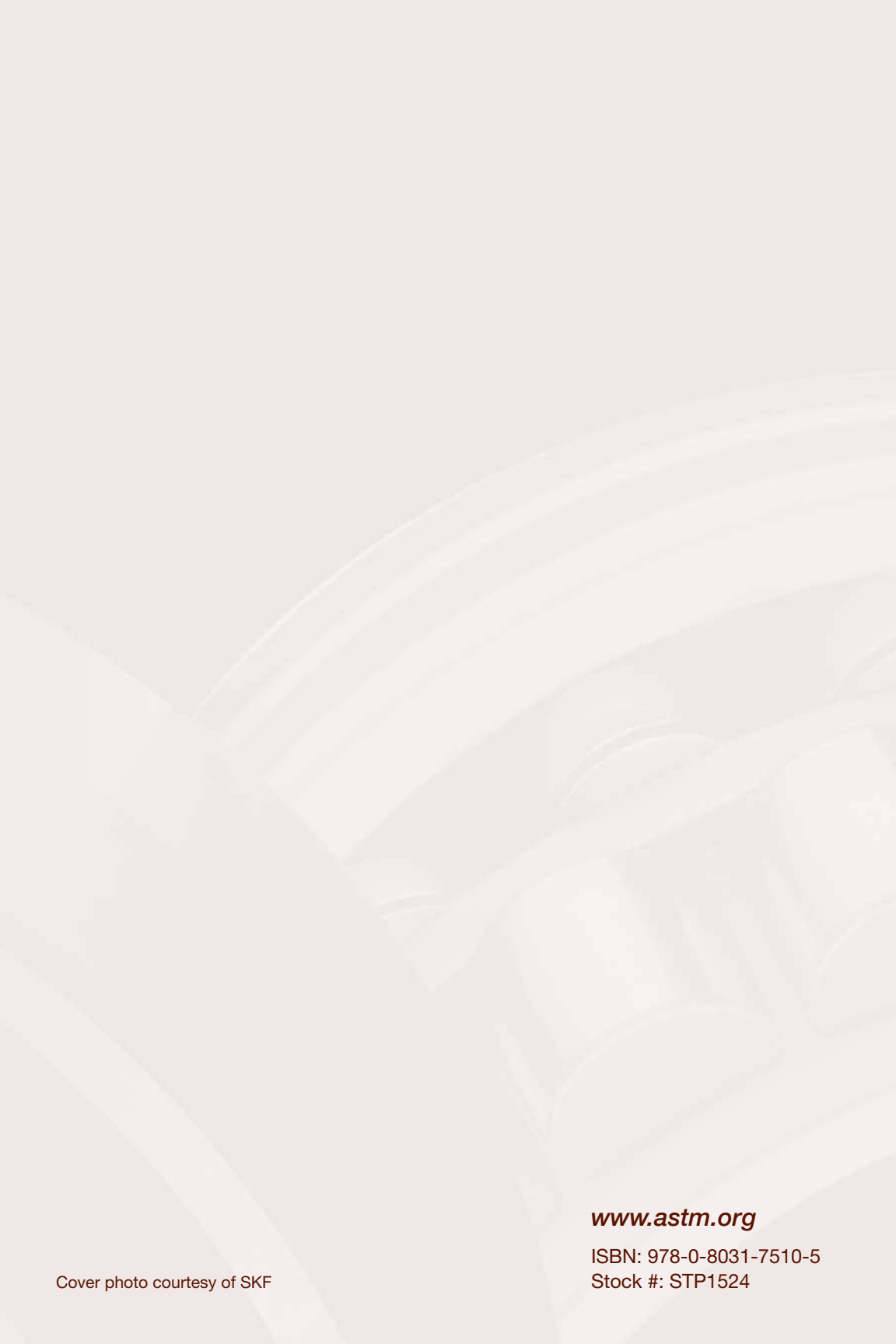
The rating principles of the EN 10247 [9] standard method are significantly different. We will compare it to the other standard methods in a future study.

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