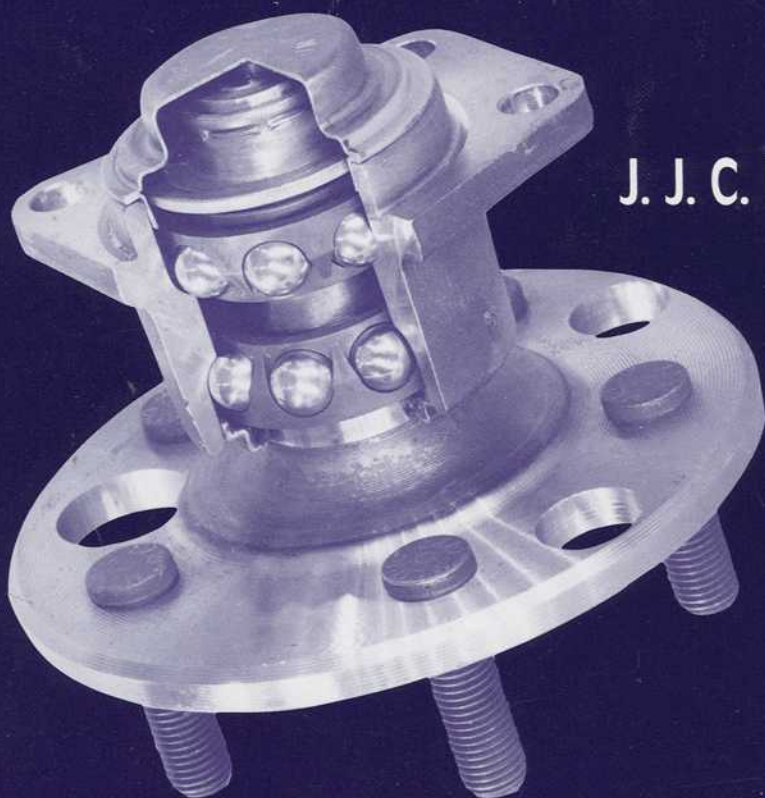


Effect of Steel Manufacturing Processes on the Quality of Bearing Steels



J. J. C. Hoo
editor

ASTM STP 987

STP 987

***Effect of Steel Manufacturing
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 ASTM
1916 Race Street
Philadelphia, PA 19103

Library of Congress Cataloging-in-Publication Data

Effect of steel manufacturing processes on the quality of bearing steels/J. J. C. Hoo, editor.

(STP; 987)

Papers from the Symposium on Effect of Steel Manufacturing Processes on the Quality of Bearing Steels, held at Phoenix, Ariz., Nov. 4-6, 1986 and sponsored by the Subcommittee A01.28 on Bearing Steels of the Committee on A01 on Steel, Stainless Steel, and Related Alloys.

Includes bibliographies and indexes.

"ASTM publication code number (PCN) 04-987000-02."

ISBN 0-8031-0999-7

I. Steel, Bearing—Fatigue—Congresses. 2. Rolling contact—Congresses. 3. Steel—Metallurgy—Congresses. I. Hoo, J. J. C. II. Symposium on Effect of Steel Manufacturing Processes on the Quality of Bearing Steels (1986: Phoenix, Ariz.) III. American Society for Testing and Materials. Subcommittee A01.28 on Bearing Steels. IV. Series: ASTM special technical publication; 987.

TA473.E42 1988

620.1'73—dc19

88-19876

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The quality of the papers in this publication reflects not only the obvious efforts of the authors and the technical editor(s), but also the work of these peer reviewers. The ASTM Committee on Publications acknowledges with appreciation their dedication and contribution of time and effort on behalf of ASTM.

Foreword

The symposium on Effect of Steel Manufacturing Processes on the Quality of Bearing Steels was presented at Phoenix, Arizona, 4–6 November 1986. The symposium was sponsored by Committee A01 on Steel, Stainless Steel, and Related Alloys and Subcommittee A01.28 on Bearing Steels. J. J. C. Hoo, General Bearing Corporation, served as chairman of the symposium and editor of this publication.

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Overview

The fatigue life of rolling bearings has experienced a significant increase in the past several years. The improvement is attributed to the lowering of oxygen content in the bearing steels to the level of less than 10 parts per million by weight. This accomplishment is the direct result of new ladle degassing practices adopted in the steel making processes. Associated with this progress the methods used to evaluate the quality of bearing steels have been also improved, refined, and in many instances newly developed.

It is very timely for the American Society for Testing and Materials Committee A01, Subcommittee A01.28 on Bearing Steels to sponsor an international symposium on the theme of effect of steel manufacturing processes on the quality of bearing steels. The symposium was held on 4 to 6 Nov. 1986 in Phoenix, Arizona. This is the third symposium sponsored by ASTM Subcommittee A01.28 since May 1974 on bearing steels. We have set up a target that for every five to six years we will provide a forum for bearing steel producers and users to get together to present their latest research results, the newest state of the art, and to discuss the direction for future development. The response from the scientific and engineering community in the world has been very enthusiastic. This symposium in Phoenix received papers from Canada, France, Germany, Italy, Japan, Netherlands, Sweden, United Kingdom, and the United States. Almost all the major rolling bearing and bearing steel manufacturers in the developed countries participated. We are extremely gratified for both the quality and quantity of the papers received.

This book has collected 22 papers presented at the symposium. It is divided into three sections. In the first section there are nine papers discussing quality requirements for better bearing steels. In this section the phenomenal improvement in rolling bearing fatigue life in recent several years is unambiguously shown in facts and figures obtained both in laboratories and in the field. The significant contribution of a new quantitative metallographical method, the SAM method, adopted by ASTM A01.28 to assess the nonmetallic inclusion content in bearing steels is also clearly demonstrated. This method is detailed in Supplementary Requirement S2 of ASTM Standard Specification for High-Carbon Ball and Roller Bearing Steel (A 295-84). It has proved once more how an improved test method can stimulate the progress of manufacturing processes. Since the rating of nonmetallic inclusions was the major theme of the first bearing steel symposium held in Boston, Massachusetts in 1974 and *ASTM STP 575, Bearing Steels: The Rating of Nonmetallic Inclusion* was the first book published anywhere dedicated entirely to that subject, we proudly feel that the symposium has become an integral part of that progress. The second bearing steel symposium was held in Phoenix, Arizona in 1981 and was published as *ASTM STP 771, Rolling Contact Fatigue Testing of Bearing Steels*.

The second section of this book contains five papers dealing with new methods to evaluate the quality of better bearing steels. Attention should be paid to papers discussing the analytical method for the determination of oxygen in steel, and the latest round-robin tests conducted jointly by ASTM Subcommittees A1.28 and E03.01 on Ferrous Metals. Again we are proud to show the world that ASTM has taken a leading role in the standardization of a new test method that has become very important in its application. It is equally significant that papers in this section do not respond to the papers in the first section of this book in that there is an urgent need to develop a practical and meaningful test method for the evaluation of macro-inclusions in bearing steels. This is no doubt an indication to the scientists and engineers of a future research subject.

2 EFFECT OF STEEL MANUFACTURING PROCESSES

The third section of this book has eight papers covering the new processes to produce better bearing steels. We designed the sequence of the papers in this book in such way so that the readers can appreciate what the quality requirements of good bearing steels are, and how they are tested before reading the new manufacturing methods. Judgment must be made on sound and thorough understandings. Only after good bearing steel is defined can the reader judge how good a new manufacturing method is. All eight papers in this section are presented by leading specialty steel manufacturers in the world.

Both rolling bearings and bearing steel industries are highly competitive. Many research results are tightly guarded as trade secrets. Moreover, because bearing steel fatigue testing is very time consuming and usually a set of good comparative tests lasts several years, the academic institutions have shown little interests in taking up the research. We feel this book adds valuable information to the science and technology of bearing steels.

Joseph J. C. Hoo

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Blauvelt, NY 10913; symposium chairman
and editor.

Quality Requirements for Better Bearing Steels

Selection of Rolling-Element Bearing Steels for Long-Life Applications

REFERENCE: Zaretsky, E. V., "Selection of Rolling-Element Bearing Steels for Long-Life Applications," *Effect of Steel Manufacturing Processes on the Quality of Bearing Steels*, ASTM STP 987, J. J. C. Hoo, Ed., American Society for Testing and Materials, Philadelphia, 1988, pp. 5-43.

ABSTRACT: Nearly four decades of research in bearing steel metallurgy and processing has resulted in improvements in bearing life by a factor of 100 over that obtained in the early 1940s. For critical applications such as aircraft, these improvements have resulted in longer-lived, more reliable commercial aircraft engines. Material factors such as hardness, retained austenite, grain size and carbide size, number, and area can influence rolling-element fatigue life. Bearing steel processing such as double-vacuum melting can have a greater effect on bearing life than material chemistry. The selection and specification of a bearing steel is dependent on the integration of all these considerations into the bearing design and application. The paper reviews rolling-element fatigue data and analysis, which can enable the engineer or metallurgist to select a rolling-element bearing steel for critical applications where long life is required.

KEY WORDS: bearing steel, carbide, fatigue, grain, hardness, lubricant, retained austenite, rolling element, vacuum

Through the use of improved technology, rolling-element bearing life and reliability have increased dramatically over the past four decades. A chart showing the major advances contributing to these life improvements is shown in Fig. 1 [1]. The major reason for these advances has been the rapidly increasing requirements of aircraft jet engines from the early 1950s to the present.

Starting in the early 1940s, new developments in the making of bearing steels began. The improved steelmaking developments were primarily initiated by the acceptance of a comprehensive material specification for AMS 6440 and AISI 52100 steel (A, Fig. 1). New heat-treatment equipment became available in 1941 which incorporated improved temperature controls and recorders. The use of neutral atmospheres during heat treatment eliminated, for all practical purposes, surface decarburization (B, Fig. 1).

As the requirement for bearing steel increased, large electric arc furnaces were installed which produced larger size billets. These larger billets necessitated working the material to reduce the billets to size for tubing or individual forgings. The working of the bearing steel refines the steel grain and carbide size and reduces the size of the materials inclusions and segregates (C, Fig. 1). This trend toward larger furnace size has continued to this time [1].

Major advances in melting practice evolved over a period between 1952 and the early 1970s. Immersion thermocouples, introduced in 1952 (D, Fig. 1), permitted better control of steel melting [1].

Some significant manufacturing process changes were made in the 1950s. Shoegrinding (E, Fig. 1) was introduced about 1953. This method improved race surface quality and tolerance.

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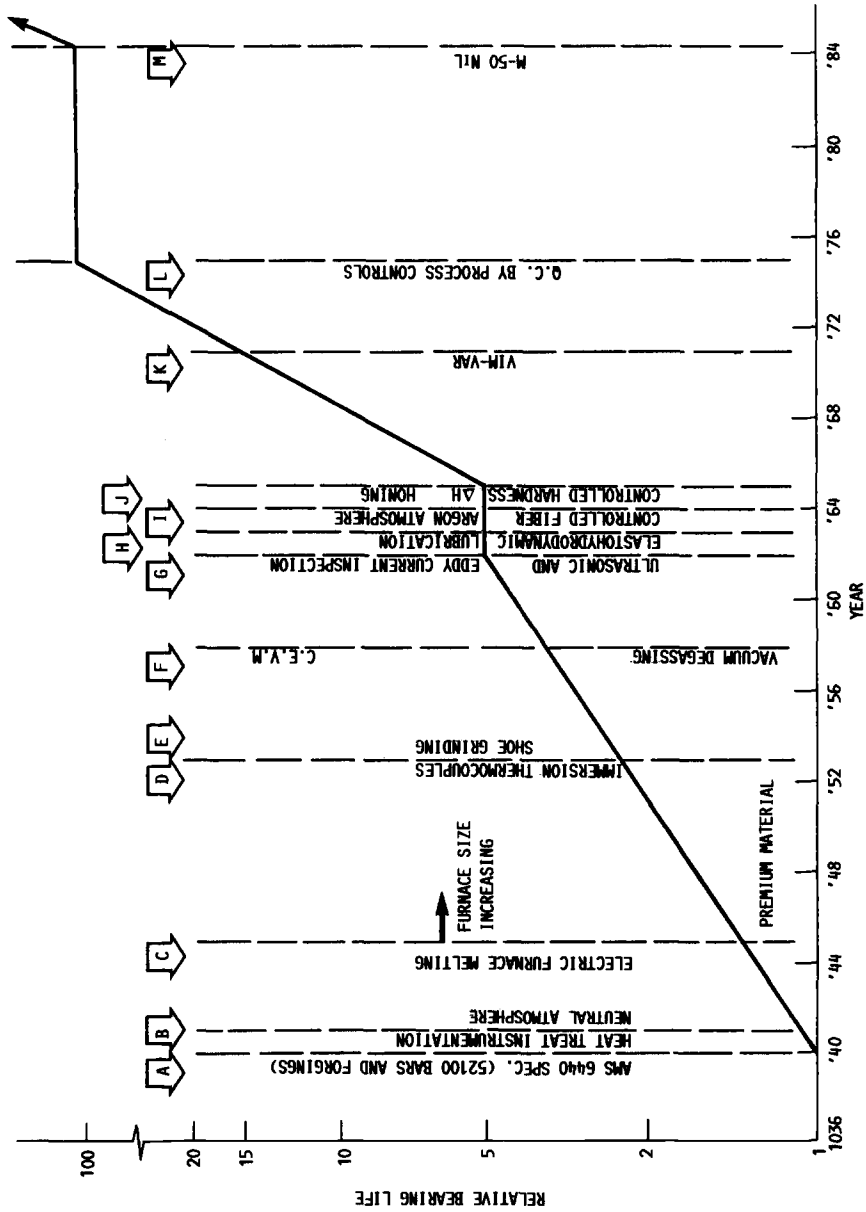


FIG. 1—Major advances contributing to rolling-element bearing life improvement over the past four decades.

With this grinding method, it is practically impossible to grind eccentricity and face runout into the bearing race. Also, the transverse radii of the races, controlled by the grinding wheel dresser, are more consistent [1].

The vacuum degassing and vacuum melting processes were introduced to the bearing industry in the late 1950s. Consumable-electrode vacuum melting (CEVM) was one such process (F, Fig. 1). Vacuum melting releases entrapped gases and reduces the quantity and alters the type of inclusions and trace elements present in the steel.

In order to assure clean steel with the vacuum-melting processes, nondestructive testing, using eddy current and ultrasonic methods, was applied to billets, bars, and tubing (G, Fig. 1). This assured the quality of the steel for the bearing manufacturing process.

In rolling-element bearings, the elastically deformed rolling-element surfaces are separated by a thin lubricant film referred to as an elastohydrodynamic (EHD) film [2]. The concept of EHD lubrication, while recognized in 1949 [3], was further recognized as a significant factor in affecting bearing fatigue life and wear (H, Fig. 1). By controlling the EHD film thickness through lubricant selection and control of operating conditions together with the improvements in surface finish, rolling-element bearings were able to operate at higher temperatures and for longer times [4].

In the 1960s, argon atmosphere protection of the molten steel during teeming was introduced (I, Fig. 1). Drastic improvement in micro- and macroscopic homogeneity and cleanliness with a resultant improvement in fatigue was realized [1].

Prior to the 1950s, as-ground races were hand polished to improve finish and appearance. Overly aggressive polishing could create a thin layer of plastically displaced or smeared material which was softer and more prone to fatigue failure. This manual process was replaced by mechanized honing in which all parts are smoothed in a more uniform manner (J, Fig. 1).

In 1958, the National Aeronautics and Space Administration (NASA) published its results of controlled fiber or grain on the effect of bearing life [5,6]. Controlled fiber can be obtained by forging to shape the raceway of angular-contact ball bearings. Forged raceways with controlled fiber orientation was introduced in 1963 (I, Fig. 1). This innovation improved the life of angular-contact ball bearings.

Work performed by NASA beginning in the late 1950s on material hardness effects culminated with the discovery of the differential hardness principle or controlled hardness (J, Fig. 1) [7]. Prior to this time, significant variations between rolling-element and race hardnesses could result in significant reduction in bearing life.

Combining improved surface finishes obtained by honing, improved lubricants whose selection was based upon EHD principles, controlled fiber and hardness, consumable-electrode vacuum melted (CEVM) AISI M-50 steel, as well as improved nondestructive inspection of the steel billet, relative bearing life of approximately 13 times the 1940 standard was achieved in 1975 [4]. The NASA research culminated by using, for the first time, vacuum-induction melted, vacuum-arc remelted (VIM-VAR) AISI M-50 (K, Fig. 1) demonstrating lives in excess of 100 times the 1940 standard at speeds to three million DN [8]. The improvement in lives with the VIM-VAR process was accompanied by improved product consistency by reducing human element variability through better process controls and audits (L, Fig. 1) [1].

In 1983, Bamberger [9] at General Electric Co. developed a significantly improved AISI M-50 steel which he called M-50NiL. This new steel was capable of being case-hardened and exhibited lives in excess of through hardened VIM-VAR AISI M-50 (M, Fig. 1).

The steel technology for long-life bearing application, over the last 20 years, has reached a 20-fold increase in life potential. The object of this paper is to review rolling-element fatigue data and analysis which can enable the engineer or metallurgist to select and specify a rolling-element bearing steel for critical application where long life is required.

Material Cleanliness

Rolling-element fatigue is a mode of failure that occurs in rolling-element bearings such as ball and roller bearings. It is a cyclic-dependent phenomenon resulting from repeated stresses under rolling-contact conditions. Fatigue can be affected by many variables, such as rolling speed, load, material, sliding within the contact zone, temperature, contact geometry, and type of lubricant. The fatigue failure manifests itself initially as a pit which, in general, is limited in depth to the zone of resolved maximum shearing stresses and in diameter to the width of the contact area (Fig. 2).

Research performed by Bear, Butler, Carter, and Anderson [5,6,10] substantiated the early findings of Jones [11] that one mode of rolling-element fatigue is due to nonmetallic inclusions. These inclusions act as stress raisers similar to notches in tension and compression specimens or in rotating beam specimens. Incipient cracks emanate from these inclusions, and enlarge and propagate under repeated stresses, forming a network of cracks which form into a fatigue spall or pit. In general, the cracks propagate below the rolling-contact surface approximately 45 deg to the normal; that is, they appear to be in the plane of maximum shearing stress (Fig. 3). Carter [10] made a qualitative generalization that the location of an inclusion with respect to the maximum shearing stress is of prime importance. Based on observations of inclusions in AISI 52100 and AISI M-1 steels, Carter concluded that

1. Inclusion location is of primary importance.
2. Size and orientation are also important.

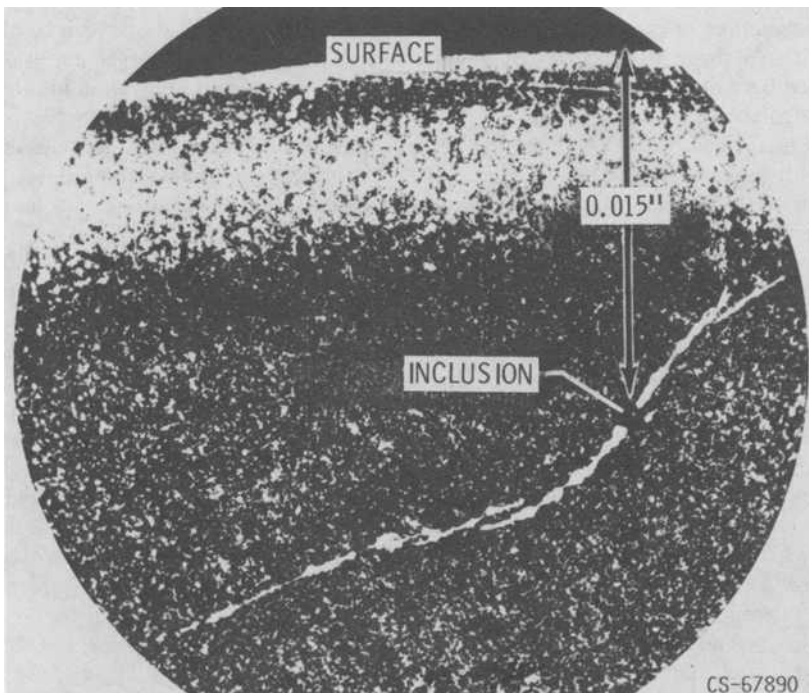


FIG. 2—Fatigue crack emanating from an inclusion.

3. The oxides and larger carbides are more harmful than the softer sulfide inclusions.
4. Inclusions, carbides, and irregular matrix conditions appear slightly less harmful to fatigue life in AISI 52100 than in AISI M-1.

Carter's conclusions were substantiated by Johnson and Sewell [12], whose results are summarized in Fig. 4. They show that as the total number of alumina and silicates increase, fatigue life decreases. However, they indicate that the increase in sulfides may have a positive effect upon fatigue life. In addition to inclusions, material defects such as microcracks, trace elements, or unusual carbide formations present in the material can contribute to failure. An attempt was made by NASA in the early 1960s to manufacture 12.7-mm (0.5-in.) diameter AISI 52100 steel balls with increased sulfur content. This effort resulted in balls having incipient cracks in their matrices.

One method for increasing rolling-element reliability and load capacity is to eliminate or reduce nonmetallic inclusions, entrapped gases, and trace elements. Improvements in steel-making processing, namely melting in a vacuum, can achieve this. These vacuum-melting techniques include vacuum induction melting (VIM) and CVM or vacuum-arc melting (VAR), as well as vacuum degassing.

It is possible with any of these melt techniques to produce material with a lower inclusion content than air-melted material, particularly those inclusions which are generally considered to be more injurious, such as oxides, silicates, and aluminates. These inclusions in part are the result of standard air melt deoxidation practice which involves the use of silicon and aluminum. Exposing the melt to a vacuum permits deoxidation to be performed effectively by the carbon. The products formed when using carbon as a deoxidizer are gaseous and thus

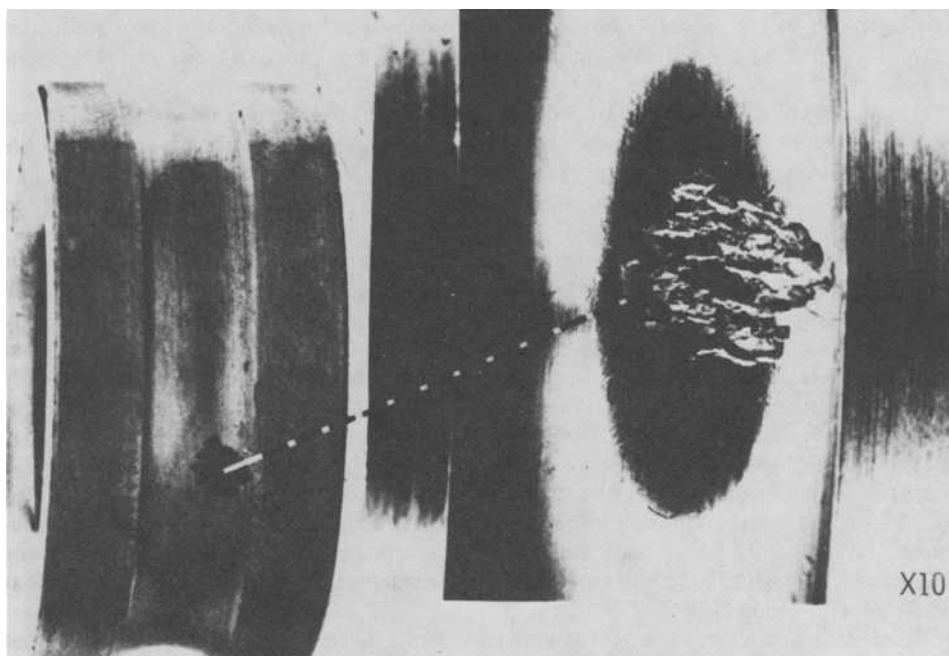


FIG. 3—Typical fatigue spall in bearing race.

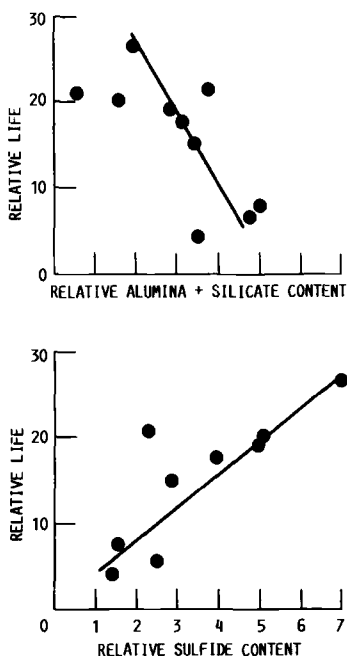


FIG. 4—Relationship between life and inclusion content.

are drawn off in the vacuum. Further, these techniques permit extremely close control of chemistry and also permit production of variations in chemical analysis which was at one time impractical.

Fatigue tests of 6309-size deep-groove ball bearings made from two heats of AISI M-50 steel produced by the CVM process resulted in an average 10% life (L_{10}) of 4.2 times the catalog life of 10 million revolutions. Additional fatigue tests of the same type of bearings made from a single heat of air-melted AISI M-50 steel resulted in a life of only 0.4 times the catalog rating [13].

The improvement in life of bearings made of vacuum-melted steels does not appear to be commensurate with the improvement in cleanliness. This, of course, supports the long-held theory that cleanliness is not the only factor involved in bearing fatigue. Even in exceptionally clean materials, nonmetallics are present to some degree and, depending on the magnitude and location in relation to the contact stresses, can be the nucleus of fatigue cracks as previously discussed. A single heat of primary air-melted AISI 52100 steel was processed through five successive consumable-electrode vacuum remelting cycles. Groups of 6309-size bearing inner-races were machined from material taken from the air-melt ingot and the first, second, and fifth remelt ingots for evaluation; they were then heat treated and manufactured as a single lot to avoid group variables. With each remelt, a progressive reduction of nonmetallic content occurred. Endurance results, summarized in Fig. 5, show that the L_{10} life appears to increase for successive remelting with the fifth remelt material reaching a life approximately four times that of the air melt group [14].

Based upon the above, it becomes apparent that significant increases in rolling-element fatigue life, and thus bearing life and reliability, can be achieved through the use of successive remelting of the bearing steel. Recognizing this fact, Bamberger, Zaretsky, and Signer [8] had

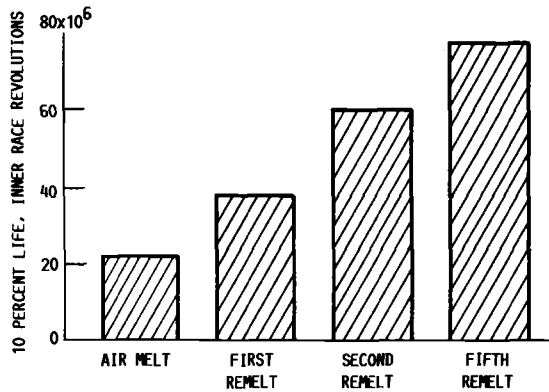


FIG. 5—Life of 6309-size bearing inner races made from air melt and successive consumable remelts of the same heat AISI 52100 steel.

120-mm bore angular-contact ball bearings manufactured from a single heat of VIM-VAR AISI M-50 steel. This was perhaps the first time this double-vacuum melting process was used for aircraft quality bearings. Two groups, each comprising 30 of these bearings, were endurance tested at a speed of 12 000 and 25 000 rpm (1.44×10^6 and 3.0×10^6 DN, where DN is a speed parameter determined by multiplying the bearing bore in millimetres by the bearing speed in rpm), respectively. Test conditions for these tests are given in Table 1. At 1.44×10^6 and 3.0×10^6 DN, 84 483 and 74 800 bearing test hours were accumulated, respectively. The results of these tests are shown in Fig. 6 and summarized in Table 1. Bearing lives at speeds of

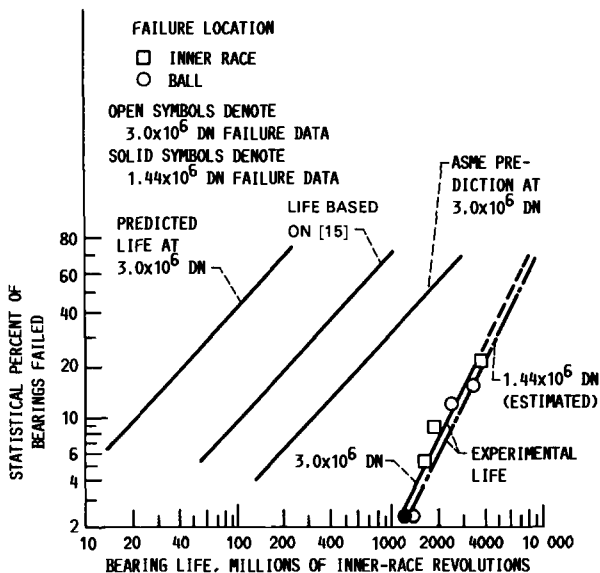


FIG. 6—Endurance characteristics of 120-mm (4.8 in.) bore angular-contact ball bearings. Thrust load, 22 241 N (5000 lb); temperature, 492 K (425°F); material VIM-VAR AISI M-50 steel; lubricant, Tetraester.

TABLE 1—Summary of endurance tests with 120-mm bore angular contact ball bearings (material, VIM-VAR AISI M-50 steel; temperature, 492 K (425°F); contact angle, 24 deg).

Speed, rpm (DN)	Thrust Load, N (lb)	Maximum Hertz Stress, N/m ² (ksi)		Predicted Life ^a rev. × 10 ⁻⁶ (h)		Experimental Life, rev. × 10 ⁻⁶ (h)		Experimental Weibull Slope	Failure Index ^c
		Inner Race	Outer Race	L_{10}	L_{50}	L_{10}	L_{50}		
12 000 (1.44 × 10 ⁶)	66 721 (5000)	2048 × 10 ⁶ (297)	1731 × 10 ⁶ (251)	29 (40)	157 (216)	b_{2700} (3750)	b_{6975} (9687)	$b_{2.1}$	1 out of 30
25 000 (3.0 × 10 ⁶)	66 721 (5000)	1965 × 10 ⁶ (285)	2096 × 10 ⁶ (304)	21 (14)	113 (76)	2400 (1600)	6200 (4133)	2.1	6 out of 30

^aCentrifugal effects included.

^bEstimated.

^cIndicates number of failures out of total number of tests.

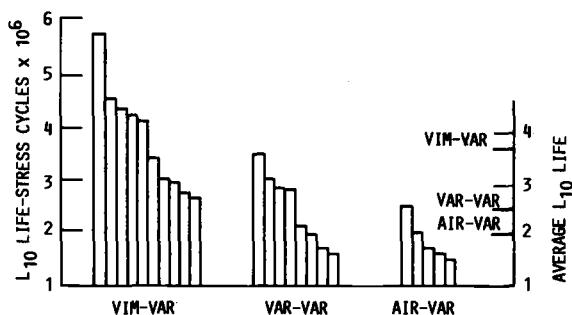


FIG. 7—Rolling-contact (RC) rig evaluation of AISI M-50 melting procedures. (Each bar represents 20 tests.)

3×10^6 DN with the VIM-VAR AISI M-50 were nearly equivalent to those obtained at lower speeds. These test results were compared with similar bearings made from CVM AISI M-50 steel run under the same conditions. At 3.0×10^6 DN, the life was in excess of 44 times that predicted on the basis of air-melt steel, approximately 23 times that using CVM steel [15], and 7.6 times that using the ASME life adjustment factors [16] based upon CVM AISI M-50 steel (see Fig. 6). Further work performed at General Electric Co. [17] using the rolling-contact (RC) tester which compared air-melted and vacuum-arc remelted (AM-VAR), double-vacuum arc remelted (VAR-VAR), and VIM-VAR AISI M-50. The results of these tests are shown in Fig. 7. The VIM-VAR material produced average 10% lives 1.9 and 1.5 those of the AM-VAR and VAR-VAR materials, respectively. These tests which are run at 4.83×10^9 N/m² (700 000 psi) and others of a similar nature tend to compress life differences because of the high Hertz stress at which they are run. However, rolling-element fatigue tests run in the NASA five-ball fatigue tester at 5.52×10^9 N/m² (800 000 psi) with 12.5-mm (0.500-in.) diameter balls made from VIM-VAR AMS 5749 and VIM AMS 5749 resulted in the VIM-VAR material having a life 14 times that of VIM AMS 5749 [18].

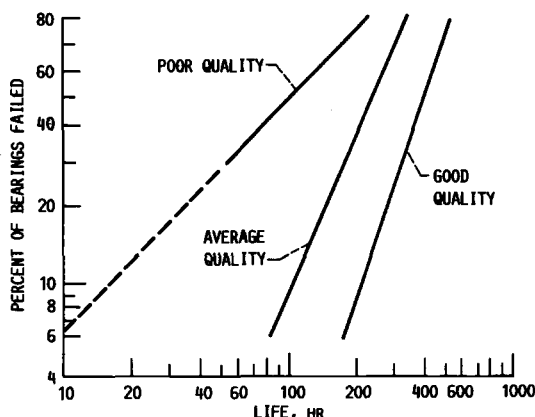


FIG. 8—Fatigue life distribution of 309-size roller bearings made from AISI 8620 steel grouped into three quality ratings based on ultrasonic response.

Even with the best quality assurance in steel melting, it is always possible for large inclusions or segregates to be encased within the billet or in the final end product. Nondestructive testing of the billet or bearing is a final step to assure the life and reliability of the end product. The most successful use of ultrasonics as an inspection tool was reported by Koved and Rospond [19]. Roller bearings made of AISI 8620 were arranged into three groups based on differences in the size and the frequency of ultrasonic indications. The individual bearings were rated by a method that assigned a numerical value to their ultrasonic pattern. The number was based on the frequency of indications and included a weighting factor for size. These numerical ratings were subsequently converted into qualitative ratings.

The fatigue life distributions for the ultrasonically rated groups of bearings are shown in Fig. 8. The three performance curves separate nicely in the same order as their ultrasonic classification. The group characterized as ultrasonically poor demonstrated the worst performance, average quality was intermediate, and good quality clearly exhibited the best fatigue-life distribution. Thus, ultrasonic inspection has the capability of differentiating, on a statistical or group basis, relative material quality in terms of life performance.

Material Hardness

Heat treatment can significantly influence several rolling-element bearing material properties. Most bearing procurement specifications do not designate heat treatment but rather call for certain material characteristics such as grain size and hardness, which are controlled by the heat treat cycle. Hardness is the most influential heat treat induced variable in rolling-element fatigue [20,21]. In general, the higher the material hardness, the longer the life. A relationship was introduced by Zaretsky [4] based upon the work reported in Ref 7, which approximates the effect of bearing material hardness on fatigue life.

$$\frac{L_2}{L_1} = e^{m(R_{c2} - R_{c1})} \quad (1)$$

where L_1 and L_2 are the bearing 10% lives at bearing hardnesses of R_{c1} and R_{c2} , respectively, and m is a material constant which can be taken as 0.1. It is assumed for the purpose of this relationship, which was obtained for AISI 52100, that all components in the rolling-element bearing, that is, the rolling elements and races, are of the same hardness. It was further assumed that this equation can be extended to other bearing steels.

In the 1960s, it was assumed by the bearing industry that materials which had higher amounts of alloying elements would have a better hardness retention at elevated temperature. It was reasoned that this would also result in a higher ambient temperature hardness as well as longer bearing life. However, no systematic study was performed and published which would confirm these assumptions. Beginning in the early 1970s such a study was undertaken at NASA [23–25]. The results of this study were most revealing, completely changing those assumptions previously held.

Short-term hot hardness measurements were made for groups of through hardened specimens of AISI 52100, AISI M-1, AISI M-50, Halmo, WB-49, AISI 440C, WD-65, and Matrix II. Measurements were also made of case-hardened specimens of Super Nitalloy (5Ni-2Al), AISI 8620, CBS 600, CBS 1000, and Vasco X-2. The results for the through hardened materials and for the case hardened materials were normalized and are shown in Figs. 9 and 10, respectively. These normalized data show that regardless of the initial hardness, the hot hardness of the individual materials shows the same functional dependence. That is, the changes in hardness with increasing temperature are independent of material composition or room temperature hardness.

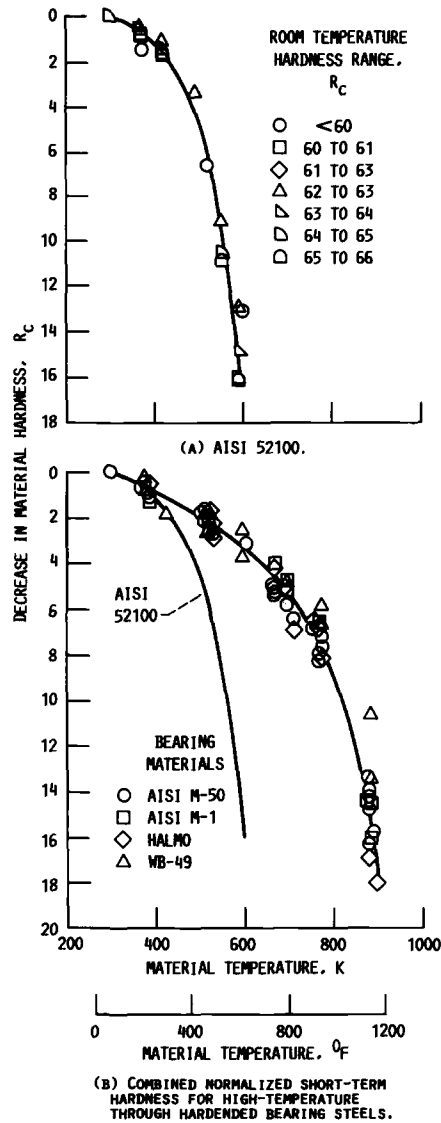


FIG. 9—Summary of short-term hot hardness through hardened bearing steel.

The data of Figs. 9 and 10, when plotted on log-log coordinates [23], can be represented by a straight line having the form

$$(Rc)_T = (Rc)_{RT} - \alpha \Delta T^\beta \quad (2)$$

where

- $(Rc)_T$ = Rockwell C hardness at operating temperature,
- $(Rc)_{RT}$ = Rockwell C hardness at room temperature,
- ΔT = change in temperature, $T_T - T_{RT}$, K, ($^{\circ}\text{F}$),
- T_T = operating temperature, K ($^{\circ}\text{F}$),
- α = temperature proportionality factor, $\text{K}^{-\beta}$ ($^{\circ}\text{F}^{-\beta}$), and
- β = exponent.

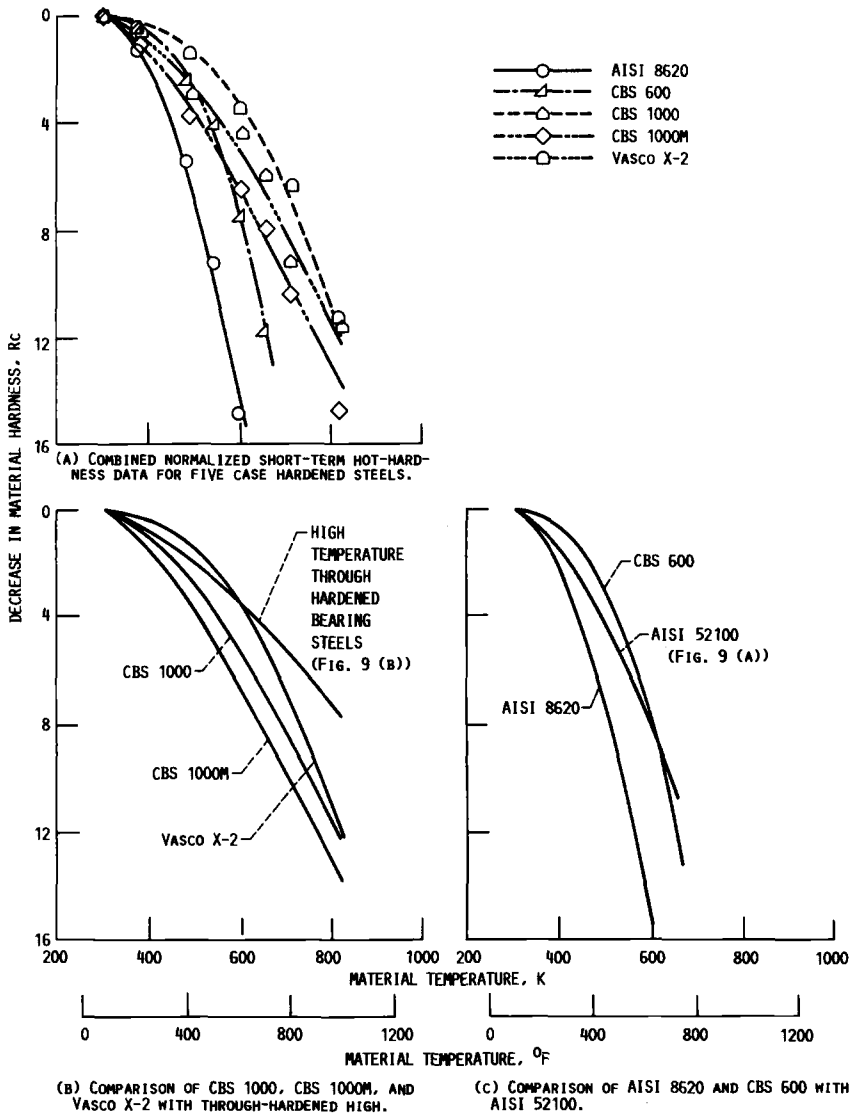


FIG. 10—Summary of case hardened steel short-term hot-hardness data and comparison with through-hardness high-speed tool steels and AISI 52100.

Values for α and β for various materials are given in Table 2. From Eq 1, let L_1 be the bearing life calculated according to Ref 26. Hence

$$L_1 = \left(\frac{C}{P} \right)^n \quad (3)$$

where C is the basic dynamic load rating of the bearing and P is the equivalent load [26]. Assume, based upon experience, that the basic load rating is based upon a material hardness

of Rockwell C60. As a result, Eq 1 can be written

$$L_2 = e^{m[(Rc)_T - 60]} \left(\frac{C}{P} \right)^n \quad (4)$$

Combining Eq 2 and Eq 4

$$L_2 = e^{m\{[(Rc)_{RT} - 60] - \alpha(T_T - T_{RT})^\beta\}} \left(\frac{C}{P} \right)^n \quad (5)$$

where $T_{RT} = 294$ K (70°F).

From the above equation, the life of a rolling-element system can be determined as a function of room temperature hardness and operating temperature for a particular steel.

Long-term hot-hardness studies were performed with five vacuum-melted steels tempered to various room-temperature hardnesses: AISI 52100 and the tool steels AISI M-1, AISI M-50, Halmo, and WB-49 [27]. Hardness measurements were taken at both room temperature and the soak temperature at regular intervals until 1000 h of soak time were accumulated. AISI 52100 was tested at temperatures to 478 K (400°F), and the other bearing steels were tested at temperatures to 700 K (800°F). With the exception of the AISI 52100, none of the steels tempered (permanently lost hardness) during soaking. The AISI 52100 steel that was initially hardened to Rockwell C 62.5 or 64.5 lost hardness during the first 500 h of the 1000-h soak tests at temperatures greater than 394 K (250°F), but it maintained its hardness during the final 500 h of soaking. Similarly, AISI 52100 that was initially hardened to Rockwell C 60.5 lost hardness during the first 500 h of the 1000-h soaking at temperatures greater than 422 K

TABLE 2—Temperature proportionality factors α and exponents β for bearing steels
 $[(Rc)_T = (Rc)_{RT} - \alpha \Delta T^\beta]$.

Material	Temperature Range		α		β	
	K	$^\circ\text{F}$	K	$^\circ\text{F}$	K	$^\circ\text{F}$
AISI 8620	294 to 589	70 to 600	73×10^{-5}	26×10^{-5}	1.7	1.7
CBS 600	294 to 589	70 to 600	0.75×10^{-5}	0.18×10^{-5}	2.4	2.4
Vasco X-2	294 to 811	70 to 1000	1.4×10^{-5}	0.38×10^{-5}	2.2	2.2
CBS 1000	294 to 811	70 to 1000	93×10^{-5}	38×10^{-5}	1.5	1.5
CBS 1000M	294 to 811	70 to 1000	340×10^{-5}	160×10^{-5}	1.3	1.3
Super	294 to 769	70 to 620	1.3×10^{-5}	0.33×10^{-5}	2.3	2.3
Nitralloy						
AISI 52100	294 to 533	70 to 500	92×10^{-5}	34×10^{-5}	1.6	1.6
AISI M-50	294 to 811	70 to 1000	133×10^{-5}	54×10^{-5}	1.4	1.4
AISI M-1	294 to 811	70 to 1000	133×10^{-5}	54×10^{-5}	1.4	1.4
AISI M-2	294 to 811	70 to 1000	133×10^{-5}	54×10^{-5}	1.4	1.4
AISI M-10	294 to 811	70 to 1000	133×10^{-5}	54×10^{-5}	1.4	1.4
AISI M-42	294 to 811	70 to 1000	133×10^{-5}	54×10^{-5}	1.4	1.4
AISI T-1	294 to 811	70 to 1000	133×10^{-5}	54×10^{-5}	1.4	1.4
(18-4-1)						
Halmo	294 to 811	70 to 1000	133×10^{-5}	54×10^{-5}	1.4	1.4
WB-49	294 to 811	70 to 1000	133×10^{-5}	54×10^{-5}	1.4	1.4
WD-65	294 to 811	70 to 1000	133×10^{-5}	54×10^{-5}	1.4	1.4
Matrix II	294 to 811	70 to 1000	133×10^{-5}	54×10^{-5}	1.4	1.4
AISI 440C	294 to 811	70 to 1000	133×10^{-5}	54×10^{-5}	1.4	1.4
AMS 5749	294 to 811	70 to 1000	133×10^{-5}	54×10^{-5}	1.4	1.4
M-50NiL	294 to 811	70 to 1000	133×10^{-5}	54×10^{-5}	1.4	1.4

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TABLE 3—Hardness of AISI 52100 steel initially hardened to three hardness values (*Rc* 60.5, *Rc* 62.5, and *Rc* 64.5) after exposure to indicated soak temperatures for 100 and 500 h.

Soak Temperature		Time at Soak Temperature, h			
		Room-Temperature Hardness (<i>Rc</i> at 297 K (75°F))		Hot Hardness (<i>Rc</i> at Soak Temperature)	
		100	500 to 1000	100	500 to 1000
K	°F				
366	200	^a	^a	^a	^a
394	250	63.2 ^b	61.5 ^b	62.1 ^b	61.2 ^b
422	300	61.0	60.0	60.0	58.8
450 to 478	350 to 400	60.0	58.7	58.0	56.5

^aNo change from initial values.

^bMaximum value attainable; if initial hardness is below this value, time at this temperature will have no effect.

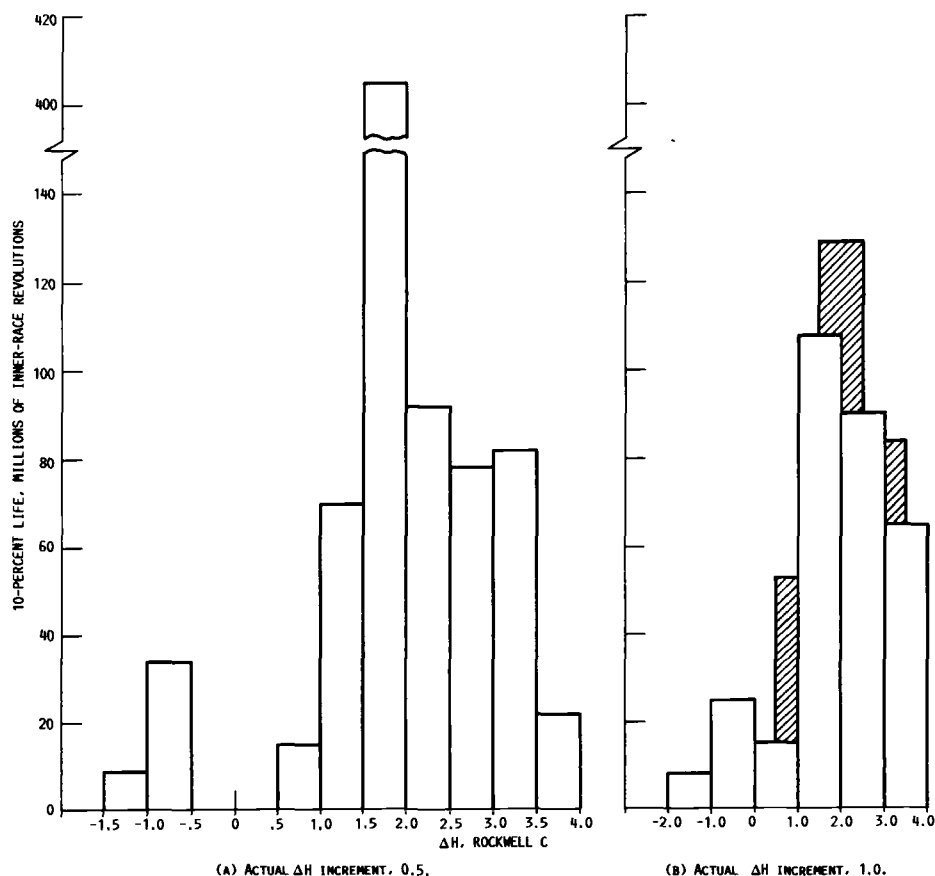


FIG. 11—Ten-percent life as function of ΔH (difference in Rockwell C hardness balls and races) for AISI 52100 207-size deep-groove ball bearings.

(300°F), but it maintained its hardness during the final 500 h of soaking. The results of these tests for AISI 52100 are summarized on Table 3. Like and similar results may be expected for other lower tempering temperature-bearing steels.

The effect of component hardness combinations on the fatigue life of AISI 52100 rolling elements subjected to repeated stresses applied in rolling contact was studied in the NASA five-ball fatigue tester [7,28]. Groups of upper test balls (analogous to the inner race of a bearing) with nominal Rockwell C hardnesses of 60, 63, and 65 were run against lower test balls (analogous to the balls of a bearing) of nominal Rockwell C hardnesses of 60, 62, 63, 65, and 66. These results indicated that, for a specific upper test ball (race) hardness, the rolling-element fatigue life and load-carrying capacity of the test system increased with increasing lower test ball hardness to an intermediate hardness value where a peak life was attained. The peak life-hardness combination occurred for each of the three lots of upper test balls (races) in which the hardness of the lower test balls was approximately 1 to 2 points Rockwell C greater than that of the upper test ball (race). According to these results, for AISI 52100 steel, a maximum bearing fatigue life should occur when the balls of the bearing are 1 to 2 points harder than the races.

Rolling-element fatigue tests were then performed on AISI 52100 207-size deep-groove ball bearings with inner and outer races from the same heat of air-melt material tempered to nominal Rockwell C hardnesses of 63 and balls from a second heat of air-melt material tempered to nominal Rockwell C hardnesses of 60, 63, 65, and 66 [28,29]. Subsequent to testing, the bearings were regrouped according to their actual values of ΔH for Rockwell C hardness increments of 0.5 and 1.0, where ΔH is the difference between the actual hardness of the rolling elements in the bearing and the actual hardness of the inner race. The results of these tests are shown in Fig. 11. Other bearing data taken from Ref 30 were reanalyzed and plotted in Fig. 12. Both sets of bearing data exhibited a maximum life at a ΔH of approximately 1 to 2 points Rockwell C. These results correlated with those obtained with the five-ball fatigue tester.

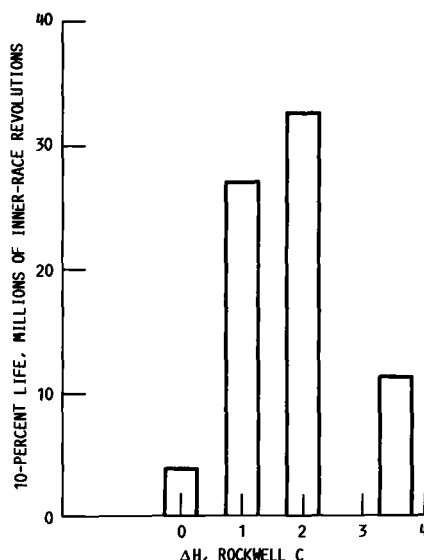


FIG. 12—Ten-percent life as function of ΔH (difference in Rockwell C hardness between balls and races) for AISI 52100 207-S-size deep-groove ball bearings.

The research of Refs 7 and 28 was repeated by Zaretsky for CVM AISI M-50 steel. The results of this research are unpublished. In general, the AISI M-50 results produced increasing life with increasing hardness and ΔH to values of ΔH equals zero. At values beyond zero, life remained unchanged in contrast to the results of the AISI 52100 material where life peaked at ΔH values of 1 to 2 points Rockwell C and then began to decrease. Based upon these tests, it may be reasonably concluded that, for elevated temperature operation, care should be taken to match the hardnesses of the rolling elements and the raceways while maintaining the highest material hardness for all components.

Carbide Effects

From the data of Refs 31 to 33 summarized in Table 4, it was speculated that an interrelation existed among median residual carbide size, number of residual carbide particles per unit area, and the percent area of residual carbides and rolling-element fatigue life. Residual carbides are those carbides that do not go completely into solution during austenitizing and that are a function of the alloying elements and heat treatment. This is opposed to the hardening carbide precipitates, which precipitate upon aging at the tempering temperature. The carbides referred to in the following will be the residual carbides.

If the carbides in a material are the nucleation site of an incipient fatigue failure, then the probability of survival S for a lot of specimens or a single specimen can be expressed as a function of three variables: percent area of carbides a ; median carbide size (length) m ; and total number of carbides per unit area n [34,35]. Based upon a statistical analysis a carbide factor C' can be derived where

$$C' = \frac{L_3}{L_2} \quad (6)$$

In Refs 31 to 33 AISI 52100 is the material which produced the highest life values. Therefore, the values for the carbide factor C' can be normalized with respect to the AISI 52100 material. Thus

$$C' = \frac{1}{K_1 \left(\frac{m}{0.26} + \frac{718}{n} + \frac{a}{9.54} + \frac{K_4}{K_1} \text{Co} \right)} \quad (7)$$

From experimental data, K_1 and K_4 were empirically determined to be $\frac{1}{3}$ and $\frac{4}{3}$, respectively [34,35], and Co is the weight percent of cobalt in the material.

Table 4 summarizes the carbide factors for the data of Refs 31 to 33. In Fig. 13, the relative material L_{10} lives are plotted against the carbide factor C' . There appears to be a reasonable correlation between the carbide factor C' and relative life. For the individual groups of data and the combined groups the confidence number was found to be 0.87 and 0.83, respectively. This means that the carbide factor C' can give a reasonable prediction of relative life under identical conditions of material hardness and lubrication mode. Thus, the carbide parameter seems to transcend such variables as heat treatment, chemical composition, and hardening mechanism to predict the lives of individual lots. The carbide parameter may be also applicable to conventional fatigue (that is, bending, rotating-bending, etc.).

Research reported in Ref 36 for AMS 5749 material further substantiated the negative effect of large carbide size and banded carbide distribution on rolling-element fatigue life. Endurance tests were run with ten 46-mm bore split-inner ring angular-contact ball bearings made from VIM-VAR AMS 5749 steel. The inner and outer rings were machined from bars. No forging

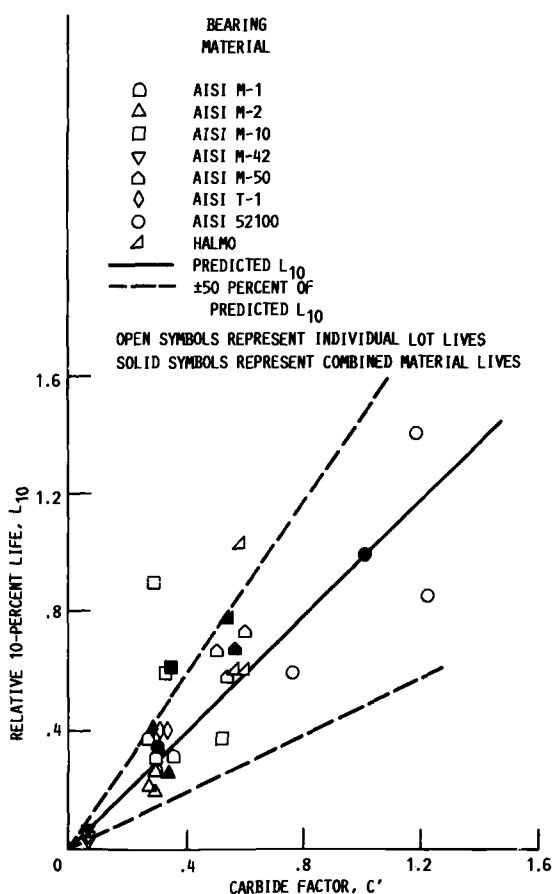


FIG. 13—Individual and combined 10% life for eight bearing materials as a function of carbide factor C' .

process was used. Test conditions included a thrust load of 4890 N (1100 lb) at a speed of 42 000 rpm and an oil in temperature of 366 K (200°F). The resultant 10% fatigue life of the bearings was 32 h as compared with a predicted life of 1900 h, a reduction in life of 98%.

Metallurgical examination of the raceways revealed large-size carbides and severe banding of the carbides in the raceways. Also, the carbide stringers or grain flow were oriented at an angle to the contact surfaces, since the bearings were run with a thrust load and the balls run at an angle to the bearing axis. The early failures in these bearings were attributed to the undesirable carbide structure and, in particular, to the large carbide stringers that were at or very near the rolling surface [36]. Hence, while previous testing [18] had shown AMS 5749 steel to have rolling-element fatigue lives at least equivalent to VIM-VAR AISI M-50, the presence or orientation of the carbides severely reduced the bearing life [36].

There appear to be two distinct criteria in the selection of a through-hardened rolling-element bearing steel. First, rolling-element fatigue life is a function of material hardness. Second, life is a function of carbide size, area, and number as represented by the carbide factor. These two criteria can be combined to both determine and evaluate without extensive testing the fatigue life of a bearing material or groups of materials. Combining Eqs 5, 6, and 7,

TABLE 4—Fatigue results with groups of 12.7-mm ($\frac{1}{2}$ in.) diameter balls run in five-ball fatigue testers. Maximum Hertz stress, 5.5×10^{-9} N/m² (800 000 psi); contact angle 30 deg; shaft speed, 10 300 rpm; race temperature, 534 K (150°F).

Material	Heat Treatment Lot	Average Hardness, R_c	Retained Austenite, Volume % ^e	Austenitic Grain Size ^b	Cleanliness Rating ^c		Life, Millions of Upper Ball Stress Cycles		Slope	Failure Index ^e	Carbide Factor, C''
					Class ^d	Type	L_{10}	L_{50}			
52100	A	62.5	4.90	13	B1	heavy	18.5	114	1.04	22 out of 29	1.22
	B	62.0	4.10	13	D1	thin	30.1	130	1.29	22 out of 29	1.17
	C	62.5	.80	13	D1	thin	12.9	84	1.00	19 out of 25	0.76
	combined	...	...	...	...	...	21.2	109	1.15	63 out of 83	1.00
Halmo	A	60.8	0.60	8	D2	heavy	22.0	74	1.56	25 out of 30	0.58
	B	60.8	1.00	8	D1	heavy	12.9	57	1.26	28 out of 30	0.64
	C	61.1	1.70	8	D2	heavy	12.9	68	1.14	26 out of 30	0.69
	combined	...	...	...	...	...	16.4	66	1.35	79 out of 90	0.63
M-50	A	62.6	1.90	10	B1	heavy	15.3	35	2.29	29 out of 29	0.60
	B	62.2	2.90	9	D2	heavy	12.3	36	1.73	24 out of 30	0.54
	C	62.3	1.50	10	D1	heavy	14.2	48	1.55	26 out of 29	0.50
	combined	...	...	...	...	...	14.4	39	1.89	79 out of 88	0.56
M-10	A	62.2	1.10	9	D3	heavy	19.4	65	1.56	26 out of 30	0.36
	B	62.0	2.40	6	D2	heavy	8.3	46	1.11	30 out of 30	0.39
	C	61.8	1.60	6	D1	thin	13.1	42	1.62	29 out of 30	0.35
	combined	...	...	...	...	...	13.2	50	1.40	85 out of 90	0.37

T-1	A	61.4	7.30	11	BI	heavy	6.6	50	0.92	26 out of 30	0.26
	B	61.4	5.20	9	D1	thin	8.4	59	.97	26 out of 30	0.33
	C	61.0	9.50	10	D1	heavy	8.5	74	.87	23 out of 30	0.30
	combined	...	...	...	...	...	8.6	59	.98	75 out of 90	0.29
M-1	A	63.3	2.90	10	B2	heavy	8.2	43	1.13	29 out of 30	0.32
	B	63.4	3.30	9	A1	heavy	5.9	37	1.02	29 out of 30	0.34
	C	63.5	1.00	8	A2	heavy	6.7	33	1.18	29 out of 29	0.38
	combined	...	...	...	...	...	7.6	38	1.18	87 out of 89	0.33
M-2	A	63.4	1.70	6	BI	heavy	5.8	35	1.05	28 out of 30	0.36
	B	63.4	2.40	10	D1	thin	5.5	26	1.23	28 out of 29	0.36
	C	63.4	2.30	9	D1	heavy	4.2	31	.95	29 out of 30	0.35
	combined	...	...	...	...	...	5.7	30	1.13	85 out of 89	0.36
M-42	A	61.8	1.00	9	A1	thin	1.0	6.6	0.97	30 out of 30	0.07
	B	61.3	4.40	10	D1	heavy	1.7	8.9	1.12	27 out of 30	0.07
	C	61.3	4.90	8	D1	heavy	1.4	5.8	1.33	30 out of 30	0.07
	combined	...	...	...	...	...	1.4	7.0	1.18	87 out of 90	0.07

^aDetermined from the integrated peak intensities of (220) γ and (200) α planes.

^bASTM E 112-63.

^cASTM E45 63, Method A (table shows predominate inclusion class and type).

^dInclusion classes: A-Sulfides, B-Alumina, C-Silicates, D-Globular oxides.

^eIndicates number of failures out of total number of tests.

dropping the subscripts 2 and 3, and letting $L_A = L_3$ where L_A is the expected bearing life at a 90% probability of survival produces

$$L_A = kC'e^{m\{[(Rc)_{RT} - 60] - \alpha(T_T - T_{RT})\beta\}}\left(\frac{C}{P}\right)^n \quad (8)$$

where k is a factor which combines processing and environmental factors which can affect bearing life [35]. The values of m can reasonably be assumed to be equal to 0.1. Values for α and β can be obtained from Table 2. For ball and roller bearings, n is equal to 3 and $\frac{10}{3}$, respectively.

A method to reduce the size of the carbides in a material is called "ausforming" [37]. The ausforming process consists of an isothermal "warm working" operation performed while the material is in a metastable austenitic condition. The austenite is subsequently transformed in either the lower bainite or the martensite transformation region of the time-temperature-transformation (TTT) curve. If the ausforming method is to be applied, the steel must have a sluggish transformation behavior in the temperature range where "warm working" is to take place. AISI M-50 is such a steel [37].

Initial rolling-element fatigue tests were performed with cylindrical rollers made from ausformed AISI M-50 material having 40, 70 and 80% deformation [37]. The results of these tests are shown in Fig. 14. It can be seen that the ausformed material is superior to the normally processed AISI M-50. In addition, there is a relation between the amount of deformation during ausforming and fatigue life. Tests with ausformed AISI M-50, 35-mm bore, single-row radial ball bearings [38] having 80% deformation produced fatigue lives approximately eight times conventionally forged AISI M-50 bearings.

Ausformed balls were fabricated from AISI M-50 bar material which was extruded to a cross-sectional area 20% of the original area (80% reduction in cross-sectional area) while the material was in a metastable austenitic condition. The 10% fatigue life of this group of ausformed balls was three and four times that of two groups of conventionally processed AISI M-50 balls [39].

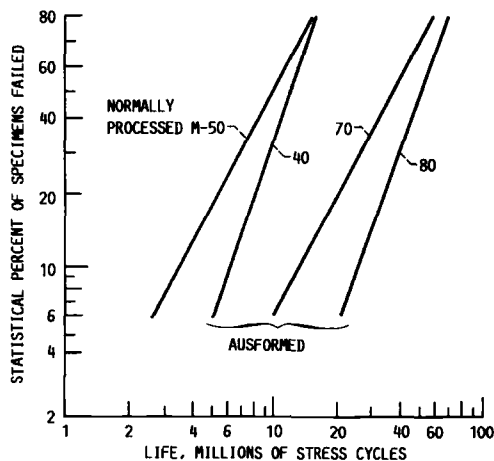


FIG. 14—Effect of ausforming on fatigue life for AISI M-50 bearing steel.

It is believed that the primary mechanism causing the improved fatigue life is "strain-induced precipitation." In essence, sufficient energy is imparted to the material during ausforming, which results in precipitation of the carbides in the material. This results in smaller and more uniformly dispersed carbides because of the availability of a larger amount of carbide nucleation sites during the ausforming process [37].

Unpublished data from NASA tests performed with VIM-VAR AISI M-50 120-mm-bore (4.8 in.) angular-contact ball bearings made by ausforming showed that forging laps were induced in the raceways of the bearing because of the relatively low forging temperature. These forging laps acted as nuclei for fatigue spalls in a rather short period of time. The conclusion reached is that ausforming can result in improved rolling-element fatigue life, but problems with forging, particularly the large, massive parts and the costs thereof, far exceeded its benefits.

Controlled Fiber

A technique which has been used to improve bearing life is the manipulation of the material fiber orientation. The races and rolling elements of most bearings are forged. Any metallic object formed by forging generally possesses a fiber-flow pattern which reflects the flow of metal during the forging operation. Carbides and nonmetallic inclusions are progressively and directionally oriented during each forming operation from the ingot to the final bearing-element shape. Although the desired microstructure is obtained by heat treatment, the carbides and inclusions generally retain their processing-induced directionality. This pattern is fibrous in appearance when the part is macro-etched, hence the term "fiber-flow lines." In addition, the entire grain pattern is preferentially oriented in the same manner.

The type of forging used to produce rolling-element bearing components will determine the fiber pattern which exists in these bearing parts. Steel balls are usually fabricated by upsetting between hemispherical dies. This fabrication technique produces a fiber-flow pattern with two diametrically opposed areas having fibers oriented approximately perpendicular to the surface, as illustrated in Fig. 15. These areas are commonly known as the poles. The excess metal

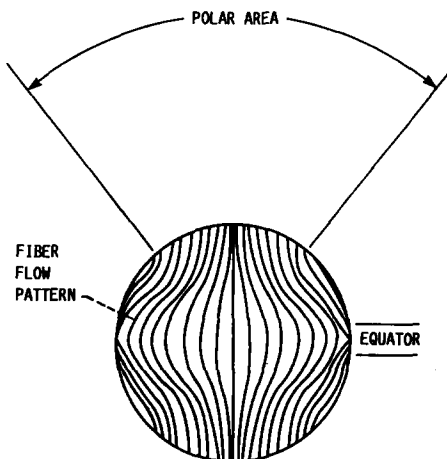


FIG. 15—Fiber orientation in a bearing ball.

extruded from between the two dies produces a thin band of perpendicularly oriented fiber when the flashing at the die parting line is removed. This line, when present, is commonly termed the "equator." Thus, a typical ball has several surface areas with varying fiber-orientation. The initial effect of fiber orientation on fatigue life was reported in Refs 5 and 6. Two lots of AISI 52100 balls were modified during manufacture to predetermine the axis of rotation and to enable one lot to be run over the poles and the other lot over the equator. The results obtained indicated a significant improvement in fatigue life when the test track passed over areas other than the poles.

Fatigue data obtained with ten different ball materials were also examined with respect to the location of specific spalls on the ball test specimens. Each of the ball specimens was destructively etched after the test to indicate the location of the failure relative to the pole areas. These data showed that a small increase in failure density occurs at the equator where the thin band of perpendicular fiber exists and that a very significant increase in failure density occurred in the polar areas or regions where fiber is essentially perpendicular to the test surface.

Attempts were made to control fiber flow in balls [40], but these were either unsuccessful or did not produce any significant improvement in fatigue life.

While the feasibility of controlling fiber flow in balls is questionable, the opposite appears to be true in bearing races. Additional research [10] with controlled fiber flow was performed by machining race cylinders from a billet of AISI T-1 steel at various angles to the direction of forging, as illustrated in Fig. 16. Three cylinders (races) were machined with axes parallel to, at 45 deg to, and perpendicular to the direction of fiber flow. The first cylinder (race) had fiber flow parallel to the test surface, the second had fiber orientation ranging continuously from parallel to 45 deg to the test surface, and the third had fiber orientation ranging continuously from parallel to perpendicular to the test surface. The 0-deg cylinder (race) had the best life while the 0 to 45-deg and the 0 to 90-deg cylinder (race) indicated that five of the six lowest lived failures occurred in the 81 to 90-deg zone. The 10% life for perpendicular fiber was found to be about 1.25 million stress cycles as compared with 4.7 million stress cycles for parallel fiber.

Experiments confirming the results of Refs 5 and 6 were reported in Ref 41. Bearings were manufactured by two methods [41], as shown in Figs. 17 and 18. One method produced races with conventional fiber orientation (Fig. 17). The other method incorporated forging techniques which produced parallel fiber flow (Fig. 18). Fatigue results from these two methods show at least a ten-fold increase in life of the bearings having side grain or fiber flow parallel to the race as opposed to bearings made with the end grain races or fiber nearly perpendicular to the running track. It was also determined that the same heats of steel which gave poor life when forged with end grain in the race performed very creditably when forged with side grain in the race. The life of parts seemed less sensitive to steel quality variations when forged with the fiber flow parallel to the race.

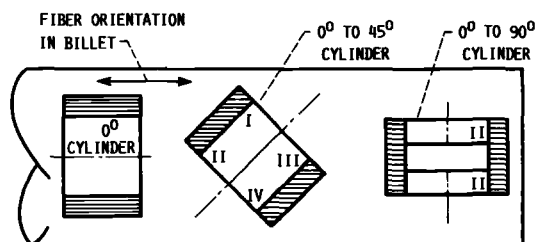


FIG. 16—T-1 tool steel cylinder orientation in billet stock.

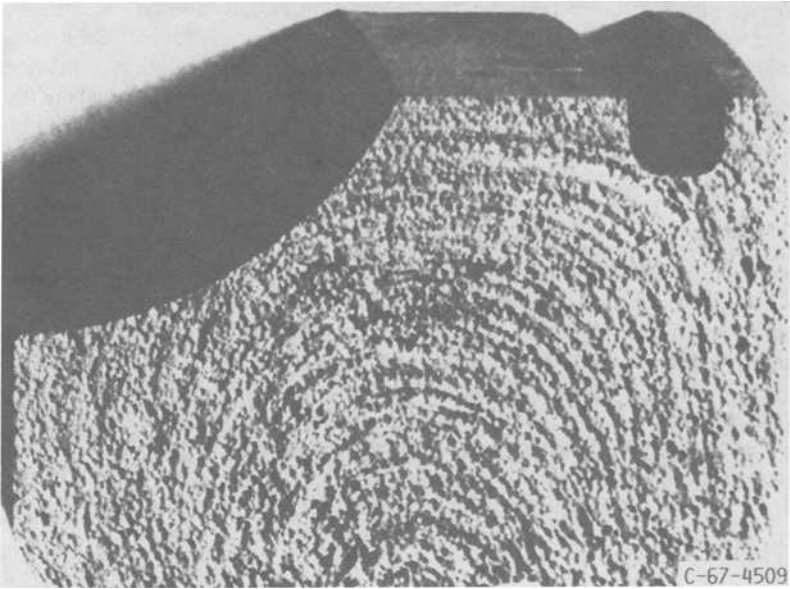


FIG. 17—*Bearing race showing fiber flow primarily perpendicular to raceway.*

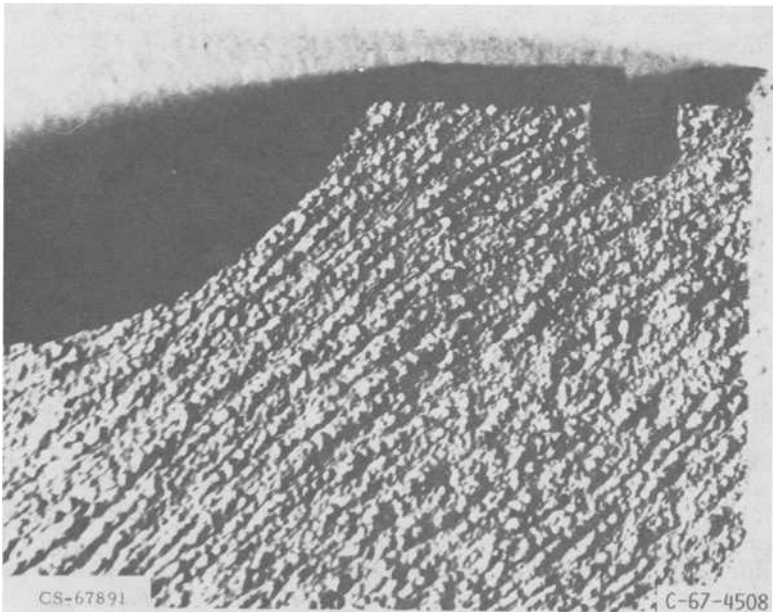


FIG. 18—*Forged bearing race showing fiber flow parallel to raceway.*

If bearings with side grain (parallel fiber flow) were radially loaded, no difference in life would be expected between these bearings and those conventionally forged. The reason for these results is that the conventionally forged bearing will generally produce parallel fiber flow in the portion of the race groove which is subject to the radial load. However, where the material is laden with large carbides, at or near the surface, the effect of fiber orientation or orientation of the carbide bands will have an insignificant effect on rolling-element fatigue life [36].

Residual Stresses

An analysis presented in Ref 42 indicates that a compressive residual stress that exists at the depth of the maximum shearing stress can decrease the maximum shearing stress. A similar analysis for superimposed stresses was subsequently reported in Refs 43 and 44. Although observations [45] would indicate that the maximum orthogonal shearing stress is the critical stress in the initiation of fatigue cracks, there also is evidence [10,11,46,47] that the maximum shearing stress is the most significant stress in the fatigue process. Thus, if the maximum shearing stress for a given Hertz stress could be decreased by compressive residual stresses, rolling-element fatigue life could be increased [42].

Residual stress can either increase or decrease the maximum shearing stress according to the equation

$$(\tau_{\max})_r = -\tau_{\max} - \frac{1}{2} (\pm S_r) \quad (9)$$

where

$\tau_{\max}$ = maximum shearing stress,
 $(\tau_{\max})_r$ = maximum shearing stress modified by the residual stress, and
 S_r = residual stress, the positive or negative sign indicating a tensile or compressive residual stress, respectively [42].

Accordingly, a compressive residual stress would reduce the maximum shearing stress and increase fatigue life according to the inverse relation of life and stress to the 9th power where

$$L \propto \left[\frac{1}{(\tau_{\max})_r} \right]^9 \quad (10)$$

Compressive residual stresses induced beneath the surface of ball-bearing race grooves were found beneficial to rolling-element fatigue life [48,49]. Ball bearing lives were increased by a factor of 2 when metallurgically induced ("prenitrided") compressive residual stress was present in the inner rings (group B, Fig. 19) [48]. Compressive residual stresses induced by unidentified "mechanical processing" operations were also found to be beneficial to the fatigue life of ball bearings [49].

Koistinen [50] reported a method of producing compressive residual stresses in the surface of AISI 52100 steel by austenitizing in an atmosphere containing ammonia. Stickels and Janotik [51] also induced compressive residual stress in the surfaces of AISI 52100 steel rolling-element specimens by austenitizing them in a carburizing atmosphere, even though the austenitizing temperature was below that needed to dissolve all primary carbides. The depth of the compressed residual surface stress was 0.3 mm (0.012 in.) having a maximum compressive residual stress of 0.6 N/m² (87 ksi). The carburized case (surface layer) contained a larger volume fraction of primary carbides and more retained austenite and was slightly harder than

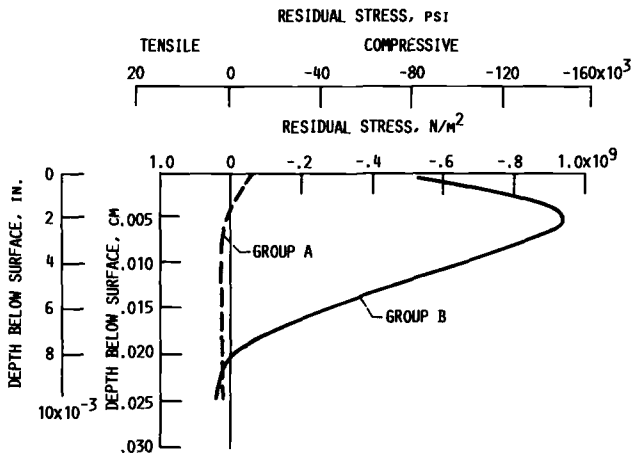


FIG. 19—Tangential residual stress patterns for two groups of 40-mm (1.6 in.) bore-size bearing races.

the core. Rolling-element fatigue tests in the RC fatigue tester at a maximum Hertz stress of $5.03 \times 10^9 \text{ N/m}^2$ (729 ksi) having a depth to the maximum shearing stress of 0.015 mm (0.006 in.) resulted in the AISI 52100 which was treated in the carburizing atmosphere having a life approximately $1\frac{1}{2}$ times greater than the untreated AISI 52100 [52]. This increase in life can be attributed to the induced residual stresses in the region of resolved maximum shearing stresses in spite of the increased presence of carbides at or near the surface. From Ref 52 it is apparent that the distribution of the induced compressive residual stresses is a fraction of the carbon potential. For ball and roller bearings the zone of maximum resolved shearing stresses due to Hertzian loading occur at a depth from 0.10 to 0.25 mm (0.004 to 0.010 in.) below the surface. With a carbon potential of 0.9 weight % carbon in iron, effective compressive residual stresses were available to a depth of 0.3 mm (0.012 in.). This depth was sufficient to have a beneficial effect. For the carbon potentials less than 0.7 weight % carbon high, tensile residual stresses were present at depths to 0.2 mm (0.008 in.). Hence, without taking due care, it is also possible to reduce the fatigue life potential using the carburizing process to induce residual stresses.

Compressive residual stresses can be induced as a result of the cyclic concentrated contact in rolling-element bearings [46,53–56]. These stresses depend on plastic deformation of the microstructure. They tend to reach a maximum at a depth of several mils beneath the rolling surface corresponding to approximately either the depth of the maximum shearing stress [43,55,56] or the depth of the maximum orthogonal shearing stress [46]. The magnitude of the residual stress tangential to the surface in the direction of rolling is dependent on both the applied load and the number of load cycles. Research reported in Ref 53 indicated a threshold load below which significant residual stresses are not induced except for very long running times.

Changes in microstructure (phase transformations) have been reported to occur in the same areas as the maximum induced residual stress [53,55]. Under some conditions of very high contact stresses, no microstructural alteration was apparent where significant residual stresses were induced in a few cycles [53]. The correlation of induced residual stress with these microstructural alterations is not clear. In Ref 56, they are proposed to be independent phenomena. However, it is probable that some or all of the induced residual stress may be attributed to the transformation of retained austenite to martensite.

It was reported in Ref 46 that there were significant effects of residual stress induced during fatigue testing on rolling-element fatigue life. In this work, maximum compressive residual stresses were induced and maximum fatigue life resulted when the ball hardness was 1 to 2 points Rockwell C greater than the race hardness ($\Delta H = 1$ to 2 points Rockwell C). An identical effect is reported in Ref 56.

It was hypothesized that such a beneficial compressive residual stress could be induced by prestressing a ball-bearing inner race, for example, by running the bearing at a load greater than the threshold load [53] for a prescribed number of cycles. The bearing, when subsequently run under more nominal service loads, would then be expected to experience a longer fatigue life.

In order to confirm the above hypothesis, Parker and Zaretsky made residual stress measurements on several 207-size deep-groove ball bearings that were run for different time periods to determine a prestress cycle suitable for inducing significant compressive residual stresses in the inner-race-ball groove [57]. The results of this prestressing are shown in Fig. 20. Compressive residual stresses in excess of $0.69 \times 10^9 \text{ N/m}^2$ (100 ksi) were included in the region of the maximum shearing stress in the bearing inner race run for 25 h at a maximum Hertz stress of $3.3 \times 10^9 \text{ N/m}^2$ (480 ksi) and a shaft speed of 2750 rpm. Twenty-seven bearings

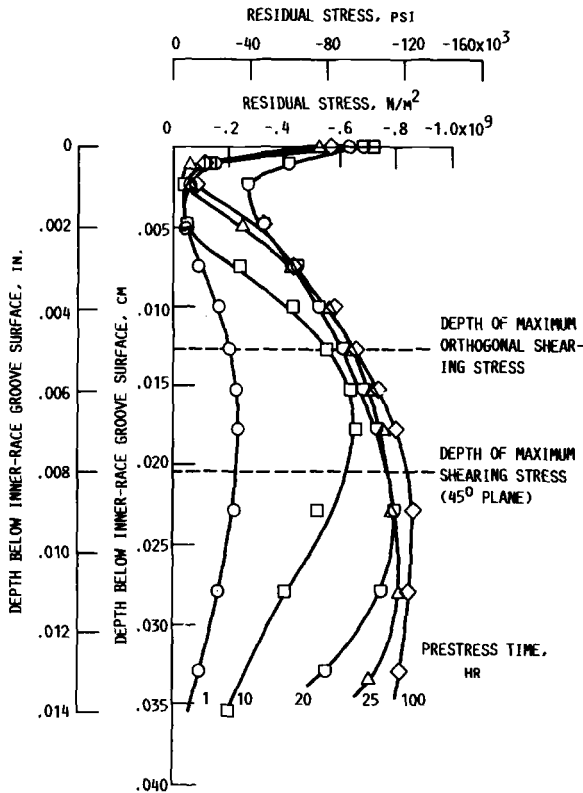


FIG. 20—Tangential residual compressive stress as function of depth below center of inner-race groove for various pre-stress time cycles for 207-size deep-groove ball bearing. Radial load, 13 800 N (3100 lb); maximum Hertz stress, $3.3 \times 10^9 \text{ N/m}^2$ (480 ksi); shaft speed, 2750 rpm.

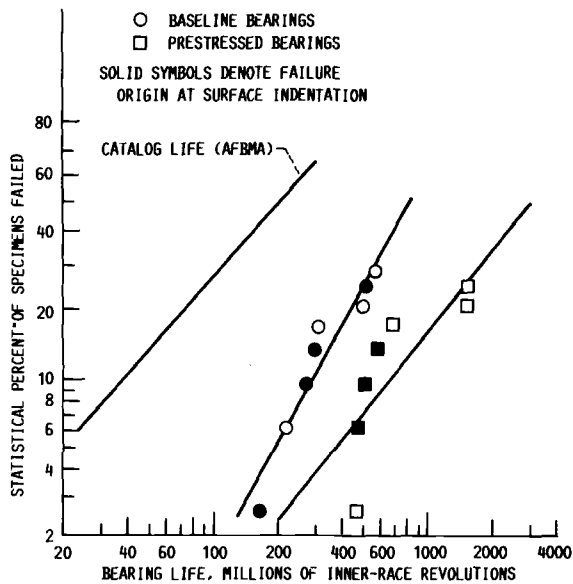


FIG. 21—Results of fatigue tests with 207-size ball bearings. Radial load 5860 N (1320 lb); maximum Hertz stress, $2.4 \times 10^9 \text{ N/m}^2$ (350 ksi); shaft speed, 2750 rpm; lubricant, refined naphthenic mineral oil.

were prestressed for 25 h at this condition and fatigue tested at a maximum Hertz stress of $2.4 \times 10^9 \text{ N/m}^2$ (350 ksi). The results of these tests are shown in Fig. 21. The results of these tests are compared with results of baseline tests without a prestress cycle at identical test conditions. The 10% fatigue life of the prestressed ball bearings was greater than twice that of the baseline bearings. In addition, the differences between the measured residual stress in the prestressed bearings and in the baseline bearings after 3000 to 4000 h of testing are small (Fig. 22) [57].

Retained Austenite

For three decades the effect of retained austenite on rolling-element fatigue has been openly discussed at technical meetings without the presentation of definitive data. In general, for a given through-hardened material, the amount of retained austenite generally increases with increasing material hardness. For case hardened materials, however, large amounts of retained austenite will be present in the case. It is well known that retained austenite will transform to martensite under Hertzian cyclic stress conditions and even at no load ambient conditions resulting in dimensional instability of the bearing component. For this reason, a low level of retained austenite is desirable for critical bearing applications. Typical maximum levels are in the range of 2 to 5%. The effect of retained austenite cannot be easily separated from the effect of material hardness. For this reason, little work has been directed toward studying the effect of retained austenite on rolling-element fatigue.

It had been speculated that increased amounts of retained austenite can result in increased life if material hardness was maintained at a relative constant level. AMS 5749 steel, because of its alloy content, tends to retain higher levels of retained austenite, but levels as low as 5% are attainable with suitable heat treatment. This characteristic allows for development of test

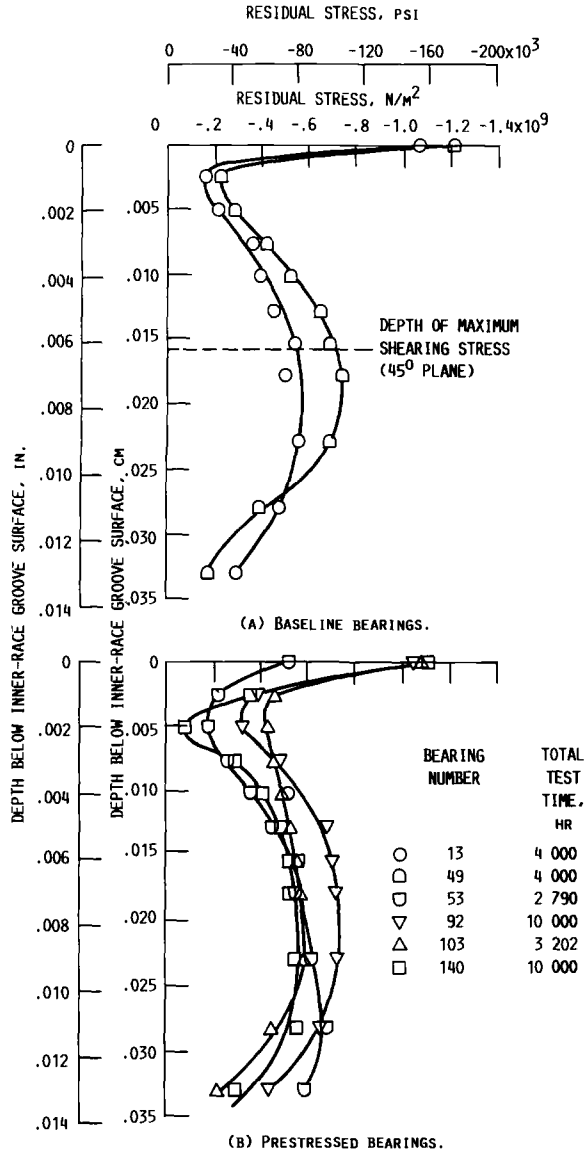


FIG. 22—Tangential residual compressive stress as function of depth below center of inner-race groove measured after 207-size ball bearings were fatigue tested. Radial load, 5860 N (1320 lb); maximum Hertz stress, 2.4×10^9 N/m² (350 ksi); shaft stress, 2750 rpm; lubricant, super-refined naphthenic mineral oil.

material with varied retained austenite while other characteristics such as hardness remain constant [58].

Rolling-element fatigue tests were conducted in the NASA five-ball fatigue tester with three lots of VIM-VAR AMS 5749 12.7-mm ($\frac{1}{2}$ in.) diameter balls at a maximum Hertz stress of 5.52×10^9 N/m² (800 ksi), a contact angle of 30 deg, and a shaft speed of 10 000 rpm at 339 K (150°F). Each lot of the AMS 5749 had a different amount of retained austenite. The results

of these tests are summarized in Fig. 23 [58]. A comparison of the 10% lives of the three lots of VIM-VAR AMS 5749 does not present a clear effect of retained austenite on fatigue life. The intermediate level (Lot B, 11.1%) gave the longest life with apparent statistical significance when compared with the highest level, where the confidence number is 97%, or greater than 2σ confidence. For the lowest level of retained austenite (Lot C), the confidence number is 87%. Thus, it is apparent that the high level of retained austenite, 14.6% (Lot A), is significantly detrimental to the 10% fatigue life. Therefore, an optimum level of retained austenite is suggested for maximum fatigue life. However, because of the limits imposed by dimensional stability, such levels are probably not practical for bearings for critical applications such as in aircraft turbine engines.

It should be noted that an unusually high Weibull slope exists in the Lot B failure distribution. The reason for this high slope along with the relatively long life of this lot is not understood; however, it is of interest to note that relatively early failures did not occur at the 11.1% retained austenite level. Thus, the 50% life data suggest a trend toward increased life with decreased retained austenite level. However, where high reliability is of importance, the main interest is in early failures, which is indicated best by the 10% life data [58].

Rolling-element fatigue tests were conducted with two rollers as specimens made from a carburized 18Cr-2Ni-4WA (0.18% carbon, 1.50% chromium, 4.25% nickel, and 1.0% tungsten) at a maximum Hertz stress of $3 \times 10^9 \text{ N/m}^2$ (435 ksi) [59]. The rollers had retained austenite of 7, 12, 40, and 50% in the roller case. Hardness measurements were only reported for the 7 and 50% retained austenite rollers. These were maximum Vickers hardness of 800 and 730, respectively. While only three or four roller pairs were tested at each austenite level, there was a significant increase in fatigue life with increasing retained austenite. However, austenite does not appear independent of hardness. The results of this work shown in Fig. 24 suggest that the deformation-induced martensite transformation results in the creation of compressive residual stresses, thus increasing life [59]. These results would also suggest that retained austenite may play more of a role in the results discussed previously relating hardness to life wherein induced compressive residual stresses are related in whole or in part to the transformation of the austenite into martensite. These transformations are also related to metallurgical changes which have been previously reported [10]. Many investigators have sought a correlation of the

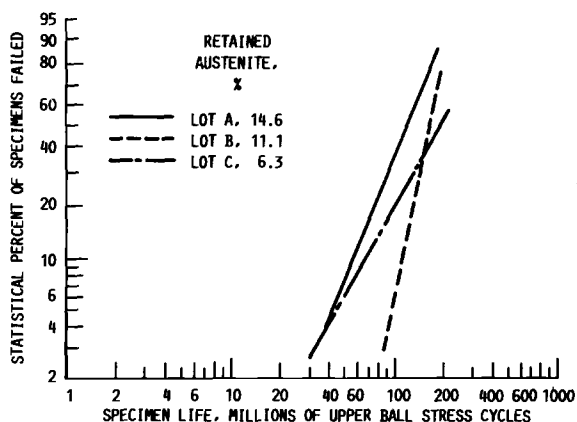


FIG. 23—Rolling-element fatigue lives of VIM-VAR AMS 5749 having various amounts of retained austenite in five-ball fatigue tester. Maximum Hertz stress, $5 \times 52 \times 10^9 \text{ N/m}^2$ (800 ksi); contact angle, 30 deg; shaft speed, 10 000 rpm; race temperature, 339 K (150°F).

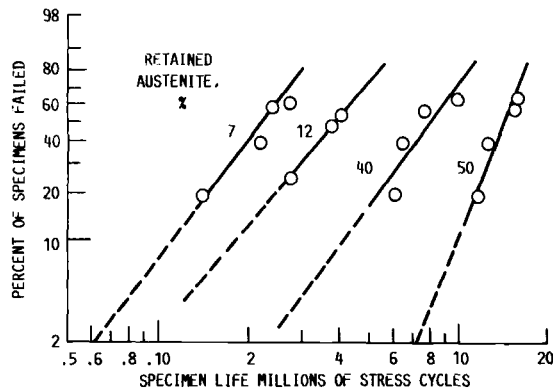


FIG. 24—Effect of retained austenite on rolling-element fatigue for two disks in contact. Maximum Hertz stress, $3 \times 10^9 \text{ N/m}^2$ (435 ksi); material, 18Cr-2Ni-4WA; speed, 2400 rpm for driver, 2400 rpm for driven; sliding speed, 1.05 m/s (41.3 in./s). (Specimens with 7% and 12% retained austenite were tested at maximum Hertz stress of $2.8 \times 10^9 \text{ N/m}^2$ (406 ksi).)

so-called “white bands” in AISI 52100 induced during Hertzian contact with time and rolling-element fatigue. The “white bands” may be indicative of the residual stresses present due to the austenite transformation.

While retained austenite appears to improve rolling-element fatigue life, it can at the same time have a deleterious effect. It was reported in Refs 60 and 61 that for tapered roller bearings made from AISI 4320 steel shallow surface distress was occurring on the raceway and roller taper surfaces. This type of distress is called peeling [61] and appears as a very shallow area, uniform in depth. Typical peeled areas in these bearings, approximately 0.008 mm (0.0003 in.) deep, are shown in Fig. 25. The peeling tended to initiate at minor surface defects such as deeper surface scratches or indentations [61].

The peeling also tended to be concentrated near the axial center of the raceways and roller with a slight bias toward the roller large end. This effect may be expected since the rollers are slightly crowned and the contact stress is somewhat higher in the center of the raceway. Profile traces across the raceways and along the roller tapers of the initial test bearings revealed that the roller crown radius had been decreased to approximately one-third of its original value. This exaggerated crown caused further stress concentration at the center of the roller and further aggravated the peeling [61].

This exaggerated crown was also observed in similar tests reported in Ref 60 and was attributed to an uneven transformation of retained austenite. The level of retained austenite in the case of the AISI 4320 bearing steel is approximately 30%. Since austenite is a relatively unstable phase, it transforms to martensite at a rate that depends on temperature and stress conditions. As it transforms, a growth of the material occurs. Stresses and temperatures are higher near the center of the roller raceway contact due to higher Hertzian loading, greater transformation, thus growth occurs there, and it becomes a self-aggravating condition. Some measurements of retained austenite on the rollers from the initial test bearings indicated that nearly all of the austenite had transformed near the center. It is believed that the growth from this transformation could account for the exaggerated crown that was measured [61]. The cup and cone raceways also experienced some crown increase but to a lesser extent than the rollers [61].

Experience has shown that test rollers made from AISI 52100 of Rockwell C hardnesses greater than 63 will have sufficient austenite transform during rolling contact to alter the

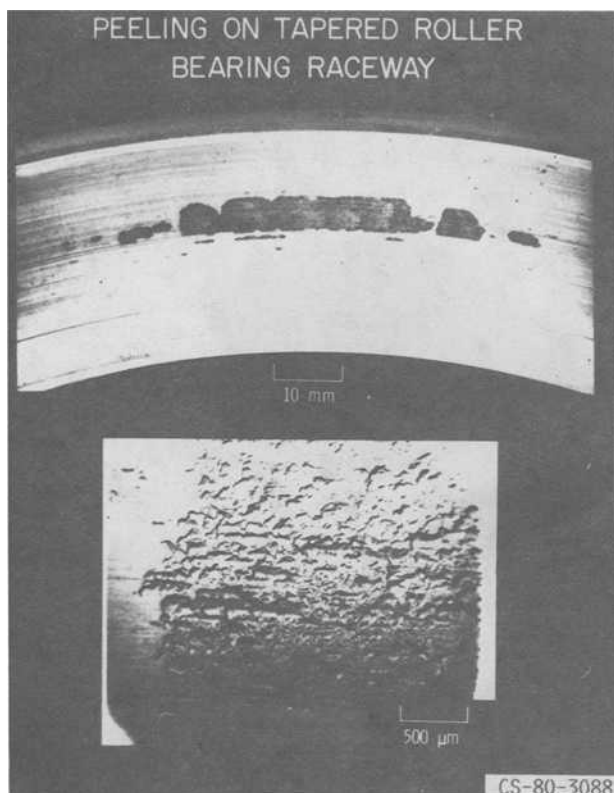


FIG. 25—Peeling failure on cup raceway surface after 569 h with standard design bearing run at 12 500 rpm.

surface waviness and cause early spalling of the surfaces. Even for unrun bearings made from AISI M-50, growth of the bore of the inner ring can occur at room temperature with time where high amounts of retained austenite were initially present to render the bearing unfit for use.

Fracture Toughness

Table 5 lists both through-hardened and case-hardened bearing materials and their chemistry. It has been standard practice in the tapered roller bearing industry to manufacture tapered roller bearings from case-hardened steels. The reason for this is quite simple: because of the higher loads which tapered roller bearings are subjected to, higher interference fits between the inner ring and shaft are required to prevent the ring from turning on the shaft. These higher interference fits induce rather high hoop (tensile) stresses in the ring which can result in fracture of the inner ring if through-hardened steels were used. In addition, high thrust loads on the flange of the tapered roller bearing can result in fracture failure of the flange if the flange were through hardened. As a result, through-hardened materials such as AISI 52100 and AISI M-50 have been rarely used for tapered roller bearing application. For temperatures under 608 K (300°F), case-hardened materials such as AISI 4320 and AISI 9310 have been

TABLE 5—Chemical composition of bearing steels.

Material	Alloying Element, % by Weight (Balance Fe)										
	C	P(max)	S(max)	Mn	Si	Cr	V	W	Mo	Co	Other
AISI 52100	1.00	0.025	0.025	0.35	0.30	1.45	...	...	...	...	...
MHT	1.03	0.025	0.025	0.35	0.35	1.50	...	...	...	...	1.36 Al
HALMO	0.56	0.003	0.008	0.36	1.12	4.84	0.53	...	5.18	...	...
AISI M-1	0.80	0.030	0.030	0.30	0.30	4.00	1.50	...	8.00	...	...
AISI M-2	0.83	0.030	0.030	0.30	0.30	3.85	1.90	...	5.00	...	...
AISI M-10	0.85	0.030	0.030	0.25	0.30	4.00	2.00	...	8.00	...	...
AISI M-42	1.10	0.012	0.007	0.15	0.17	3.77	1.15	1.66	9.51	7.99	...
AISI M-50	0.80	0.030	0.030	0.30	0.25	4.00	1.00	...	4.25	...	...
T-1 (18-4-1)	0.70	0.030	0.030	0.30	0.25	4.00	1.00	18.4	...	...	...
T-15	1.52	0.010	0.004	0.26	0.25	4.70	4.90	12.5	0.20	5.10	...
440C	1.03	0.018	0.014	0.48	0.41	17.30	0.14	...	0.50	...	...
AMS 5749	1.15	0.012	0.004	0.50	0.30	14.50	1.20	...	4.00	...	...
VASCO Matrix II	0.53	0.014	0.013	0.12	0.21	4.13	1.08	1.40	4.80	7.81	...
CRB-7	1.10	0.016	0.003	0.43	0.31	14.00	1.03	...	2.02	0.32	...
AMS 5900	1.10	0.014	0.007	0.40	0.30	14.00	1.0	...	2.0	...	0.25 Nb
AISI 9310 (C) ^a	0.10	0.006	0.001	0.54	0.28	1.18	...	...	0.11	...	...
CBS 600 (C)	0.19	0.007	0.014	0.61	1.05	1.50	...	...	0.94	...	...
CBS 1000M (C)	0.14	0.018	0.019	0.48	0.43	1.12	...	...	4.77	...	...
VASCO X-2 (C)	0.14	0.011	0.011	0.24	0.94	4.76	0.45	1.40	1.40	0.03	...
Bower 315 (C)	0.13	...	...	0.50	...	1.55	...	...	5.00	...	...
AISI 8620 (C)	0.21	...	...	0.80	...	0.50	...	...	0.20	...	...
AISI 3310 (C)	0.11	...	...	0.52	...	1.58	...	...	...	...	...
AISI 4320 (C)	0.20	...	...	0.55	...	0.50	...	...	0.25	...	...
AISI 4620 (C)	0.20	...	...	0.55	...	...	...	...	0.25	...	...
AISI 4720 (C)	0.20	...	...	0.55	...	0.45	...	...	0.20	...	...
M-50 NtL (C)	0.13	0.030	0.030	0.30	0.25	4.00	1.20	...	4.25	...	...

^a(C) = Carburizing grades.

commonly used. Within the last two decades, for higher temperature application, case carburized steels such as CBS 600, CBS 1000M, and VASCO X-2 have been used. Life tests were performed with 120.65-mm (4.75-in.) bore tapered-roller bearings made from CVM AISI 4320 and CVM CBS 1000M at 12 500 rpm, a bearing temperature of 391 K (245°F), a thrust load of 53 400 N (12 000 lb), and a radial load of 26 700 N (6000 lb) [61]. Twelve bearings of the AISI 4320 material ran to a 1100-h cutoff time without failure of any type. At these conditions the rated catalog life of this bearing design is 102 h, so that the experimental 10% life is greater than 10 times the catalog life [61].

Sixteen bearings made from CVM CBS 1000M material were run to spalling fatigue failure or to the 1100-h cutoff time. Twelve of the bearings ran to 1100 h without failure. Three bearings experienced spalls on the cup or cone raceways. One bearing was suspended at 820 h without spalling failure. Based on the three failed bearings, the life of the CBS 1000M bearings was estimated to be 600 h, or about six times the rated catalog life [61].

A comparison of the results of the two materials showed that the CVM CBS 1000M bearing life is less than the AISI 4320 bearing life. However, a quantitative estimate of the difference was not possible from these results since no fatigue failures occurred with the AISI 4320 bearings. Metallurgical analysis of the CBS 1000M bearings revealed that the cup and cone materials had coarser than desired grain structure. These bearings were made from the first heat of the material and were exposed to forging temperatures later found to be excessive. In general, prior austenite grain size should be ASTM 8 or finer, and individual grains should not exceed ASTM 5 [1].

Ten tapered roller bearings of a modified design made from VIM-VAR AISI M-50 steel were run at the same loads as the AISI 4320 and CBS 1000M bearings, but at a speed of 18 500 rpm. The rated catalog life for these bearings was 46 h. For various reasons related to rig malfunction, testing of six of these ten bearings were suspended before the cutoff time of 1100 h. Two bearings reached the 1100-h cutoff time without failure. Two bearings developed cracked cones after 188 h.

The average tangential tensile hoop stress, based on the assumption of thin rings, was estimated for the 18 500-rpm condition to be approximately $0.145 \times 10^9 \text{ N/m}^2$ (21 000 psi). This calculated stress also includes effects of the cone-shaft interference fit. This stress is at the lower end of the range where critical crack size can be readily reached in through hardened steel [61].

For ball and cylindrical roller bearings through hardened steels are commonly used. Most of these bearings are run at speeds less than 2 million DN with some bearings in main shaft turbojet engine applications running at 2.2 million DN. As engine speeds increase to achieve improved efficiency and better specific fuel consumption, bearing speed will also increase. It is expected that bearing speeds will approach 3 million DN in the next decade. At bearing speeds greater than 2.3 million DN, bearings using conventional through-hardened bearing steels such as AISI 52100, AISI M-50, or 18-4-1 (AISI T-1), will be subject to race fracture either independently or as a result of a fatigue spall. This was graphically illustrated in the test reported in Ref 8 with a 120-mm (4.8 in.) bore angular-contact ball bearing run at 3 million DN in which a defect was induced at the bearing inner race to cause spalling. The expected inner-race spalling at the induced defect occurred after 6 h and 17 min. Testing was continued until an obvious severe fragmentation fracture occurred to the inner ring, terminating the test, 7.5 min after detection of the spalling (Fig. 26). Post-test examination showed that the inner ring had fractured into eight discrete segments [8].

From these tests it became obvious that, as with tapered roller bearings, materials with improved fracture toughness, such as case carburized steels, must be used. Comparison of the fatigue lives of CVM AISI M-50, CVM AISI 9310, and CBS 600 in the RC fatigue tester at $4.8 \times 10^9 \text{ N/m}^2$ (700 ksi) is shown in Fig. 27. The AISI 9310 and CBS 600 had equivalent

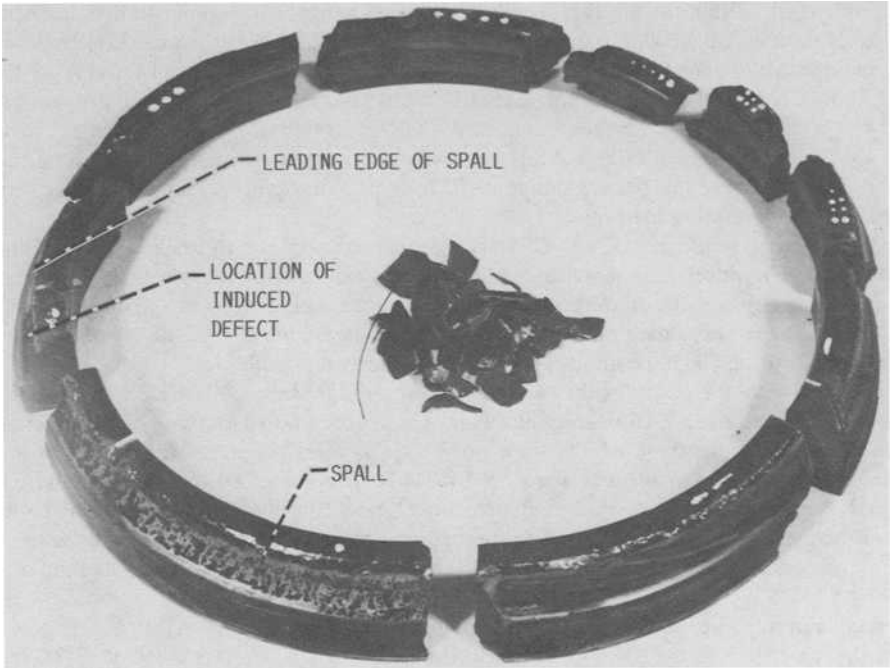


FIG. 26—Fractured bearing inner race caused by initiation of rolling-element fatigue spall.

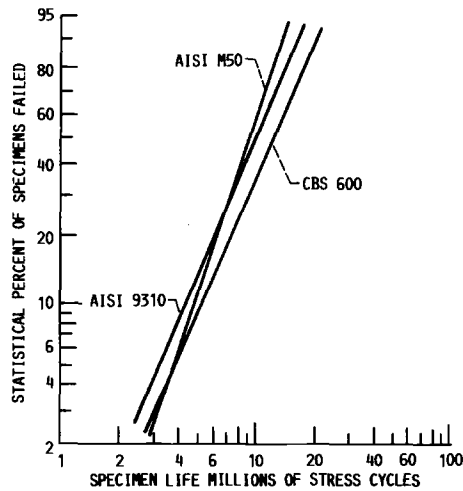


FIG. 27—Rolling-element fatigue lives of CVM CBS 600, CVM AISI M-50, and CVM AISI 9310 in RC fatigue tester. Maximum Hertz stress, $4.8 \times 10^9 \text{ N/m}^2$ (700 ksi); speed, 10 000 rpm; temperature, 483 K (74°F).

fatigue lives to the AISI M-50 [17]. However, materials such as AISI 9310 do not maintain their hot hardness, while materials such as CBS 600 require much more quality assurance and control during processing to achieve the required rolling-element fatigue life.

Fracture toughness of a material is inversely proportional to carbon content and hardness. The amount of carbon present also determines hardness. A means of improving the fracture toughness without affecting hardness is to add nickel. When present in high-chromium, low-carbon steels, nickel causes the steel to become fully austenitic at temperatures above 1359 K (1650°F), where the steel is heat treated or carburized. This also influences carbide size and distribution within the steel which affects fatigue life.

Recognizing the above, E. N. Bamberger, at General Electric Co. [9], modified the chemistry of AISI M-50 steel by decreasing the amount of carbon and increasing the amount of nickel [9]. He called this modified material M-50NiL (the "Ni" referring to increased nickel and the "L" to low carbon).

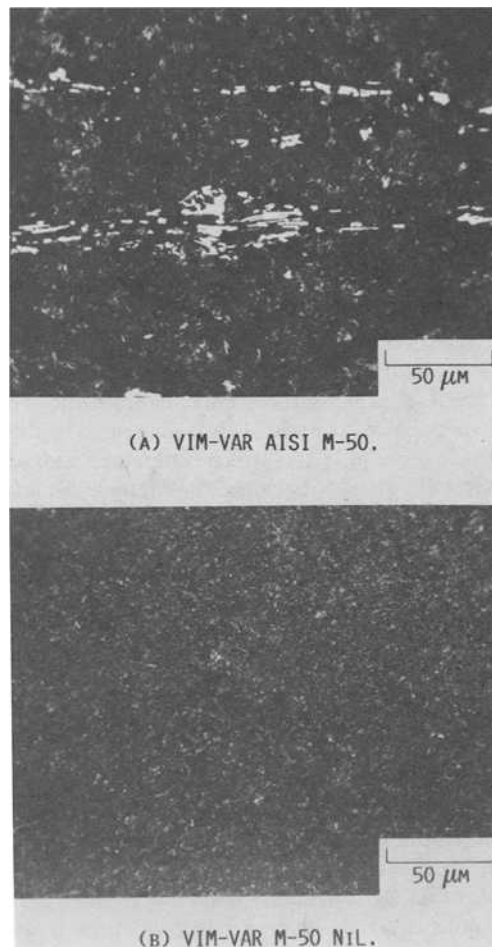


FIG. 28—Comparison of microstructure of AISI M-50 NiL. (M-50 NiL has smaller, more uniform carbide structure.)

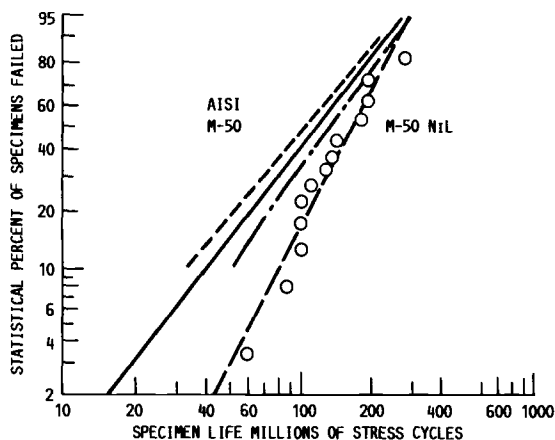


FIG. 29—Rolling-element fatigue test in RC fatigue tester of VIN-VAR AISI M-50 and carburized, VIM-VAR M-50 NiL. Maximum Hertz stress, $4.8 \times 10^9 \text{ N/m}^2$ (700 ksi); speed, 50 000 rpm; temperature, 483 K (74° F).

The M-50NiL, which is case carburized, has a core which has a fracture toughness K_{Ic} of over $60 \text{ MPa} \sqrt{\text{m}}$ (50 ksi $\sqrt{\text{in.}}$) and an average core hardness of Rockwell C 43 to 45 as compared to AISI M-50 of $29 \text{ MPa} \sqrt{\text{m}}$ (20 ksi $\sqrt{\text{in.}}$) [62]. The carbide structure of M-50NiL, shown in Fig. 28, has fine carbides (compounds of carbon and various alloying elements) dispersed evenly within the steel's microstructure as compared to the courser carbides found in the standard AISI M-50 [62].

Compressive residual stresses on the order of $0.2 \times 10^9 \text{ N/m}^2$ (30 ksi) are induced in the zone of maximum resolved shearing stresses during the carburization process [62]. These residual stresses combined with the fine carbide structure will improve the rolling-element fatigue life of the material over conventional AISI M-50. This is illustrated by the rolling-element fatigue test results of the RC fatigue tester (Fig. 29), which compares standard VIM-VAR AISI M-50 and VIM-VAR M-50NiL. The M-50NiL had a 10% fatigue life of over twice that of the standard AISI M-50 [9]. Bearing tests reported in Ref 9 show that under conditions which resulted in fracture of standard AISI M-50 inner races, bearing races made from M-50NiL and selected carburized steels did not show any signs of incipient fracture or even a tendency to fracture. This assures that bearing steel requirements for the next decade and beyond can be met by these most recent carburized materials.

Conclusion

During the last four decades, significant technical advancements in the bearing and steel industry have allowed for the design and manufacture of rolling-element bearings having lives over 100 times that which could be achieved in the 1940s and at speeds to 3 million DN. The results of these achievements can be categorized into improved material cleanliness, controlled material hardness, smaller and evenly dispersed carbide structure, induced compressive residual stresses, and improved fracture toughness. While the effect of material grain size was not discussed, it is generally accepted in the bearing industry that prior austenite grain size should be ASTM Number 8 or finer and individual grains should not exceed ASTM Number 5. It is further apparent that the chemical composition of a bearing steel does not have the affect on material hot hardness or life that it once was thought to have. What appears to be of

paramount importance is the steel melting process such as VIM-VAR and an optimized heat treating process, together with its related process controls, in order to achieve a uniform and consistent end product. The vast differences which were once thought to distinguish through-hardened materials from case carburized materials really never existed but resolves to differences in fracture toughness necessary for either highly loaded or very high-speed applications or both. Case carburized M-50NiL appears to incorporate all the desirable qualities discussed herein into a material whereby optimal life and reliability can be achieved not only for current bearing applications but for those of the next decade and beyond.

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Impact of Steel Quality on Integrated Automotive Wheel Bearing Performance

REFERENCE: Tayeh, G. S. and Woehrle, H. R., "Impact of Steel Quality on Integrated Automotive Wheel Bearing Performance," *Effect of Steel Manufacturing Processes on the Quality of Bearing Steels*, ASTM STP 987, J. J. C. Hoo, Ed., American Society for Testing and Materials, Philadelphia, 1988, pp. 44-60.

ABSTRACT: Quality and performance concerns have increased dramatically over the last ten years in the automotive industry. Demands on bearing steel internal quality as a means of improving bearing performance have reflected this trend. Meaningful measures of internal quality were vital to promote improvement, and SAM ratings of microinclusions were developed partly to serve this purpose. In addition, AMS 2301 magnetic particle inspection procedures were applied to measure macroinclusions. Cooperative work with numerous steelmakers led to significant cleanliness improvement as measured by these tools. Ongoing laboratory testing of thousands of bearings over the past seven years shows continuous bearing performance improvement attributable to many factors, among which internal steel quality plays a significant role. The internal laboratory bearing test data are also confirmed by long-term field performance as determined by a mathematical model, taking into account total bearings in the field, accumulated service, service sales, and service inventories. This model is in turn confirmed by failure mode analysis of failed components retrieved from the field.

KEY WORDS: inclusions, bearing alloys, ratings, fatigue (materials), antifriction bearings, bearing reliability, wheel bearings, bearing field performance

Changes in attitudes and philosophies about quality and performance in the automotive industry over the last ten years have been enormous. It was only natural that automotive components manufacturers also would feel the pressure to improve the quality and performance of their products. The bearings are critical components. As automobile bearing suppliers, we reacted to this challenge and significantly improved the quality of our product.

Material properties are one of the most important factors determining performance of mechanical components in general and bearings in particular. This paper will discuss work directed toward improving bearing steel internal quality and consequent improvements obtained in bearing performance.

Background

The product around which this development work has centered is the New Departure-Hyatt "wheel spindle bearing" which was introduced in 1978 on General Motors' new front-wheel-drive cars. These cars represented the first wave of a major change in the basic vehicle design of one of the world's major car-producing companies at a time of intense competition from both domestic and off-shore producers. Simultaneously, it was a time of changing customer demands and needs resulting from escalating energy shortages and costs as well as customer

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perceptions of quality advantages available from off-shore producers. In this environment, quality demands were (and remain) enormous—there was no truly “acceptable” failure rate, and the only genuinely acceptable performance level was “the life of the vehicle.” For this product, the changing ideas regarding quality became apparent very quickly.

The “wheel spindle bearing” comes in two basic variations, illustrated in Figs. 1 and 2. Both variations provide a sealed and lubricated double-row angular contact ball bearing integral with the hub, which provides for mounting on the vehicle. In addition, the spindle provides a flange for direct mounting of brake components and the wheel. For the driven-wheel version, shown in Fig. 1, the spindle is hollow and provides a bore spline through which power is supplied to the wheel. The nondriven-wheel version incorporates a solid spindle, as shown in Fig. 2.

Two bearing steels are used in the design. The hub, with its two integral raceways, and the spindle, with one integral raceway, are made from a Society of Automotive Engineers/American Iron and Steel Institute (SAE/AISI) 1070 grade modified with additional manganese. The single separable inner ring and the balls utilize SAE/AISI 52100. Chemistries for both grades

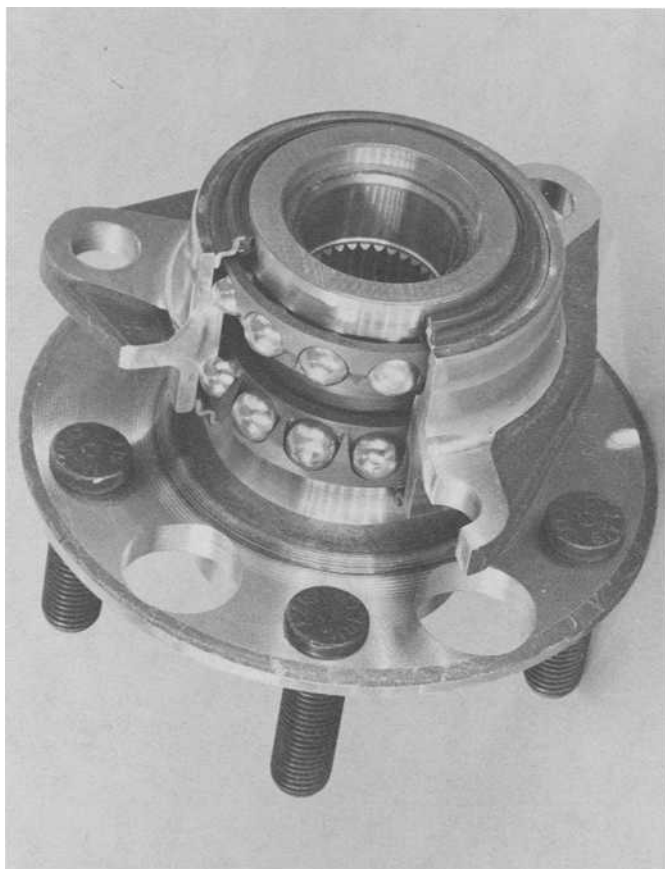


FIG. 1—Typical “wheel spindle bearing” for driven wheel application.

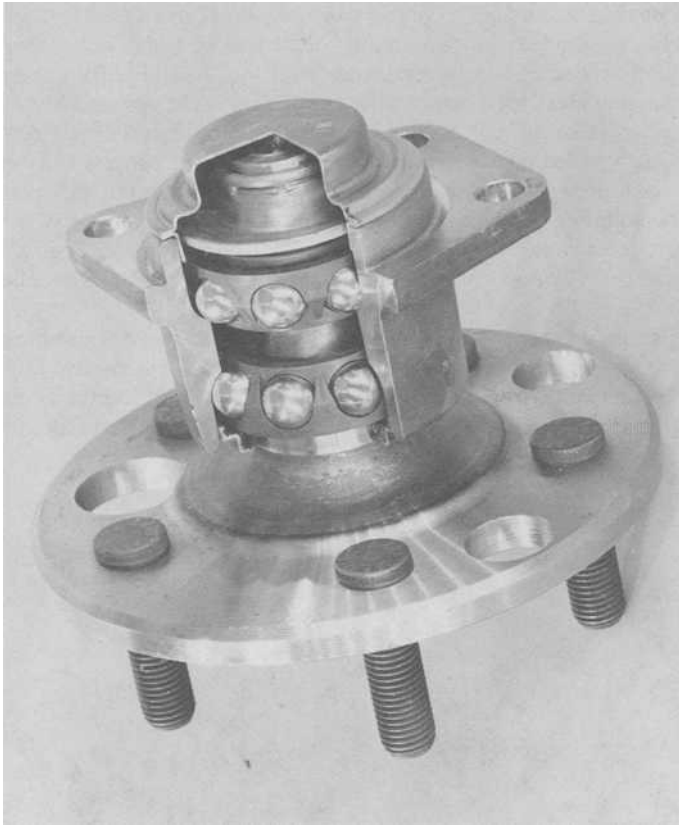


FIG. 2—Typical “wheel spindle bearing” for nondriven wheel application.

are listed in Table 1. Quality requirements for both steels were initially defined by the then-current issue of the ASTM Specification for High-Carbon Ball and Roller Bearing Steel (A 295-84). In spite of best efforts to the contrary, some fatigue failures were encountered, most of which were attributable to nonmetallic inclusions in the steel. The minimization and, hopefully, elimination of these failures became our challenge and the challenge of our steel suppliers.

TABLE 1—Composition ranges of steels evaluated, %.

Grade	1070 Mod.	52100
Carbon	0.65 to 0.75	0.98 to 1.10
Manganese	0.80 to 1.10	0.25 to 0.45
Phosphorus (max)	0.025	0.025
Sulfur (max)	0.025	0.025
Silicon	0.15 to 0.35	0.15 to 0.35
Chromium	0.20 max	1.30 to 1.60
Nickel (max)	0.25	0.25
Copper (max)	0.35	0.35
Molybdenum (max)	0.06	0.10

Internal Quality Measurement

Evaluation of Microinclusions

Fundamental to any improvement process is the ability to measure, in a significant and quantitative manner, the properties of concern. Bearing steel inclusion ratings historically have most often been directed toward microinclusions—principally manganese sulfides and the oxides resulting from deoxidation treatments applied to the liquid steel. Certainly in the United States, and to a significant degree in many other parts of the world, the most common rating method has been that defined as Method A of the ASTM Practice for Determining the Inclusion Content of Steel (E 45-85), which in turn borrows heavily from the “Jernkontoret Method” developed many years ago in Sweden [1,2]. This method most commonly applied using six 160 mm² (0.25 in.²) specimens representing the top and bottom, respectively, of three designated ingots are examined at a magnification of $\times 100$ and compared with reference plates available from ASTM. Inclusions present are characterized as one of four morphological types—sulfides, aluminates, silicates, and globular oxides—with each type further divided into two series based on inclusion width. A rating is arrived at by recording the “worst” field of each type and series for each specimen.

As bearing failures were studied, along with cleanliness rating samples representing related heats of steel, one factor became apparent; there was often an evident visually perceptible difference in the apparent cleanliness of the steel which was not reflected in the so-called J-K microcleanliness ratings. The reason for this was fairly obvious; the “worst-field” ratings did not take occurrence frequency into account in any way. Therefore, a specimen with a single inclusion of Severity X received a rating identical with a specimen containing numerous fields of equal or nearly equal severity. While the traditional worst-field method of rating had proven adequate for many years, the improved quality levels of recent steels had outstripped the ability of the technique to detect differences.

The need for a more discerning approach to inclusion rating of bearing steels was evident. Any new method would have to take into account inclusion frequency as well as severity. The method would have to reflect metallographically observable indifferences in cleanliness. If industrial acceptance of the method were to be gained, the time required for execution of an evaluation should not be exorbitantly longer than the time required by current methods. Finally, the method must allow reproducibility of results by different operators in different laboratories.

A method which is believed to meet the above objectives was developed at New Departure-Hyatt in 1981 and put into commercial service during that same year. This method was called the SAM rating system and is detailed in Supplementary Requirement S2 of ASTM A 295-84, which is included in the Appendix. The method concentrates on measurement of the frequency of ratable fields of Type B aluminates and Type D globular oxides as defined by ASTM E 45-85 because these are the inclusion types associated with rolling contact fatigue failures. Type A manganese sulfides are disregarded because they are not considered significant in bearing fatigue [3,4]. Type C silicates are also ignored because they occur infrequently in bearing steels.

In order to keep the time required for rating down to a reasonable level, as well as preventing the approach from becoming unduly tedious, we limited frequency determination to the larger inclusions. For Type B aluminates, thin series fields rated 1.5 or higher and heavy series fields rated 1.0 or higher are counted. For Type D globular oxides, only the heavy series ratable fields are counted.

A numerical SAM rating for Type B aluminates is obtained by summing the field ratings obtained and normalizing the results by dividing by the area evaluated. Similarly, a SAM rating for Type D globular oxides is obtained by adding “units” assigned to each rated field and again normalizing for the area evaluated. Details and examples of the calculations appear in the Appendix.

The development of the SAM rating method provided the tools necessary to measure adequately microcleanliness with respect to inclusion types known to be related to bearing performance and, therefore, provided the basis for progressive work with steel suppliers directed toward improving this aspect of internal quality. There is, however, another class of inclusions that contributes to bearing failure which is not adequately addressed by microcleanliness evaluation.

Evaluation of Macroinclusions

For purposes of this paper, "macroinclusions" are taken to be the relatively rare and frequently large inclusions either of exogenous nature or, more commonly, the result of reoxidation of deoxidizing or grain refining additions (principally aluminum) or both during steel teeming.

It is remarkable that internal quality assessment of bearing steels in the United States has been so heavily preoccupied with microcleanliness that larger, although rarer, macroinclusions have been relatively ignored, at least for purposes of quantitative evaluation. Larger inclusions obviously pose a greater threat to bearing performance than do the smaller ones. It may be speculated that the emphasis on microinclusions is the result of the much higher frequency and uniformity of distribution of this inclusion type compared to macroinclusions. With emphasis on higher reliability, reflected by discussion of B-1 and B-0.1 bearing lives of 160 000 km (100 000 miles) and more, the small number of early failures related to macroinclusions took on an increasingly significant role in determining acceptability of the bearing, and therefore the bearing steel, for the application.

Macroinclusions have been assessed in aircraft steels for many years using, among other methods, the Society of Automotive Engineers' Aerospace Material Specification, "Aircraft Quality Steel Cleanliness—Magnetic Particle Inspection Procedure" (AMS 2301G). An essentially equivalent method is also described in ASTM E 45-85 under the heading "Magnetic Particle Method." Because of the familiarity of steel suppliers with AMS 2301, they elected to adopt this method to evaluate macroinclusions in bearing steels for the wheel spindle application. The "quarter-section" sampling method was selected because it provided assessment of inclusions occurring both near the surface and near the center of steel sections. The sampling choice is significant because macroinclusions appear not to occur at the section mid-radius sampling location prescribed by ASTM E 45-85 for microinclusion rating. At least, it is the experience of the authors that a macroinclusion has virtually never been observed in a mid-radius sample.

With the addition of the AMS 2301 macroinclusion rating system to the SAM rating method for microinclusions, a "tool kit" for quantitative evaluation of bearing steel internal quality was reasonably complete. The job then became one of applying these tools to attain the quality improvements necessary to permit increased bearing life.

Internal Quality Improvement

The approach taken to pursue internal quality improvements was simply one of gathering and organizing copious data employing the described tools and then using these data as the basis for a constructive and cooperative effort with the steel suppliers involved. The data were summarized on a roughly annual basis and presentations made to as many suppliers as possible, preferably at their mill locations in order to enhance the number of mill personnel exposed to the information as well as to include mill personnel close to the actual steelmaking operations. Coded competitive information for the relevant grades was included. Presentations were followed by working discussions aimed at establishing programs targeting specific

internal quality problem areas as appropriate for the given mill. The effects of these programs were tracked and reviewed at the following year's meeting. Where circumstances dictated, review meetings were more frequently.

Microinclusions

The progress made in microcleanliness over the past four and one-half years as measured by SAM ratings is illustrated by Figs. 3 and 4 for 1070 Model steel and by Figs. 5 and 6 for 52100 steel. Figures 3 and 5 show the progress made with Type B aluminates, while Figs. 4 and 6 similarly illustrate progress for Type D globular oxides. Several aspects of these Figures and the data they represent are worth noting as follows:

1. Each year's data reflect the aggregate performance of from five to eight steelmakers. The 1070 Modified data largely represents North American producers with some European participation. The 52100 data includes sources representing Europe, Japan, and North America. The 52100 data in particular, therefore, represent a reasonable estimate of world capability with respect to the characteristics measured.
2. The number of heats of steel studied is fairly large: 2242 heats of 1070 Modified and 933 heats of 52100 over the four and one-half year period.
3. All the SAM rating data reported was generated by the authors' organization so that there is no bias in the data due to laboratory-to-laboratory variation in techniques.

Over the period of this effort, the average SAM rating of Type B aluminates has been reduced by 55% for 1070 Modified and by 73% for 52100. Average Type D globular oxides SAM ratings have been similarly curtailed by 61% for 1070 Modified and 60% for 52100.

Steelmaking process consistency also shows improvement over the study period as evidenced by reductions of 55% and 59% in the SAM rating standard deviations for Type B inclusions for 1070 Modified and 52100, respectively. Standard deviations for SAM ratings for Type D

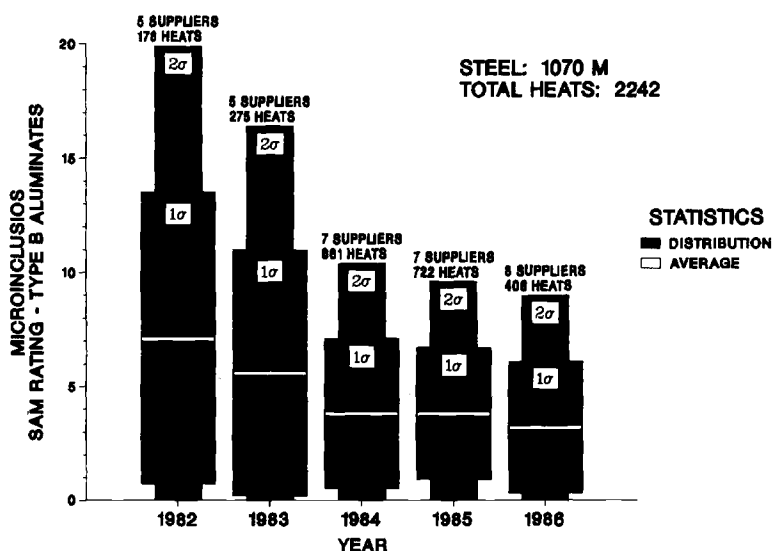


FIG. 3—History of SAM ratings for Type B aluminates in 1070 Model steel.

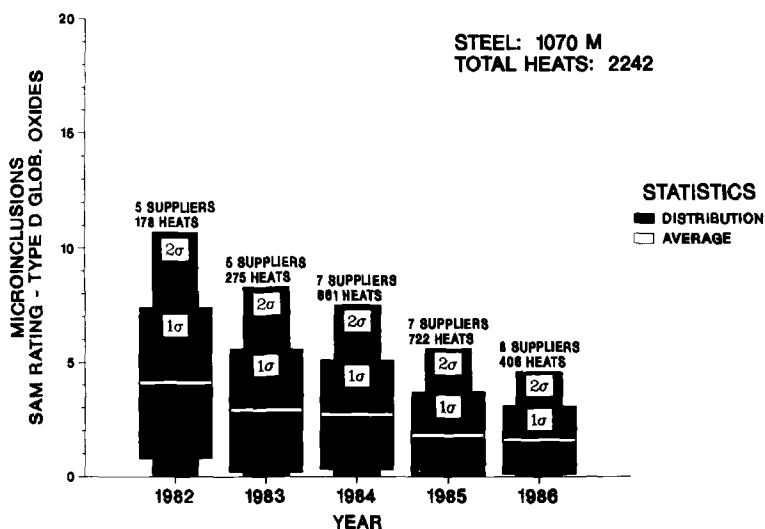


FIG. 4—History of SAM ratings for Type D globular oxides in 1070 Model steel.

inclusions show comparable reductions of 55% and 51% for 1070 Modified and 52100, respectively.

The details of steelmaking practice changes contributing to the improvement in microcleanliness are beyond the scope of this paper and are as varied as the specific circumstances of the approximately one dozen steelmakers involved. In general terms, however, efforts to improve microcleanliness centered about furnace slag controls, temperature controls throughout the process, and ladle refining and stirring practices including vacuum treatments.

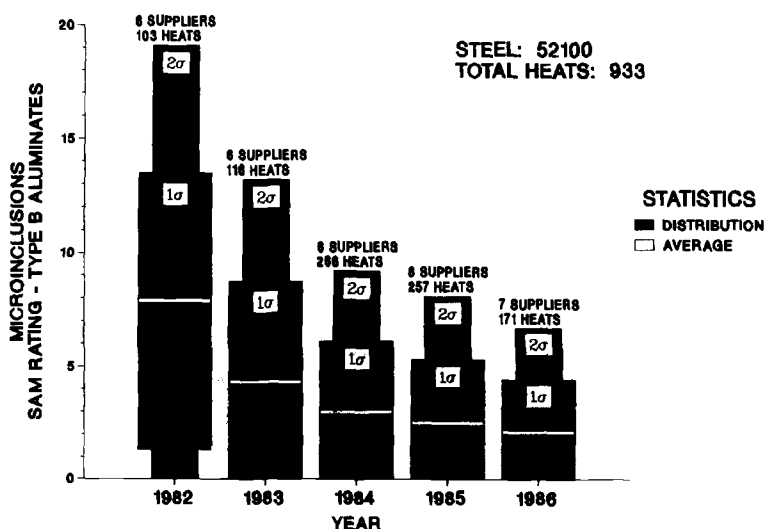


FIG. 5—History of SAM ratings for Type B aluminates in 52100 steel.

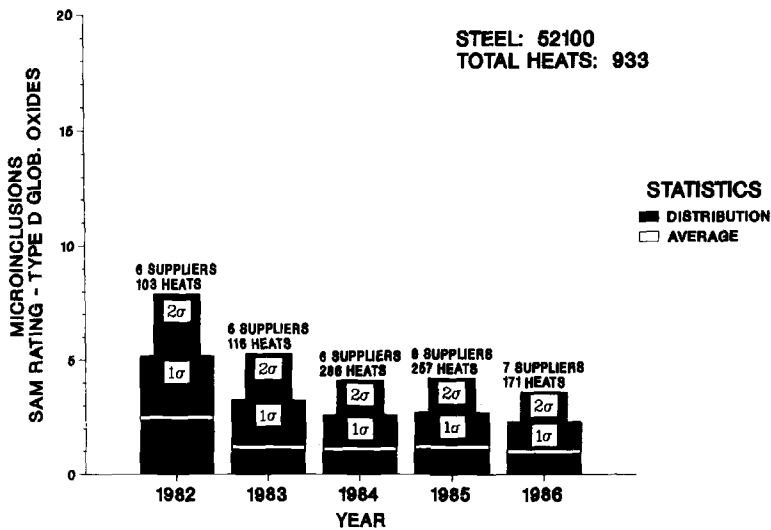


FIG. 6—History of SAM ratings for Type D globular oxides in 52100 steel.

Macroinclusions

Improvement in macroinclusion levels as measured by AMS 2301 quarter-section sample testing is summarized in Figs. 7 and 8 for 1070 Modified and 52100, respectively. The inclusion levels are not expressed in the usual “frequency and severity” parameters defined by AMS 2301, but are instead converted to total indication length per unit area examined and expressed in millimeters of indication length per square meter. It is the opinion of the authors

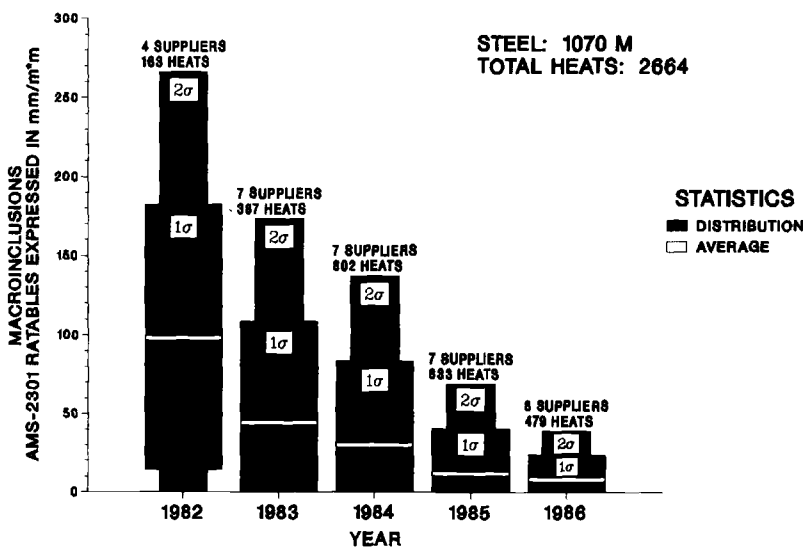


FIG. 7—History of AMS 2301 macroinclusion ratings in 1070 modified steel.

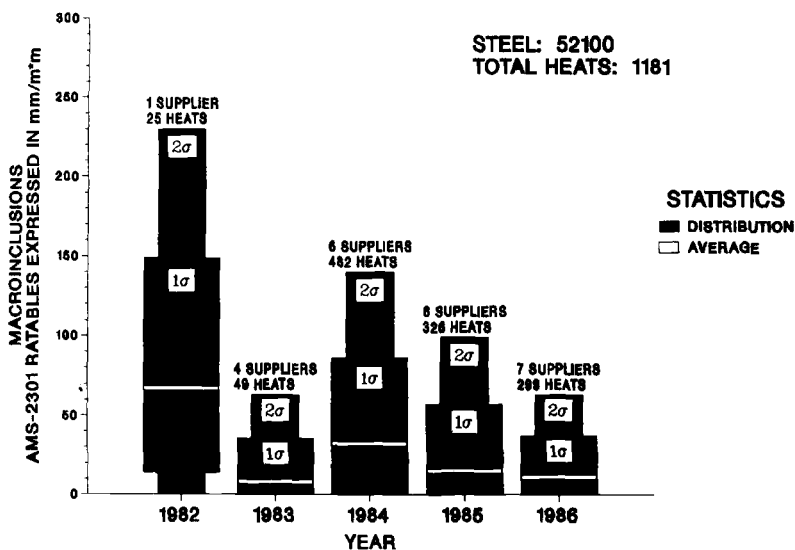


FIG. 8—History of AMS 2301 macroinclusion ratings in 52100 steel.

that the resulting single number gives a more graphic representation of data. Again, several aspects of the figures should be noted:

1. The steelmaker mix is, of course, the same as previously indicated with respect to the microinclusion data.
2. The number of heats studied is again large: 2664 heats of 1070 Modified and 1181 heats of 52100.
3. Differing from the microinclusion data, all the AMS 2301 results were generated by the supplying mills so that there is some opportunity for laboratory-to-laboratory bias in the data. Limited auditing was done at the authors' facility to assure proper sample preparation and reasonable duplication of inspection results.

From the start of 1982 through the first half of 1986 the average macroinclusion level of 1070 Modified had been reduced by 91%. Steelmaking process consistency for 1070 Modified with respect to macroinclusions had also improved, as indicated by a 81% reduction in macroinclusion level standard deviation.

With respect to 52100 macroinclusions, the number of heats studied during 1982 and 1983 is probably too small to be significant. However, even if we limit comparisons to the period extending from the start of 1984 through the first half of 1986, an improvement in macroinclusion levels of 65% is observed. Similarly, process consistency improvement is reflected by a 52% reduction in standard deviation over this same period.

Probably the most significant steelmaking practice change responsible for macroinclusion level improvement, certainly for suppliers utilizing top-poured ingots, was the adoption of argon gas shrouding of the teeming stream to minimize the opportunity for reoxidation during the pouring of ingots. Other process changes centered around pouring floor practices and hot-top designs. As is the case for the microinclusion related practices, the specific changes varied from supplier to supplier and details are beyond the scope of this paper.

Internal quality improvement, of course, is not an objective sought for its own sake. The motivation to improve internal quality stemmed from the need to improve bearing performance. The extent to which the improved cleanliness succeeded in enhancing integral wheel bearing performance is the subject of the next section of this paper.

Bearing Performance Improvement

The wheel bearing system reliability and performance standards introduced in the 1970s were different from the old standards in three significant ways: the acceptable failure rates are an order of magnitude smaller; failure rates are specified at different and higher mileage intervals, specifically, 19 200, 38 400, 80 000, and 160 000 km (12 000, 24 000, 50 000 and 100 000 miles); and failure rates and mileage intervals are based on actual customer use. Meeting these standards required changes in many areas of the wheel bearing business. Improvement in the steel internal cleanliness is one of the most significant of these areas. It is not the intent of this paper to discuss all the changes which contributed to making higher reliability standards a reality. However, it is important to note the changes which helped to reduce, and in some cases eliminate, bearing failures due to factors other than "normal" metal fatigue. Design, analysis, and testing are some of the important areas where these changes took place.

Design

At the time of the introduction of the integral wheel bearing (1979), General Motors produced mostly rear-wheel-drive cars. The front wheels of these cars were each running on two single-row tapered roller bearings (Fig. 9). Mounting, greasing, sealing, and the adjustment of the bearings' endplay were all done at the car production line. Maintenance at a dealership or service station at specific mileage intervals was also required to clean, regrease, and adjust the bearings. These operations are all critical to the safe and reliable performance of the bearings. The design of the integral wheel bearing eliminated some field problems this conventional design was experiencing. This was done by lowering the number of critical parts, reducing the number of critical surfaces which need close tolerances, eliminating the uncertainties due to press fits, providing positive and tighter control of bearing preload, and eliminating the need to service the bearing.

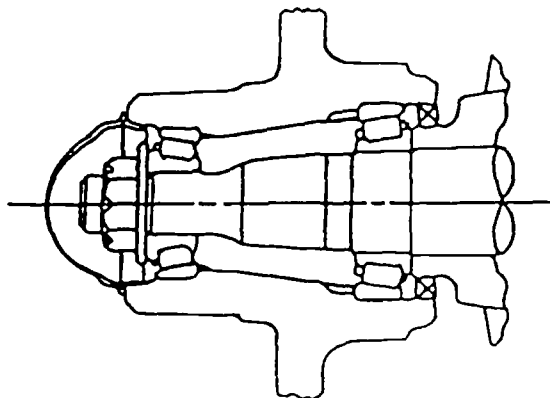


FIG. 9—Conventional taper wheel bearing arrangement.

Analysis

The new functions the integral design added to the wheel bearing system, especially the structural requirement of the rings, and the new specifications for reliability made it necessary that the analysis be based on data measured in the car under real driving conditions and that these conditions be quantitatively relatable to the overall customer use. At about the same time New Departure-Hyatt started the development of the integral wheel bearing, General Motors Proving Grounds started a program to collect data on suspension loads and distance traveled at given load levels. These data were obtained from a group of customer-owned cars during normal daily driving. The customers were randomly selected on a nationwide basis. This program lasted for 24 months and accumulated about 264 000 data kilometres (165 000 miles). Since that program, similar programs have been conducted to update this base and to make sure that it reflects the latest in car suspensions and driving habits.

This new data base made it possible to relate specific vehicle durability schedules to actual customer use. This is accomplished by instrumenting a test vehicle and collecting the suspension data during one representative cycle of the schedule. The data are then analyzed and compared on a statistical basis to the customer data base, and a ranking of the severity of that specific durability schedule is computed.

A vehicle durability schedule that produces significant wheel bearing forces is selected as the integral wheel bearing durability schedule. The schedule simulates the suspension forces of a much greater than average severity customer driving 160 000 km (100 000 miles).

A matrix of bearing loads, speeds, and the associated number of revolutions is computed from the suspension force matrix. This new matrix is used as the base for the structural analysis and bearing fatigue analysis.

Fatigue analysis of the structure is implemented using the finite-element method technique in addition to experimental work. The analysis also includes evaluation of any deformations which could alter the bearing load distribution and thus influence bearing fatigue life.

A new algorithm was written to compute bearing fatigue life. It uses the large matrix of bearing conditions as input and calculates bearing life based on the loads and dynamics of each ball in the bearing.

Testing

A large number of complete vehicles is continuously running through durability tests at General Motors Proving Grounds. The number of test vehicles is adequate to assure safety and reliability of the product. However, reliability testing of components needs to be run with larger samples to gain the statistical confidence needed, and in many instances this testing has to precede the vehicle tests. Machine testing in the laboratories is the way to meet these requirements. Most wheel-bearing test machines are designed to run accelerated testing under one or two load and speed conditions; they were not designed to run a test with a large variation of loads and speeds. This led to the development and building of new bearing test machines capable of running a large matrix of loads and speeds. These new machines are used mainly to test new integral wheel bearing designs and project field performance to a degree acceptable to the car divisions. The largest number of integral wheel bearings was tested and is still being tested on the conventional test machines, which simulate two driving conditions: straightaway and cornering, both at the same speed. Endurance test data presented in this paper are the result of these tests. Loads are kept below a level which produces 2800 MPa (400 000 psi) mean Hertzian stress.

Figure 10 shows the results of six years of endurance testing. This testing is made up of 2653 actual production integral wheel bearings and is part of a continuous program to monitor product quality. Improvement in performance is most noticeable from 1983 production to 1985. Figure 11 shows the results of engineering development tests starting in 1977. The

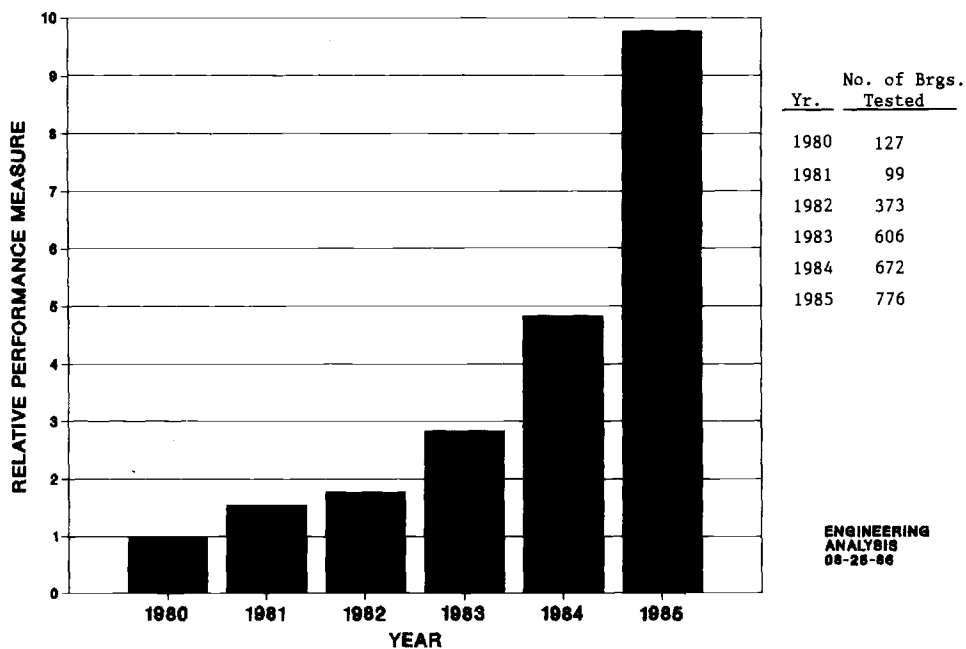


FIG. 10—Spindle bearing performance reliability testing by model year.

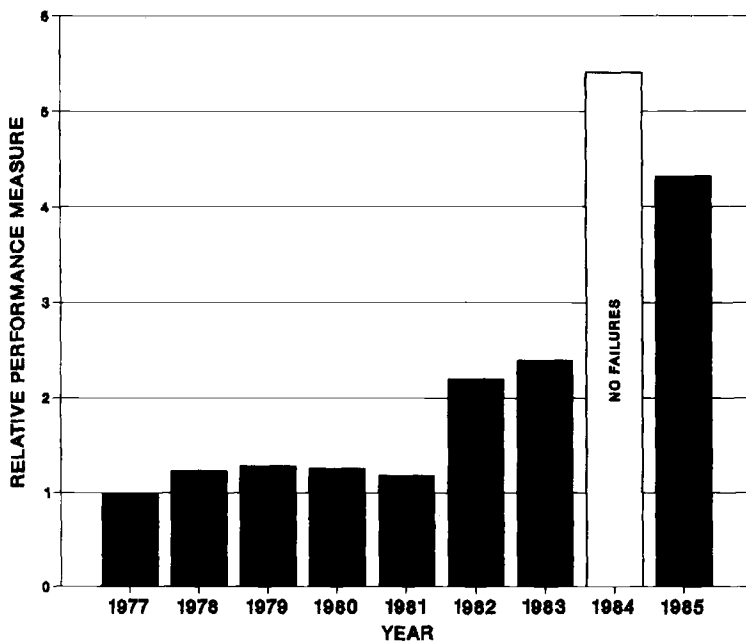


FIG. 11—Spindle bearing performance laboratory testing by model year.

number of bearings tested in this program is 1245. This graph also shows significant gains in performance for the bearings made in 1984 and 1985. The steel cleanliness data, presented in Figs. 3 through 8, shows continuous improvement from 1982 to 1986, with the majority of the indicators showing maximum improvement from 1982 to 1983. The cleanliness measurements were made at the time of the release of that specific heat of steel. This measurement usually precedes the actual production and testing of bearings, made from that heat, by at least six months. This could explain the lag in time between maximum cleanliness improvement and maximum improvement in bearing endurance. Regular production integral wheel bearings made in 1984 show 70% improvement in fatigue life over those made in 1983, and 1985 production shows 250% improvement over 1983 production.

Field Performance

It was apparent early in the development program of the integral wheel bearing that a new method to track field performance was needed. Most field performance data available at that time were for 19 200 km (12 000 miles) of customer use. The new method of tracking had to be capable of measuring performance at 80 000 and 160 000 km (50 000 and 100 000 miles). Figure 12 shows a model which was developed to measure the field performance of the integral wheel bearing. The validity of this model is based, to a large extent, on the fact that New Departure-Hyatt is the sole producer of original equipment and after-market integral wheel bearings for General Motors' cars. Figure 13 shows a typical report of this model.

The field performance model consists of three main elements: determination of mileage distributions of all cars using the integral wheel bearing; determination of the number of integral wheels bearings replaced in service; and a mathematical model which matches the two distributions and estimates the performance at any mileage level.

The vehicle mileage distribution is estimated using monthly production records for all cars using the integral wheel bearing analyzed, in conjunction with a distribution which relates vehicle mileage to age. The distribution which relates vehicle mileage to age is based on a Department of Transportation study. The study divides car mileage accumulation (monthly)

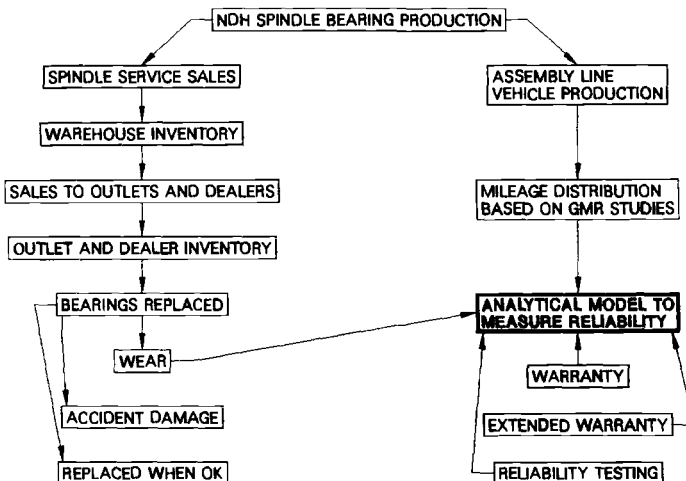


FIG. 12—NDH spindle bearing field performance analytical model.

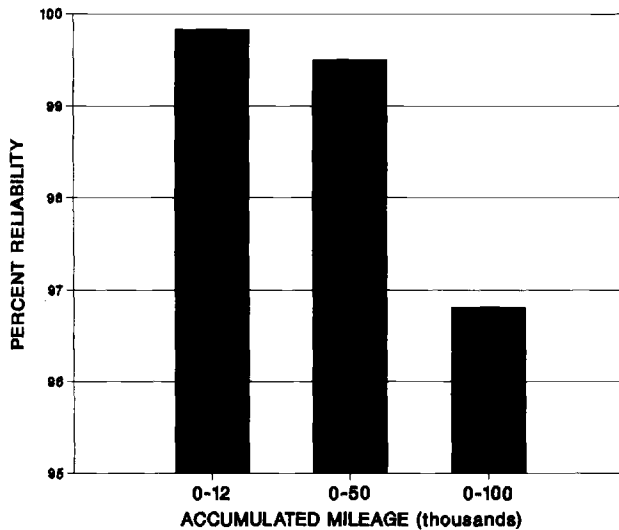


FIG. 13—NDH spindle bearing field performance analysis through 30 June 1985 by accumulated mileage (in miles).

into 25 equal categories and reports the fraction of cars in each category. The study is kept current by General Motors Transportation Research. Figure 14 shows estimates of mileage distribution of approximately 53 000 000 integral wheel bearings. The distribution in this graph is limited to four mileage categories.

The number of integral wheel bearings replaced is estimated based on after-market sales, warehouse inventory, and dealer inventory. Monthly records are kept for after-market sales

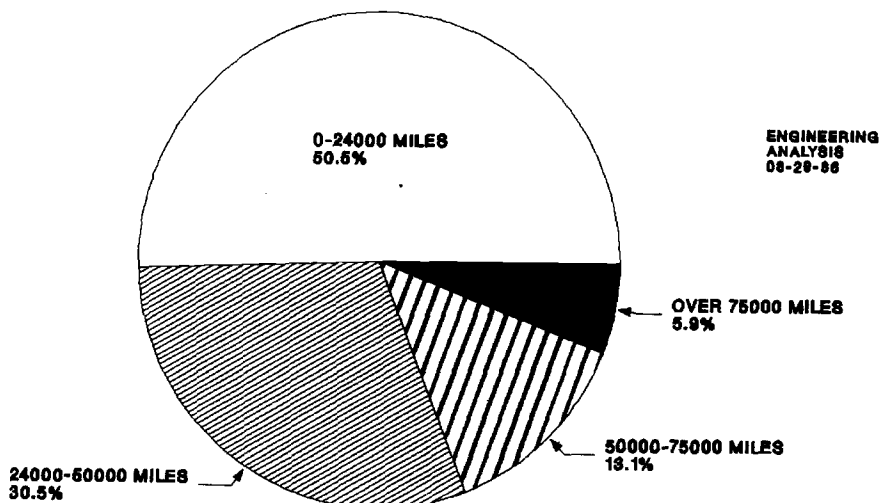


FIG. 14—NDH spindle bearing field mileage distribution through 31 July 1986 by length of service. 52 952 400 bearings in field. (in miles).

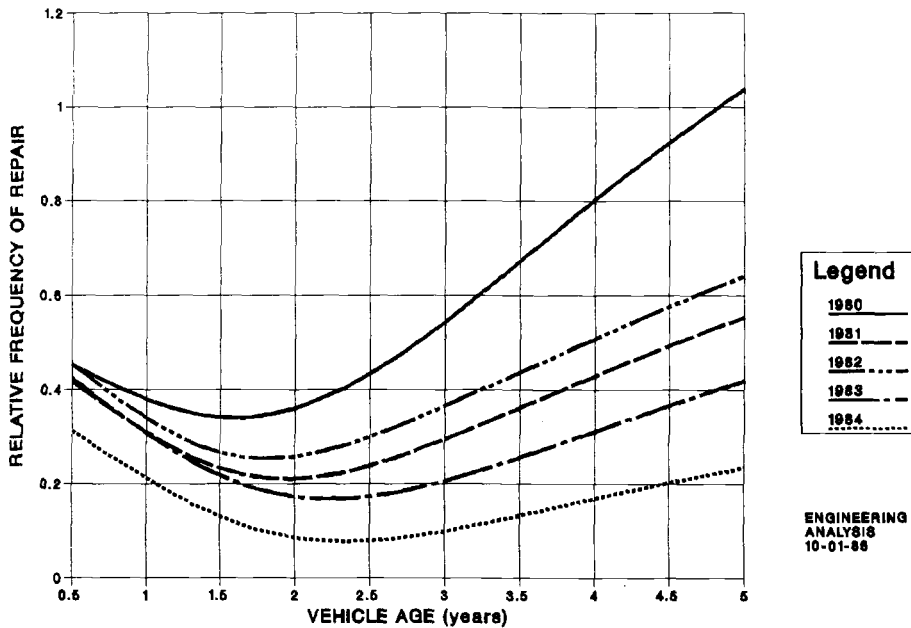


FIG. 15—Wheel bearing relative frequency of repair by model year.

and inventory. The dealer inventory is estimated by conducting a survey of approximately 140 General Motors dealerships nationwide and projecting the results across the entire dealership network. The difference between the number of integral wheel bearings sold and the number in inventory, at the dealer level, is used as the number of bearings replaced in service.

The mathematical model estimating the field performance is based on the Weibull distribution. Warranty and Continuous Protection Plan data, in addition to a range of Weibull slopes, are all used in an iteration process which produces the most likely solution and best fits the actual after-market sales data.

Figure 15 shows the performance of the integral wheel bearing over a period of five years and measured in terms of frequency of repair (replacement). All performance data were scaled relative to the 1980 level, which is taken as unity. The data show little improvement in performance for new bearings (one year old). However, older bearings, with more mileage, show significant improvement in performance, particularly from 1982 to 1983 and from 1983 to 1984. Examination of replaced bearings shows that early bearing problems (within the first year of use) are very seldom due to metal fatigue. This fact leads to the conclusion that improvement in fatigue life of the integral wheel bearing after extended service is due, to a large extent, to the reported improvement in steel cleanliness.

Conclusions

1. The SAM rating method for microinclusions and the AMS 2301 rating method for macroinclusions, when used in tandem, have proven to be effective tools for the evaluation of steel internal quality and for the facilitation of steel internal quality improvement.

2. Laboratory testing under conditions relatable to field use and mathematical modeling of service bearing consumption over time both confirm significant improvements in bearing fatigue performance correlating with improvements in steel internal quality.

It is appropriate to emphasize that when laboratory testing is designed correctly and with an understanding of field demands, it is possible for the laboratory test to predict field performance.

The steel quality and bearing performance improvements reported herein represent a response to the increasingly stringent demands of the marketplace. The authors feel that these gains could not have been achieved without the parallel development of a new level of cooperative relationships between supplier and customer based upon meaningful measurements and tests.

Acknowledgments

The authors wish to thank Ms. Sharon A. Majoy for her efforts in developing the SAM Rating System as well as for generating much of the cleanliness data presented in this paper and Mr. Thomas L. Knoll for his help in analysis of the data presented in this paper.

The authors further wish to thank the technical and steelmaking personnel of LTV Steel Central Alloy District in Canton, Ohio, and of Atlas Specialty Steels Division in Welland, Ontario, for their cooperation and leadership within the North American steelmaking community toward attaining the internal quality improvements reported in this paper.

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APPENDIX

Supplementary Requirement S2 of ASTM A 295 84

S2. "SAM" Inclusion Rating System

The purchaser may specify that the "SAM" inclusion rating system described below be used in addition to the micro-inclusion rating system described in 6.4. It is a method for measuring bearing steel cleanliness which reflects the frequency of occurrence of the larger oxide microinclusions, in addition to their severity. The method is, therefore, more sensitive to metallographically observable variations in microcleanliness than the standard system described in 6.5. Results are expressed in terms of two rating numbers reflecting Type B aluminate and Type D heavy globular oxide inclusion contents.

S2.1 Sampling—See Section 6.

S2.2 Inclusion Rating—Specimen preparation and rating are performed as in 6.4 except as follows:

S2.2.1 In order to minimize the time required for evaluation, the following inclusion types are disregarded: Type A manganese sulfides, because they are not considered significant in

TABLE 2—*SAM rating for B-type inclusions.*

No. of Observed Fields	B Thin	No. of Observed Fields	B Heavy
Not recorded	0.5	not recorded	0.5
Not recorded	1.0	2	1.0
3	1.5	1	1.5
1	2.0	0	2.0
0	2.5	0	2.5
Total area observed = 1.5 in. ²			
SAM rating = $(3 \times 1.5) + (1 \times 2) + (2 \times 1) + (1 \times 1.5) = 10 \div 1.5 = 7$.			

TABLE 3—*SAM rating for D-type inclusions.*

No. of Observed Fields	D Heavy	Units
5	0.5	(1)
2	1.0	(2)
1	1.5	(3)
0	2.0	(4)
0	2.5	(5)
Total area observed = 1.5 in. ²		
SAM rating = $(5 \times 1) + (2 \times 2) + (1 \times 3) = 12 \div 1.5 = 8$.		

rolling contact fatigue; Type C silicates, because they occur infrequently in the subject steels; and Type D thin globular oxides and the smaller ratable Type B aluminates, because of their small size. In addition, Type D heavy globular oxides are defined as those particles measuring 13 μm (0.05 in.) or larger at their widest point at 100 $\times$. Type B heavy aluminates are similarly defined as inclusions measuring 13 μm (0.05 in.) or larger in width at 100 $\times$.

S2.2.2 The "SAM" rating for B-type inclusions is obtained as follows [see Table S1 (Table 2 herewith)]: all B thin fields with a rating of 1.5 or higher are recorded for all samples; all B heavy fields with a rating of 1.0 or higher are recorded for all samples; and the above numbers are summed and normalized by dividing the total rated areas of all samples in square inches. The nearest whole number is recorded as the rating.

S2.2.3 The "SAM" rating for D-type inclusions is obtained as follows [see Table S2 (Table 3 herewith)]: all D heavy fields with a rating of 0.5 or higher are recorded for all samples. Fields of 0.5 are counted as one unit; fields of 1.0 as two units; fields of 1.5 as three units and so on. The number of units is summed and normalized by dividing by the total rated area of all samples in square inches. The nearest whole number is recorded as the rating.

Quantitative Inclusion Ratings and Continuous Casting: User Experience and Relationships with Rolling Contact Fatigue Life

REFERENCE: Hampshire, J. M. and King, E., "Quantitative Inclusion Ratings and Continuous Casting: User Experience and Relationships with Rolling Contact Fatigue Life," *Effect of Steel Manufacturing Processes on the Quality of Bearing Steels*, ASTM STP 987, J. J. C. Hoo, Ed., American Society for Testing and Materials, Philadelphia, 1988, pp. 61-80.

ABSTRACT: The limitations of the Jernkontoret (JK) method of assessing inclusions as described in ASTM E 45 are pointed out and two alternative quantitative methods of inclusion assessment presented in detail. Close relationships are demonstrated between the rolling contact fatigue life and the quantitative inclusion assessments for 25 different casts of Society of Automotive Engineers SAE 52100 type steel. From this, the SAM B-type count shows the closest relationship with fatigue life. Inclusion of the SAM count in steel specifications has improved the monitoring of steel quality from suppliers. The quality of continuous-cast bearing steel from four European suppliers is described and rules are drawn up defining the optimum method of steel manufacture and mechanical working.

KEY WORDS: bearings, fatigue life, SAE 52100 steel, SAE 4720 steel, inclusions, SAM counts, continuous casting, bearing steels

Rolling element bearings are very reliable components which allow accurate rotational motion with low friction while exhibiting consistent, high endurance limits even when transmitting high load. This consistent, high level of performance is achieved primarily by the high quality and reliability of the steel from which bearings are made. The standard steel adopted by most of the bearing producers throughout the world is based on the Society of Automotive Engineers, SAE 52100 composition, which is generally melted in electric arc furnaces (EAF) and then refined to high levels of steel cleanliness by processes which usually involve vacuum degassing and other techniques designed to reduce nonmetallic impurities to a minimum. The levels of nonmetallic inclusions are generally assessed by comparison against standard charts, of which the Jernkontoret (JK) method, as described in the ASTM Practice for Determining the Inclusion Content of Steel (E 45), Plate III, is the most widely used and has been adopted by RHP Bearings, Ltd., for specifying steel nonmetallic inclusion content for many years, using the limits as laid down in the ASTM Specification for High-Carbon Ball and Roller Bearing Steel (A 295).

However, over the last few years several workers have recognized the fact that the JK technique has serious limitations. It is the intention of the first part of this paper to highlight these limitations, to describe more quantitative techniques developed by RHP Bearings, Ltd., and workers in the United States, and to describe how these more quantitative techniques are

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related to bearing performance. The second part of this paper is devoted to the subject of continuous casting of bearing steels, which is a route currently being developed by many bearing steel manufacturers worldwide. The advantages of the process are that first, much less primary rolling has to be carried out and second, yields are improved due to much less discarded material compared to the ingot cast route where top and bottom discards account for a significant proportion of the total weight of an ingot. The current evaluation considers the quality of bearing steel produced in this way by four European bearing steel manufacturers, using different types of plant and casting blooms of different sizes.

Mechanism of Rolling Contact Fatigue

In service, bearing life is often limited by incorrect fitting, corrosion, ingress of dirt, or inadequate lubrication, but even under well-lubricated conditions, correctly selected, installed, and maintained bearings will fail eventually by breakup of the contacting surfaces, which results in rough running, excessive vibration or audible noise, and in some cases, eventual seizure.

This surface breakup is best explained by considering the complex stress systems a bearing experiences during running. As a bearing rotates, a given point beneath the raceway surface is subject to a buildup of stress as the rolling element approaches that point and a falloff as it passes (Fig. 1). Subsequent contacts repeat the process, resulting in a constant load/unload situation, the most severe feature of which is considered to be a reversing shear stress whose value is at a maximum just beneath the surface. This fluctuating stress eventually gives rise to fatigue cracking, which leads to the surface breakup, hence the concept of rolling contact fatigue.

The mode of fatigue failure under good lubrication conditions is generally considered to consist of subsurface crack nucleation in the region of peak reversing shear stress, followed by propagation of the crack to the surface (Fig. 2). Eventually, this crack propagation results in the formation of a pit at the surface (Fig. 3). The initial nucleation of the fatigue crack has been reported by many workers [1-6] to be greatly influenced by the presence of nonmetallic inclusions left over from the steelmaking process. Oxides are considered to be particularly deleterious, since they are brittle and because they are surrounded by tensile stresses due to differential coefficients of thermal expansion [7], which add to the reversing shear stress field. Figure 4, which is a section through a typical rolling contact fatigue spall, illustrates this fact, failure in this case being initiated at an alumina inclusion trail.

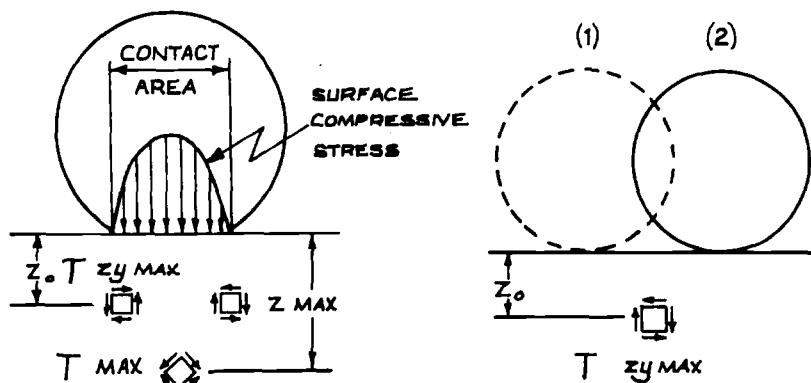


FIG. 1—Stresses in rolling contact.

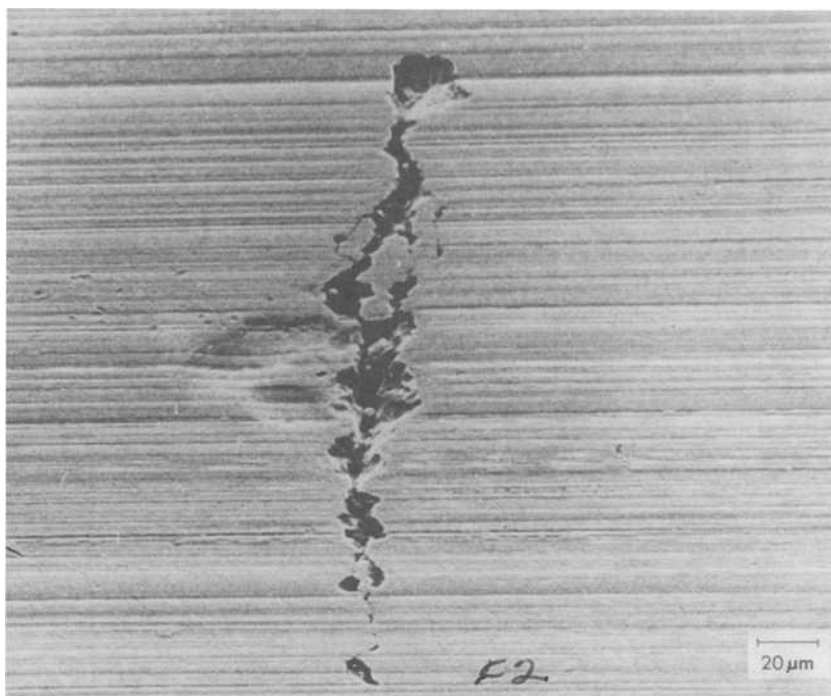


FIG. 2—Emergence of subsurface fatigue crack at surface.

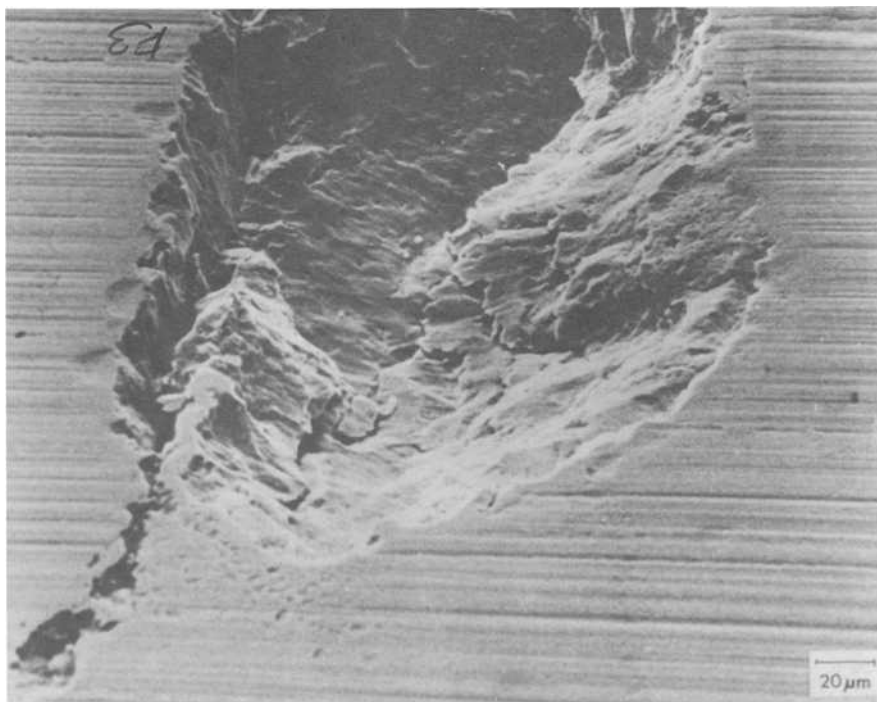


FIG. 3—Typical rolling contact fatigue spall.

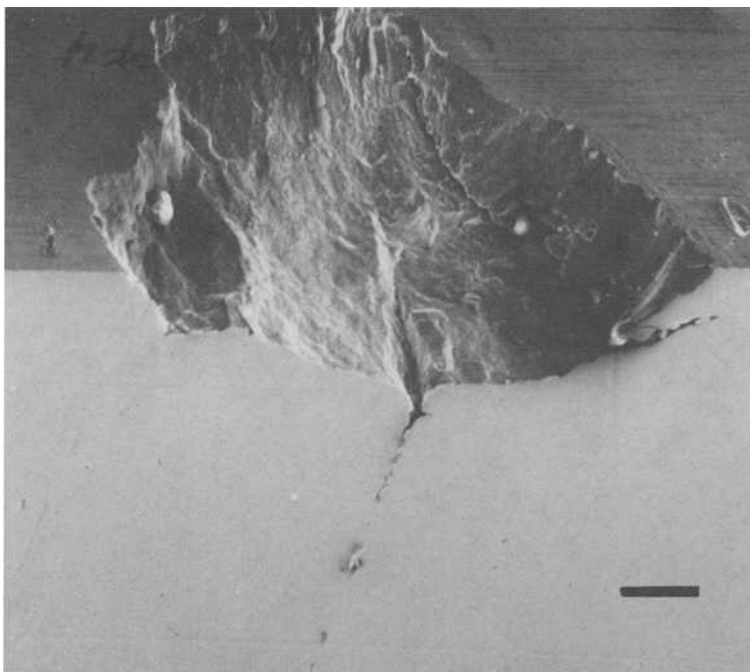


FIG. 4—Section through a fatigue spall, showing association with a B-type alumina inclusion trail (marker is 40 μm).

On the other hand, sulphide inclusions are considered much less deleterious [8] and even beneficial in increasing rolling contact fatigue life [9]. One reason put forward for this is that sulphide inclusions can encapsulate oxide inclusions and reduce the tensile stresses around these inclusions. This is especially the case with steels deoxidized with calcium silicide.

Rolling Contact Fatigue Testing

Rolling contact fatigue life is often determined under accelerated, high-load, high-speed conditions, where fatigue failure occurs in days or weeks rather than in years, as is often the case under service conditions. RHP Bearings, Ltd., treats such tests as screening tests which are relatively cheap and convenient for comparative purposes. The rigs most used by RHP Bearings, Ltd., test flat washers of 74.6 mm (2.94 in.) outside diameter at a maximum surface compressive stress level of 4170 MPa (600 000 psi). The test assembly can be seen in Fig. 5 and is described in full elsewhere [10].

One of the main uses of these rigs is in the approval and monitoring of steel suppliers and steelmaking techniques, an activity which is carried out on a regular basis by RHP Bearings, Ltd. Through this work, a database of test results has been generated, covering many of the major bearing steel manufacturers in the world, most of whom use ostensibly similar steelmaking techniques. While most of this work is related to through-hardening SAE 52100 steel, several tests have been carried out on carburized bearing steels, similar to SAE 4720, conforming to the requirements of the ASTM Specification for Carburizing Steels for Anti-Friction Bearings (A 534).

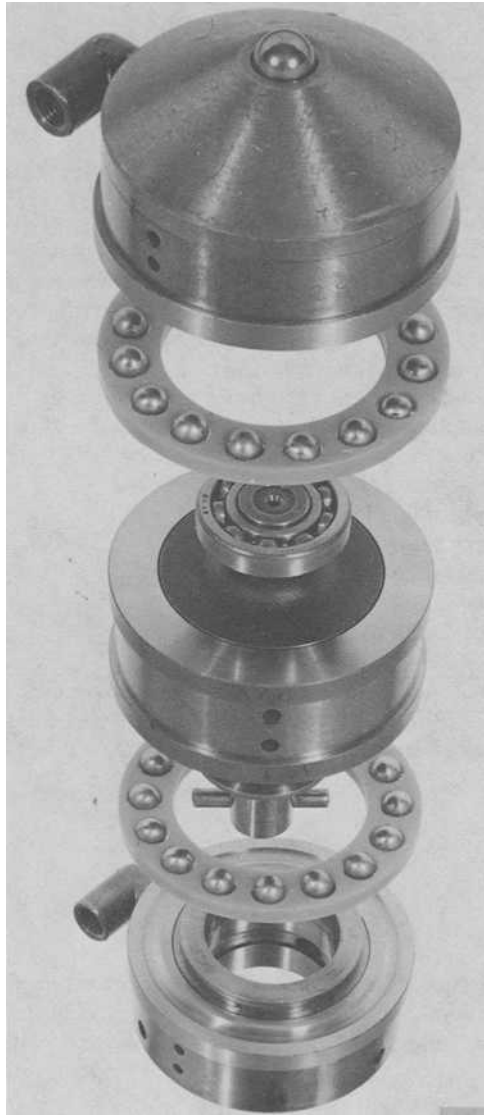


FIG. 5—RHP Bearings, Ltd., flat water fatigue rig assembly.

Fatigue test data obtained from these tests are analyzed by Weibull statistics, the main parameter recorded being the L_{10} life, which is defined as the life within which 10% of the batch will fail or, alternatively, the life which 90% of the test components will exceed. While these rigs are very good at ranking different steels, when full bearing life information is required by RHP Bearings, Ltd., a variety of rigs are available which test full bearings under stress levels closer to those encountered in normal applications [11].

Inclusion Assessment and Fatigue Life*RHP Inclusion Index Technique*

The JK method of assessing inclusions as described in ASTM E 45 rates the quality of a steel sample in terms of the worst field of 100 fields examined when compared against standard charts. The main drawback to this technique is the fact that a steel with 100 equally bad fields would rate the same as a steel with only one bad field and 99 good fields. Of these two steels, the former is obviously the dirtier and likely to give a worse fatigue performance, a fact which would not be obvious from the standard JK count.

RHP Bearings, Ltd., realized this fact several years ago when casts of steel meeting the JK requirements failed to achieve the required rolling contact fatigue performance. When these particular casts of steel were examined in more detail, high densities of inclusions were noted. Since it was obvious that a technique was needed which took into account the density of inclusions as well as the worst field, RHP Bearings, Ltd., developed the following technique: 100 Fields were examined as previously, but instead of rating only the worst field, each field was rated. A typical example of what was recorded for, say, the B-type alumina inclusions is given below (100 fields total):

- (a) 65 fields with B-type rating less than 1.0.
- (b) 20 fields with B-type rating of 1.0.
- (c) 8 fields with B-type rating of 1.5.
- (d) 4 fields with B-type rating of 2.0.
- (e) 3 fields with B-type rating of 2.5.

In this analysis, a B-thin of say 2.0 and a B-heavy of 2.0 were counted the same. A summation was then achieved by multiplying the number of fields of any particular severity by the severity number, ignoring any fields with ratings less than 1.0. In the above example, the summation would be

$$\begin{aligned}
 20 \times 1.0 &= 20 \\
 8 \times 1.5 &= 12 \\
 4 \times 2.0 &= 8 \\
 3 \times 2.5 &= 7.5
 \end{aligned}$$

Thus, the B-type RHP inclusion index is 47.5. This same procedure was carried out for both B-type and D-type inclusions, counting "thin" and "heavy" inclusions of both types. In this way the inclusion content of any steel sample could be quantified with an inclusion index.

Relationship between RHP Inclusion Index and Fatigue Life

Armed with this technique of quantitatively assessing inclusion content, RHP Bearings, Ltd., assessed the different casts of steel which had been fatigue tested on the RHP Bearings, Ltd., flat washer rigs over the previous few years. Several samples were taken from each cast and average inclusion index values obtained. Figure 6 shows the relationship between the rolling contact fatigue L_{10} life and the B-type alumina inclusion index for both SAE 52100 type steels and SAE 4720 type steels. In this figure, the L_{10} fatigue life axis is relative life, where the calculated life based on Lundberg and Palmgren theory [12] for this particular bearing under these particular operating speeds and load conditions is unity.

From this figure two important points can be noted: first, as the B-type Inclusion Index decreases, the fatigue life increases and second, at any particular level of cleanliness, the life of the SAE 4720 carburizing steel is a factor of almost two times that of the mean life for the

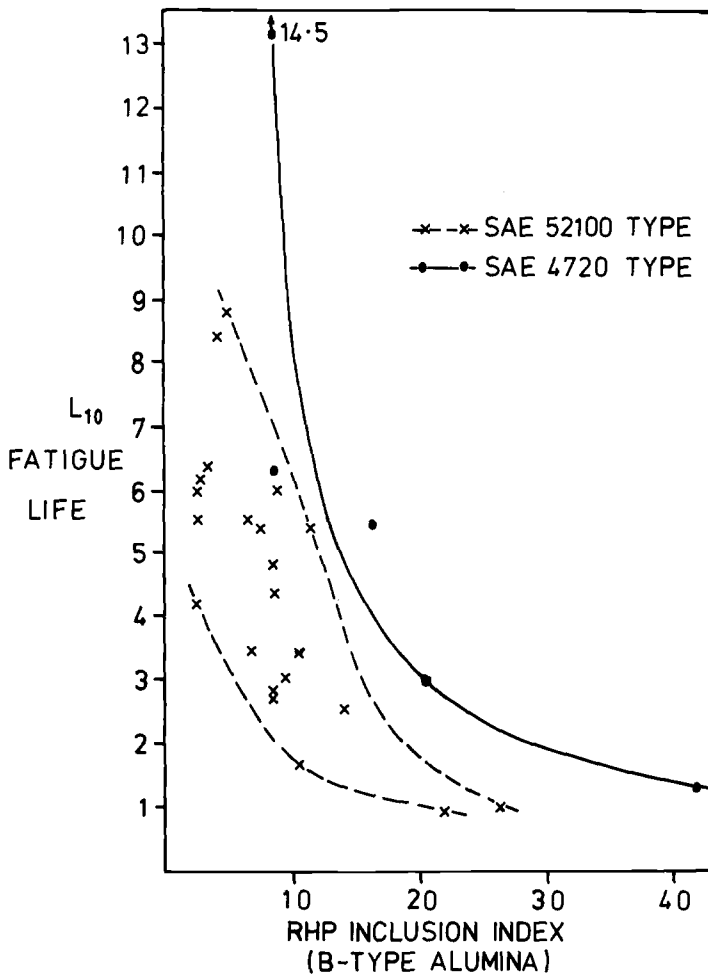


FIG. 6—Relationship between rolling contact fatigue L_{10} life and B-type alumina RHP inclusion index.

through-hardened SAE 52100, a fact observed by other workers [13]. The steel casts in both types of steel which exhibited the longest lives were produced by the electro-slag refined (ESR) route, the ESR carburizing steel producing the longest life of any standard bearing steel tested on the RHP Bearings, Ltd., rigs. The likely reason for the superior life of the carburized steels compared to through-hardening steels is the large residual compressive stresses produced in the surface layers of carburized components.

However, the salient point as far as manufacturers of through-hardened steel are concerned is the fact that low-carbon carburizing steels are inherently more dirty than high-carbon steels, due to better carbon-deoxidation during the steelmaking process in the case of the high-carbon steels. In the experience of RHP Bearings, Ltd., fatigue performance of typical casts of the two different types of steel are broadly the same.

Figure 7 shows the relationship between L_{10} fatigue life and the D-type globular oxide inclusions, L_{10} life being related to the calculated life as in Fig. 6. Obviously, the relationship between the density of D-type inclusions and fatigue life is very loose in the case of the through-hardened steels, although a relationship does exist for the carburizing steels. Again,

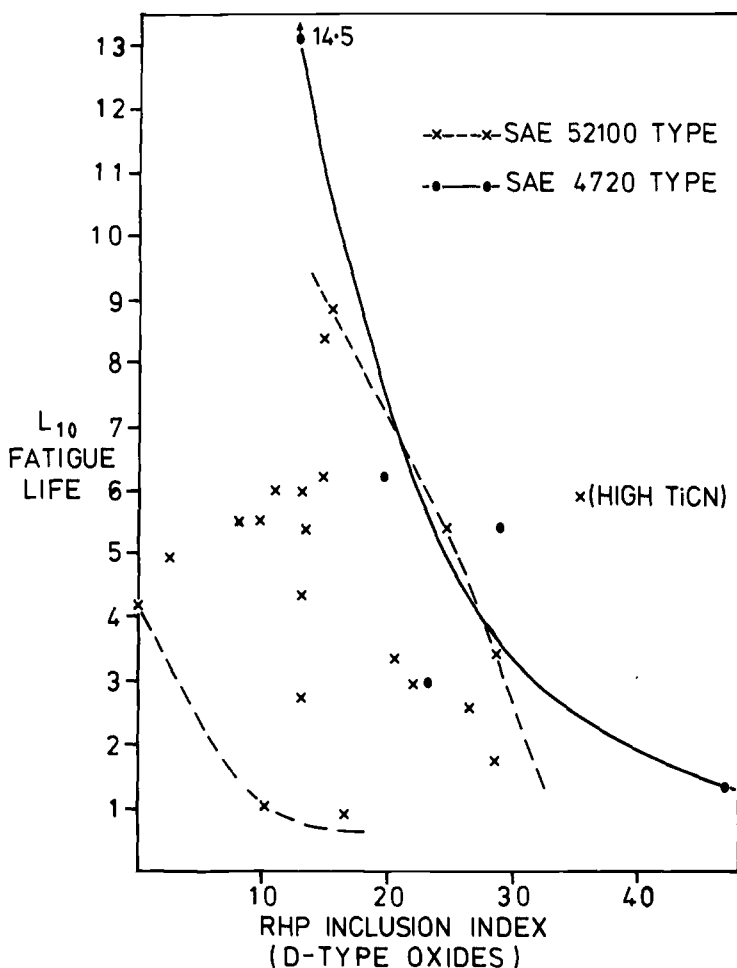


FIG. 7—Relationship between rolling contact fatigue L_{10} life and D-type globular oxide RHP inclusion index.

the carburizing steels exhibited a higher fatigue life at any particular cleanliness value. One difficulty in assessing D-type inclusions is in how to rate the titanium carbide-nitride particles, and RHP Bearings, Ltd., now ranks these particles as D-type inclusions. Of interest in Fig. 7 is that one particular cast with a high content of titanium carbonitride particles, and thus a high D-type count, gave a very good fatigue life. This suggests that titanium carbonitride particles are much more benign in their effect on fatigue life than oxide inclusions. The likely reason for this is that the differential coefficients of thermal expansion between the matrix and the different types of inclusions are by far greater for the alumina particles than for the titanium carbonitride particles, resulting in relatively low stresses around titanium carbonitride particles compared to alumina particles [7]. From this and from the results presented, it would appear that fatigue life is controlled largely by the content of alumina particles, which confirms the findings of other workers [8,9].

Since the RHP inclusion index was found to quantify the quality of steel in this way, the technique was to be included, specifying maximum values, in the RHP Steel Specification in 1983. However, at that time the SAM count technique was being developed by New Departure

Hyatt Bearings in the United States and had been put forward for inclusion in the revised ASTM A 295 specification. In view of the fact that RHP Bearings, Ltd., had always adopted the ASTM A 295 requirements in its specifications, it was decided to compare the performance of the RHP inclusion index and the SAM count before a final decision was made as to which technique RHP Bearings, Ltd., should adopt in its steel specifications.

SAM Count Technique

This technique is similar to the RHP inclusion index method in that it rates each field examined rather than just the worst field and then multiplies the number of fields by their severity. However, it differs from the RHP inclusion index method in that it ignores B-thins of severity less than 1.5, it ignores all D-thins, and it includes all D-heavies down to a severity of 0.5. The multiplication factor for D-heavy inclusions is one for a severity of 0.5, two for a severity of 1.0, and so on. The area examined is measured, the inclusion summations obtained and the summations are then equated to an area of 645 mm² (1 in.²) to obtain the SAM value.

Relationship Between SAM Count and Fatigue Life

All the various casts of steel tested up to the present day on the RHP Bearings, Ltd., flat washer fatigue rigs have been assessed using the SAM count technique. In the majority of cases, SAM counts were taken on sections cut from the fatigue test pieces which had produced

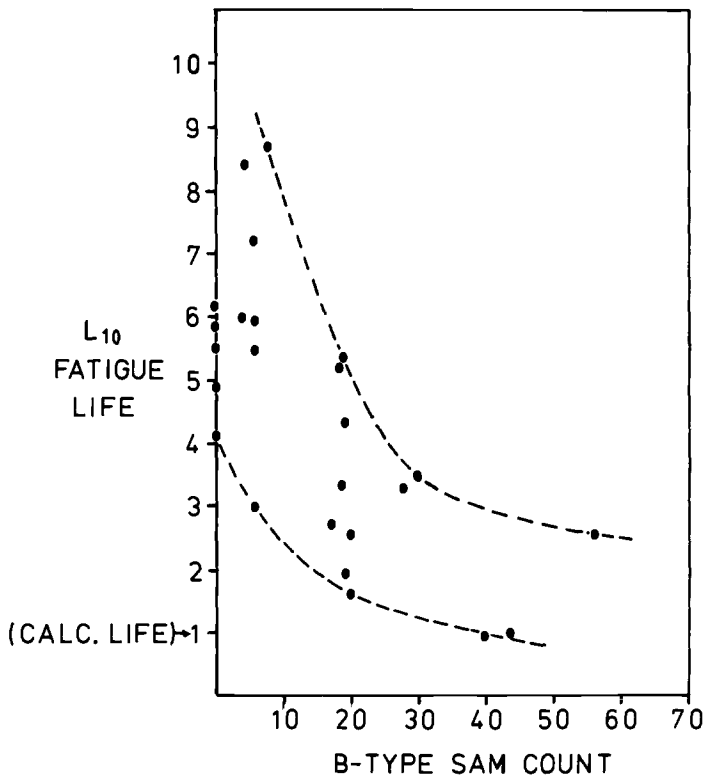


FIG. 8—Relationship between rolling contact fatigue L_{10} life and B-type alumina SAM count for SAE 52100 type steel.

the lowest three lives from that particular series. Since the fatigue test samples were machined from rolled bars of approximate diameters 85 to 100 mm (3.5 to 4.0 in.), these samples represented mid-radial sections.

The relationship between the L_{10} fatigue life and the B-type alumina SAM count can be seen in Fig. 8, which includes 25 different casts of mainly SAE 52100 type steel with a few casts of SAE 1070 type steel, all produced using the vacuum degassing technique, with the exception of the highest life result, which was produced by the ESR route. The L_{10} life axis is again based on the calculated life for the RHP Bearings, Ltd., fatigue test conditions having a value of 1.0. In this figure it is very clear that the relationship between L_{10} fatigue life and B-type alumina inclusion SAM count is very close, bearing in mind the variability of the rolling contact fatigue test.

The relationship between the L_{10} fatigue life and the D-type SAM count can be seen in Fig. 9 for the same 25 counts of steel included in Fig. 8. Although a loose relationship between L_{10} life and D-type SAM count exists, there are two results which fall well out of the scatter band, with L_{10} lives approximately six times the calculated life. However, the high D-type count in these casts is again complicated by the fact that they contained high contents of titanium carbonitride particles. These results add weight to the fact that titanium carbonitride particles have a benign effect on fatigue life. Alumina particles again appear to be the major factor controlling fatigue life.

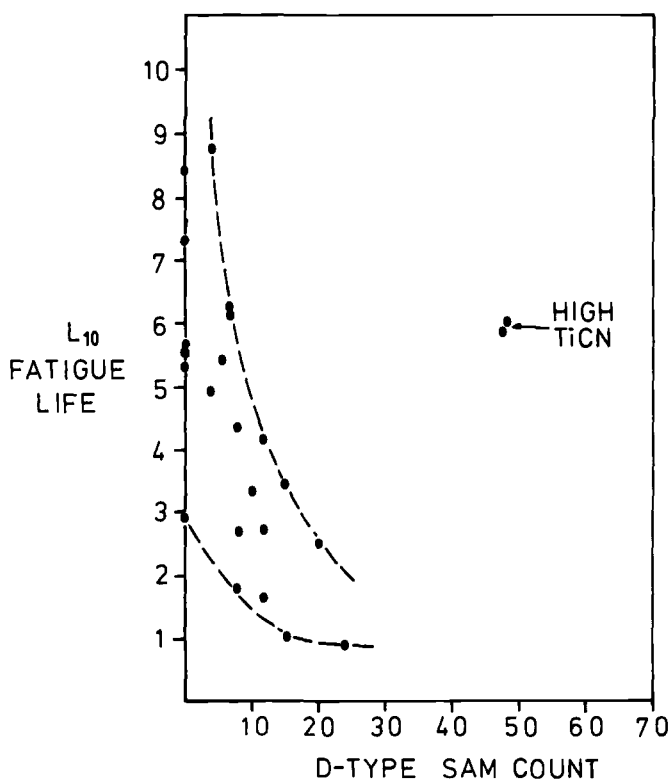


FIG. 9—Relationship between rolling contact fatigue L_{10} life and D-type globular oxide SAM count for SAE 52100 type steel.

Figure 10 shows the relationship between the L_{50} life and B-type SAM count for the same 25 casts of steel. Although the relationship is very loose, it can be seen that at high B-type SAM counts, L_{50} life is always short, whereas at low B-type SAM counts, L_{50} life is very variable. From this, it appears as though the L_{10} fatigue life is the most sensitive to B-type inclusion content, while L_{50} is more likely to be dependent on the bulk properties of the steel such as carbide morphology, carbide banding, and chemical composition.

Carburizing steel samples tested on the RHP Bearings, Ltd., fatigue test rigs have also been assessed by the SAM count method, and the relationship between L_{10} life and B-type SAM count can be seen in Fig. 11, compared against the results for the through-hardened steels. The carburizing steels exhibited fatigue lives as good as or better than any of the through-hardened steels, at all values of B-type SAM count, which agrees very well with the relationship in Fig. 6. The reasons for the superior performance of the carburized steels have been described earlier in relation to Fig. 6.

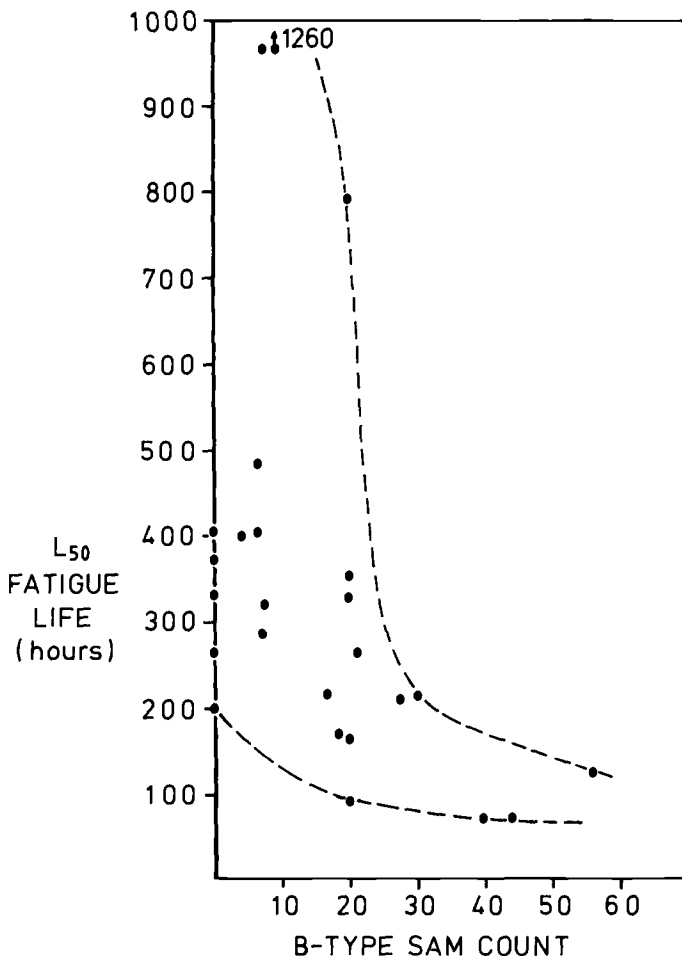


FIG. 10—Relationship between rolling contact fatigue L_{50} life and B-type alumina SAM count for SAE 52100 type steels.

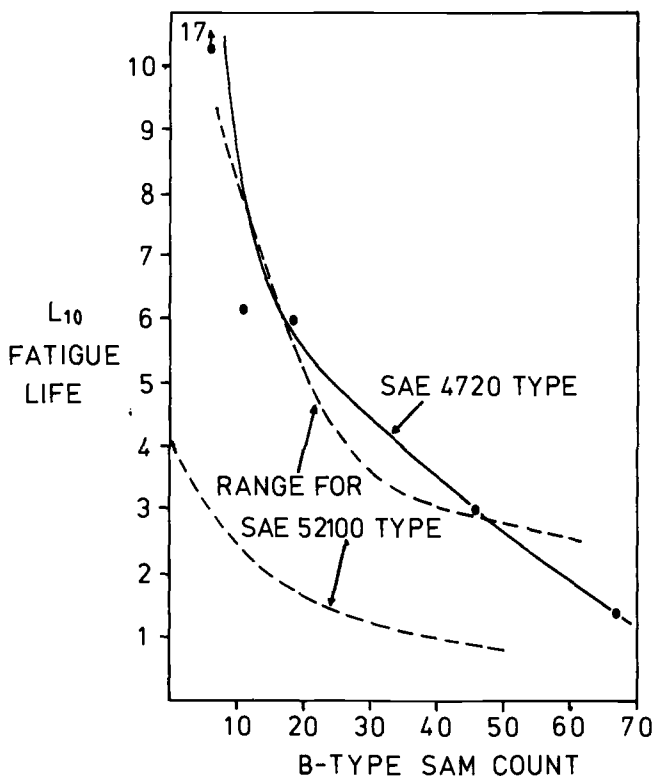


FIG. 11—Relationship between rolling contact fatigue L_{10} life and B-type alumina SAM count for carburizing steels.

Comparison of RHP Inclusion Index and SAM Count Methods

Both of the quantitative inclusion counting methods described accurately assess steel quality with regard to rolling contact fatigue performance in that they both relate B-type alumina inclusion content to L_{10} fatigue life. In addition, the SAM D-type count also relates to fatigue life apart from the complication of the titanium carbonitrides. In view of this and in view of the fact that the SAM count has now been adopted for inclusion in ASTM A 295-84, RHP has adopted the SAM count in preference to the RHP inclusion index for its steel specifications. To meet the latest RHP bearing catalogue life requirements, a minimum L_{10} fatigue life equating to a value of 2.5 times the calculated life is required. To meet this requirement, maximum SAM counts should ideally be no greater than

B-type alumina 10
D-type globular oxide 5

Practical Use of the SAM Count Method

SAM counts have been mandatory on all steel supplied to RHP Bearings, Ltd., since September 1984, and a large data bank of information has been built up since that time, especially on the major steel supplier, Stocksbridge Engineering Steels (previously British Steel

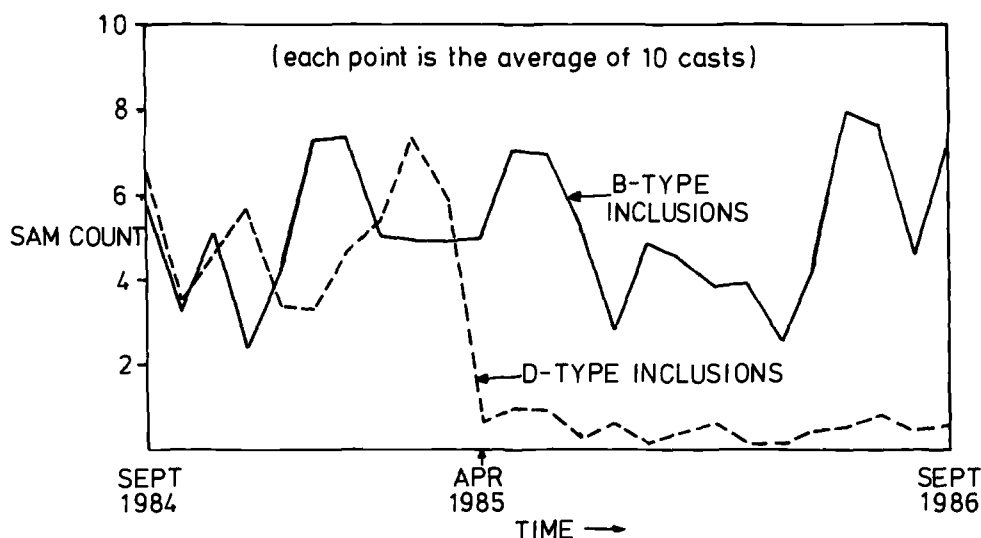


FIG. 12—SAM counts of steel delivered from major steel supplier.

Corporation-Stocksbridge Works). The results of the SAM counts over this period of time can be seen in Fig. 12, where each point is the average result from ten consecutive bearing steel casts, the overall quantity of steel included in this exercise being over 25 000 tonnes. The major feature of this data is the very marked decrease in D-type SAM counts in April 1985, which coincided with the introduction of a ladle steelmaking facility at Stocksbridge [14]. Individual D-type SAM counts since that time have very rarely exceeded a value of three.

The introduction of the ladle steelmaking facility did not have such a dramatic influence on the content of B-type alumina inclusions, which have remained constant over the two-year period. Nevertheless, the average levels of the B-type alumina SAM count have been between three and eight over this period, which equates to a very good rolling contact fatigue life (Fig. 8). Individual B-type SAM counts have rarely exceeded a value of 10 and have never exceeded a value of 15 over this period.

This collection of SAM count data forms the basis for very close cooperation between the steelmaker and user, since trends can very quickly be identified, such as the large decrease in D-type inclusions in April 1985 and the sudden increase in B-type inclusions in mid-1986. In addition, cases like the latter can quickly be brought to the attention of the steelmaker and corrective action undertaken.

In this way, far firmer control can be exercised over steel suppliers, and much better steel consistency can be achieved than ever before when using only the JK worst-field analysis. A further advantage is that the number of expensive rolling contact fatigue tests carried out for approval, and consistency assessment reasons can be reduced and more reliance placed on SAM count assessments.

Continuous Casting of Bearing Steel

Continuous casting of bearing steels offers economies of manufacture since less primary rolling is required and discard levels are reduced compared to the ingot route. However, continuous casting of high-carbon bearing steels presents many technical difficulties for the

steelmaker in controlling porosity and carbide segregation levels. RHP Bearings, Ltd., have evaluated the products of four European bearing steel manufacturers in the form of billets, tube, and rolled bar, with special reference to steel cleanliness, carbide segregation, and rolling contact fatigue life.

Steelmaker 1

This unit consists of a four-strand caster with a bloom size of 355 by 305 mm (14 by 12 in.). At the time the evaluation was carried out, there was no electro-magnetic stirring of the liquid/solid zone. Blooms were rolled down to approximately 90-mm-diameter (3.5 in.) billets, which equated to a 17 : 1 reduction in area, and microsections were taken from several billets representing different positions along the length of the cast. In general, structures were good, with only light carbide segregation in the central zones.

Rolling contact fatigue tests carried out on flat washers machined from the rolled billets gave a good life, and microcleanliness levels were in general quite good. It should be noted, however, that the rolling contact fatigue samples were machined from the outer part of the billets and not from the central region which contained the carbide segregation.

Billets were rolled to tube and sections of tube were examined. The main feature was the evidence of light carbide banding adjacent to the bore surface, as can be seen in Fig. 13. After hardening, the banding still persisted. Rolling to this particular tube size represented a total reduction in area from the bloom size of 95 : 1. There were no major defects such as cracks in this tube.

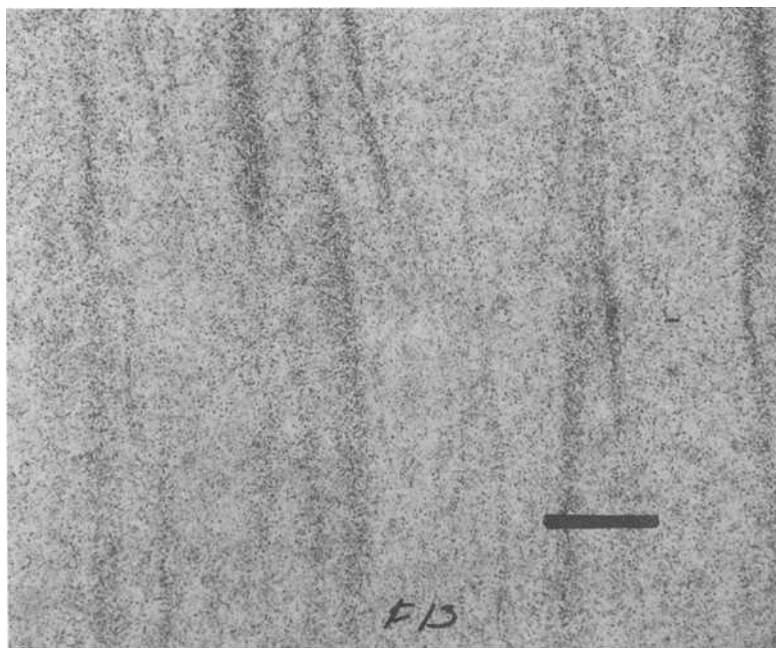


FIG. 13—Light carbide banding adjacent to bore of tubes manufactured from Steelmaker 1 continuous-cast steel (marker is 100 μm).

Steelmaker 2

This unit consists of a six-strand caster with a bloom size of 320 mm by 250 mm (12.5 by 10.0 in.). The initial evaluation was carried out on steel that had not been electromagnetically stirred. Billets which had been rolled to approximately 90 mm (3.5 in.) diameter (13 : 1 reduction) were sectioned, in which carbide segregation and some porosity was evident in the central region.

Rolling contact fatigue tests carried out on samples machined from this billet on the flat washer rigs gave an acceptable life, and microcleanliness levels were good. However, it should be noted that the fatigue samples were again machined from the outer parts of the billets. Tube of two sizes was rolled from these billets. In the larger tube, heavy carbide banding was noted adjacent to the tube bores, with cracks leading from the segregation to the bore in some places (Fig. 14). Rolling of this particular tube size represented a total reduction in area from the bloom size of 41 : 1. The smaller tube size, which had a total reduction in area of 85 : 1, was relatively free of banding, was not cracked, and possessed a good uniform spheroidized structure.

A second evaluation of this steelmaker was carried out on tube produced from electromagnetically stirred blooms. The total reduction ratio in this case was 41 : 1. Although this tube had a much better carbide structure than the nonstirred material, carbide banding still persisted in the area adjacent to the bore, although no evidence of cracking could be found. Again, after hardening, banding was still apparent.



FIG. 14—Bore crack in tube manufactured from Steelmaker 2 continuous-cast steel (marker is 200 μm).

Steelmaker 3

This unit consists of a four-strand caster with a bloom size of 320-mm (12.5 in.) diameter and incorporates electromagnetic stirring. Material rolled to approximately 90-mm-diameter (3.5-in.) billet (13 : 1 reduction) was inspected and found to be free from any major porosity, but carbide segregation was quite noticeable near the bore, and cellular carbide existed throughout the billet. Nonmetallic inclusion content was quite low.

Rolling contact fatigue testing of flat washers produced from these billets gave a poor life, with fatigue failure sites associated with the cellular carbide networks.

To date, no tube from this supplier has been evaluated.

Steelmaker 4

This unit consists of a four-strand caster, producing small rounds and squares in the size range 100 to 200 mm (4.0 to 8.0 in.) and incorporates electromagnetic stirring.

Steel from two casts of SAE 52100 type composition has been examined, which was produced as 100- and 115-mm-diameter (4 and 4.5 in.) blooms. These blooms were rolled directly to tube with total reductions in area of 2.8 : 1 and 2.2 : 1. Tube structures were very poor, with heavy, striped carbide segregation, especially in the area adjacent to the tube bores; there were serious bore defects associated with this segregation, as seen in Figs. 15 and 16. Rolling contact fatigue tests carried out on samples of this steel gave a very poor performance.

Steel from a cast of SAE 1070 type composition continually cast by this steelmaker has been evaluated in the form of rolled bar. Total reduction of area in this case was 91 : 1. Banding was

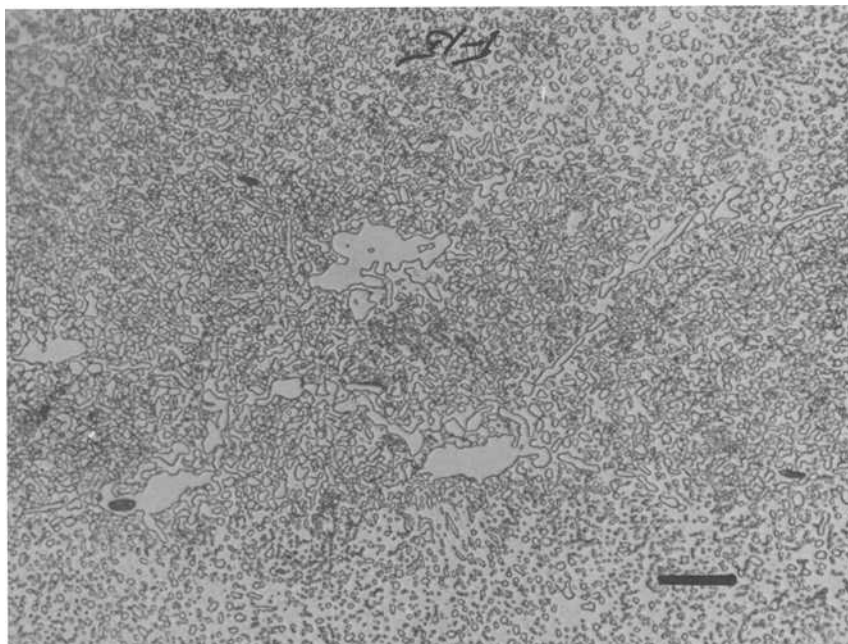


FIG. 15—Heavy carbide segregation in tube manufactured from Steelmaker 4 continuous-cast steel (marker is 10 μm). (Photograph courtesy of TI Desford, Ltd.).



FIG. 16—Bore cracking associated with carbide segregation in tube manufactured from Steelmaker 4 continuous-cast steel (marker is 200 μm). (Photograph courtesy of TI Desford, Ltd.).

again very pronounced in the central region (Fig. 17), and heavy slag inclusions were also found in the central area (Fig. 18).

Comment on Continuous-Cast Quality

Of the four manufacturers evaluated, steel from Steelmaker 4 was clearly not acceptable at all, proving that bearing steel billets cannot be direct cast successfully by this manufacturer. It is also doubtful whether any caster would be successful in casting such small billets in bearing steels. The other three suppliers who manufactured relatively large blooms which were subsequently rolled to billet all produced steel which was relatively clean, and good fatigue lives were recorded on material that was free of serious carbide banding/network. However, carbide banding adjacent to the bores of tubes was a problem with all suppliers, especially when no electromagnetic stirring was employed, unless very high levels of reduction were employed to produce small diameter tubes. Carbide segregation has been recognized as a severe problem in continuous-cast bearing steels. A number of technical papers have been published on this subject and the effects of electromagnetic stirring [15–20].

In general, because of the incidence of tube bore defects, continuous-cast material will not be used for bearing outer raceways manufacture by RHP Bearings, Ltd., due to the fact that the bearing track in this case would fall in the worst-affected zone. Use of this steel for inner raceway manufacture is being considered, since the bearing track in this case will be in the most homogeneous part of the steel. The RHP Bearings, Ltd., rules will specify that large blooms of at least 80 000 mm² (124 in.²) in cross-sectioned area must be used, electromagnetic stirring

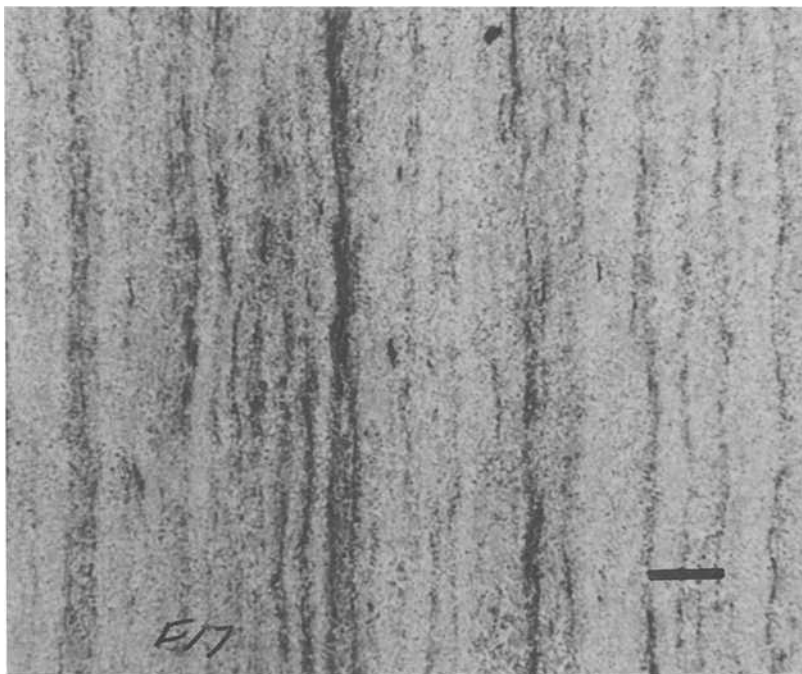


FIG. 17—Carbide banding in central zone of bar rolled from Steelmaker 4 continuous-cast steel (marker is 100 μm).

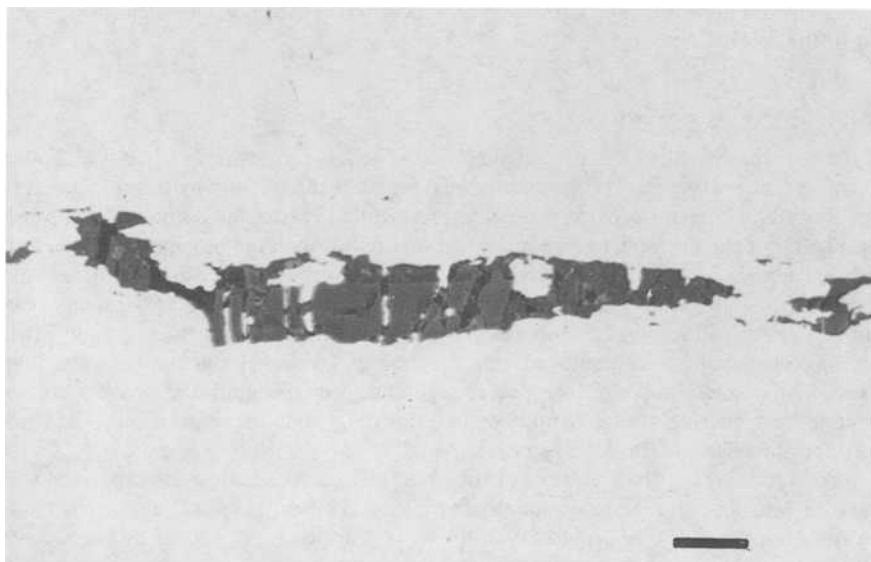


FIG. 18—Heavy slag inclusions in central zone of bar rolled from Steelmaker 4 continuous-cast steel (marker is 25 μm).

must be employed, and ideally, minimum total reduction ratios must be on the order of 80 : 1. Even then, until confidence in the continuous-cast product is gained, a high level of product inspection will be maintained.

To date, continuous-cast bearing steels evaluated by RHP Bearings, Ltd., cannot be classed as being indistinguishable from the ingot-produced steels, a situation which must be the ultimate goal of bearing steel steelmakers.

Conclusions

1. The standard method of describing microcleanness in bearing steels, namely ASTM Recommended Practice E 45 Plate III, does not quantify a steel sufficiently to predict its rolling contact fatigue performance.
2. The RHP inclusion index is a quantitative inclusion assessment method and relates closely to rolling contact L_{10} fatigue life, especially with respect to B-type alumina inclusions.
3. The SAM count qualitative inclusion assessment method accurately describes steel quality, especially with respect to B-type alumina inclusions and relates closely to rolling contact L_{10} fatigue life. SAM count values for both B- and D-type inclusions, which relate to acceptable fatigue life, can be determined from these relationships.
4. Titanium carbo-nitride particles which were earlier thought to have a significant effect on fatigue life have been found to be benign in this test program.
5. The SAM count technique is a very useful tool with which to monitor the quality and consistency of steel produced by different manufacturers.
6. Continuous-cast bearing steels have been evaluated from four steelmakers, and severe defects have been observed in billet, bar, and tube. Ground rules have been laid down which should result in sound material being produced, but in general, continuous-cast steel quality from the suppliers concerned is still somewhat below that of ingot cast steel.

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Effects of Material Properties on Bearing Steel Fatigue Strength

REFERENCE: Schlicht, H., Schreiber, E., and Zwirlein, O., "Effects of Material Properties on Bearing Steel Fatigue Strength," *Effect of Steel Manufacturing Processes on the Quality of Bearing Steels, ASTM STP 987*, J. J. C. Hoo, Ed., American Society for Testing and Materials, Philadelphia, 1988, pp. 81-101.

ABSTRACT: If rolling bearings are subjected to rolling contact stressing in case they are operated under elastohydrodynamic conditions, the formation of spalling in the raceway and consequently the failure of the rolling bearing are a result of fatigue of the material. Material properties have different effects on the rolling bearing life, depending on the contact pressure. Stressing conditions, therefore, must be precisely defined in order to describe these effects. A hypothesis, based on information obtained from the published literature and from our own research, is presented on the failure mechanism of bearings, the formation of spalling after subsurface cracks. The effect of material properties such as material discontinuities, strength, and residual stresses are discussed.

KEY WORDS: rolling contact, bearings, failure mechanism, material stress, structural alterations, material properties, nonmetallic inclusions, residual stresses, spalling, subsurface cracks, S-N curve, fatigue strength, endurance limit

With rolling elements cycling on one another, the force is transmitted through the surfaces, which are more or less separated by the lubricant. In a very complex way, the material stress of the individual elements is dependent on the size and shape of the element, the curvature ratio, the contact pressure (normal forces), the conditions in the lubricating gap (and consequently the tangential forces transmitted in the surface due to friction), the tribomechanical properties of the lubricant, the operating temperature, and other factors.

The stressing conditions must be precisely defined for a description of the effect of material properties on the life of rolling bearing steel. For the case of stressing of elements in rolling contact under elastohydrodynamic (EHD) conditions, the material stressing explained with the various stressing hypotheses is compared with the failure phenomena. The formation of sub-surface cracks can obviously be explained only by accepting the validity of the distortion energy hypothesis and the alternating shear stress hypothesis. A hypothesis is presented in this paper on the failure mechanism of bearings. Using this hypothesis as a starting point, the effects of nonmetallic inclusions, of material strength, and of residual stresses on the pattern of the S-N curve are discussed.

Structural Changes

For rolling bearings operating under EHD conditions, the formation of spalling in the raceway is due to fatigue of the material, that is, due to plastic deformation, generation of

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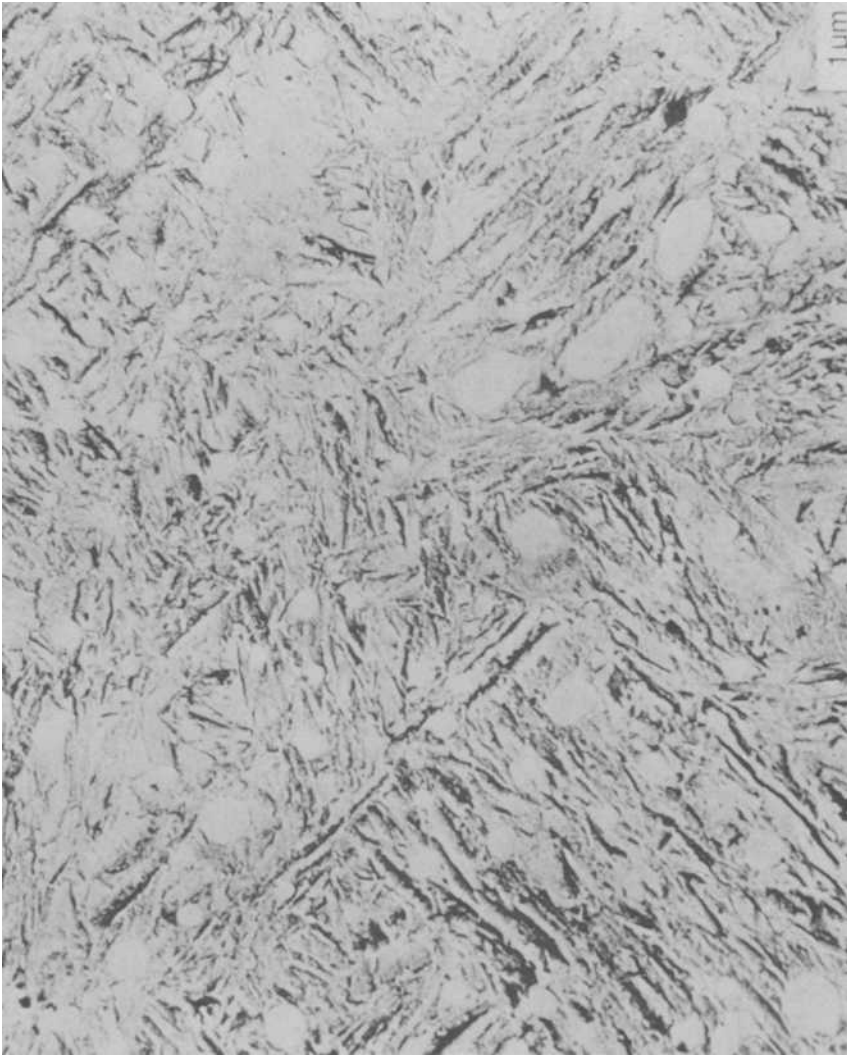


FIG. 1—First stage of fatigue—slip motions in martensite structure.

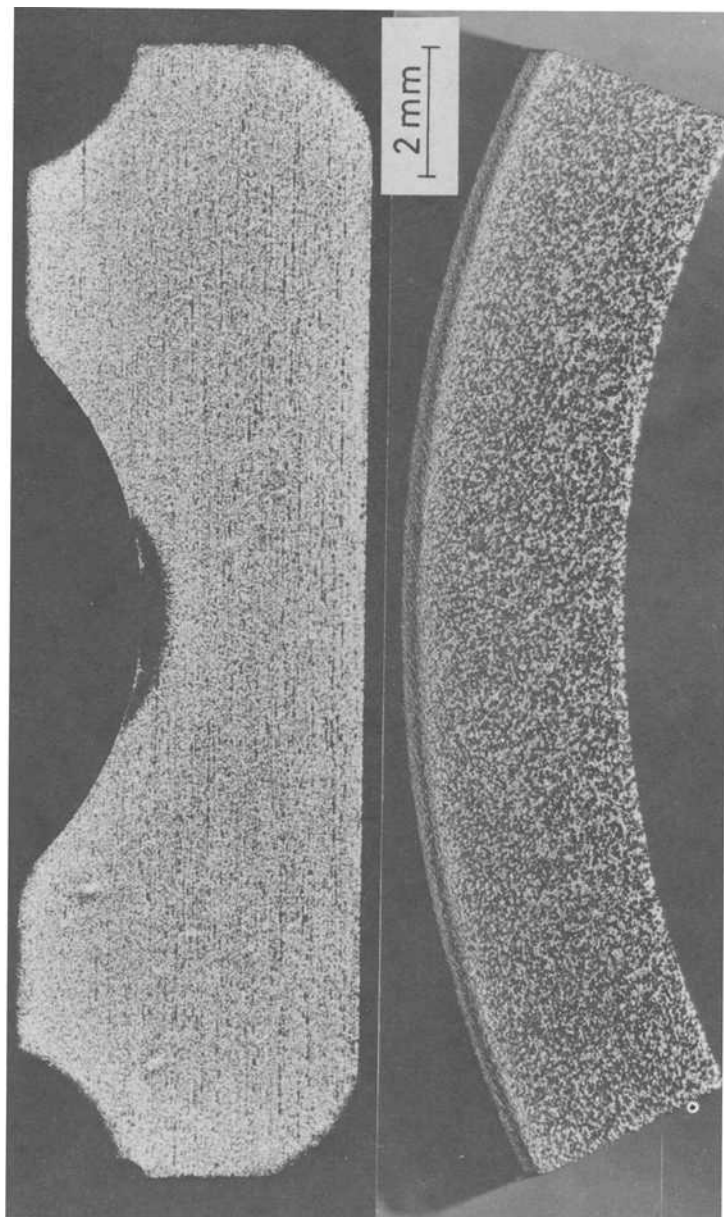


FIG. 2—Dark etching areas in axial and parallel sections of deep-groove ball-bearing inner ring.

incipient cracks, formation of cracks, propagation of cracks, and finally, spalling of material particles. Therefore, the observation of structural changes which occur under the raceway surface due to rolling contact stressing is very important.

If elements in rolling contact are subject to such a high material stressing that the yield point is locally exceeded, the material is plastically deformed. Material changes of this type were frequently found [1-18]. Figure 1, for example, shows signs of slip motions in the martensitic structure of a rolling bearing ring. The ring was cut parallel to its face. It is typical for this phenomenon that the slip marks are located orthogonal to each other and form an angle of about 45 deg with the surface.

Figure 2 shows so-called dark etching areas (DEAs). In zones which appear dark after etching, slip motions have occurred in the martensitic structure. The darker these areas appear, the more slip motions have occurred in the martensite. From the position and darkness, conclusions can be drawn about the heaviest stressing of the material under the raceway surface.

Figure 3 shows the results of analysis by X-ray diffraction of elements in rolling contact which were stressed in the described way. The plastic deformation of the martensite reduces the half value breadth; that is, there is a reduction of the lattice distortion. Furthermore, in the area of plastic deformation, residual compressive stresses parallel to the raceway were detected [19]. The generation can be explained by means of the model shown in Fig. 4 [11]: In the moment of plastic deformation generation, when the contact pressure reaches its maximum value and consequently pressure is highest in the z -direction, a "shortening" of the material in the z -direction and a "lengthening" parallel to the surface occurs.

Frequently observed structural changes under the raceway of elements in rolling contact are the so-called butterflies. Figure 5 shows white zones which resist etching. These zones resemble wings of butterflies. The "body" of the butterfly, which appears dark, is usually a nonmetallic inclusion in the steel. From the direction of the "wings," it can be assumed that the butterflies were generated due to the maximum shear stress. The depth under the raceway at which most butterflies are found, coincides with the zone of maximum stressing.

Many theories for butterfly formation have been advanced [16,20-23]. We explain the generation of butterflies as follows: nonmetallic inclusions are defects in the material and can generate tensile and shear stresses in the surrounding zones. Load stresses are superimposed on this static stress condition under rolling contact; in this case a hydrostatic stress condition and shear stresses are in operation simultaneously (see Figs. 8a and 8d). If the material is overstressed, the elastic energy concentrated in the area of nonmetallic inclusions is transformed into energy needed for crack generation and into deformation energy. The crack extends in the direction of the main shear stress. Due to hydrostatic pressure, the crack surfaces are pressed against each other during growth of the crack, deformed plastically, and consequently heated to austenitizing temperature. Carbides dissolve, and due to self-quenching, martensitic highly deformed rehardening zones are generated [1,16]. Consequently, the mechanism of butterfly generation is largely identical to the generation mechanism of the "adiabatic shear bands" observed with impact of projectile [24-26].

The so-called "white bands" are another phenomenon. After longer running times they were observed on ball bearing rings. The bands in the left part of Fig. 6 form an angle of 30 to 35 deg with the surface tangent. The white bands extending across one another, shown in the right part of Fig. 6, form an angle of 75 to 80 deg with the surface tangent. They develop after a still longer running time penetrating the previously formed flat white bands. These structural changes are indicative of the more or less pronounced decay of martensite. References 2, 13, and 14 report on the mechanism that generates the white bands.

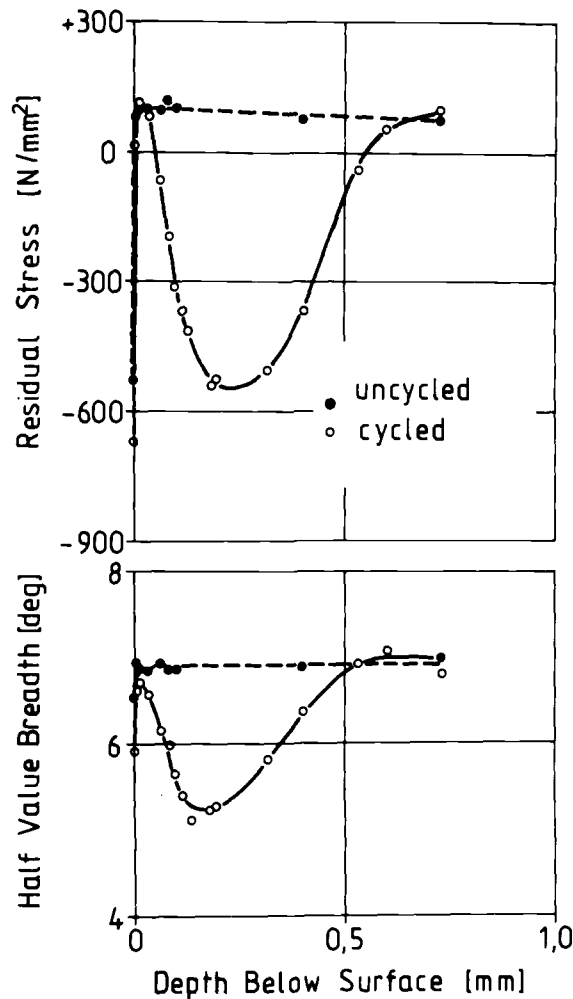


FIG. 3—Residual stress and half-value breadth after cycling.

Structural Changes and Material Stressing

The results from a number of cycling tests are shown in Fig. 7. Accordingly, the development of plastic deformations and white bands depends on the magnitude of the specific load and the running time. Below a certain stress limit, the existence of structural changes cannot be proved.

If DEAs and white bands occur before raceway spalling and if a limiting load exists for structural changes and consequently for the formation of spalling, it can be concluded that spalling under rolling contact stressing is also due to material fatigue. As a consequence, technical endurance strength can be expected also for rolling bearings. In fact, endurance strength of rolling bearings is being confirmed by experiments under improved testing conditions [14,17,27].

If the generation of the described structural changes depends on the magnitude and time of stressing and if the structural changes occur at specific depths and form specific angles with the

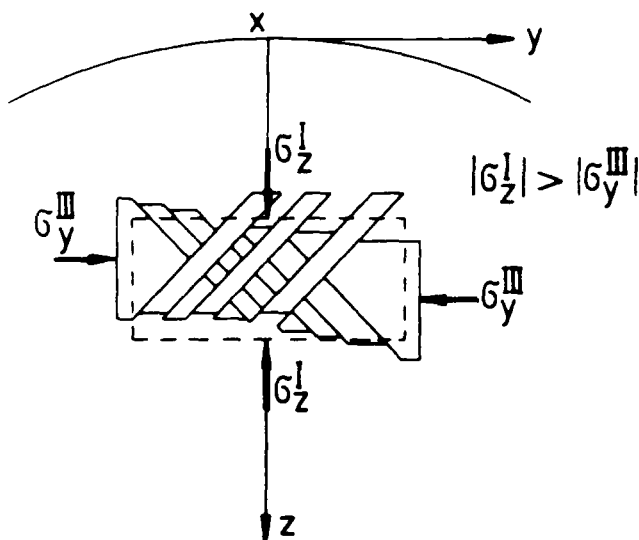


FIG. 4—Anisotropic deformation at region of maximum shear stress.

raceway—and this applies also to butterflies—then it must be possible to relate the generation of the structural changes and the process of fatigue to a mechanomathematical model.

The material stressing under rolling contact, the main shear stress hypothesis, the distortion energy hypothesis, and the alternating (orthogonal) shear stress hypothesis will be discussed. For the interpretation of the structural changes, idealized assumptions were made: the material is in an ideal condition, namely, homogenous, isotropic, and elastic; and the stress distribution in the contact area calculated by Hertz is valid.

Figure 8 shows a comparison of various material stressing hypotheses. The maximum values of the material stressing below the contact area, relative to the Hertzian contact pressure P_0 , differ by their absolute value and depth. The maximum stressing according to the distortion energy hypothesis and the main shear stress hypothesis occur below the surfaces of the center of contact area (Fig. 9) at a depth of 0.7 to 0.78b (b = half the width of the contact area). The main shear stress forms an angle of 45 deg with the tangent of the raceway, while the orthogonal shear stress acts parallel or orthogonal to the raceway; what is more, the orthogonal shear stresses are alternating stresses which, with $\tau_{zy \max} = \pm 0.25p_0$, occur at a distance of

$$y = \pm 0.87b \text{ from the center of the contact area}$$

Hypothesis on the Failure Mechanism at Rolling Contact

According to our research, which agrees with other researchers, the distortion energy hypothesis is best suited to describe the highest possible stressing of the material under rolling contact and the generation of structural changes. The depths and the direction of the phenomena largely coincide with the values obtained with the distortion energy hypothesis. Minor differences are explained by the fact that existing residual stresses and residual stresses resulting from overrolling, and above all tangential stresses created by hydrodynamics, cause a displacement and rotation of the field of stresses.

The structural changes are generated during the first phase of the fatigue process in the moment of highest shear stresses under approximately 45 deg at $\tau_{45 \max}$ and $y = 0$, that is,

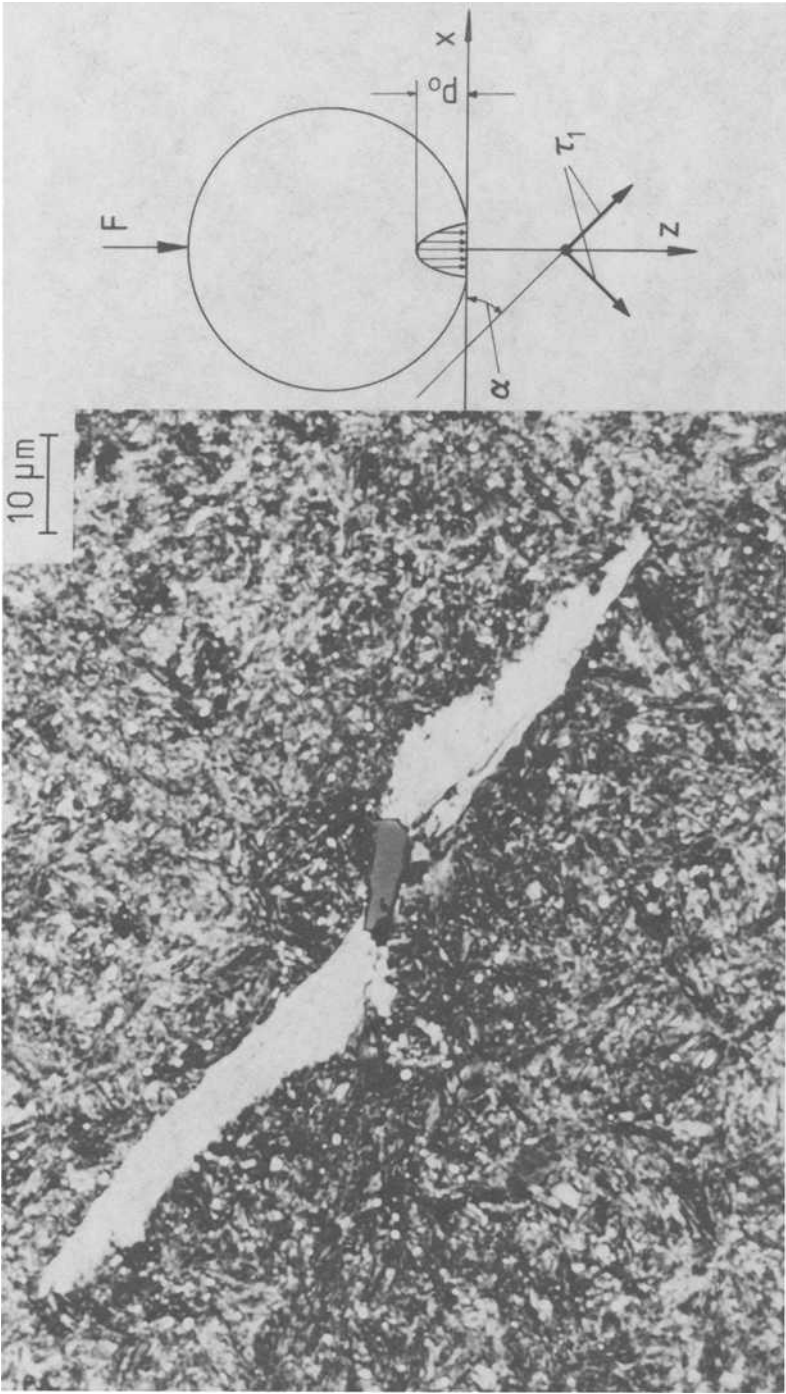


FIG. 5—Butterfly in a cycled component in rolling contact.

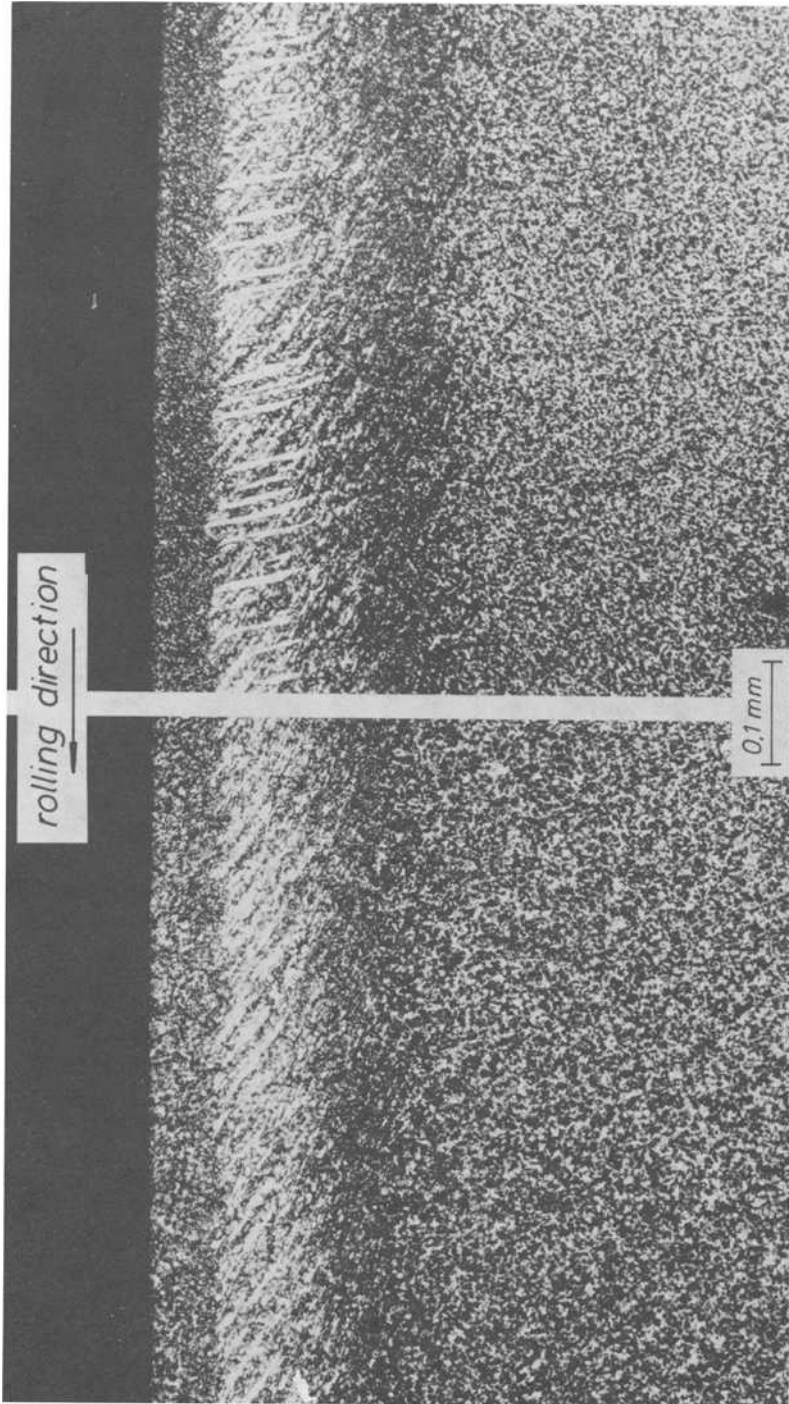


FIG. 6—White bands in cycled deep-groove ball-bearing inner rings (parallel section in groove bottom).

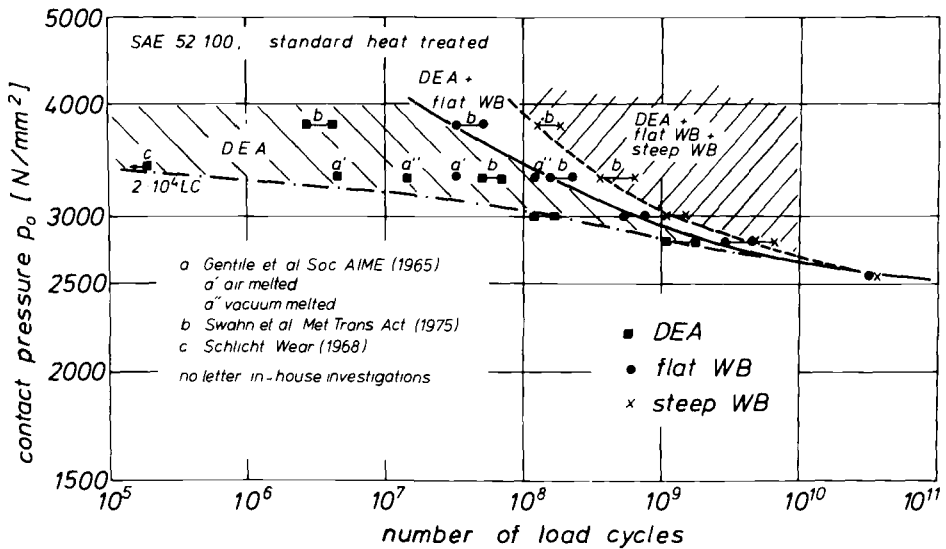


FIG. 7—Correlation between contact pressure, number of load cycles, and development of structural changes.

under the influence of p_0 . Nevertheless, it is hard to imagine that cracks develop at the moment of largest calculated material stressing, that is, under the maximum contact pressure p_0 acting directly on the surface above. This would mean that the incipient cracks are able to grow at the moment of maximum hydrostatic pressure (Fig. 8d).

Hydrostatic pressure on the one hand makes the material more ductile [28]. On the other hand, hydrostatic pressure influences the energetic conditions during the generation of a crack. Cracks caused by shear stresses under the simultaneous action of hydrostatic pressure must originate from fairly large incipient cracks, which means a higher degree of plastic deformation. Furthermore, for crack propagation, more deformation energy is needed. The hypothesis is that spalling forms in two clearly separate phases [29].

The previously mentioned structural changes occur in the first phase. With the plastic deformation of the martensite, incipient cracks are generated which are related to a local loss of strength. At the same time, compressive residual stresses are generated parallel to the raceway. With increasing number of load cycles, the degree of plastic deformation increases. Preconditions for the generation of a crack are being continually enhanced. The structural changes are caused by the material stressing determined according to the distortion energy hypothesis.

The second phase of fatigue includes the generation of the subsurface crack until the formation of spalling. We believe that subsurface cracks are generated during the relief phase of the load cycle; while in one point of the raceway $p = \max$ is changing to $p = 0$, the microcrack below the raceway exceeds its critical size and can begin to grow. At this moment the hydrostatic pressure is lower. We assume that the alternating shear stresses and the multiaxial residual stresses play an important role in this second phase of fatigue.

The plastic deformations in the first phase of fatigue prepare the ground for the generation of a crack. According to the distortion energy hypothesis, the simultaneously occurring compressive residual stresses are parallel to the raceway but reduce the material stressing. The residual stresses have little influence on the alternating shear stresses; so, during the fatigue process, the material stressing remains unchanged according to the alternating shear stress hypothesis. According to the distortion energy hypothesis, material stressing is continually reduced.

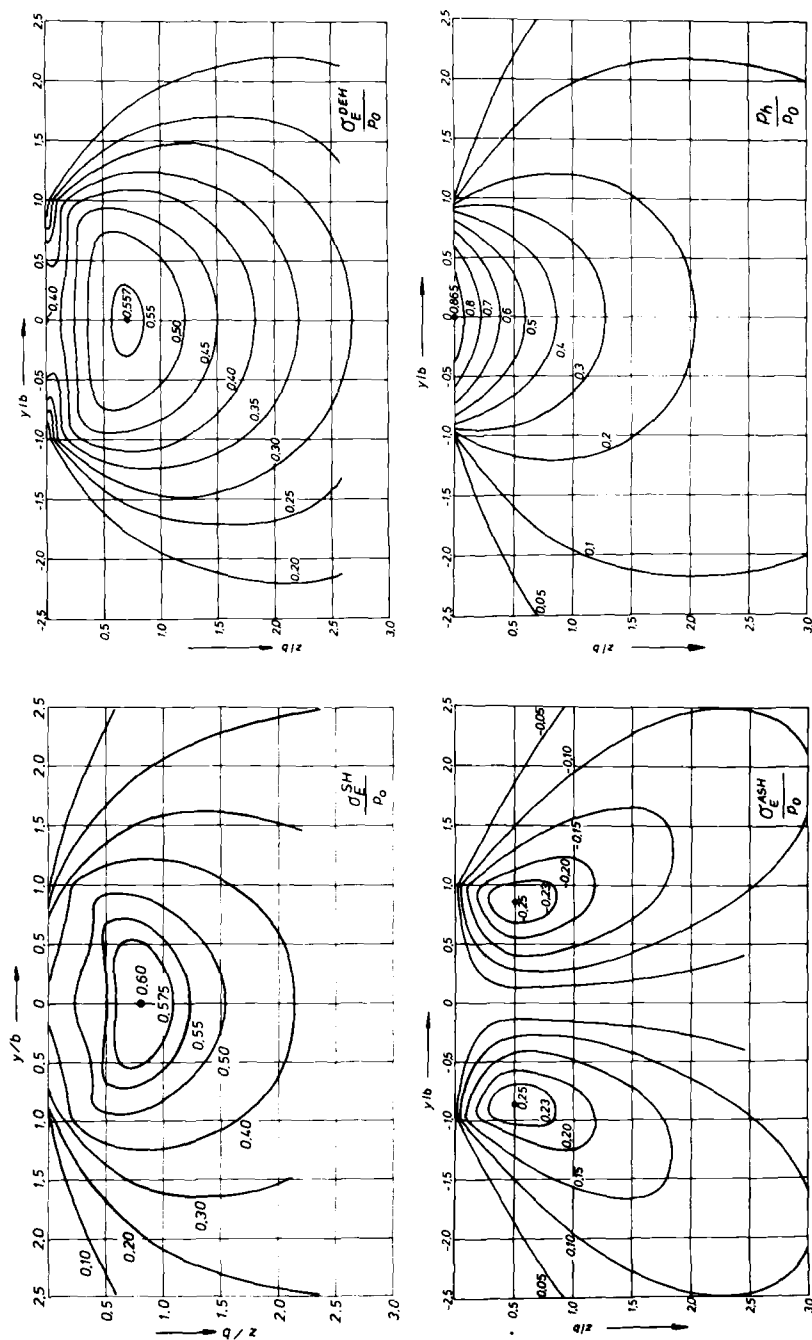


FIG. 8—Dimensionless equivalent stresses (σ_E/p_0) according to shear stress, distortion energy, and alternating shear stress hypotheses and also the ratio of the hydrostatic pressure p_h to maximum Hertzian pressure p_0 .

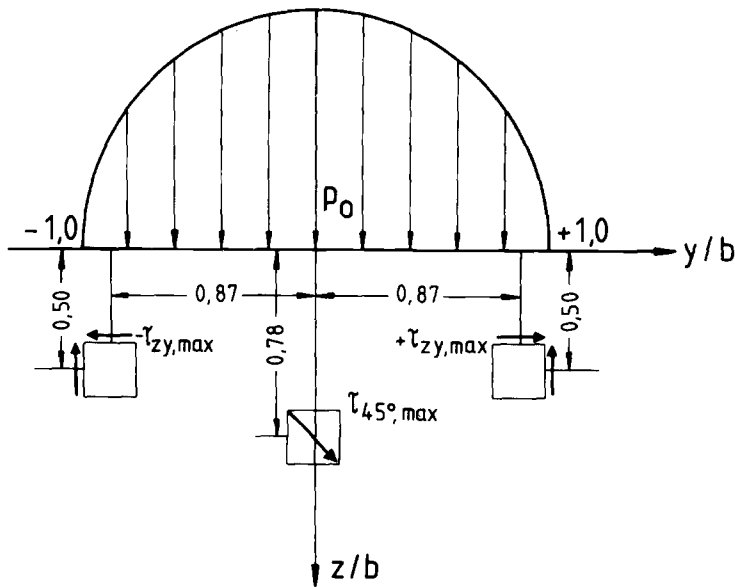


FIG. 9—Shear stresses below surface.

S-N Curve at Rolling Contact

The representation of the life of elements in rolling contact by an S-N curve is possible only under certain restrictions because contact pressures of p_0 larger than 3000 N/mm^2 cause clear macroscopic plastic deformations [14]. The deformations change the geometry of the contact zone, which leads to reduced contact pressure and modified relative motions in the contact zone. Plastic deformations also cause considerable alterations in the material (material properties and residual stress condition), and the material is stressed in a different way from material in an undeformed condition. Nevertheless, the S-N curve will be used in simplified presentation as a basis for the discussion of the fatigue process under different contact pressures and of the material properties.

The plastic deformations under the raceway of elements in rolling contact depend on the magnitude of load and the number of load cycles. In case of a macroscopic inspection, the more or less marked darkness of the DEAs can falsely appear as a uniform deformation of the martensitic structure. Microscopic inspections show, however, that the slip motions can be first restricted to smaller areas (Fig. 1); the martensitic structure (carbon content in the matrix, acicular structure), grain boundaries, inclusions, segregations, and residual stresses of the first and second kind play an important role (Fig. 10).

At high contact pressures and correspondingly higher shear stress amplitudes, slip motions occur, and they are evenly distributed over the entire highly stressed material volume. The deformation no longer occurs in a certain area. Accordingly, incipient cracks resulting from the slip motions [30–32] are more numerous and their distribution more homogenous. These procedures which, according to our working hypothesis, belong to the first phase of the fatigue process take place during what is by far the longest part of the life, thereby determining the length of the fatigue process.

The speed of crack propagation in the second phase of fatigue is probably very high. Figure 11 shows a subsurface crack which appears under the surface of a rolling bearing ring as a

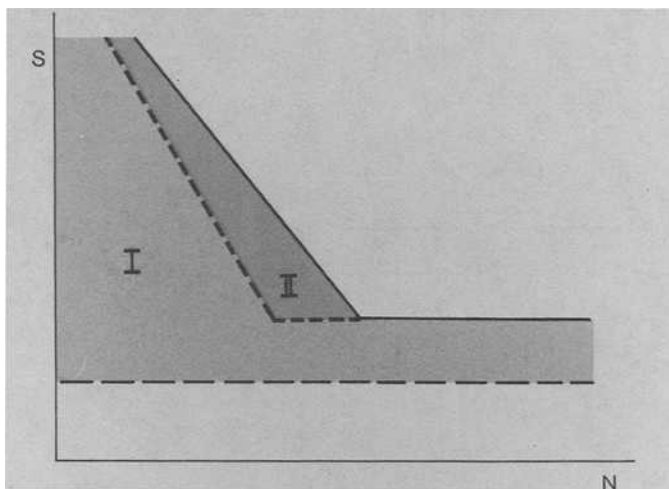


FIG. 10—*S-N curve, Phases I and II of fatigue.*

nonetching band. The nonetching areas are generated by the mechanism described previously for butterflies and adiabatic shear bands. That means that the crack passes at high speed the area under hydrostatic pressure and that during this process its surfaces are pressed against each other, plastically deformed, and heated up.

This phenomenon is frequently observed in highly stressed elements in rolling contact. To date, the origin of the crack could not be found. It is not known whether the origin is located in the nonetching area, that is, in the field of stress below the contact ellipse, or whether the nonetching areas are the end of the crack which, by itself, has separated surfaces. It would be interesting to find an answer to these questions, but it is not relevant to this discussion.

The fatigue processes are, in principle, the same for lower contact pressures. Due to the smaller shear stress amplitude, discontinuities in the material are of increasing importance for local plastic deformations. Fewer incipient cracks are generated, and plastic deformations are restricted to small areas where microcracks form and later can grow.

Plastic deformations under the raceway of elements in rolling contact can be detected also after loading of a load level below the endurance strength limit. They reduce local notch stresses. Under this stress condition, microcracks do not develop or they do not grow.

Effect of Nonmetallic Inclusions on Fatigue Behavior

The adverse influence of nonmetallic inclusions on the rolling contact fatigue of rolling bearings was proved in numerous investigations [14,33–39]. The questions still being discussed—some of them controversially—refer to the determination and evaluation of the degree of contamination, the correlation of nonmetallic inclusions and the life, and the existence of a critical size of nonmetallic inclusions. Another important question is the dependence of these parameters on contact pressure. An attempt to answer these questions will show that they are correlated. Figure 12 shows a correlation between nonmetallic inclusions and the life [14]. The results were obtained from investigations on an angular contact ball bearing test rig with a contact pressure of approximately $p_0 = 2600 \text{ N/mm}^2$. Eleven different lots were tested over a period of several years. Therefore, empirical answers to most of the above questions may not be expected.

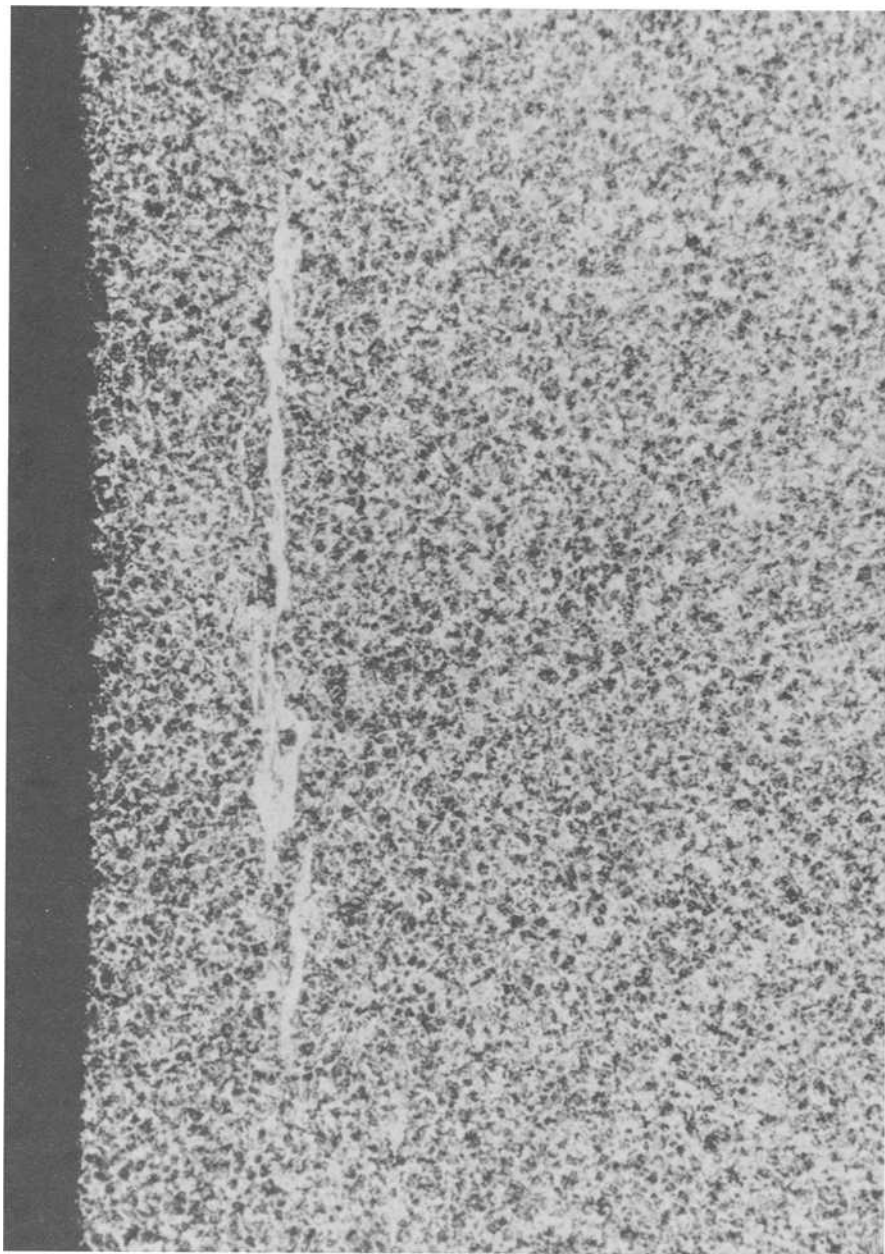


FIG. 11—Subsurface cracks with unetched areas.

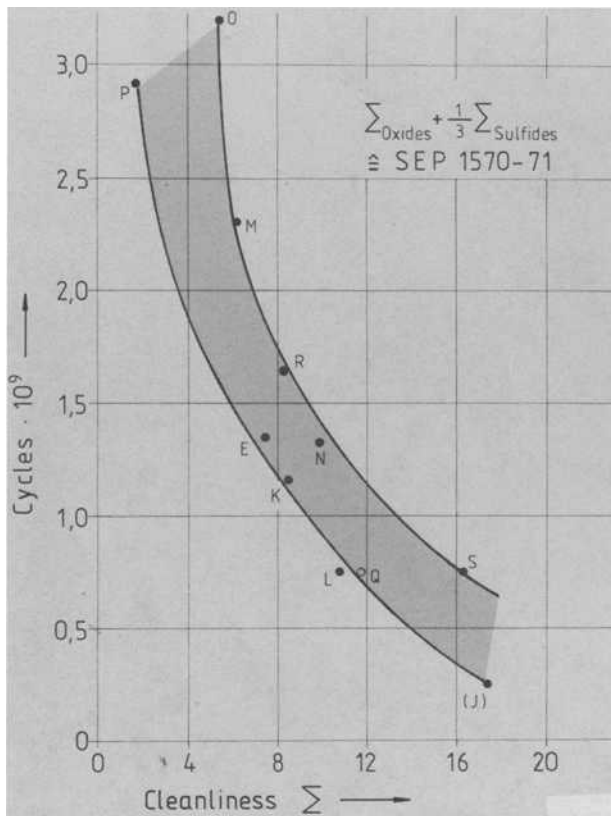


FIG. 12—Inclusion content and rolling contact fatigue of various heats of 100 Cr 6.

Contrary to soft-annealed, unalloyed, or low-alloy steels, it is assumed that the fatigue life of high-alloy steels is substantially affected by material discontinuities. Nonmetallic inclusions in the raceway subsurface of elements in rolling contact are internal stress risers which influence the fatigue process in two respects: in the first phase they enhance the plastic deformation of martensite by locally reducing the yield strength, and they facilitate the formation of microcracks; in the second phase of fatigue, the incipient cracks at the nonmetallic inclusions preferentially begin to grow.

Figure 13 shows the results of our studies. At high specific contact pressure, the entire highly stressed material volume is plastically deformed during the stress cycle. The high hydrostatic pressures increase the ductility of the material and reduce, particularly in the first phase of fatigue, the adverse influence of the nonmetallic inclusions. Due to the interaction of slip bands and nonmetallic inclusions, microcracks develop and generate, in the second phase of fatigue, subsurface cracks. The residual stress condition existing at that moment plays an important role.

With decreasing contact pressure, the effect of the nonmetallic inclusions on the life increases. The plastic deformations concentrate increasingly on areas with nonmetallic inclusions, and the nonmetallic inclusions act as incipient cracks.

Nonmetallic inclusions reduce the life in the fatigue strength range as well as the endurance

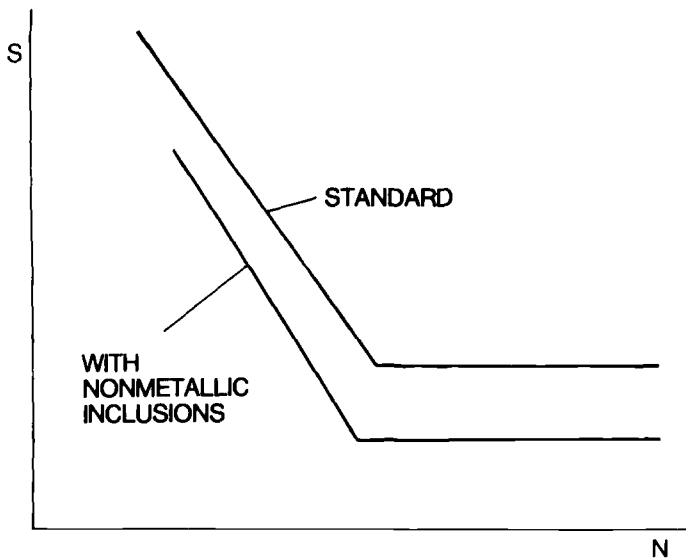


FIG. 13—*S-N curve, effect of nonmetallic inclusions.*

limit under rolling contact stressing. If, by chance, life tests are run at the load level as the investigations according to Fig. 12, the endurance limit of extremely clean material may be above the testing load.

The critical material stressing, which causes the failure of an element in rolling contact, is restricted to a narrow zone below the raceway. The existence of a "critical" nonmetallic inclusion in this highly loaded zone must be considered under the laws of statistical probability. This must be a guideline for discussions about the influence of type (chemistry), size, shape, distribution, and direction of nonmetallic inclusions.

In the fatigue strength load-range, one somewhat less critical nonmetallic inclusion can cause the failure of an element in rolling contact. The endurance limit during rolling contact fatigue is determined by the most critical nonmetallic inclusion which happens to be located in the zone of highest material stressing. That means that each element in rolling contact has its own endurance limit. Major scattering in life in the lower load range of the fatigue strength range and in the load range of the endurance limit is due to this fact.

Effect of Material Strength on Fatigue Behavior

Earlier investigations into rolling contact were carried out under very high contact pressures [14,39-42]. It is shown that life is increased with increasing hardness of the elements in rolling contact.

Figure 14 illustrates that with a contact pressure of $p_0 = 2800 \text{ N/mm}^2$, rings which were differently tempered and were within a hardness range of 58 to 64 HRC had practically the same life [14]. This appears plausible if the nonmetallic inclusions are considered to have a very important effect. Therefore, strength and nonmetallic inclusions cannot be discussed independently of each other. Figure 15 illustrates our concept of the effect of material strength.

We assume that nonmetallic inclusions reduce life to a different extent, depending on the specific contact pressure. At high contact pressures, the zones of highest material stressing are plastically deformed to such a high degree that the entire stressed material volume is largely

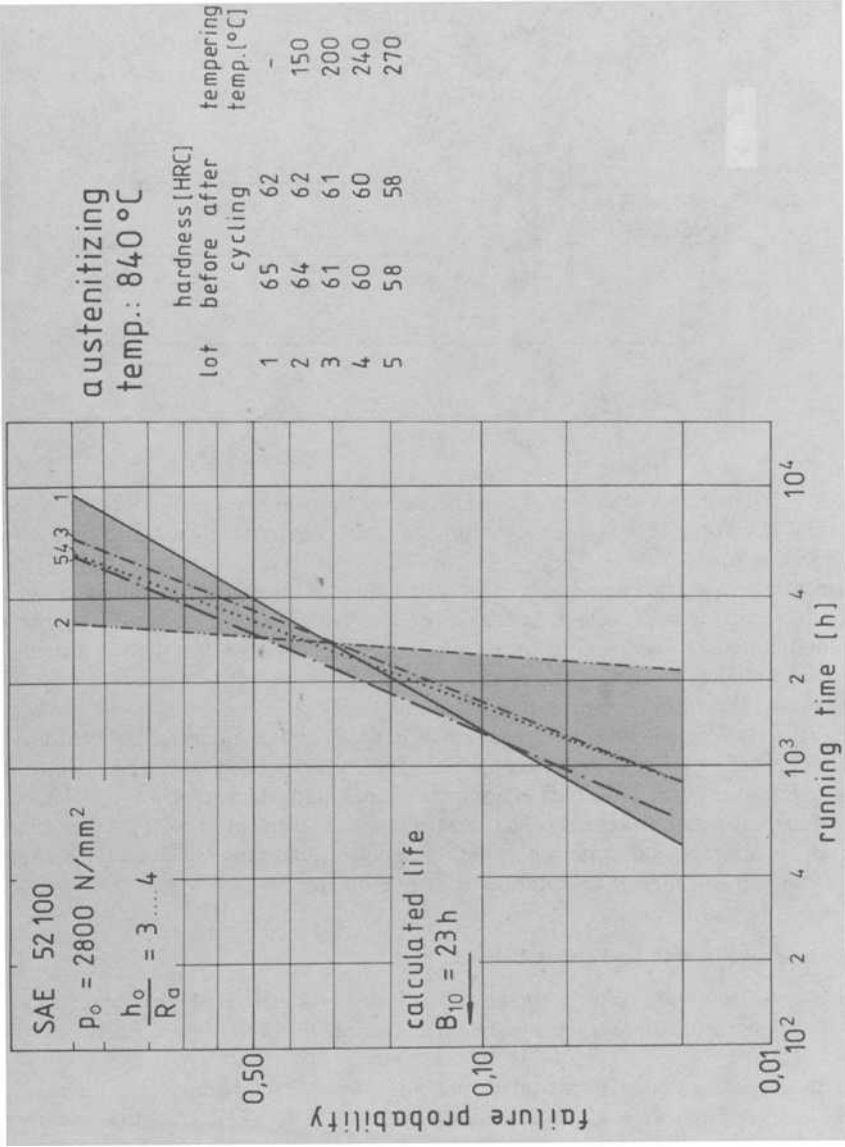


FIG. 14—Running times of ball bearings of different heat treatments.

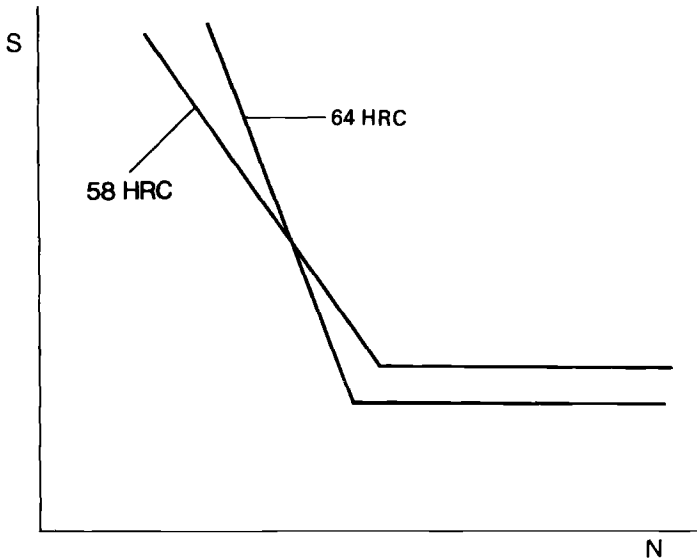


FIG. 15—*S-N curve, effect of material strength.*

homogenous. This causes on the one hand a reduction of the notch stresses at the nonmetallic inclusions and on the other hand a reduction of the strength. With increasing hardness, the resistance to plastic deformation rises, which explains the dependence of the life on the hardness.

The lower the contact pressure, the more the plastic deformation locally concentrates on nonmetallic inclusions. High yield strength retards the plastic deformation in the first phase of fatigue. Ductility makes the formation of cracks more difficult in the second phase of fatigue. If these statements are right, an increase of the yield strength and an improvement of the ductility cause an increase in life in the lower fatigue strength range and an increase of the rolling contact endurance limit. The more inclusions in a steel, the greater the positive effect of high ductility becomes.

Influence of Residual Stresses on Fatigue Behavior

Investigations have shown that compressive residual stresses can increase the life of elements in rolling contact [11,33,43,44]. This is explained in the distortion energy hypothesis by a reduction of the material stressing [45]. This is simplified in Fig. 16, where the residual stresses are assumed to extend parallel to the surface in the x - and y -directions at a constant depth. Tensile residual stresses increase the equivalent stress.

The effect of compressive residual stresses on the structural changes can be proved also by experiment (Fig. 17). Rolling contact elements of nearly the same hardness were subjected, on a roller test rig, to 10^8 load cycles under a contact pressure of $p_0 = 3300 \text{ N/mm}^2$. Subsequent X-ray diffraction analysis showed that the plastic deformations (illustrated by the reduction of half-value breadth) and the change of the residual stress condition were very much retarded by compressive stresses of about 200 N/mm^2 [11].

Since, on the one hand, the alternating shear stress hypothesis cannot take into account the influence of residual stresses [46] and, on the other hand, the residual stresses have a strong effect on the procedure in the first phase of fatigue, we expect that (Fig. 18) the compressive residual stresses produce an increase of the rolling contact endurance limit and, at least in the

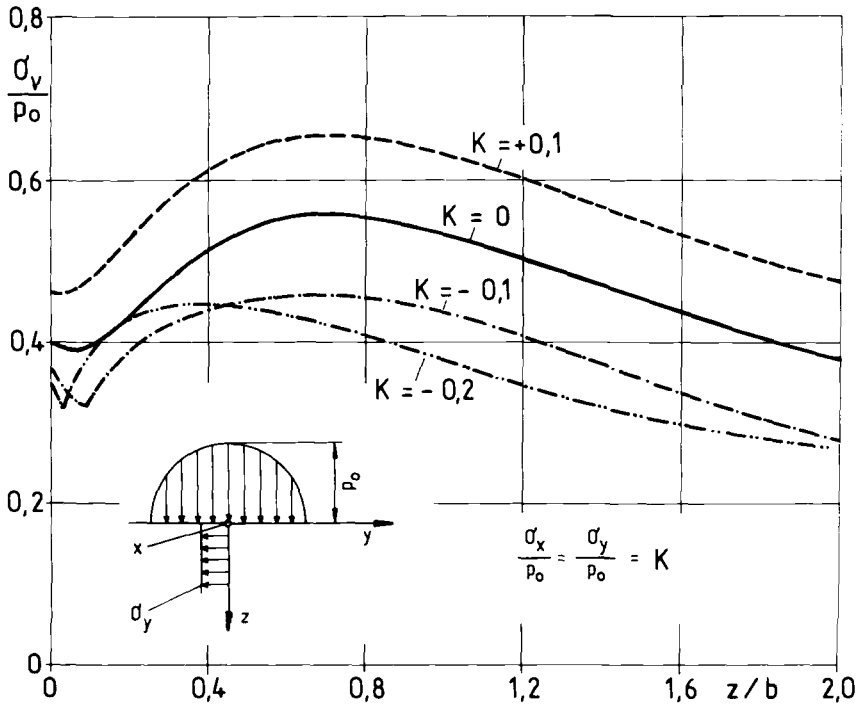


FIG. 16—Effect of residual stresses on equivalent stresses.

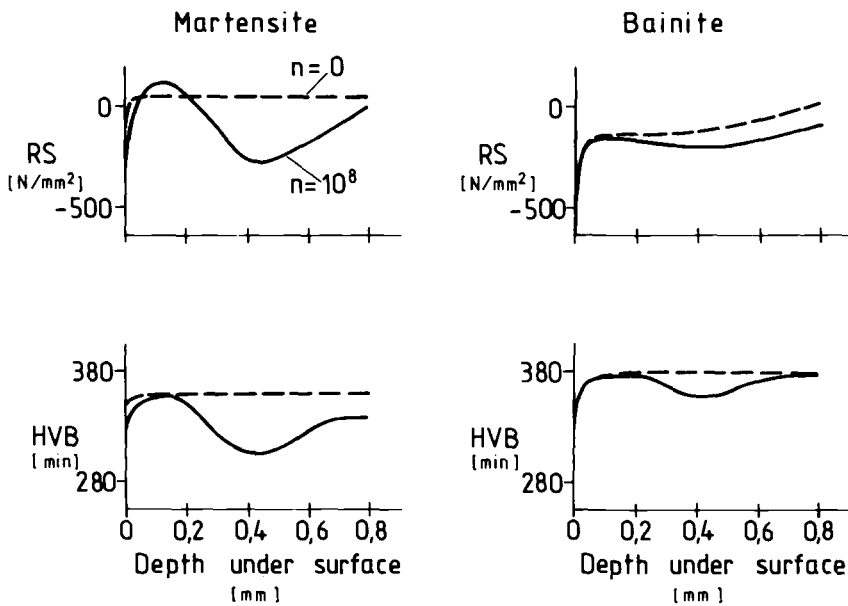


FIG. 17—Changes of residual stress and half-value breadth through cycling.

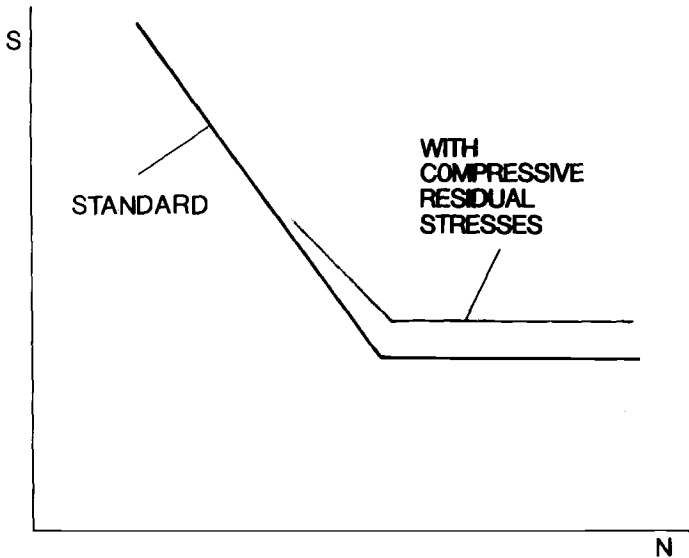


FIG. 18—*S-N curve, effect of compressive residual stresses.*

lower fatigue strength range, in a substantial increase of the life. It can be assumed that the heavy plastic deformations occurring under very high contact pressures largely cancel the effect of residual stresses.

Summary

The formation of spalling in the raceway of elements in rolling contact operating under elasto-hydrodynamic conditions is due to material fatigue. From our own present research work and from published research, we developed a hypothesis on the fatigue process. Starting from this hypothesis, the influence of material properties on bearing life was explained, which is summarized below.

1. The fatigue process takes place in two phases. Phase 1 is characterized by material stressing, which can be explained by means of the distortion energy hypothesis. Plastic deformations of the martensite cause residual stresses and the formation of microcracks. These are preconditions for the generation of subsurface cracks in Phase 2 of fatigue. The procedures in Phase 2 of fatigue are decisively influenced by the alternating shear stresses and the multiaxial residual stresses. Phase 1 is decisive for the duration of fatigue.

2. Nonmetallic inclusions accelerate the plastic deformations in Phase 1. They are a main cause for the formation of cracks in Phase 2. The effect of nonmetallic inclusions on the life depends on the contact pressure. The rolling contact endurance limit is reduced, and fatigue strength is reduced particularly in the lower load range.

3. The effect of the material strength cannot be separated from the effects of nonmetallic inclusions. A high-yield strength and high ductility (hardness between 60 and 58 HRC) help, at low contact pressures, to retard the plastic deformations and to prevent the growth of cracks; so they increase the rolling contact endurance limit and the life in the range of the low fatigue strength. At high contact pressures, high hardness is superior.

4. Compressive residual stresses influence the fatigue process in Phase 1 by reducing the material stressing thereby retarding plastic deformations. They increase the rolling contact

endurance limit and prolong the life in the lower range of fatigue strength. At high contact pressures, the effect of residual stresses can be neglected.

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Failsafe Rating of Ball Bearing Components

REFERENCE: Voskamp, A. P. and Hollox, G. E., "Failsafe Rating of Ball Bearing Components," *Effect of Steel Manufacturing Processes on the Quality of Bearing Steels, ASTM STP 987*, J. J. C. Hoo, Ed., American Society for Testing and Materials, Philadelphia, 1988, pp. 102–112.

ABSTRACT: Methods used traditionally throughout the bearing industry to calculate the life of rolling bearings predict a finite life under all conditions of application, even at very low contact stresses. More recently, endurance tests on heavily loaded 6309 deep-groove ball bearings have demonstrated that virtually infinite fatigue life is a practical reality for some applications, provided that bearings are accurately manufactured from high-quality steel. Accordingly, a new method to calculate rolling contact fatigue life has been developed in which an effective fatigue limit has been introduced. Metallographic and X-ray analysis of the highly stressed subsurface volumes of tested bearings have shown that a critical stress exists for a given material and test condition, below which microplastic deformation, transformation of retained austenite, and the development of residual stresses and texture do not take place, consistent with the concept of a fatigue limit.

KEY WORDS: fatigue, bearings, X-ray, retained austenite, residual stress, texture, preferred orientation, microstructural changes, life calculation

The most widely used material for the manufacture of rolling bearings is SAE 52100 type high-carbon, chromium-alloyed steel. This steel is normally given a martensitic heat treatment by which it is hardened to the range 58 to 65 HRC. Among the factors essential to attaining high structural integrity and long fatigue life are correct design, freedom from surface and other manufacturing defects, and a low level of injurious nonmetallic inclusions in the steel, particularly oxides and carbonitrides, as well as control of the microstructure in heat treatment. This paper will concentrate, in particular, on understanding the rolling contact fatigue process.

Bearing Fatigue Life

Methods traditionally used throughout the bearing industry to calculate the life of rolling bearings, embodied, for example, in the International Standards Organization (ISO) standard 281/1-1977 (E), are based on the work performed by Lundberg and Palmgren [1,2] in the 1930s and 1940s. The life calculation methods remain representative of the materials, manufacturing quality, lubrication, and test methods of those times. While these methods have been modified by introducing adjustment factors to accommodate improvements in life, particularly from steel quality improvements, they retain the essential feature of predicting a finite life under all conditions of application, even at very low contact stresses.

Many modern rolling bearing applications, however, suggest that service failures are rarely related to fatigue. Moreover, endurance tests on heavily loaded deep-groove ball bearings (6309) have demonstrated [3] that virtually infinite life is a practical reality, provided bearings are accurately manufactured from high-quality steel.

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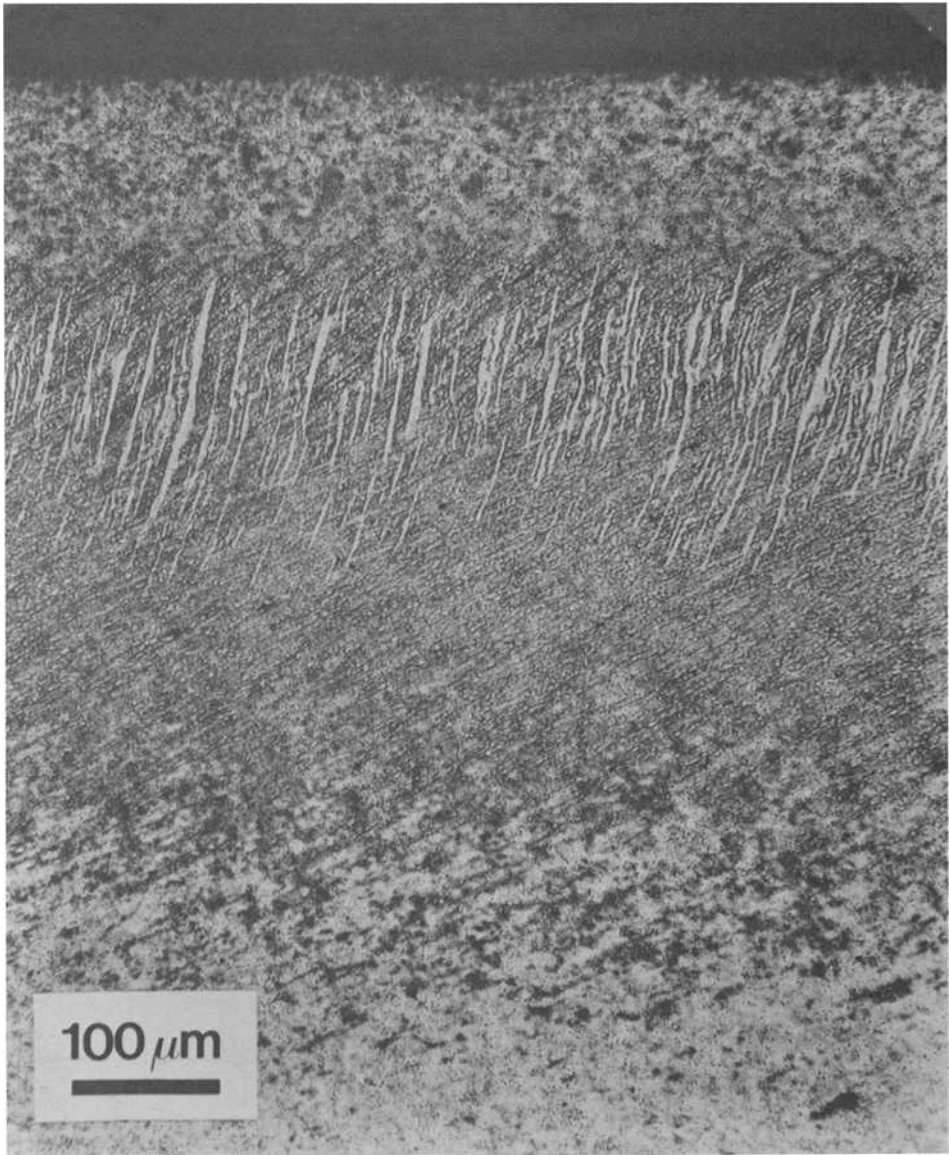


FIG. 1—Optical micrograph of axial section through center of ball track of tested, unfailed 6309 DGBB inner ring, showing zones of structural changes. Etchant: Nital. Test conditions: maximum contact stress, 3800 MPa; inner ring rotational speed, 6000 rpm; operating temperature, 53°C; full film oil lubrication. Test time was equivalent to 660 million revolutions [7].

As a consequence of these observations, a new life-prediction model has been developed by Ioannides and Harris [4] to take into account a number of limitations in the Lundberg-Palmgren theory. In their theoretical approach, they consider an array of relatively small subsurface material elements making up the stressed volume, V . These volume elements are exposed to a varying danger of fatigue in accordance with the prevailing local stresses above a threshold value, below which failure does not occur. As discussed further by Ioannides et al. [5], if we consider the orthogonal shear stress amplitude as an example of a fatigue criterion, that is, as used by Lundberg and Palmgren, the probability for failure ($\ell n 1/S$) given by the Ioannides-Harris relationship becomes

$$\ell n \frac{1}{S} \sim \frac{N^e (\tau_0 - \tau_u)^c V}{Z_0^h}$$

where

- N = fatigue life,
- τ_0 = maximum orthogonal shear stress,
- Z_0 = depth at which this occurs, and
- e , c , and h = constants.

In this expression, τ_u is the threshold orthogonal shear stress amplitude before failure can occur, this being dependent on factors including steel cleanness, condition of heat treatment,

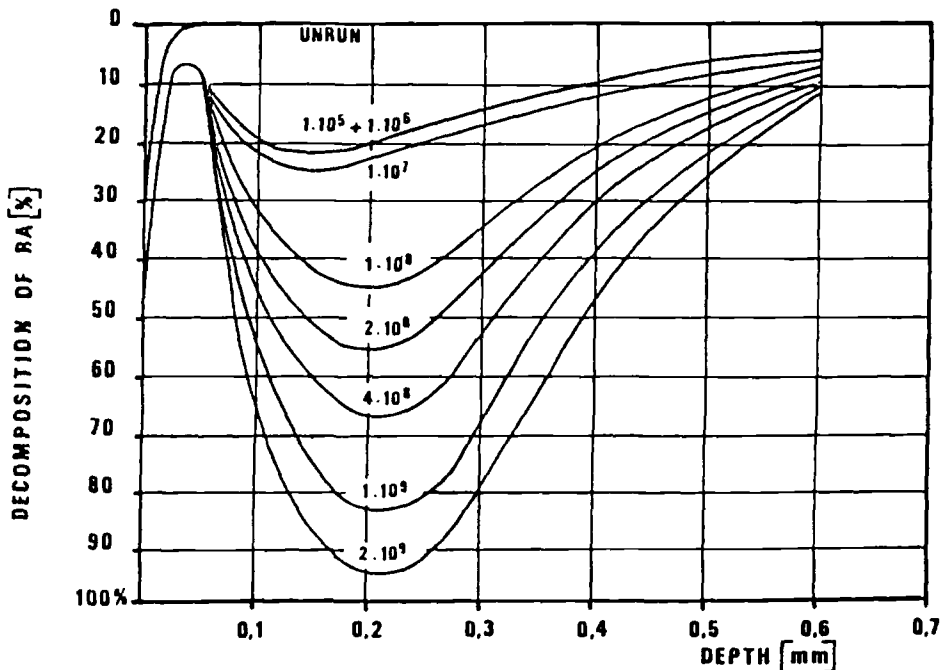


FIG. 2—Decomposition of retained austenite versus depth beneath 6309 DGBB inner ring raceway tested for various number of revolutions at a maximum contact stress of 3300 MPa.

and bearing operating temperature. The use of this approach predicts that the probability for failure asymptotically approaches zero at τ_w , that is, rolling contact fatigue exhibits a characteristic "S-N" (stress versus number of cycles to failure) curve, with a form similar to that well-known in structural fatigue [5].

Subsurface Changes in Rolling Contact

Microstructural changes in the subsurface regions of bearings were first reported in 1946 by Jones [6]. In a localized region under the raceway, corresponding approximately in depth to the highly stressed subsurface regions, differences in etching response of the microstructure are observed, shown in an advanced stage in Fig. 1 [7].

This phenomenon has been discussed in more detail by Voskamp [7]. Retained austenite measurements are a sensitive monitor for the onset of microstructural alterations. Long before any visible change occurs in the response of the subsurface regions to etching, retained austenite, present as a consequence of normal martensitic heat treatment, starts to decay. An example of the changes as a function of depth below the surface for an increasing number of revolutions in 6309 deep-groove ball bearing inner rings, tested at a maximum contact stress of 3300 MPa, is shown in Fig. 2, supporting the observations of a narrow localized band of transformation in the subsurface regions.

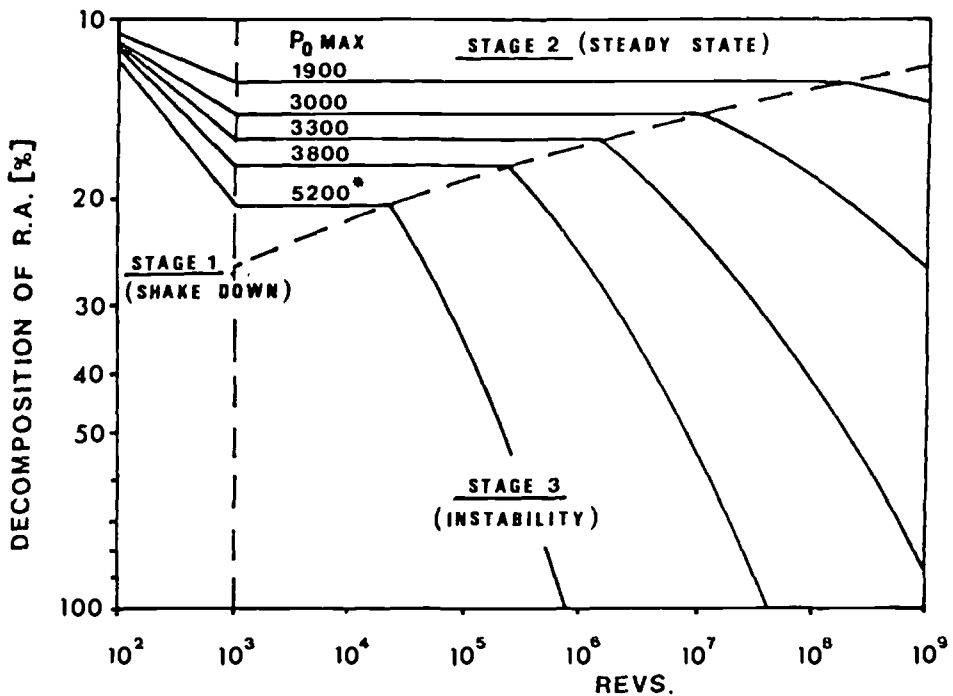


FIG. 3—Retained austenite decomposition at 0.2 mm below center of the ball track in a 6309 DGBB inner ring raceway measured versus number of revolutions for different maximum contact stresses P_0 (MPa). The solid lines represent the material response to rolling contact loading during the shakedown, steady-state, and instability stages. The diagram represents the material response for 6309 DGBB inner rings with a raceway radius of 9.05 mm at a rotational speed of 6000 rpm, under full film oil lubrication and at a bearing operating temperature of 53°C. *Modified raceway radius = 10.38 mm.

The result of systematic analysis of retained austenite decomposition at different contact stresses is shown in Fig. 3. The decomposition may be divided into three stages:

- Stage 1: Rapid decrease in retained austenite during the first few hundred revolutions of testing (shakedown),
- Stage 2: Period of retained austenite stability (steady state), and
- Stage 3: Further decomposition of retained austenite to levels below which it cannot be detected (instability).

The changes in retained austenite level as a result of the contact stresses are not the only measure of the changes in the subsurface regions. As discussed by Voskamp [8], compressive residual stresses develop in a plane parallel to the bearing track in directions parallel to and perpendicular with the rolling direction; these are shown in Fig. 4. In a direction normal to the surface, tensile residual stresses exist in localized subsurface regions of tested bearings, Fig. 5.

In addition to this, the existence of preferred orientation or texture has been shown to be a significant feature in Stage 3 of microstructural alterations [8]. The relative intensity of the (211) X-ray diffraction line with depth is shown in Fig. 6. Full pole-figure analysis confirms that the main preferred orientation developed corresponds with a (100) plane of bcc Fe parallel with the raceway surface, with [110] parallel to the rolling direction. Consequently, other corresponding {100} planes becomes aligned perpendicular to the raceway surface at 45 deg to the rolling direction.

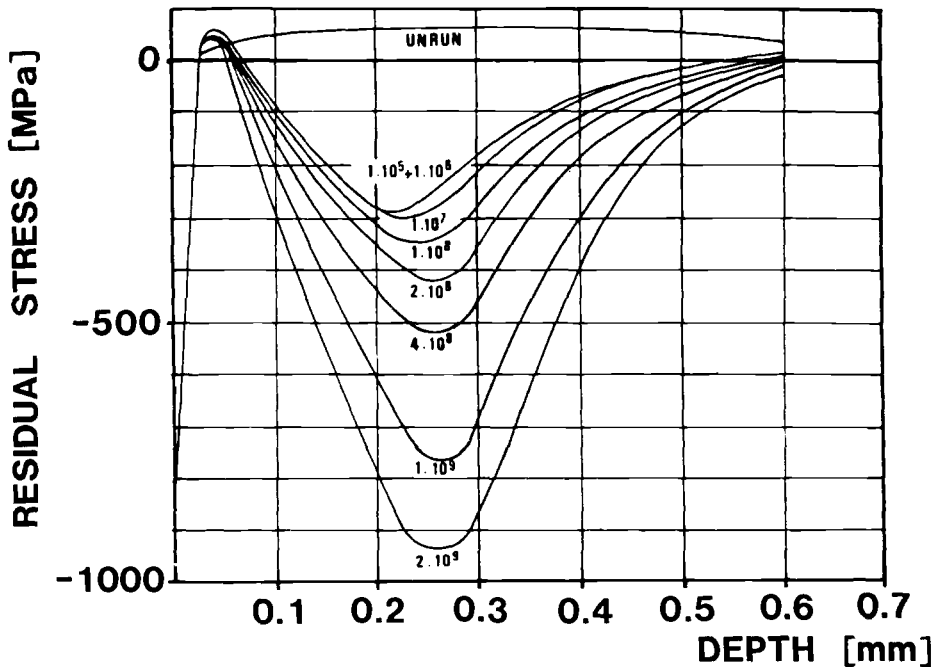


FIG. 4—Residual stress as measured in the circumferential direction versus depth beneath the inner ring raceway of 6309 DGBB inner rings tested for various numbers of revolutions. Test conditions: maximum contact stress, 3300 MPa; inner ring rotational speed, 6000 rpm; operating temperature, 53°C; and full film oil lubrication.

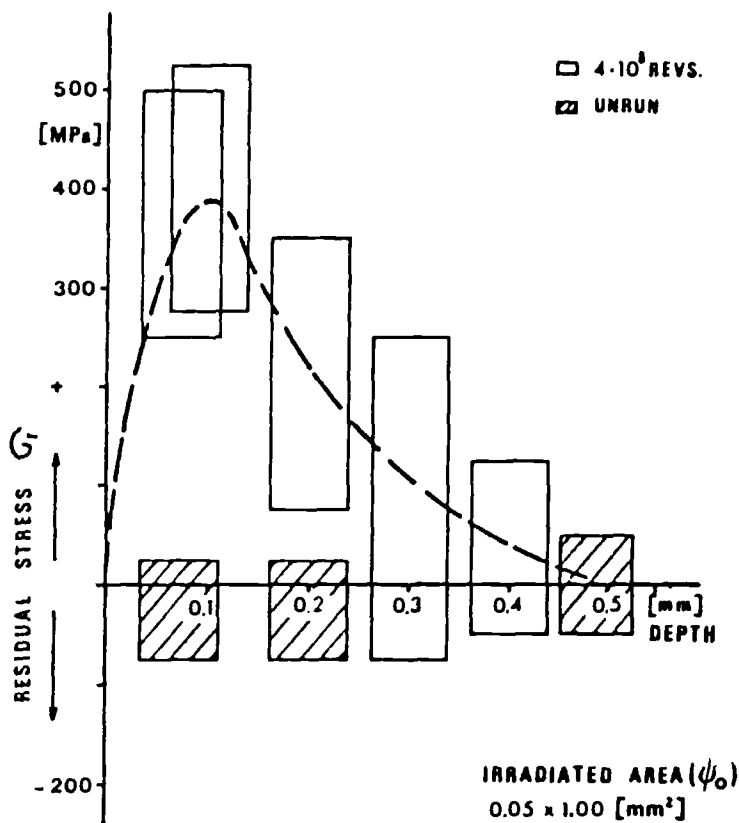


FIG. 5—Residual stress as measured in the radial direction versus depth beneath the raceway of unrun and overrolled 6309 DGBB inner rings. Test conditions as given in Fig. 1.

The tensile residual stresses in combination with the existence of a preferred orientation in a localized subsurface region of tested bearings is a powerful driving force for crack growth, particularly since the preferred orientation is such that the common fracture plane, $\{100\}$, in steel is aligned parallel to the surface and perpendicular to the developed tensile stresses. This mechanism is summarized in Fig. 7.

In clean steels, subsurface cracks have been observed under extreme pressures extending over large distances (Fig. 8) without causing a spalling fatigue failure. At lower stresses of testing, in steels with a high content of oxide inclusions, fatigue spalls in early failures tend to be small and irregular, Fig. 9. In later failures, particularly in cleaner steels, microstructural alterations allow spalls to propagate more extensively, and they are relatively "flat," Fig. 10. Over 95% of all failures in our experiments on such clean steels show advanced texture developments. Moreover, they tend to have a pronounced V-shape, with the sides of the V near 45 deg to the rolling direction, consistent with cracks developing perpendicular to the rolling plane along the additional fracture plane orientations, as a result of the pronounced texture development.

To account for these observations, it is suggested that crack initiation may start from several locations in the vicinity of stress-concentrating oxide or carbonitride inclusions, to give an irregular spall shape. In clean steels containing fewer and smaller inclusions, microplastic

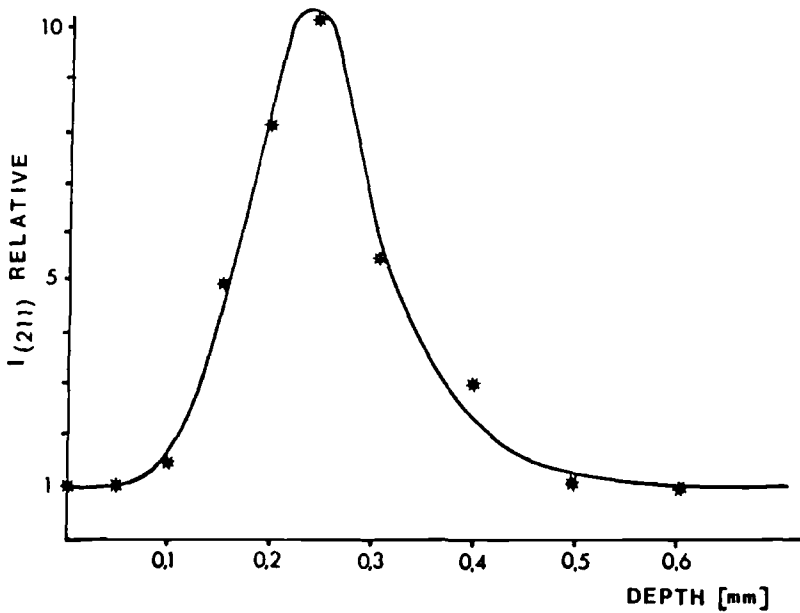


FIG. 6—The relative intensity of the (211) X-ray diffraction line, as measured in specimens prepared from tangential sections through tested 6309 DGBB inner rings at several depths below the center of the ball track.

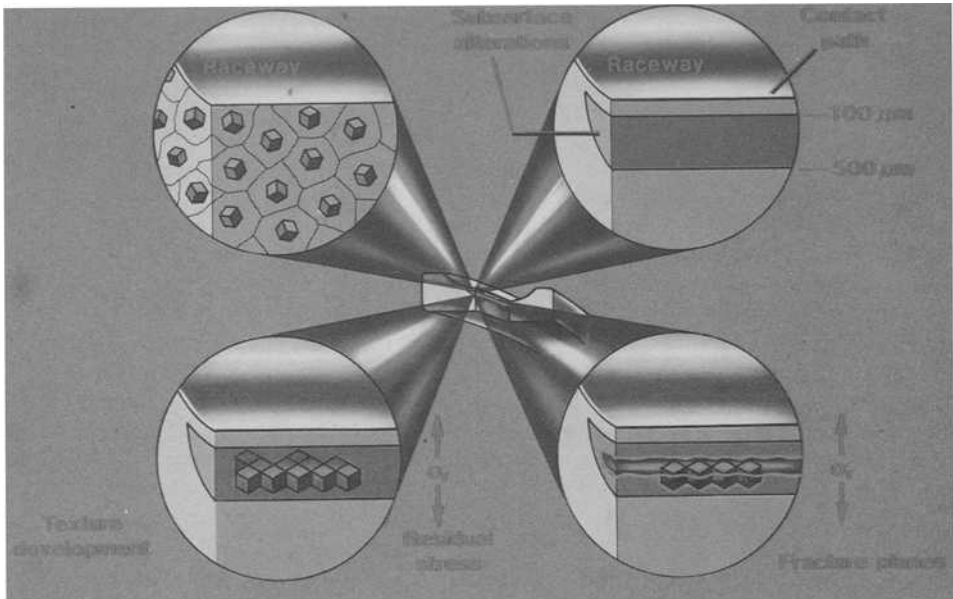


FIG. 7—Schematic representation of the interaction between the subsurface material changes developed in rolling contact. A tensile residual stress develops in combination with strong preferred orientation or texture development by which the common fracture plane of steel is oriented perpendicular to that stress.

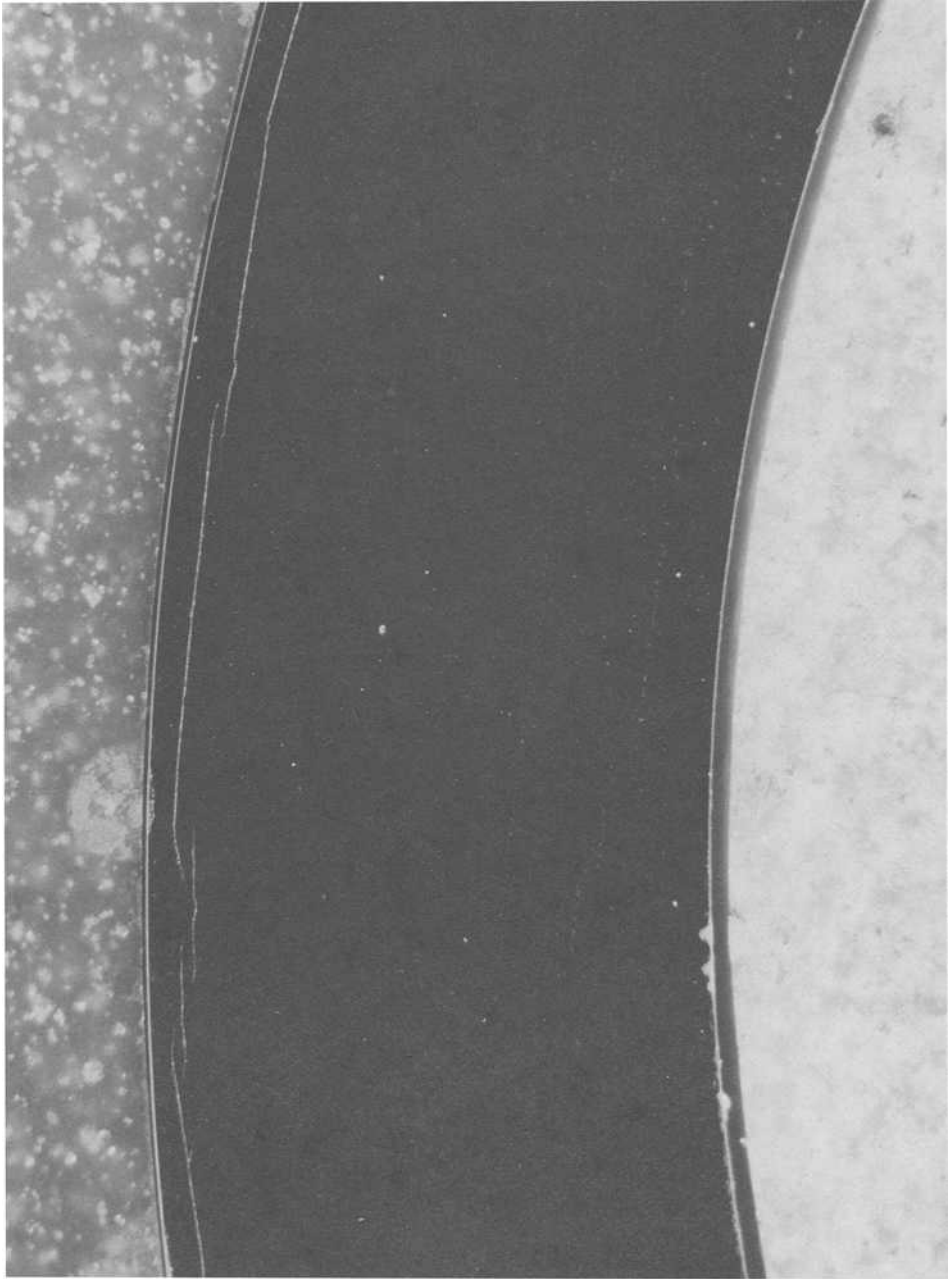


FIG. 8—Optical micrograph of axial section through center of ball track of a tested 6309 DGBB inner ring, showing several cracks developed at about 0.25 to 0.35 mm below the raceway without causing a spalling failure. Test conditions: maximum contact stress, 5200 MPa*; inner ring rotational speed, 6000 rpm; operating temperature, 53°C; total number of revolutions, 7.10^6 .

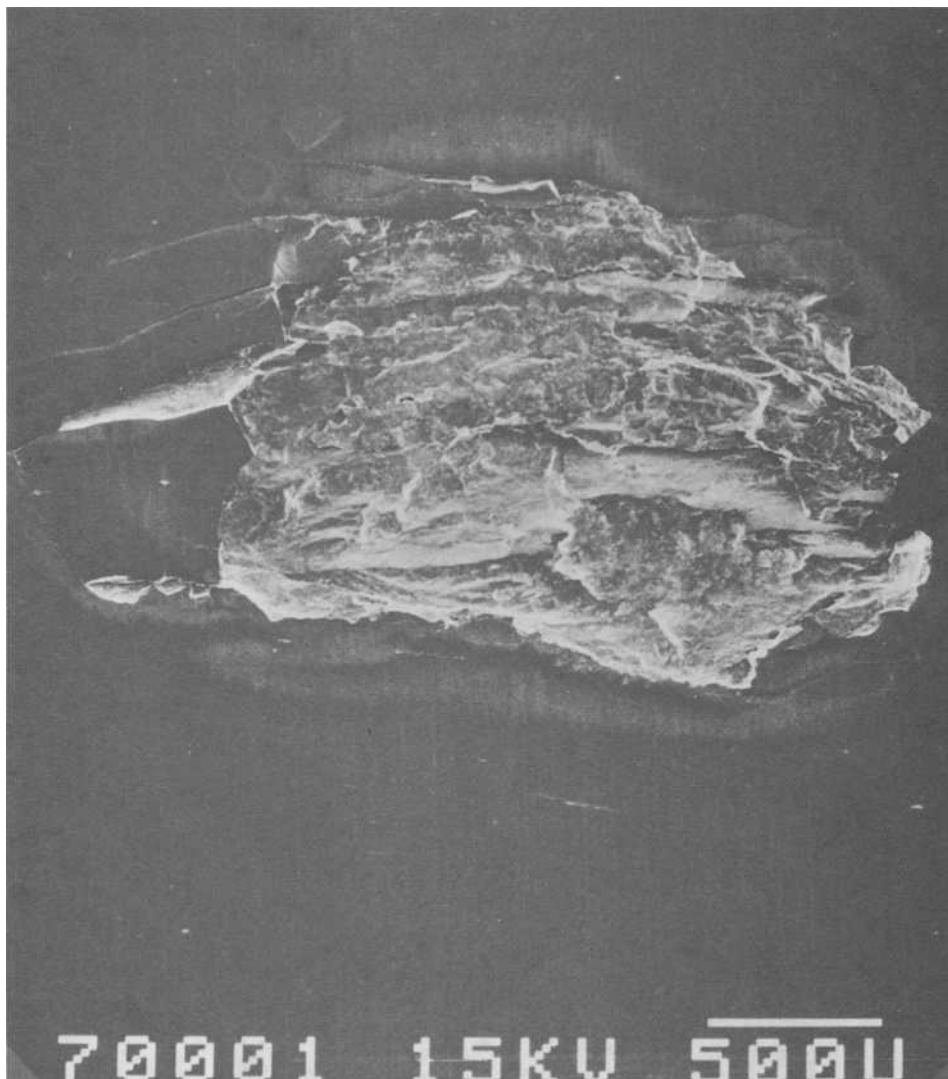


FIG. 9—Fatigue spalls in early failures, particularly in less clean steels, indicating their small and irregular nature.

deformation to Stage 3 instability and the development of preferred orientation allow the formation of more regular-shaped spalls in the region of discontinuities which were previously of noncritical size. Current test programs indicate that susceptibility to failure increases with Stage 3 development.

The local softening and microstructural changes in the subsurface regions of rolling bearings imply a different resistance to fatigue failure from that in untested bearings. The new model [4] can accommodate such variations by considering that the fatigue limit is a function of number of cycles of testing, depending, for example, on the combination of stress, frequency of stressing, and temperature.

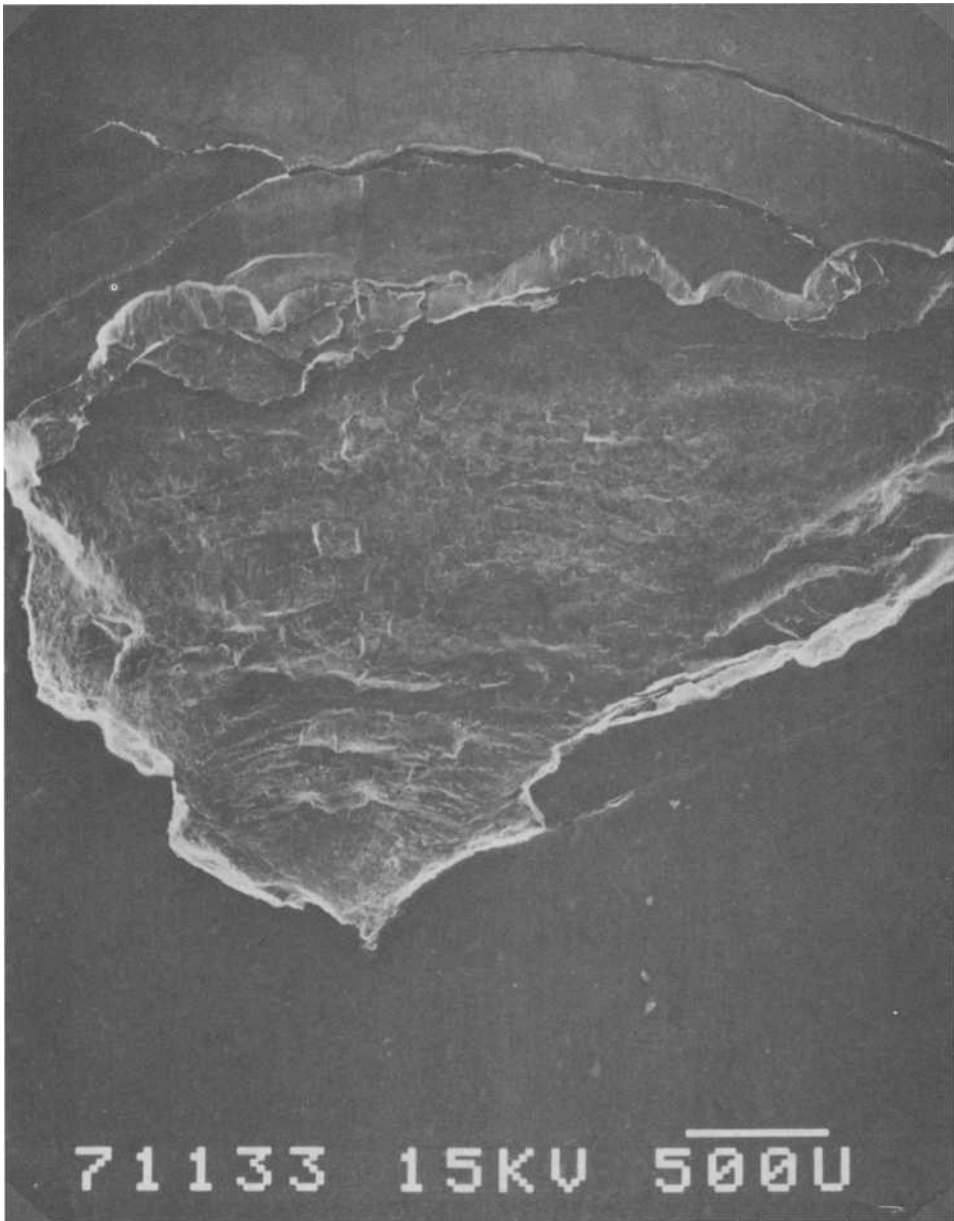


FIG. 10—Fatigue spalls in late failures, particularly in clean steels, indicating their flat-bottomed appearance as a result of texture development and their tendency to propagate more than in less clean steels. The pronounced V-shape is consistent with cracks developing at 45°C to the rolling direction along the additional fracture plane orientation.

The changes in the subsurface regions of tested bearings are a consequence of the accumulation of microplastic deformation by the effects of stress cycling. The onset of the decay is dependent on a number of factors, including load, frequency, and temperature. It is accompanied by local softening of the matrix. Of particular significance, the duration of a steady-state

stability decreases with increasing load. Conversely, at low contact stresses, the transition to instability only just commences beyond 10^8 revolutions.

This mechanism is consistent with the concept of a threshold stress for fatigue, in agreement with the analytical predictions of the Ioannides and Harris model [4].

Acknowledgments

The authors wish to thank the Managing Director of SKF Engineering and Research Centre for his kind permission to publish this paper. The contributions of Dr. E. Ioannides and Mr. H. Verschoor to this work are gratefully acknowledged.

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The Role of Carbides in Performance of High-Alloy Bearing Steels

REFERENCE: Pearson, P. K. and Dickinson, T. W., "The Role of Carbides in Performance of High-Alloy Bearing Steels," *Effect of Steel Manufacturing Processes on the Quality of Bearing Steels*, ASTM STP 987, J. J. C. Ho, Ed., American Society for Testing and Materials, Philadelphia, 1988, pp. 113-131.

ABSTRACT: Larger carbides inherent in high-alloy steels can initiate surface peeling failures and cause other raceway distress which affects bearing performance under thin lubrication film operating conditions. They can also contribute to the microstructural alteration in conventional subsurface rolling contact fatigue. Carburizing low-carbon, high-alloy bearing steel produces fine carbides comparable to low-alloy steels.

KEY WORDS: carbides, bearing steels, M-50 steel, rolling contact, surface failures, M-50NiL, peeling

Carbides are an essential phase in high-carbon, high-alloy steels used for rolling element bearings. Carbides provide wear resistance, inhibit grain growth, and are the reservoir of alloy which enables the steels to develop the desired properties during heat treatment. Increasing carbide content, however, lowers toughness and ductility, and carbide segregation found in high-alloy steels can, in extreme cases, cause erratic size and microstructural behavior during heat treatment. Investigators of bearing fatigue performance have long intuitively believed that the normally larger, more angular, and segregated carbides in high-alloy steels are less desirable than the finer, uniformly dispersed carbides in low-alloy, high-carbon bearing steels. Many private specifications by producers and users of bearing steels exist to control size and distribution of carbides in microstructures, although no ASTM or other industry-wide standard exists at this time.

The principal steel discussed in this paper is M-50, used extensively in American mainshaft gas turbine bearings, helicopter transmission bearings, and other aerospace applications. Other steels referred to are 18-4-1, used in many British mainshaft bearings; 440C, the most common stainless bearing steel; M-50NiL, a low-carbon, carburizing modification of M-50 developed to produce bearings with improved fracture toughness; and 52100, the most frequently used low-alloy, high-carbon steel for bearing components. Chemical compositions of these steels are listed in Table 1. The microstructures of M-50, 18-4-1, and 52100 in Fig. 1 illustrate the differences in carbide morphology between the high- and low-alloy steels. The carbide size and distribution shown for 52100 are typical of properly processed steel. In high-alloy steels, the size, shape, and segregation of carbides vary greatly with alloy content, steelmaking, solidification, and subsequent processing, especially reduction from ingot to finish size. Figure 1 and other photomicrographs of M-50 in this paper are examples of typical carbides, but not indicative of all of the variations that can occur.

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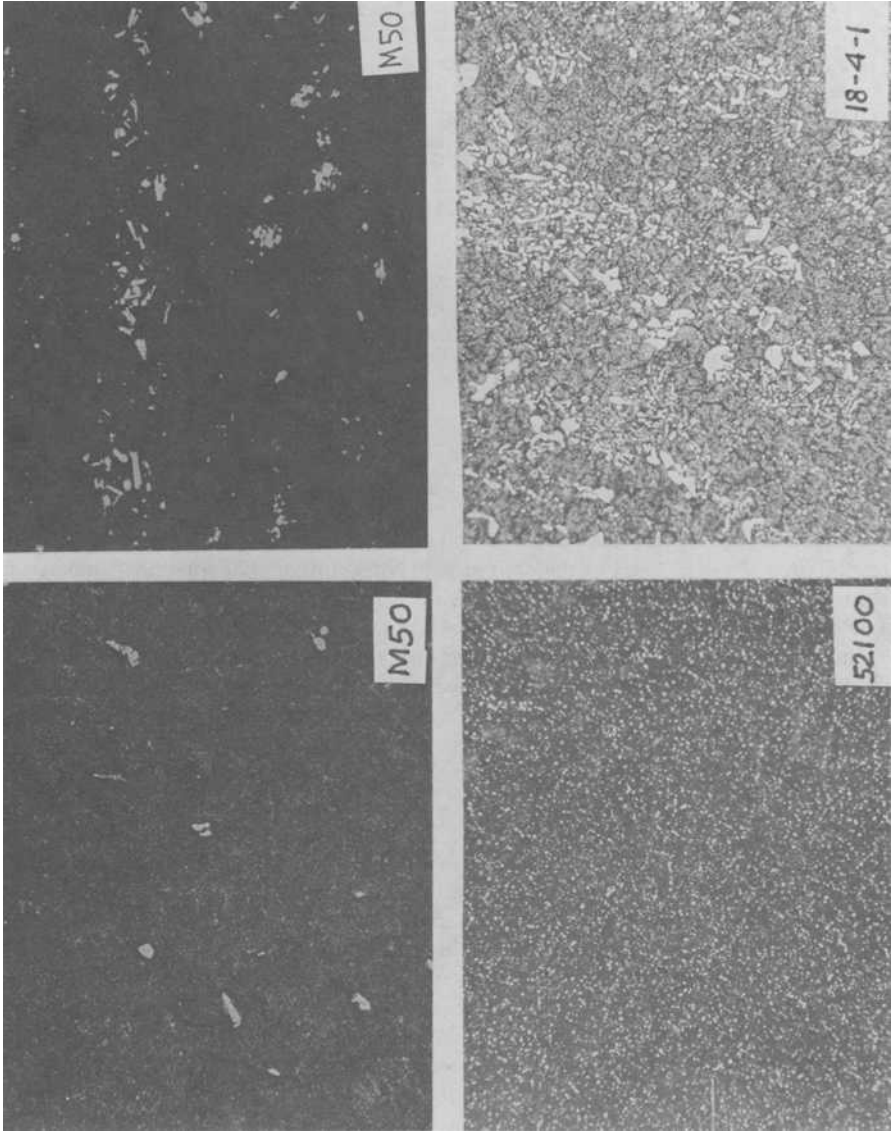


FIG. 1—Carbides in bearing steel microstructures. Nital etch, $\times 319$.

TABLE 1—Chemical composition, weight percent.

Steel	C	Cr	Mo	W	V	Ni
M-50	0.82	4.12	4.25	...	1.0	...
18-4-1	0.75	4.25	...	18.0	1.2	...
440C	1.05	17.25	...	...	...	...
M-50NiL	0.13	4.12	4.25	...	1.2	3.4
52100	1.04	1.45	...	...	...	...

The success of M-50 and 18-4-1 steel bearings in many applications proves that carbides are not defects in the sense of nonmetallic inclusions, and any effect of carbides on fatigue life is subtle. Rescalvo and Averbach [1] have determined that K_{Ic} fracture toughness depends mostly upon matrix properties with only a small influence of undissolved carbides. Chevalier, Zaretsky, and Parker [2] examined steels tested with the National Aeronautics and Space Administration (NASA) five ball fatigue tester and derived a carbide factor indicating increasing life with fewer, smaller, and better distributed carbides. Butterfield and McNelley [3] observed voids the order of one micron adjacent to carbides in M-50. Vincent, Coquillet, and Guiraldenq [4] found carbides at the fracture origin of M-50 steel in rotating beam tests.

This paper presents observations of tested bearings illustrating mechanisms by which carbides can affect performance and describes differences in carbides which relate to their behavior in bearings.

Experimental Procedure

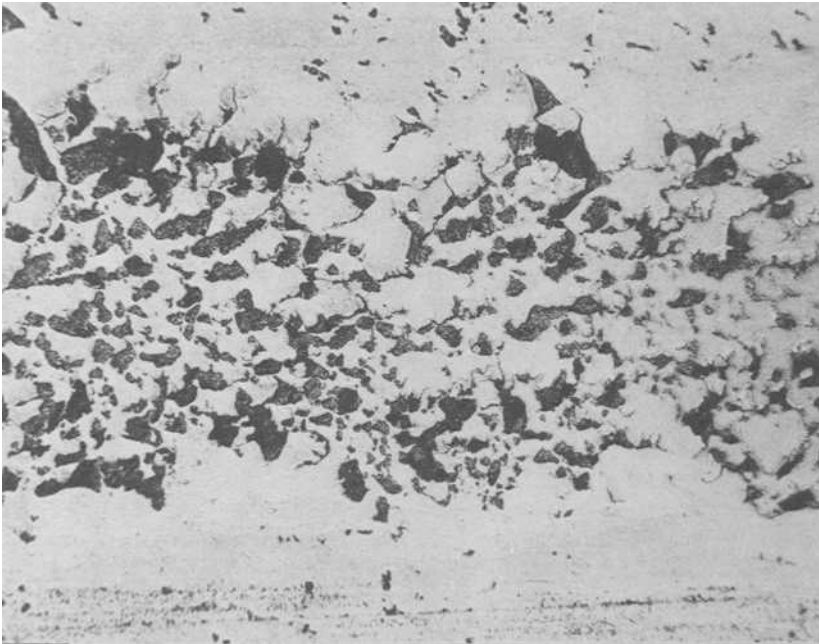
Observations from three test bearing sizes are included in this discussion. The 150 and 40-mm bore bearings were run in thrust with speed and lubrication to simulate mainshaft turbine bearing operating conditions. The 35-mm bore bearings were radially loaded in a conventional test of fatigue life. The bearings and testing conditions are described in Table 2. Testing included multiple steps of running for specific time periods with careful examination of surfaces and mapping of topographical features. Raceways and balls were examined at high magnification with optical and scanning electron microscopy (SEM), both methods being necessary to describe all details of the surfaces.

One manner in which carbides may effect performance is a contribution to surface roughness. Carbides are harder than the matrix steel and can stand up in relief. This effect is usually not observed on raceways finished by honing, but can be seen on the more highly polished surfaces of lapped balls. Carbide heights are measured with interference microscopy, estimated to one half an interference fringe, about 0.1 μm .

The primary carbides in M-50 steel have been identified by Bridge, Maniar, and Philip [5] as M_2C containing a high percentage of molybdenum and MC considered to be essentially vanadium carbide. M_2C is estimated to have a Rockwell C hardness in the mid 70s, MC in the mid 80s [6]. Metallographic techniques were used to distinguish the two types. For optical microscopy, specimens were first etched with Marble's reagent to darken the matrix and leave the entire $\text{M}_2\text{C} + \text{MC}$ carbide structure, followed by 4% sodium hydroxide (NaOH) saturated with potassium permanganate (KMnO_4) to darken the M_2C and leave only the MC unetched. With SEM, differences in electron absorption cause the M_2C to appear white and the MC gray with a back-scattered electron image. Microprobe analyses were also made to identify carbide type and composition.

TABLE 2—*Bearing and test parameters.*

Test Bearing Number	208	9130	207
Bore by O.D. by width	18 by 40 by 80 mm	35 by 150 by 225 mm	17 by 35 by 72 mm
Curvature, inner/outer	0.52/0.52	0.52/0.52	0.52/0.52
Pitch diameter	60 mm	185 mm	53.5 mm
Number balls, size	11 to 12.7 mm	22 to 22.2 mm	9 to 11.1 mm
Nominal contact angle	15 deg	30 deg	0 deg
Thrust load	907.2 kg	7711 kg	0
Radial load	0	14.5 kg	1111 kg
Speed	10000 rpm	7500 rpm	3000 rpm
AFBMA life	67.0 h	27.0 h	30.6 h
Inner maximum contact stress	2.30 GPa	2.16 GPa	3.62 GPa
Oil-in-temperature	148°C	94°C	54°C
Bearing O.D. temperature	148°C	148°C	66°C
Lubrication	MIL-L-7808 synthetic 2.0 CS @148°C	MIL-L-23699 synthetic 5.5 CS @94°C	petroleum 66 CS @52°C
Theoretical central film thickness	0.0762 μm	203 μm	0.711 μm
Lambda	0.59	1.57	5.5


FIG. 2—*Peeled raceway, $\times 100$.*

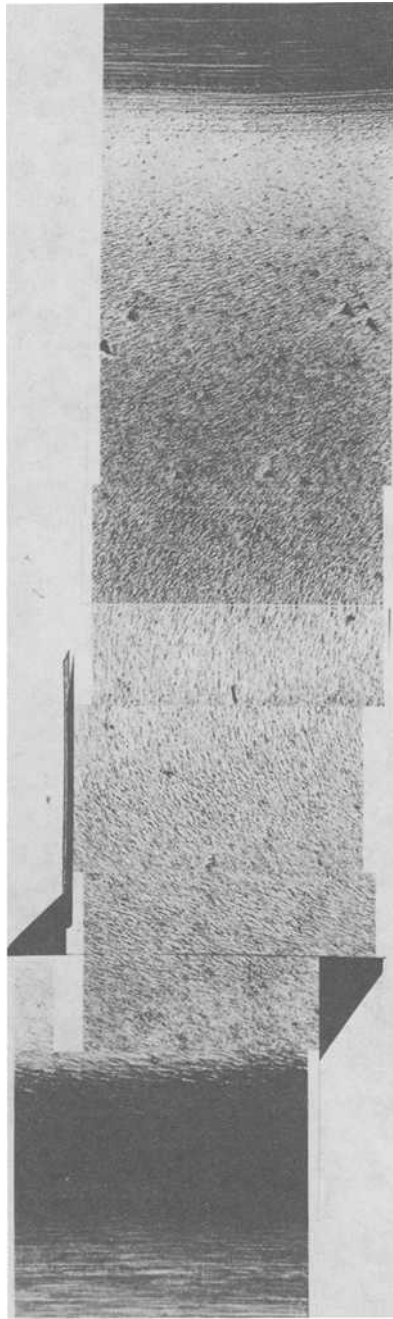


FIG. 3—Swirl pattern in raceway caused by action of carbides on balls. Beginning of peeling. (Rolling direction right to left.) $\times 35$.

Observations of Tested Bearings

The combination of imperfect rolling under thrust loading and thin lubrication film produces failure at the raceway surface by a process of peeling. Many small, shallow peels occur because surface asperities or contaminants break through the EHD film. A severely peeled raceway is shown in Fig. 2. The beginning of a peeling process is shown in a 150-mm bore inner raceway in Fig. 3. A swirl pattern has developed in the direction of ball spin, caused by carbides on the ball surface rubbing across the raceway, and peeling is developing in this zone. This bearing operated 1 hour with carbide height on balls up to $1\text{ }\mu\text{m}$.

A 150-mm bore raceway run 1 h with $\frac{3}{4}\text{-}\mu\text{m}$ carbides on balls is shown in Figs. 4 through 7. Each individual peel is triangular with a point origin (Fig. 4). The direction of peeling onward from the origin is the spin, not the rolling direction. Primary carbides of M-50 steel are at the origin of every peel. Carbide origins are shown at higher magnification in Figs. 5 and 6. The carbide cluster at the origin in Fig. 6 appears to be breaking under the contact stresses of bearing operation.

A macroview of this process is shown in Fig. 7, in which a portion of the peeled raceway has been metallographically polished and etched. The peeling pattern in the raceway can be seen following the flow of primary carbides in the microstructure.

In a related action, the carbides breaking through the lubrication film cause flow of the metal on the raceway surface, producing a glazed appearance. This is seen as a white etching layer in the microstructure (Fig. 8), a bearing run 1 hour with $1\text{-}\mu\text{m}$ carbide height on balls. A primary carbide appears to be breaking up in this worked zoned. (Note that this white layer



FIG. 4—Peeling pattern in raceway. Nital etched to show carbides at apex of triangular peeled areas. $\times 200$.

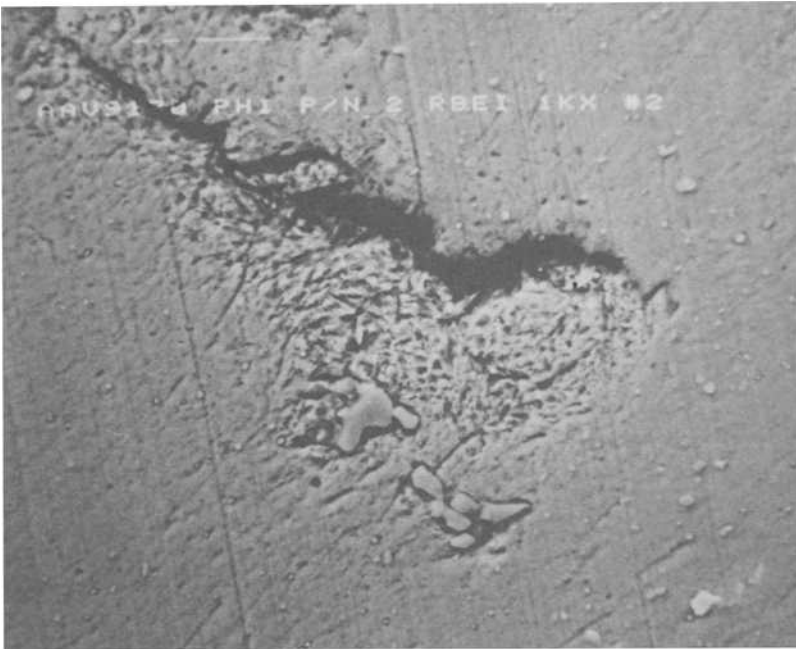


FIG. 5—Apex of peel. SEM, $\times 1000$.



FIG. 6—Cracked and broken carbides at apex of peel. SEM, $\times 3000$.

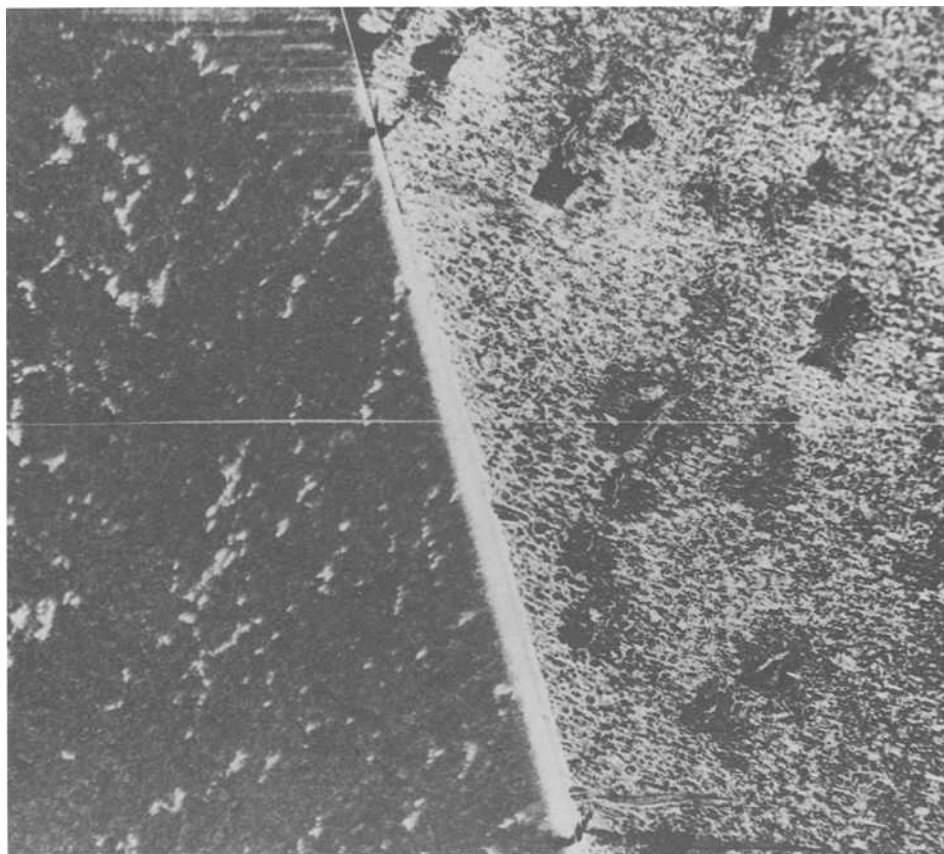


FIG. 7—Matchup of carbide pattern and peel pattern. Lefthand portion is microstructure just below raceway surface. Righthand portion is adjacent raceway surface. Nital etch, $\times 76$.

appears in the ball running path only and is far in excess of anything caused by honing of the raceways.)

Carbides on ball surfaces that cause raceway damage have been identified as the harder MC type and are shown in Fig. 9. During the initial period of bearing operation, the carbides wear, flatten out, and recede into the matrix, causing the damaging action to slow down and stop. A pattern of furrows from break-up during operation can be seen surrounding a carbide on the run ball in Fig. 9.

Peeling and glazing in the 150-mm bore bearings were created in short running times by abnormally high carbides on the ball surfaces. Carbide origins of peeling have been also found in 40-mm bore bearings with long running times, and with carbide height less than $\frac{1}{4} \mu\text{m}$ (10 millionths of an inch). Figure 10 shows isolated patches of peeling in otherwise normal raceways after 147 h operation. The primary cause of the peeling is the furrows in the honed raceways; however, magnification reveals that each peel is beginning at a carbide. Figure 11 shows a carbide cracking in the zone of peeling action.

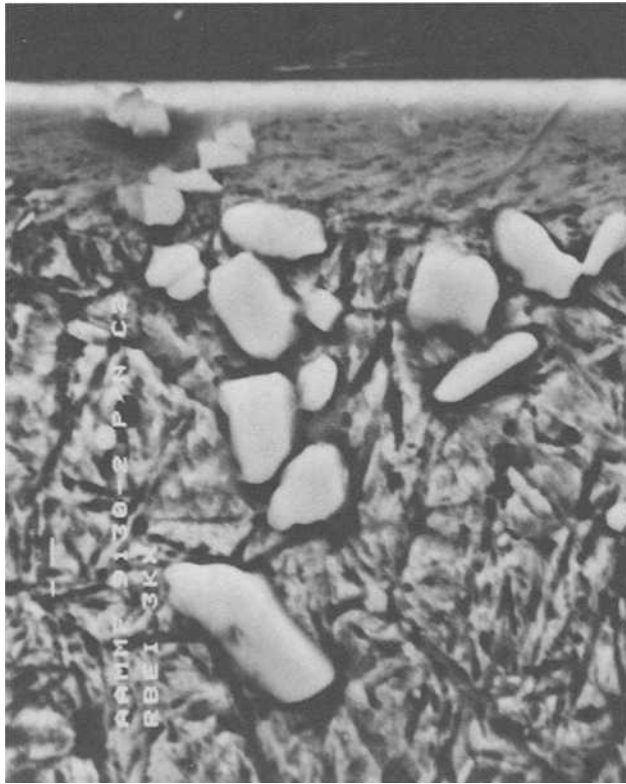


FIG. 8—Microstructure of cut section of inner with glazed raceway. Nital etch, $\times 3000$.

Carbides also can have a role in the creation of furrows. Figure 12 shows a furrow in a honed raceway caused by a carbide breaking out and scratching the surface. The $\times 1000$ magnification is much higher than is usually used to evaluate and control manufactured surfaces. Large carbides can also become embedded in honing stones and contaminate honing oil to cause scratching.

A failure from this type of furrow is shown in Fig. 13. Peeling around the furrow has led to a macro spall, the dark area at the left. The hydraulic film breaks up as it flows into the furrows leading to burnishing at the edges, peeling at carbides in this area, and eventually to macro-spalling.

Carbides can also act as stress raisers, contributing to the microstructural alteration that leads to fatigue in the subsurface zone of high contact stresses. Figure 14 is a 440C stainless steel inner ring run in the 35-mm bore test for 2100 h. The microstructural alteration is predominantly at the larger carbides. This is a long-life phenomenon at stresses not usually found in service, but it demonstrates the stress-raising effect of the carbide and may account for the negative influence of large carbides in results from highly stressed contact fatigue test rigs.

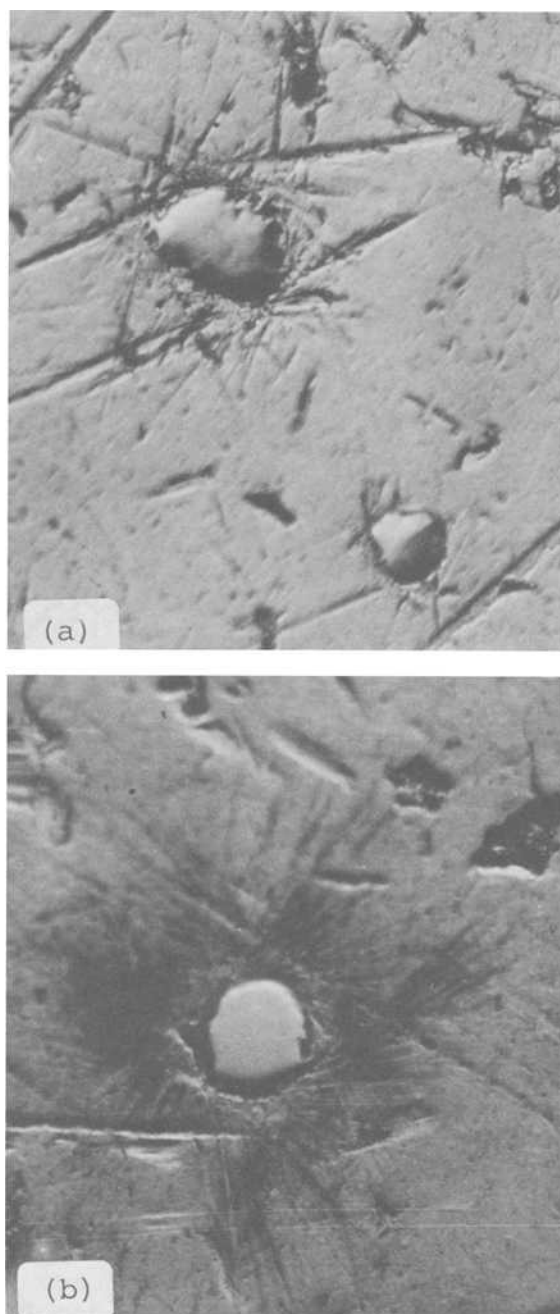


FIG. 9—Carbides on ball surfaces. (a) *Before running.* (b) *After one hour operation.* $\times 1000$.

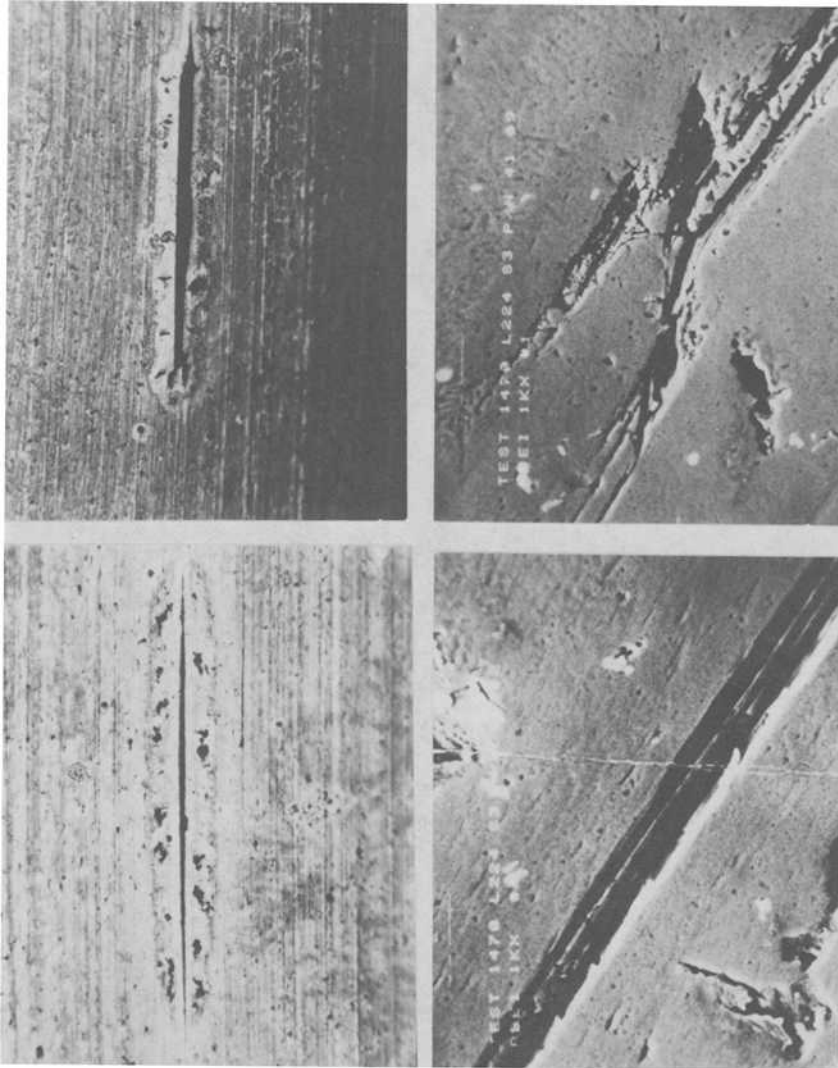


FIG. 10—Microspalls at carbides in vicinity of furrows in 40-mm bore inner raceway. Top: optical $\times 64$. Bottom: SEM $\times 640$.

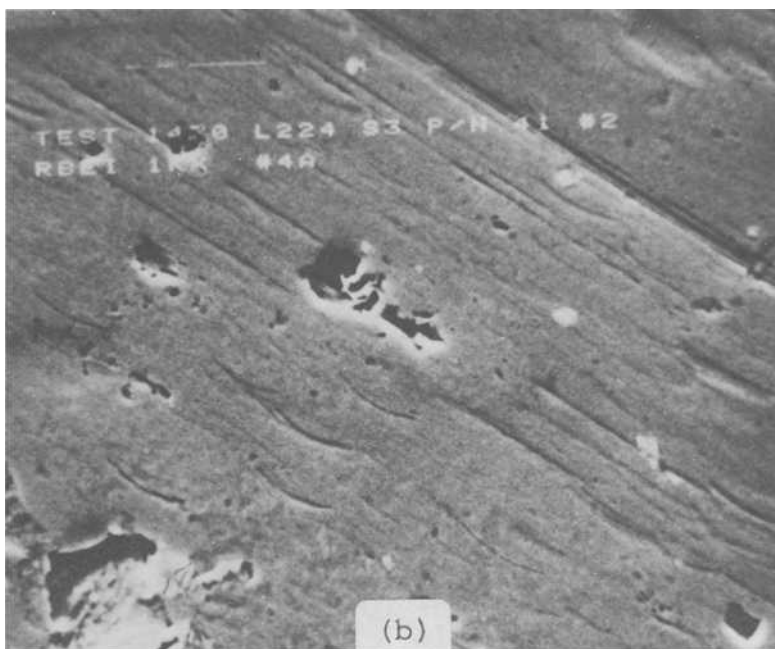
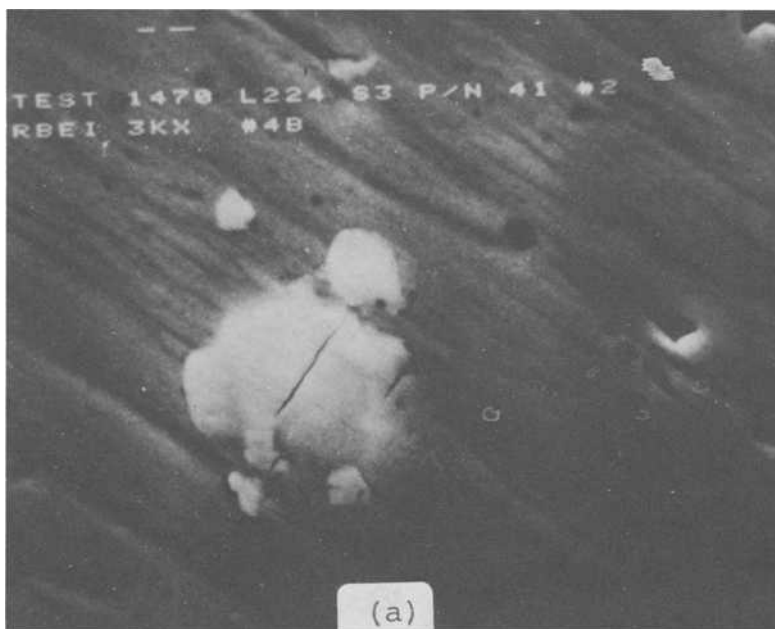


FIG. 11—Damage at carbides in 40-mm bore raceway. (a) Cracked carbide at $\times 3000$. (b) Pitting at carbides, $\times 1000$.

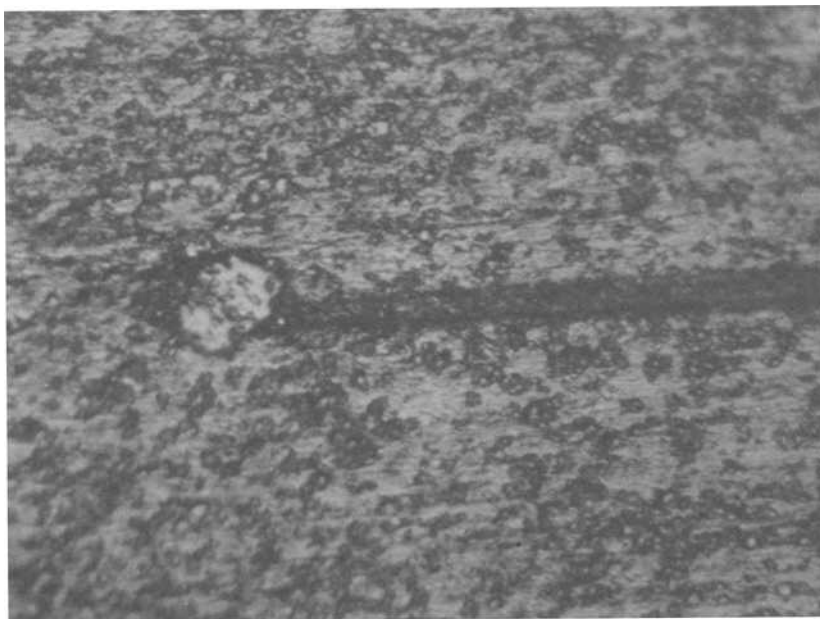


FIG. 12—Furrow in honed raceway caused by carbide. $\times 1000$.

Carbides in M-50 Steel

The carbides observed to effect bearing performance are the larger, primary carbides which are insoluble at normal hardening temperatures. Chemical compositions of primary carbides in hardened and tempered specimens from four heats of steel are listed in Table 3. M_2C carbide is by weight per cent approximately 62 molybdenum, 14 vanadium, 11 chromium, and 6 iron. MC has approximately 41 vanadium, 37 molybdenum, 6 chromium, and 2 iron except in Heat B, which had higher vanadium and lower molybdenum.

TABLE 3—Carbide compositions, M-50 steel, weight percent.

Element		Heat A	Heat B	Heat C	Heat D
MC	Fe	2.9	2.1	1.7	2.4
	Cr	5.9	5.4	5.7	5.9
	Mo	37.2	27.3	37.6	37.5
	V	41.2	45.1	40.1	42.1
M_2C	Fe	6.8	5.6	^a	5.3
	Cr	11.5	11.6	^a	10.2
	Mo	62.5	59.9	^a	64.5
	V	13.0	14.0	^a	14.4

^a M_2C carbides too small to analyze.

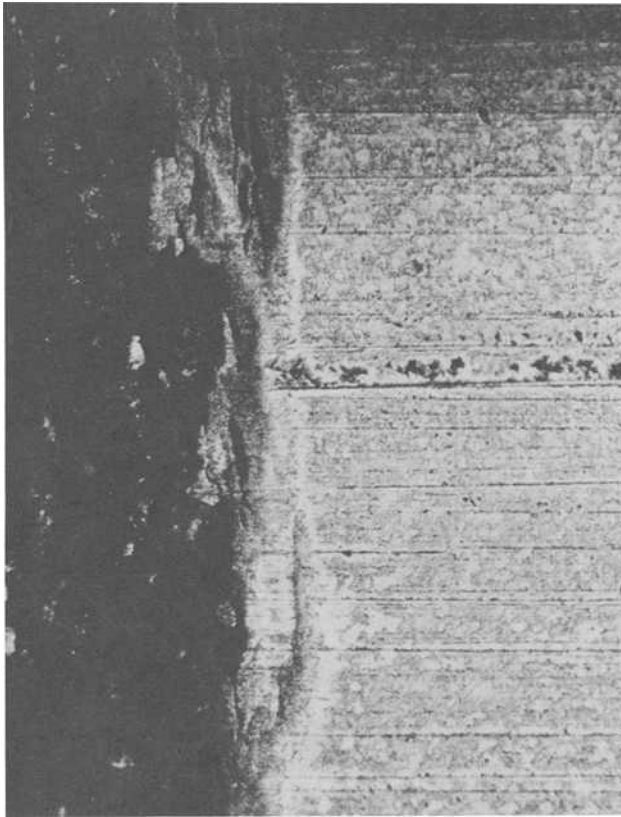


FIG. 13—Furrow with peeling and macrosplatter. $\times 50$.

The carbides in three heats of steel are shown in Fig. 15, microstructures of longitudinal specimens. Figures 15a and 15c are from 150 mm square bars, 15b from 100 mm square. The segregation pattern in Heat A is predominantly M_2C with similar size MC particles interspersed. Heat B, although from a different steelmaker, is similar to Fig. 15a. Heat C received a heat treatment to reduce segregation by which ingots were soaked for several hours at temperatures near the melting point (1175 to 1205°C). This heat treatment produces a different microstructure of predominantly MC carbides, some as large as 25 μm , with M_2C found only as very fine particles.

A fourth Heat D from a third steelmaker is shown in Fig. 16. This specimen was cut from a forged ring and is not exactly longitudinal; however, carbides appear comparable to those in Heats A and B.

Carburizing M-50NiL, a low-carbon modification of M-50, is used to produce bearing rings with improved fracture toughness while maintaining the hardness and elevated temperatures properties of M-50. Introduction of carbon during carburizing instead of during steelmaking produces fine, well-dispersed carbides comparable to low-alloy, high-carbon steels. The microstructure of an M-50NiL heat treated case with carbon content similar to standard M-50 steel is shown in Fig. 17.

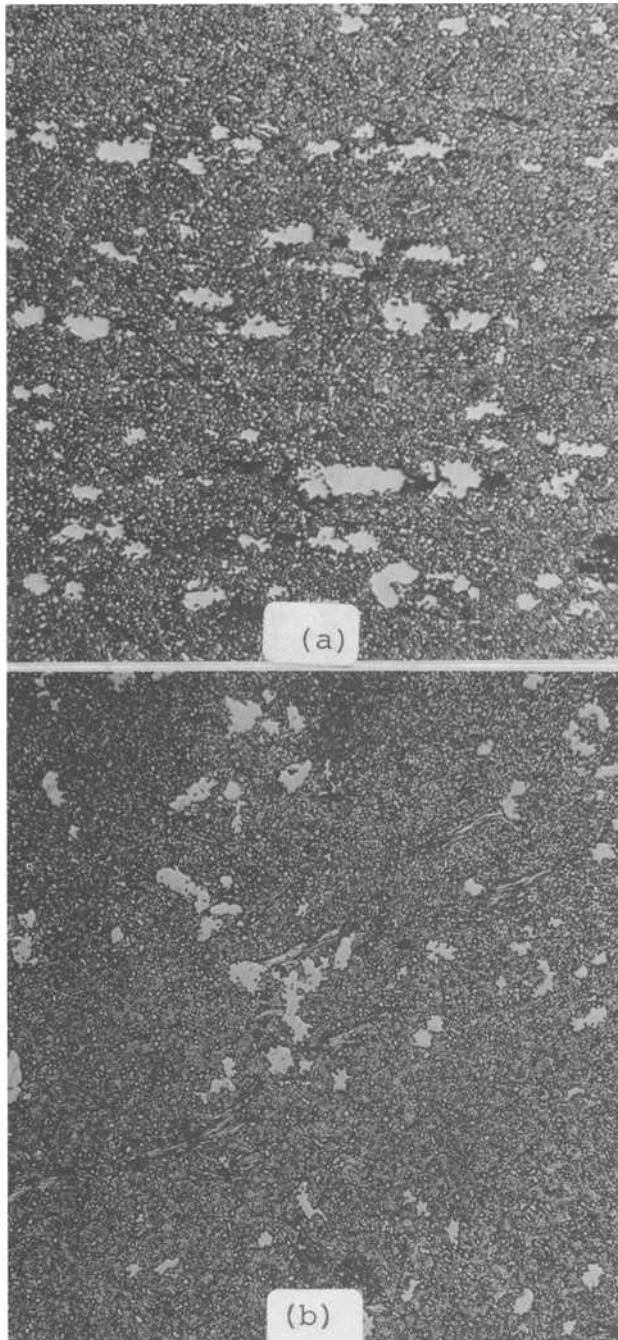


FIG. 14—440C steel, 35-mm bore inner in zone of maximum subsurface shear stresses. $\times 500$. (a) Cross section. (b) Circumferential section.

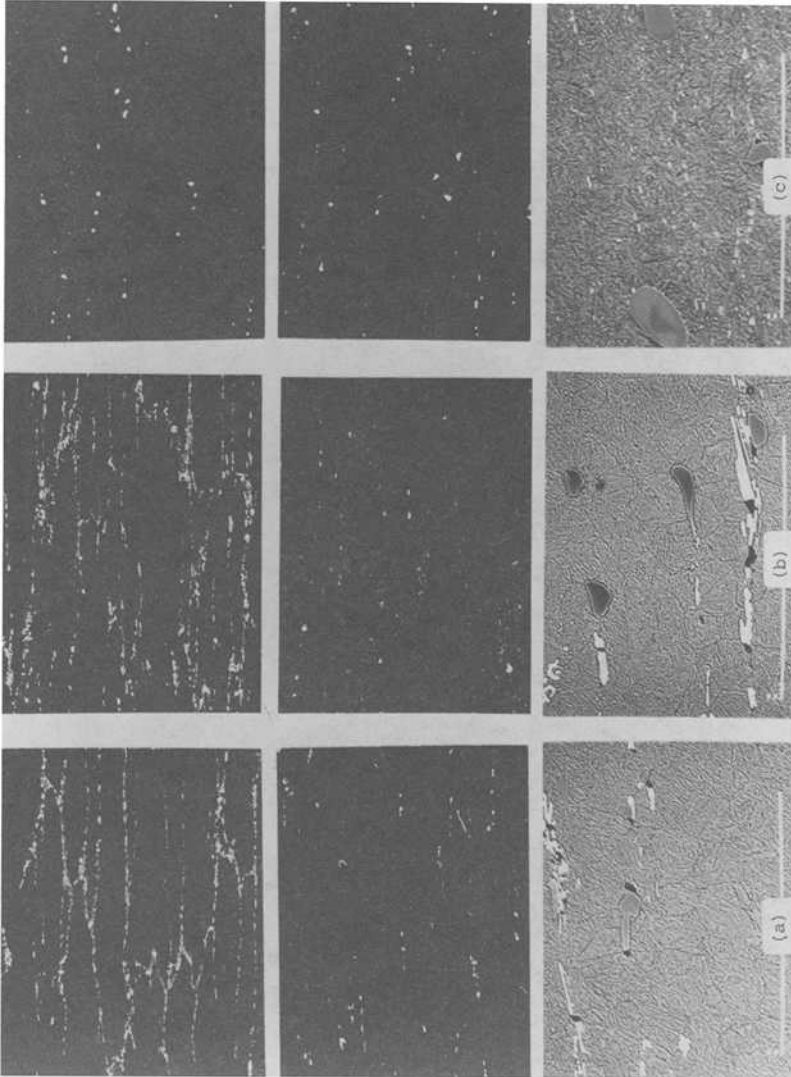


FIG. 15—Heats A (a), B (b), and C (c). Top: Carbides at $\times 60$. Middle: MC only at $\times 60$. Bottom: SEM, 100 μm marker, $\times 480$.

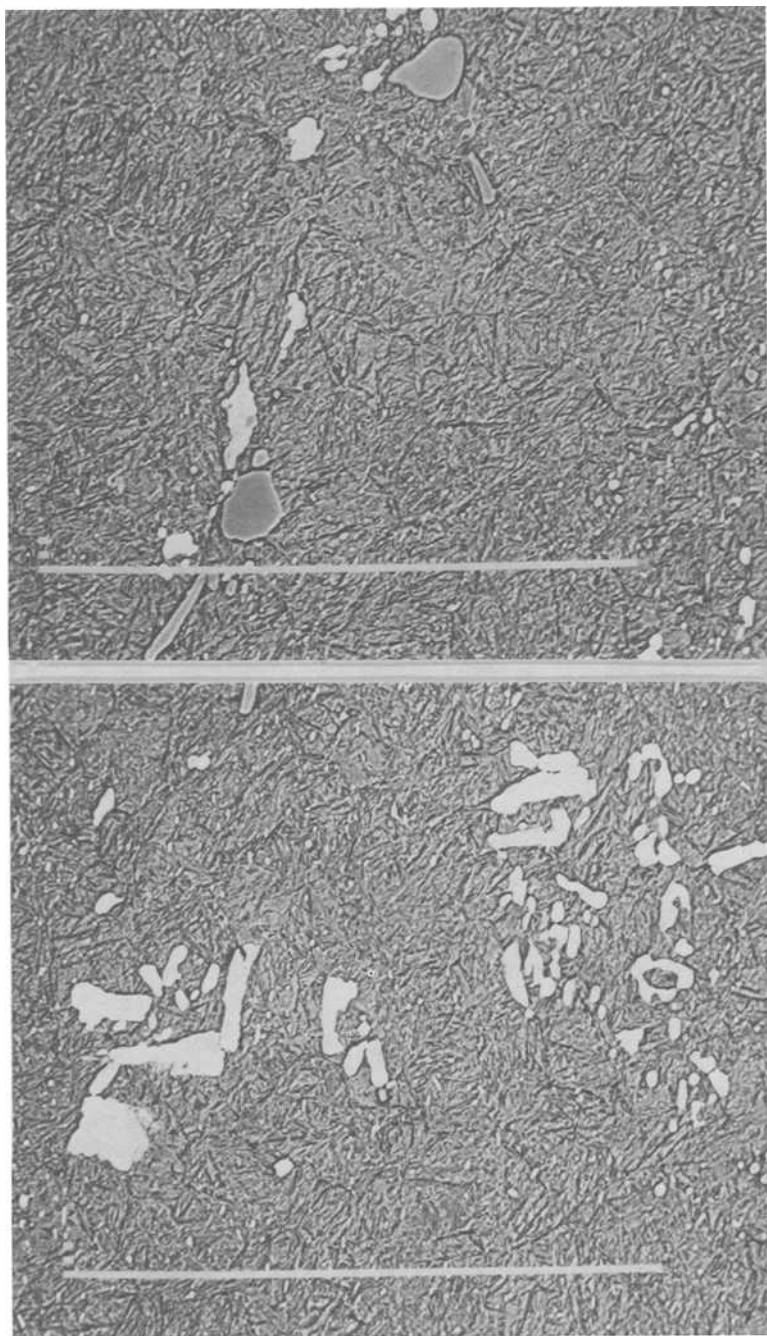


FIG. 16—Heat D. $\times 800$.

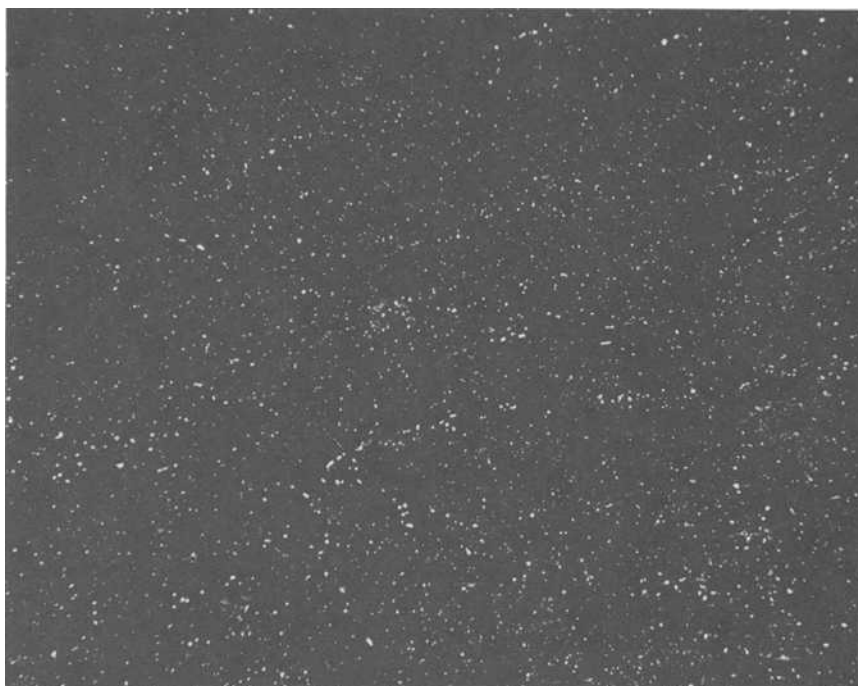


FIG. 17—*M-50NiL heat-treated case. $\times 500$.*

Discussion

Large quantities of M-50 steel from several steelmakers, with and without the heat treatment of ingots to reduce segregation, have been used successfully in service for many years. In difficult and advanced applications where improved performance is necessary, all evidence points to the desirability of fine and uniformly dispersed carbides, and several subjects relating to carbides deserve attention.

1. The heat treatment of ingots is undesirable because it creates large, very hard MC particles which can cause surface asperities. This is unfortunate because the heat treatment is otherwise successful in reducing segregation.

2. Improvement in size and distribution of all carbides, not just the MC, is important. The carbides that initiated surface peeling in our tests were predominantly M_2C . While small MC particles coexist in M_2C clusters, no evidence indicated that the peeling was related only to MC. The subsurface stress raising particles in 440C are a softer chromium carbide.

3. The effect of carbides becomes more noticeable under thin lubrication film conditions. Thin films require superior surfaces with more attention given to the isolated aberrations found by visual observation with microscopy, in addition to the usual mechanical measurements of roughness.

4. The change in carbides during heat treatment of ingots means that as cast, carbide is not an equilibrium structure. Differences in carbide composition between suppliers indicates that other factors in steelmaking have an effect. Steel producers have a very variable condition to control in order to produce quality and uniformity of carbides for improved bearing performance.

5. The voids at the carbides in Fig. 15 are similar to those reported by Butterfield and McNelley in Ref 3. This condition persisted through metallographic attempts to eliminate it and is obviously a subject for further study. It should be noted, however, that in our tests Heat D, which was free of voids, was the one which had the surface peeling failures at carbides in the 150-mm bore tests.

6. Carburizing as used in the manufacturing of M-50NiL bearings is a method to achieve fine carbides while maintaining the properties of M-50 steel and the quality of vacuum induction melting/vacuum-arc melting (VIM-VAR) double vacuum melting. Powdered metal processes are another well-known way to achieve fine carbides in high-alloy tool steels and would be desirable for many M-50 steel bearing applications if integrity and reliability of P/M steels could be made equivalent to that obtained with VIM-VAR melting.

Conclusions

The larger primary carbides in high-alloy bearing steels act as stress-raisers to initiate the surface peeling mode of failure under thin lubrication films and contribute to subsurface microstructural alteration associated with conventional rolling contact fatigue. The hard MC carbides in M-50 steel can cause surface asperities which lead to surface damage and failure. Carbon introduced by carburizing M-50NiL steel produces fine carbides comparable to low-alloy, high-carbon bearing steels.

Acknowledgments

The authors would like to acknowledge the contributions of their many colleagues at Torrington/Fafnir who contributed to this work, especially Messrs. W. J. Chmura, H. Daverio, R. E. Schomburg, R. E. Fairchild, and E. Zack for the bearing testing program and Mr. J. R. Imundo and Miss R. J. Beaulieu for the metallography of carbides. We also express our appreciation to Dr. B. L. Averbach of the Massachusetts Institute of Technology, who has been a valuable advisor throughout this research program and has also provided the electron microprobe analyses. We also appreciate several interesting discussions with Mr. E. N. Bamberger of The General Electric Company, Aircraft Engine Group. Figure 13 is from a bearing tested under contract to General Electric.

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Rolling Contact Fatigue Life of Various Kinds of High-Hardness Steels and Influence of Material Factors on Rolling Contact Fatigue Life

REFERENCE: Tsushima, N., Maeda, K., and Nakashima, H., "Rolling Contact Fatigue Life of Various Kinds of High-Hardness Steels and Influence of Material Factors on Rolling Contact Fatigue Life," *Effect of Steel Manufacturing Processes on the Quality of Bearing Steels*, ASTM STP 987, J. J. C. Hoo, Ed., American Society for Testing and Materials, Philadelphia, 1988, pp. 132-148.

ABSTRACT: The influence of material factors, such as types of steels, steelmaking process, cleanliness, fiber orientation, forging ratio, subzero treatment upon rolling contact fatigue life, was investigated for rolling bearing steels produced recently. Rolling contact fatigue life of recent steels which have high cleanliness has increased remarkably, and the influence of fiber orientation or subzero treatment upon rolling contact fatigue life is negligible, contrary to previous steels. Rolling contact fatigue life levels of each type of steels tested under three different test conditions were described. Depending on test conditions, their rolling contact fatigue strengths relative to others differed significantly in some cases.

KEY WORDS: rolling contact fatigue, bearing steels, continuous casting, heat-resistant steel, fiber orientation, forging, subzero treatment

Figure 1 shows the results of the life test of ball bearing 6206 which is conducted periodically under fixed test conditions every year. It can be seen that the bearing life increases year by year. Figure 2 shows Weibull distributions of lives of 6206 bearings in each group, which is divided at an interval of five years. Each Weibull distribution moves to the righthand side (longer-life side) with the progress of the time, but the change of Weibull slope is small. Test results using simple-shaped cylinder specimens (12 mm in diameter and 22 mm long, as described later) show the same tendency (Figs. 3 and 4). It is concluded, therefore, that the improvement of rolling contact fatigue life of rolling bearings is caused mainly by the improvement of steel quality because of progress in the steelmaking technique.

There have been many investigations [1-3] on the influence of some material factors upon rolling contact fatigue strength. It is supposed, however, that the influence of these factors on recent steels would differ from previous results because recent steels are of much better quality.

This paper describes results of investigations on the influence of material and heat treatment factors upon rolling contact fatigue life of steels produced recently, as well as their rolling contact fatigue life levels under different conditions.

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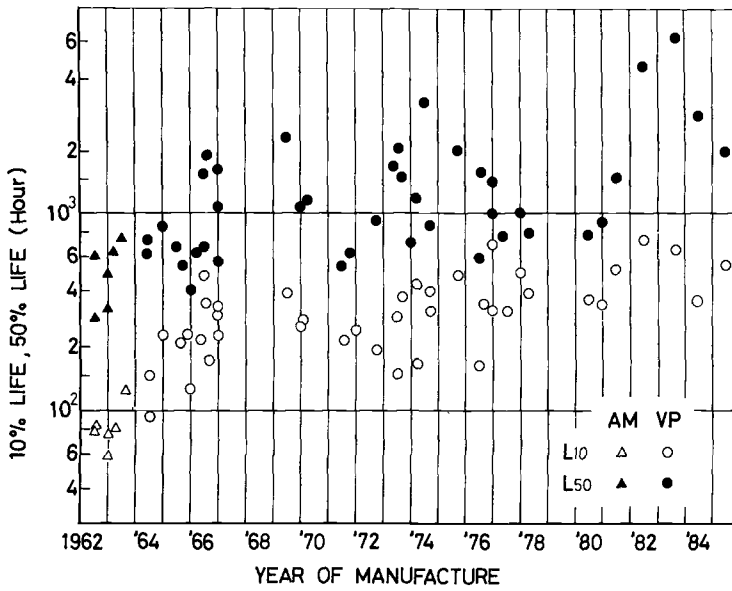


FIG. 1—Improvement in rolling contact fatigue life of ball bearing 6206.

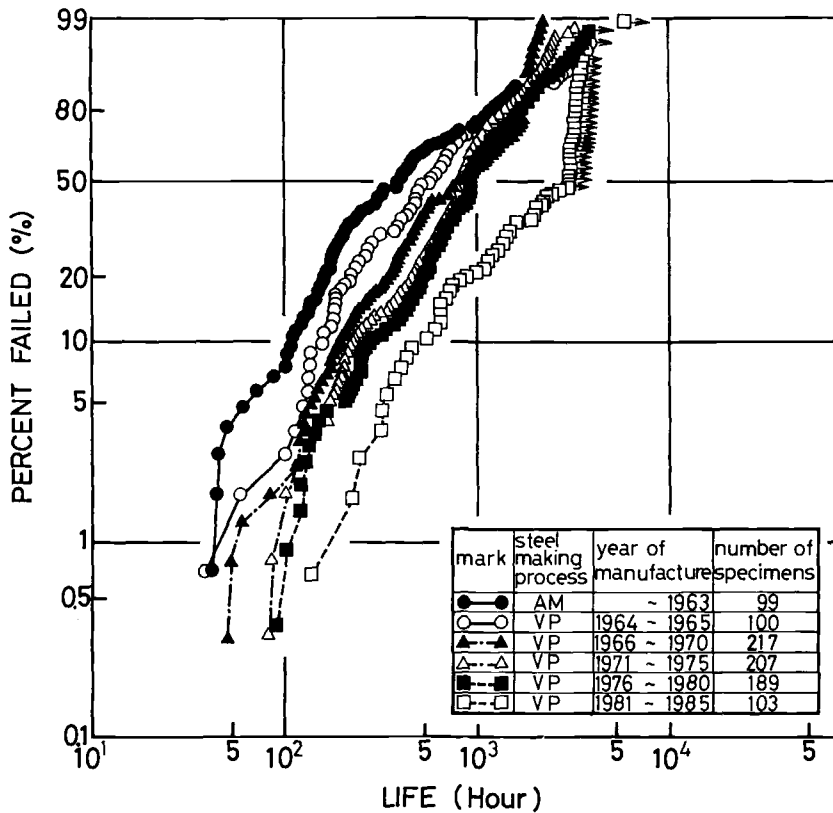


FIG. 2—Weibull distributions of lives of 6206 bearings in each group, which is divided at an interval of five years.

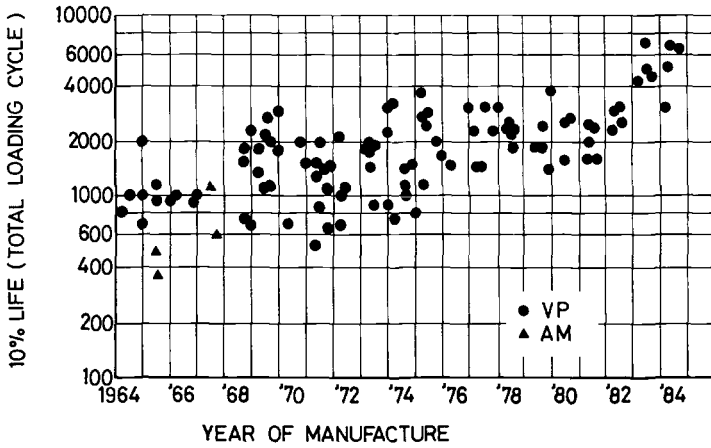


FIG. 3—Improvement in rolling contact fatigue life of through-hardening steel (SUI 2) in relation to year of manufacture.

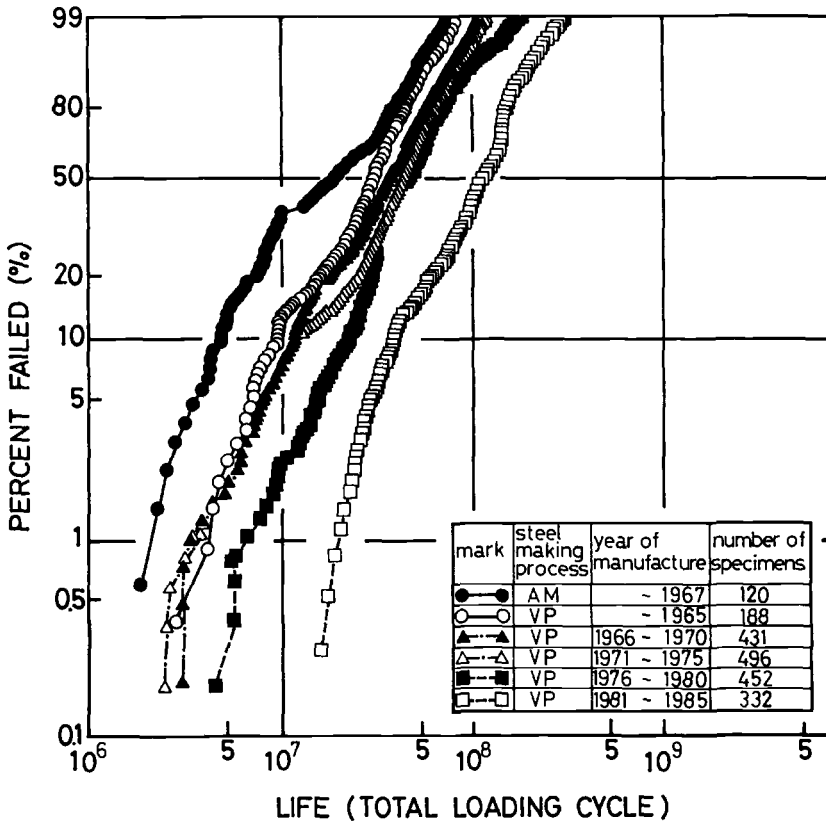


FIG. 4—Weibull distributions of lives of 12-mm-diameter point contact specimens in each group, which is divided at an interval of five years.

Test Rigs

Rigs for rolling contact fatigue tests are two of the point contact type rolling fatigue test rigs [1,2] and a line contact type rolling fatigue test rig [3,4] reported in previous papers. Schematic constructions of these rigs are shown in Fig. 5 and Fig. 6, and standard test conditions are shown in Tables 1 and 2.

In the point contact type test rig, a cylinder specimen 12 mm in diameter (or 60 mm in diameter for large-sized specimen) and 22 mm long (or 90 mm long) is rolling contacted with two partner balls 19.05 mm in diameter (or 25.4 mm in diameter). The hardness of the balls was HRC 64. The rolling contact fatigue test can be carried out twice (or 7 times) for one specimen with changing the contact position in axial direction each time.

In the line contact type test rig, a specimen, 12 mm in diameter and 12 mm long with crowning R480 mm on the raceway to avoid edge loading with mating rollers, is rolling contacted with two partner rollers whose size is 20 mm in diameter and 20 mm long. The hardness of partner rollers was HRC 62.0. Material conditions of specimens are shown in Table 3. Nonmetallic inclusion rating levels of these specimens are shown in Table 4.

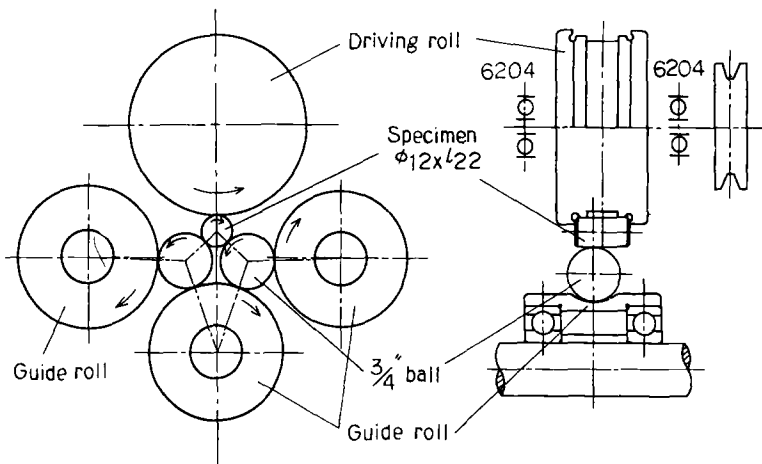


FIG. 5—Schematic construction of point contact type rolling contact fatigue test rig (12-mm-diameter cylinder).

TABLE 1—Test conditions of point contact type rolling contact fatigue test.

	φ12 Specimen	φ60 Specimen
Shape of specimen, mm	φ12 × 22	φ60 × 90
Size of mating ball, mm	φ19.05	φ25.4
Contact stress, MPa	P_{max} 5880 (600 kgf/mm ²)	
Loading speed, cpm	46240	6610
Lubrication	VG 56 turbine oil splashing	
Test per one specimen	twice	7 times

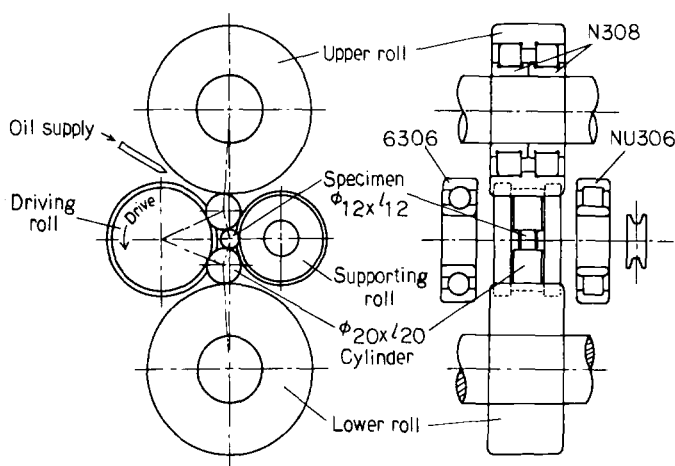


FIG. 6—Schematic construction of line contact type rolling contact fatigue test rig.

TABLE 2—Test conditions of line contact type rolling contact fatigue test.

Shape of specimen, mm	$\phi 12 \times 12$ with crowning R480
Shape of mating specimen, mm	$\phi 20 \times 120$ roller
Contact stress, MPa	$P_{\max}$ 4155 (424 kgf/mm ²)
Loading speed, cpm	20400
Lubrication	VG 56 turbine oil circulation

Test Results

Life Comparison Between Through-Hardened Steel and Carburized Steel

The Weibull distributions of the results of the rolling contact fatigue test with point contact specimens, made of through-hardened steel (SUI 2) and two kinds of carburized steels (SCr 420, SCM 420) produced recently, are shown in Fig. 7a. Compared to the data with previous steels shown in Fig. 7b, life increase is remarkable both in through-hardened and carburized steels. For this reason, it is considered that the ladle refining technique in the steelmaking process has been completely mastered. The life of carburized steels is a little longer than that of through hardening steel.

Life of Continuous Casting Steel

Around 1982, continuous casting process began to be applied to ball bearing steels in place of ingot casting process, and now, continuous casting process is commonly used. Figure 8 shows Weibull distributions of the lives of point contact specimen made of through-hardened steels (SUI 2) produced by both steelmaking processes. There is no significant difference in life between both processes. In this case, continuous cast steels tested were supplied from several different sources with continuous casting processes which were only slightly different. However, there was no significant difference in life among different continuous casting processes.

TABLE 3—Material conditions of specimens tested.

Materials	$\phi 12 \times 22$ mm, $\phi 12 \times 12$ mm Specimens			$\phi 60 \times 90$ mm Specimens		
	Through Hardened (SUJ 2)	Carburized (SCr 420, SCM 420)	Induction Hardened (S53C, 4150)	Through Hardened (SUJ 3, SUJ 4)	Carburized (SCM 420, SNCM 420 etc.)	Induction Hardened (S50C, SCM 440)
Surface Hardness, Hv	710 to 745	710 to 745	690 to 720	720 to 760	720 to 740	660 to 700
Case Depth (Hv 513), mm	...	1.2 to 1.6	2.0 to 2.6	...	3.2 to 4.0	2.9 to 3.2
Retained Austenite at Surface, %	10 to 15	25 to 30	5 to 7	10 to 20	20 to 30	5 to 7
Compressive Residual Stress in Subsurface, MPa	0 to 100	100 to 200	250 to 350	100 to 200	300 to 500	450 to 600

TABLE 4—Nonmetallic inclusion rating levels of specimens tested.

	A		B		C		D	
	T	H	T	H	T	H	T	H
Through Hardening	1.0 to 1.5	0 to 0.5	0.5 to 1.5	0 to 0.5	0 to 0	0 to 0	1.0 to 1.0	0 to 0.5
(SUJ 2)	$\bar{x} = 1.35$	$\bar{x} = 0.25$	$\bar{x} = 1.00$	$\bar{x} = 0.07$	$\bar{x} = 0$	$\bar{x} = 0$	$\bar{x} = 1.0$	$\bar{x} = 0.45$
Carburizing	1.0 to 2.0	0 to 0.5	0.5 to 1.5	0 to 0.5	0 to 0	0 to 0	1.0 to 1.0	0 to 0.5
(SCr 420, SCM 420)	$\bar{x} = 1.60$	$\bar{x} = 0.16$	$\bar{x} = 1.15$	$\bar{x} = 0.04$	$\bar{x} = 0$	$\bar{x} = 0$	$\bar{x} = 1.0$	$\bar{x} = 0.30$

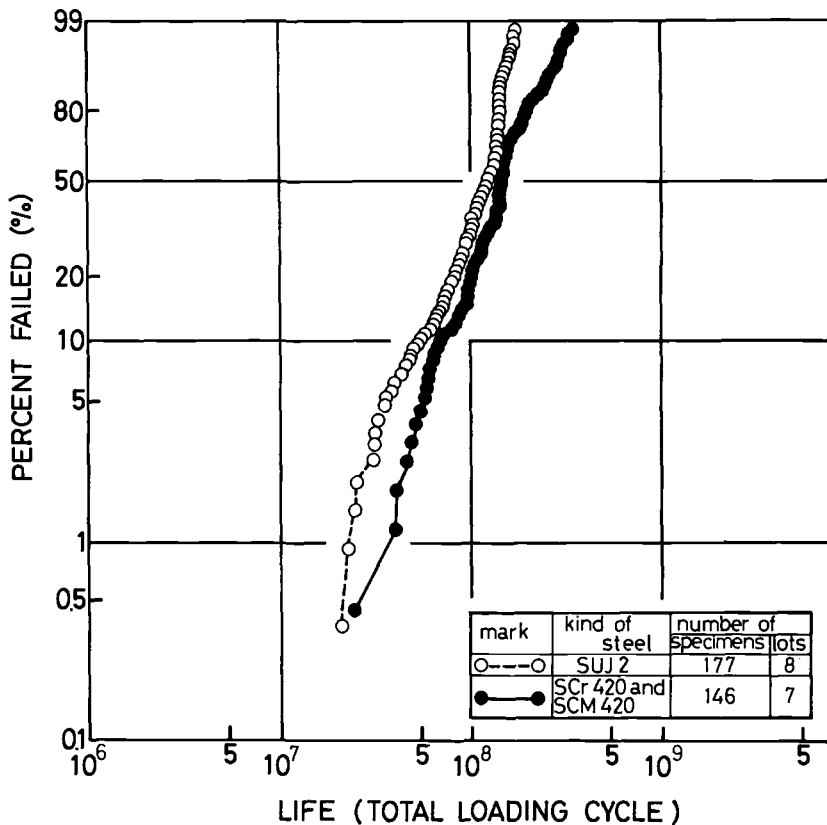


FIG. 7a—Weibull distributions of lives of recent through-hardening steels (SUJ 2) and carburizing steels (SCr 420, SCM 420) (12-mm-diameter point contact specimen).

Life of High-Purity Steel

It is well-known that the cleanliness of the material, especially oxygen content, affects rolling contact fatigue life strongly. Oxygen content level of ball bearing steels produced recently is less than 10 ppm. High-purity steel which has lower oxygen (5 ppm), low titanium (less than 20 ppm) and sulfur (less than 30 ppm) contents is currently a topic of study. Test results of these high purity steels in the point contact test are shown in Fig. 9. Life level of these high purity steels is almost equal to that of the conventional material. Since a similar result was obtained by the test of ball bearing 6203 (Fig. 10), we can consider that impurity contents such as oxygen, titanium, and sulfur less than levels of conventional materials do not influence rolling contact fatigue life significantly.

Life Levels of Various Bearing Steels

Table 5 shows the test results of small-sized point contact specimens which were made of various kinds of steels manufactured since 1982. Table 6 shows the test results with large-sized point contact specimens. In case of small-sized specimens, the life of carburizing steel is a little longer than that of through hardening steel, and that of induction hardening steel is

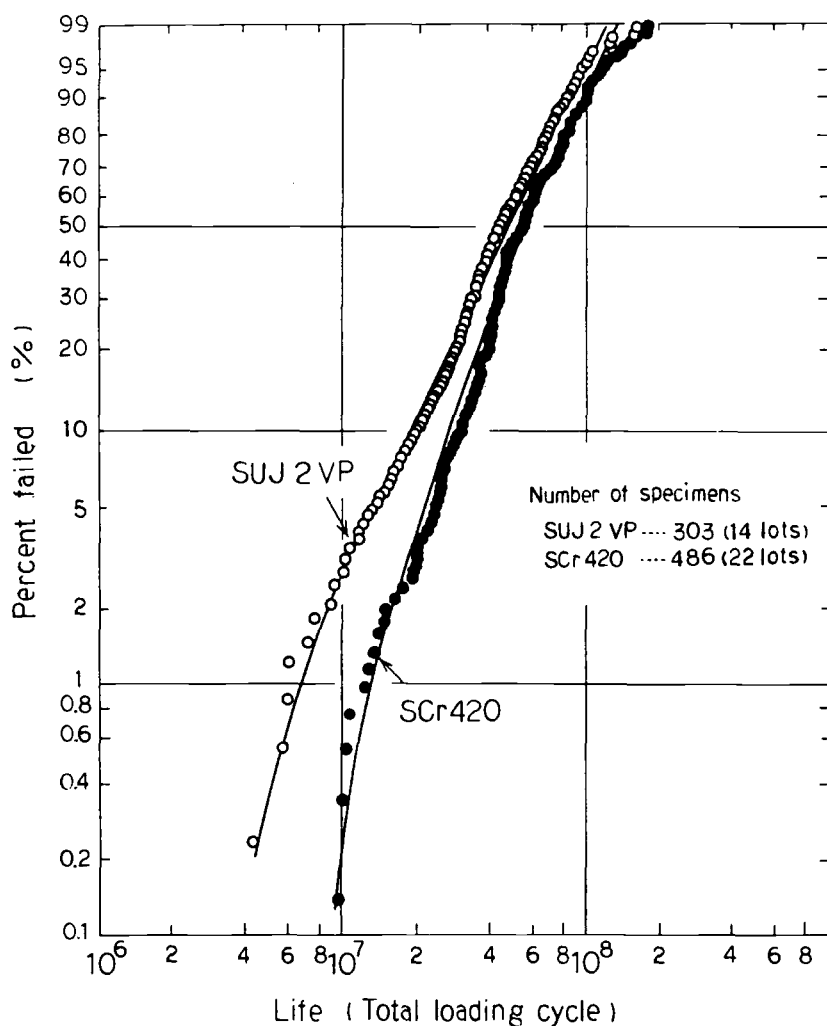


FIG. 7b—Weibull distributions of lives of previous through-hardening steels (SUJ 2) and carburizing steels (SCr 420) (12-mm-diameter point contact specimen).

considerably lower than that of through hardening steel. Among carburizing steels, nickel-chromium-molybdenum (Ni-Cr-Mo) steel shows the lowest life, which is considered not to be due to chemical compositions but due to lower cleanliness. In case of large-sized specimens, on the contrary, the lives of through-hardening steel and induction-hardening steel are nearly the same, and that of carburizing steel is three times longer. This means that the relative strength of rolling contact fatigue of each type of steels can vary, depending on size of specimens or test conditions.

There may be a favorable effect on compressive residual stress due to heat treatment which is larger for larger specimens, as shown in Table 3. With small-sized specimens, the test conditions are severer than in with large-sized specimens as predicted from microstructural changes in the subsurface of the specimen during test [5]. Therefore, steels such as induction-hardening steels, which are less heat resistant due to low carbon content, are disadvantageous in the test with small-sized specimens.

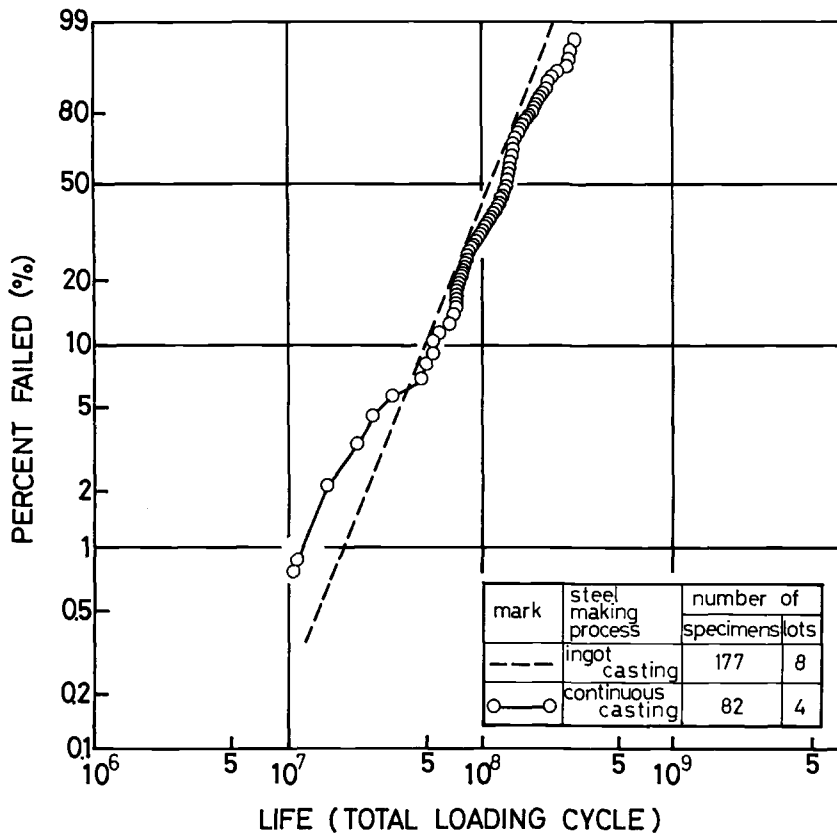


FIG. 8—Weibull distributions of lives of recent through-hardening steels (SUJ 2) made by ingot casting and continuous casting (after 1981, 12-mm-diameter point contact specimen).

TABLE 5—Life levels of recent materials with 12-mm-diameter point contact specimens.

Material	Steels	10% Life ($\times 10^4$)	
		Range Median	Number of Lots
Through hardening steel	SUJ 2 ^a	2590 to 7650 5510	8
Carburizing Cr steel	SCr 420 ^b	4580 to 10500 6150	3
Carburizing Cr-Mo steel	SCM 420 ^c	4590 to 11000 6150	4
Carburizing Ni-Cr-Mo steel	4320 8620	4740 to 6250 5590	4
Induction hardening steel	S53C ^d 4150	1060 to 1850 1320	4

^aSUJ 2 corresponds to AISI 52100.

^bSCr 420 = 0.18% to 0.23% C, 0.15% to 0.35% Si, 0.60% to 0.85% Mn, and 0.90% to 1.20% Cr.

^cSCM 420 = 0.18% to 0.23% C, 0.15% to 0.35% Si, 0.60% to 0.85% Mn, 0.90% to 1.20% Cr, and 0.15% to 0.30% Mo.

^dS53C corresponds to AISI 1053.

TABLE 6—Life levels of materials with 60-mm-diameter point contact specimens.

Material	10% Life ($\times 10^4$)	Years of Manufacture		
		1972–1976	1977–1983	
		$P_{\max}$ 5880 MPa	$P_{\max}$ 5880 MPa	$P_{\max}$ 5290 MPa
SUJ 3 ^a	range median (number of lots)	235 to 643 455 ($n = 6$)	469 ($n = 1$)	591 ($n = 1$)
SUJ 4 ^b	range median (number of lots)	294 to 581 427 ($n = 2$)	...	...
Ni-Cr-Mo steel	range median (number of lots)	1400 ($n = 1$)	1230–2920 1810 ($n = 5$)	...
Cr-Mo steel	range median (number of lots)	...	967 to 2010 1160 ($n = 3$)	...
Induction hardening steel (S50C, SCM 440 ^c)	range median (number of lots)	...	293 to 480 384 ($n = 2$)	1120 to 2620 1460 ($n = 3$)

^aSUJ 3 = 0.95% C, 0.40% to 0.70% Si, 0.90% to 1.15% Mn, and 0.90% to 1.20% Cr.^bSUJ 4 = 0.95% to 1.10% C, 0.15% to 0.35% Si, less than 0.50% Mn, 1.30% to 1.60% Cr, and 0.10% to 0.25% Mo.^cS50C corresponds to AISI 1050.^dSCM 440 corresponds to AISI 4140.

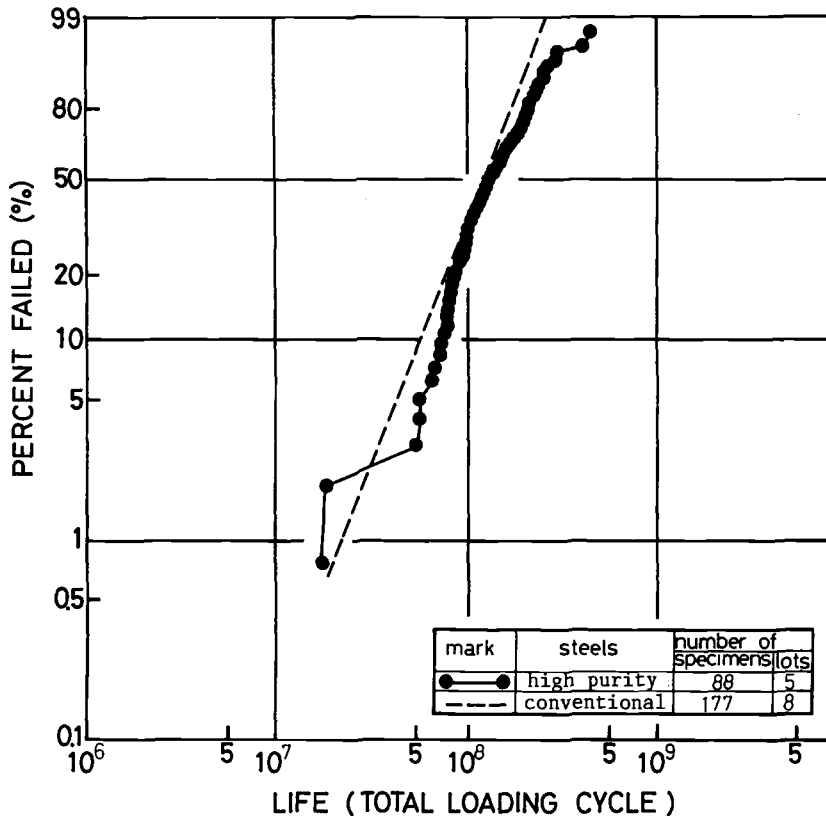


FIG. 9—Weibull distributions of lives of high-purity steel and conventional steel (after 1982, SUJ 2, 12-mm-diameter point contact specimen).

Heat-resistant steels such as M-50 are used as a bearing steel for jet-engine bearings. Table 7 shows rolling contact fatigue lives of typical heat-resistant steels. The test was conducted at room temperature, and the temperature of lubricant was 70°C for the line contact test and 85 to 90°C for the point contact test. Compared with the life levels of through-hardening steels or carburizing steels mentioned in Table 5, the life levels of some of these heat-resistant steels are extremely short. This is due to different type of failure, especially surface origin type flaking [6], which occurs frequently in high-alloyed steels in this test. The reason for this type of flaking is not known yet, though. On the other hand, heat-resistant steels produced by powder forging do not show surface origin type flaking and have rather long life. It is not yet clear why powder-forged specimens do not show surface origin type flaking.

Influence of Fiber Orientation

In the previous paper, it was described that the influence of fiber orientation upon rolling contact fatigue life was considerable [1]. Figure 11 shows the test results of the influence of fiber orientation using point contact and line contact type specimens. Fiber orientation was changed in five steps from 0 deg (fiber orientation is parallel to the specimen axis) to 90 deg (fiber orientation is normal to the specimen axis), as shown in Fig. 11a. The influence of fiber

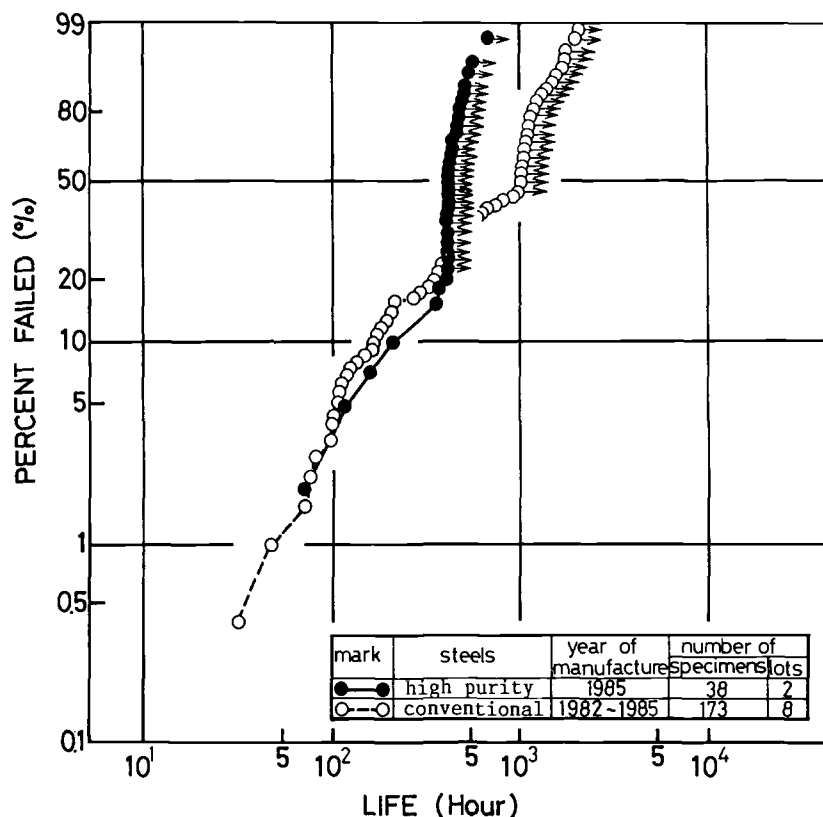


FIG. 10—Weibull distributions of lives of high-purity steel and conventional steel with ball bearing 6203 (calculated life $L_{10h} = 55.7 h$).

orientation upon rolling contact fatigue life with recent steels has decreased in contrast to previous steels. For this reason, it is considered that the harmful influence of nonmetallic inclusions which are arranged along fiber orientation is reduced because the number and size of nonmetallic inclusions are much reduced.

Influence of Forging Ratio

The significant influence of forging ratio such as bar forging or upset forging upon rolling contact fatigue life was described in the previous paper [1]. Influence of forging ratio examined with recent steels is shown in Table 8. The influence of the forging ratio from 5.8 to 13.6 upon rolling contact fatigue life is not clear. It is thought that lower nonmetallic inclusion content is the main reason.

Influence of Subzero Treatment

Generally, retained austenite is thought to have a desirable effect on rolling contact fatigue life. Therefore, the influence of subzero treatment upon rolling contact fatigue life has been considered undesirable because it decreases amount of retained austenite. The test results on

TABLE 7—Life levels of various high alloy steels with 12-mm-diameter point contact and line contact specimens.

Material	Steels	Point Contact Test			Line Contact Test		
		$L_{10}(\times 10^4)$	$L_{50}(\times 10^4)$	Lots	$L_{10}(\times 10^4)$	$L_{50}(\times 10^4)$	Lots
Heat-resistant carburizing steel	CBS600, ^a CBS1000, ^b M-50NiL, ^c PYROWEAR53 ^d	7670	36700	5	3670	9870	4
Heat-resistant steel	M-50 T1	240/ 3480	1730/ 16900	4 2	1630 ...	5160 ...	15 ...
Heat-resistant stainless steel	BG42, ^e CRB7/ ^f	126/ ^g	353/ ^g	4	864	2770	4
Stainless steel	SUS440C	327/ ^h	3240/ ^h	8	316	749	1
High speed steel made by powder forging	HAP10, ^g HAP40, ^h RT904/ ⁱ	8760	47300	3	...	...	...

^aCBS600 = 1% Si, 0.65% Mn, 1.4% Cr, and 1% Mo.^bCBS1000 = 0.45% Si, 0.6% Mn, 2.8% Ni, 1% Cr, 4.4% Mo, and 0.4% V.^cM-50NiL = 4.1% Cr, 1.2% V, 3.5% Ni, and 4.3% Mo.^dPYROWEAR53 = 0.8% Si, 2.3% Ni, 1% Cr, and 3.3% Mo.^eBG42 = 1.1% C, 14.8% Cr, 4.1% Mo, and 1.3% V.^fCRB7 = 1.1% C, 14% Cr, 2.1% Mo, 1.1% V, and 0.3% columbium.^gHAP10 = 1.4% C, 5% Cr, 6% Mo, 3% W, and 3.8% V.^hHAP40 = 1.3% C, 4.2% Cr, 5.0% Mo, 6.4% W, 3.1% V, and 7.9% Co.ⁱRT904 = 1.3% C, 4.4% Cr, 5.3% Mo, 6.6% W, and 3.2% V.^jSurface origin type flaking predominantly occurs in these steels, thus decreasing their lives markedly.

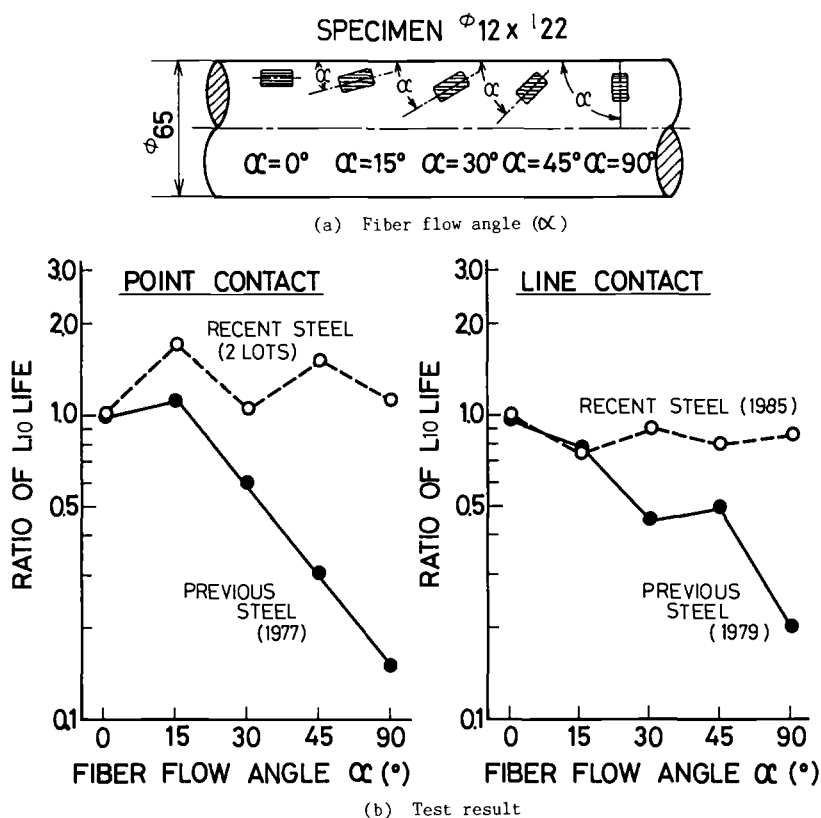


FIG. 11—Influence of the fiber flow angle on rolling contact fatigue life (SUS 2, 12-mm-diameter point contact and line contact specimens).

line contact specimens made from recent steels are shown in Fig. 12. It is clear that subzero treatment does not decrease rolling contact fatigue life. It is considered that the effect of retained austenite which moderates stress concentration around inclusions became unnecessary for recent steels, which are greatly improved compared with earlier steels.

Conclusion

According to the life test of deep-groove ball bearing 6206, conducted periodically under constant test conditions, bearing life has clearly improved over 20 years. Similar improvement in rolling contact fatigue life is recognized in the test using 12-mm-diameter cylinder specimen. These improvements in rolling contact fatigue life are considered to be due to the improvement of cleanliness in steels. However, Weibull slope, which expresses the scatter of life distribution, has not changed.

It is anticipated that the influence of material and heat treatment factors on rolling contact fatigue life changes with progress in steel quality. It was confirmed that the influences of fiber-flow orientation and subzero treatment have changed from the previous ones.

Relative strength of rolling contact fatigue of each type of steel was significantly dependent on test condition. Therefore, the so-called material factor in bearing life calculation formula is considered to be variable, depending on running conditions. It is thought that the influence of lubrication on bearing life becomes still more important with the improvement of steel quality.

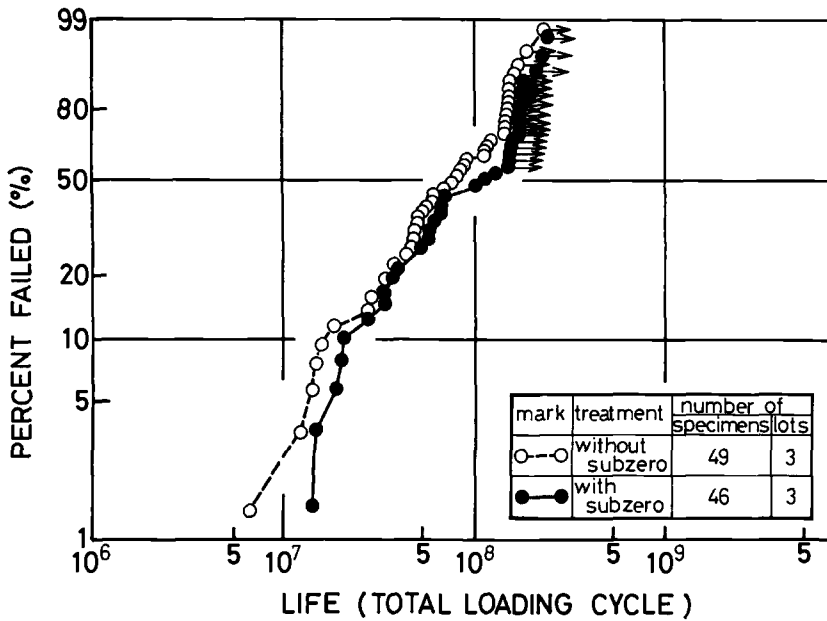
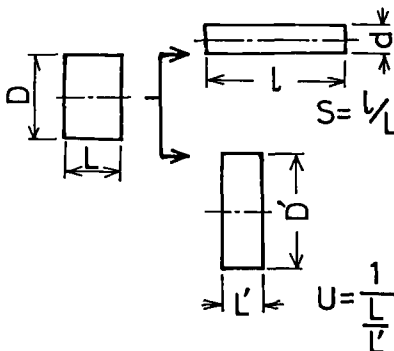


FIG. 12—Influence of subzero treatment on rolling contact fatigue life (SUJ 2, 12-mm-diameter line contact specimen).

TABLE 8—Influence of forging ratio on rolling contact fatigue life of M-50 steel with 12-mm-diameter line contact specimens.

Specimen No.	Forging ratio	Number of specimens	L_{10} ($\times 10^4$)	L_{50} ($\times 10^4$)
1F	$(1/1.5)U \Rightarrow 5.8S$	22	1980	4370
1C	$(1/1.5)U \Rightarrow 6.9S$	23	1140	4040
1D	$(1/1.5)U \Rightarrow 9.1S$	21	1200	3810
1E	$(1/1.5)U \Rightarrow 13.6S$	17	1700	6760



S: Bar forging

U: Upset forging

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Relationship of Melting Practice, Inclusion Type, and Size with Fatigue Resistance of Bearing Steels

REFERENCE: Monnot, J., Heritier, B., and Cogne, J. Y., "Relationship of Melting Practice, Inclusion Type, and Size with Fatigue Resistance of Bearing Steels," *Effect of Steel Manufacturing Processes on the Quality of Bearing Steels*, ASTM STP 987, J. J. C. Hoo, Ed., American Society for Testing and Materials, Philadelphia, 1988, pp. 149–165.

ABSTRACT: The evaluation of numerous steelmaking practices and the consequences of process modifications have been assessed by both fatigue tests and analysis of the inclusions at the origin of fracture. Increasing the average size of oxides causes the fatigue properties to deteriorate; the harmful influence of large globular Al_2O_3 -CaO inclusions was confirmed, and the role of other types such as titanium nitrides, alumina, and certain sulfides was clarified. General criteria of total oxygen or total titanium content are not related to the fatigue performance of steels manufactured by different steelmaking processes and practices; within an individual practice, however, they may serve as useful indicators of fatigue life. The poor correlation between standard cleanliness ratings and fatigue performances has been explained. The application of these results to qualification procedures and quality-control methods are discussed. Currently, the best quality assurance can be obtained through the qualification of the steelmaking practice by fatigue tests and a strict control over the reliability of the process by measurements of the critical parameters.

KEY WORDS: bearings, inclusions, rolling contact fatigue, steelmaking process, cleanliness rating, oxygen, titanium, calcium

Bearings are perhaps the most widespread type of mechanical engineering components used in all types of machinery. The load applied on a bearing determines its dimensions and, in turn, the size of associated mechanical components such as housings, shafts, and others. In the last decade there has been a steady trend to increase the allowable loads on bearings in order to build smaller, more efficient, and cheaper assemblies.

Improvements in bearing performance have been achieved through better control of several factors, including surface finish, lubricant properties, particle removal in lubricant, design and manufacture of bearings, and steel fatigue properties.

The present paper will deal only with steel properties in terms of fatigue life and the relationship to nonmetallic inclusion content, together with steelmaking processes and practices. It will also discuss how these results can influence quality control methods and qualification procedures.

Testing Methods and First Studies

The last Phoenix Symposium on rolling contact fatigue testing in 1981 revealed a diversity of methods involved in fatigue testing of bearing steel [1]. This diversity has also been observed

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in recent publications [2-3]. These observations led us to summarize the testing methods and main results of the initial work carried out at Ugine [4], which has been published only in part outside of France [5-6]. These studies were instigated by the desire to determine how the steelmaking practice used to manufacture American Iron and Steel Institute (AISI) 52100 steel can influence the service behavior of ball and roller bearings.

After trying a number of test methods and receiving advice from various sources, we oriented our research towards developing a test on finished bearings that would show the least possible susceptibility to variations in bearing manufacturing, assembly, and subsequent operating parameters. It was observed that testing a rigid, deep-groove ball bearing (6309 model) and a double-row self-aligning ball bearing (1309 model) at low speed and under heavy load gave a good evaluation of steelmaking factors. Analysis of the same test results indicated the influence of radial segregation and hot working reduction. In ball tests in particular, the distribution of spalling between the pole and the equator provided a good indication of the segregation effect.

Our observations agree with those of Zwirlein and Schlicht [7], who suggest that under different operating conditions, the intrinsic qualities of the steel may not be the major factor of influence on the fatigue life of bearings, and that they may even be more or less masked by other factors. The good evaluation of metallurgical factors provided under our test conditions was explained by the alternating shear fatigue phenomenon occurring below the contact surface.

Following this, a decision was made to perform tests in parallel on finished bearings and on rotating bending fatigue (RBF) test specimens. These comparisons clearly demonstrated that

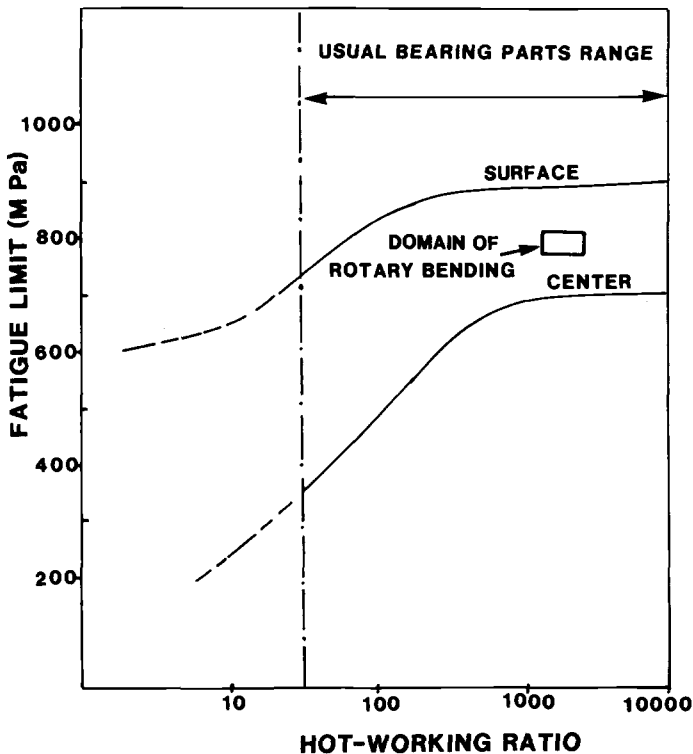


FIG. 1—Schematic representation of the influence of hot-working and radial segregation on fatigue properties.

for the different parameters tested, a very good correlation between the two types of test was achieved. We arrived at the conclusion that in spite of very different stress conditions, the same discontinuities in the steel produce the same effects in both rolling contact fatigue on bearings and in RBF test. Some of the major results of these preliminary studies are shown in Figs. 1 and 2.

Figure 1 shows a schematic representation of the influence, under constant steelmaking conditions, of hot-working reduction and radial segregation on the fatigue properties of steel. This diagram illustrates the importance of precisely situating the stressed area with respect to the as-cast material when studying the influence of steelmaking parameters. In our testing program, the vast majority of RBF tests were conducted with maximum stress at one third of the radius from the center of the rolled product and with hot-working reduction constant at approximately 1500. The RBF test specimens have the same heat treatment as the bearing parts and a polished surface finish.

Figure 2 shows the main correlation obtained between L_{10} life on a series of 6309 inner rings manufactured from bars and fatigue limits on Moore RBF specimens for different steelmaking parameters. Note that the authors obtained the same type of correlation with 6309 inner rings manufactured from tubes, and also with balls on 1309 bearings.

These results led us to increase the use of RBF tests for studying steelmaking parameters and therefore to reduce the number of finished bearing tests. There was an advantage in this choice because the RBF tests require less time, are less costly, and provide the possibility of preserving the defect at the origin of fracture for study purposes. Thrust-type tests which have a lower cost

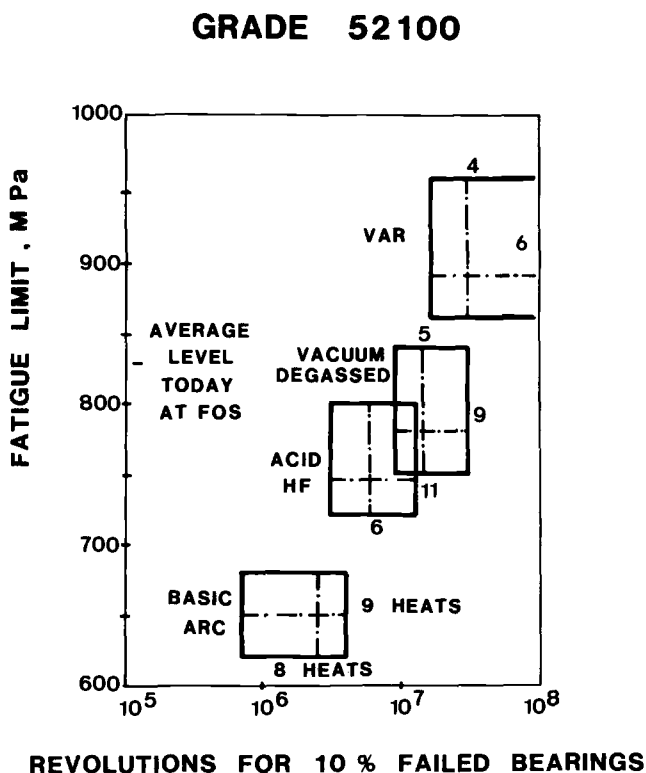


FIG. 2—Relation between RBF limit and L_{10} values on 6309 bearings.

and consume less time than bearing tests have also been developed; but although they yield results that are consistent with rolling contact phenomena, the test conditions must be rigorously controlled if abnormally scattered results are to be avoided.

Overall, the studies revealed the relative harmlessness of manganese sulfide inclusions and the particularly harmful effect of inclusions containing calcium.

The results therefore demonstrate that there are material defects which, under certain operating conditions, may have an influence on bearing life. All other things being equal, these material defects vary with steelmaking practice techniques, which justifies certain demands made by bearing manufacturers on steelmakers. It also justifies efforts that have been made in further studies to improve knowledge of steelmaking process factors.

Origin of Fatigue Fractures

In all materials, resistance to cracking decreases as mechanical properties such as hardness and tensile strength increase. Overall performance then becomes progressively more dependent on the presence of small defects that initiate cracking. This susceptibility to imperfections is well-known in ceramics, tool steels, and spring steels, and is also found in bearing steels. Nordberg demonstrated the existence of a relationship between the fatigue limits of a given type of steel, its mechanical properties, and the size of the discontinuities present in the steel [8]. The nonmetallic inclusions are the most obvious discontinuities, and their harmful influence on the fatigue limits of bearing steel has already been examined in numerous publications [9-15].

The systematic examination of fractures in failed RBF test specimens has led us to establish a general relationship, shown in Fig. 3, between the RBF limits of a heat and the average size

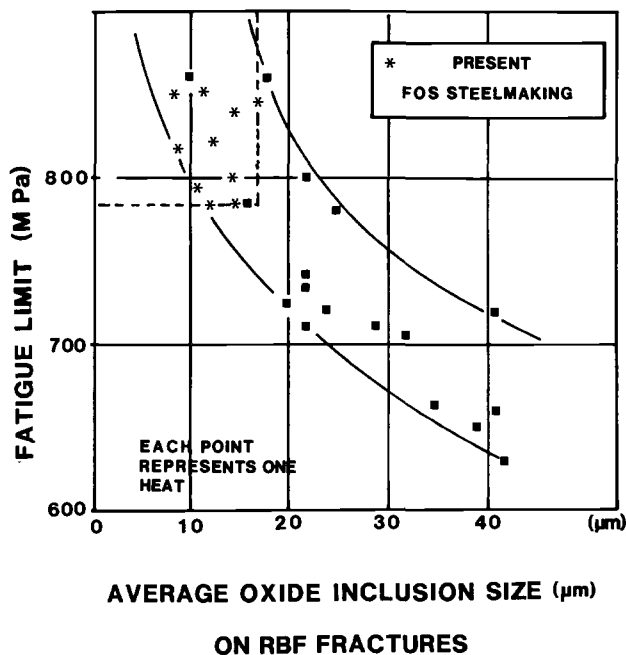


FIG. 3—Relation between the average oxide inclusion size on fractured RBF specimens and the RBF fatigue limit.

of oxide inclusions at the origin of the fracture. Because the fracture occurs in transverse conditions, the size of the inclusion is the largest transverse dimension. Oxides, indeed, were and remain most frequently associated with RBF fractures. Figure 3 represents results obtained over a long period of time and from various sources. At the lower fatigue limits (≤ 700 MPa), we have heats from older Ugine steelmaking practices already shown in Fig. 2, as well as steel samples obtained from other manufacturers. The upper fatigue limits (≥ 800 MPa) represent recent steelmaking practices implemented at Fos-sur-Mer and other nonremelted steel techniques from outside sources. In these cases, when a fracture occurs at an oxide inclusion, the size of the inclusion is much smaller than $20\text{ }\mu\text{m}$.

A scanning electron microscope (SEM) examination gives a precise image not only of the size of the inclusion, but also of its composition. Figure 4 represents a large complex oxide inclusion. Inclusions of titanium nitride (Fig. 5) and manganese sulfide enriched in calcium (Fig. 6) are also found at the origin of fracture.

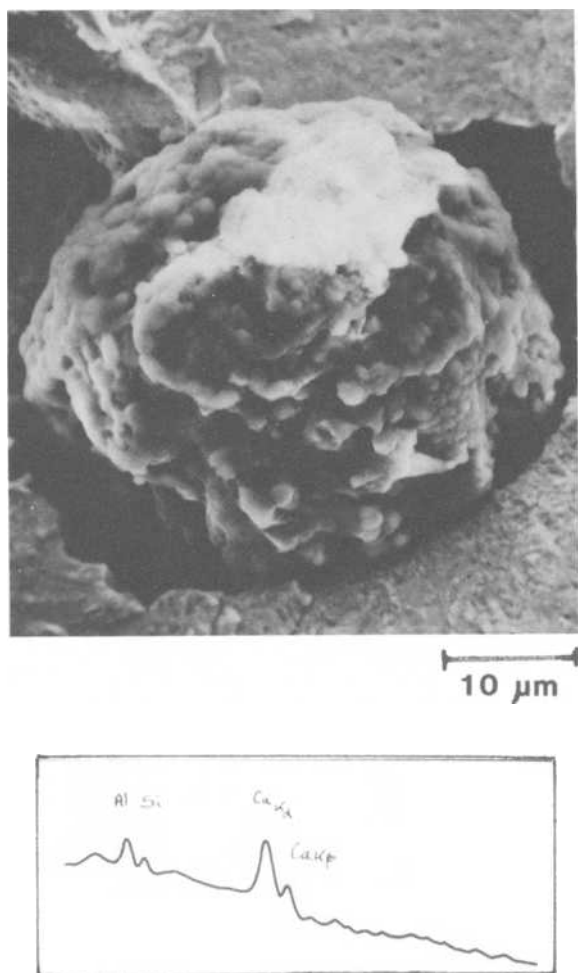


FIG. 4—Complex oxide inclusion on broken RBF specimen.

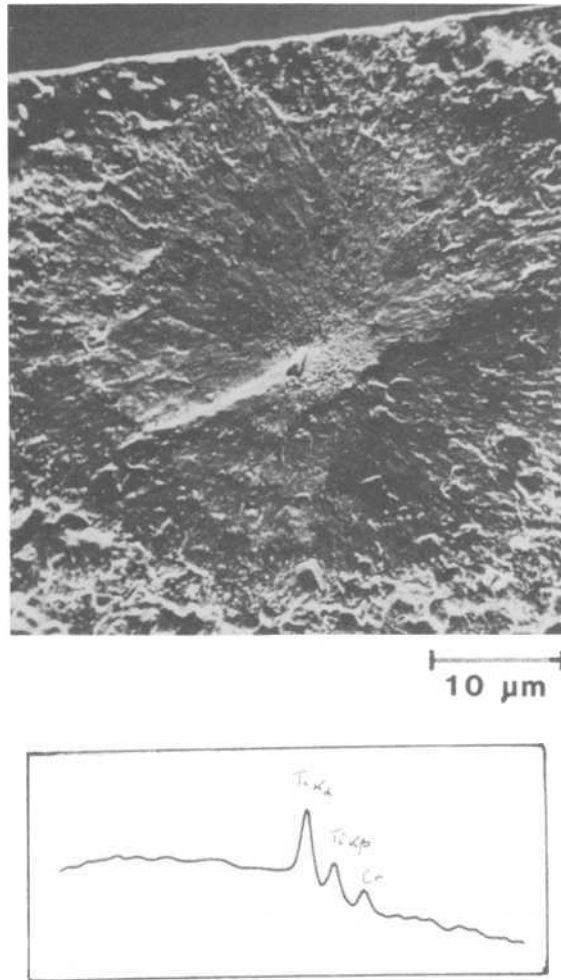


FIG. 5—Titanium nitride on broken RBF specimen.

The harmful effect of various types of inclusions and of different size ranges within the same family is presented in Fig. 7 by fatigue curves representing the stress at the inclusion position versus the fatigue life of the corresponding RBF specimen. Fatigue diagrams pertaining to titanium nitride only, or to various oxide size ranges only can be easily constructed by selecting the appropriate data; the figure therefore shows a summary of the results for nitrides and oxides. It can be seen that the fatigue life decreases sharply when the oxide inclusion size increases. The titanium nitride inclusions observed on fractures are always small, and therefore the data do not allow a similar analysis over a size range. However, it is readily apparent that, at a similar average size, nitrides are far more harmful than oxides. In fact, nitrides with an average size of 7 μm are equivalent to oxides with an average size of 25 μm in terms of the deterioration of fatigue life.

An overall schematic view of the relation between the harmful effect of inclusions on fatigue life versus inclusion size is presented in Fig. 8. Each unit increase of the index corresponds

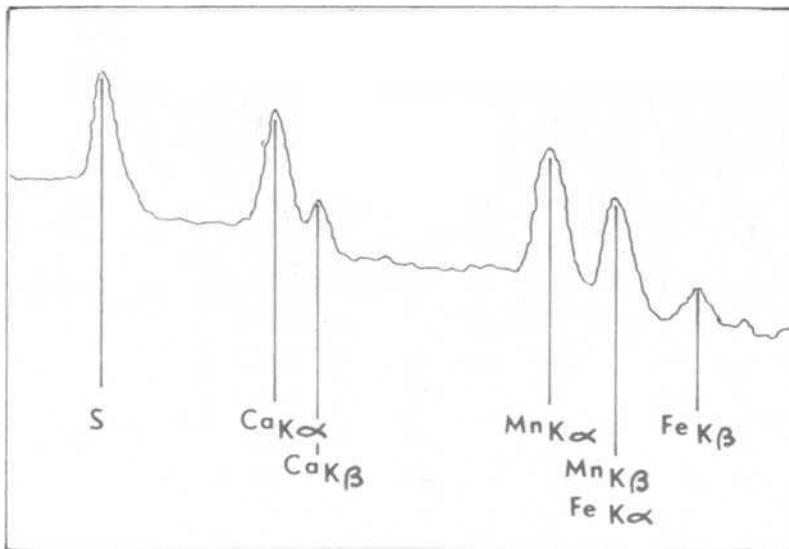


FIG. 6—Calcium-enriched sulfide on broken RBF specimen.

approximately to a 125 MPa decrease in the RBF limit (at 10^8 cycles). This figure shows that calcium-enriched sulfides, although small, compare with oxides of an equal size. Large globular inclusions are most harmful because of their size. Regardless of size, however, their intrinsic harmful effect appears close to that of Al_2O_3 oxides. Titanium nitrides clearly stand out as the most harmful type of inclusions over an equal size range.

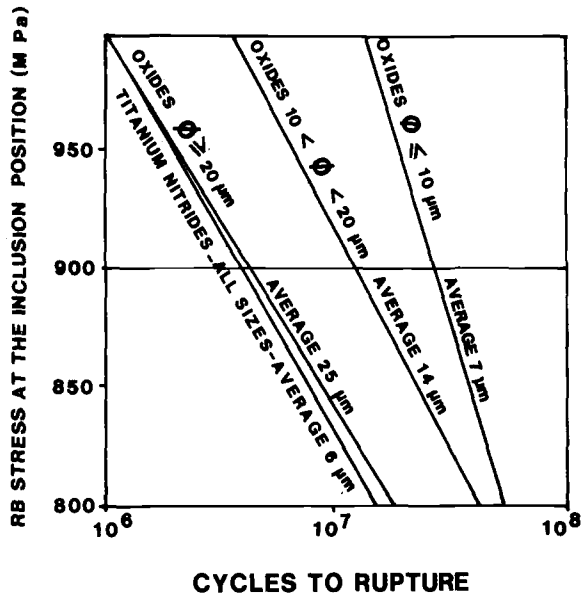


FIG. 7—Relation between local stress and fatigue life for oxide and nitride inclusions.

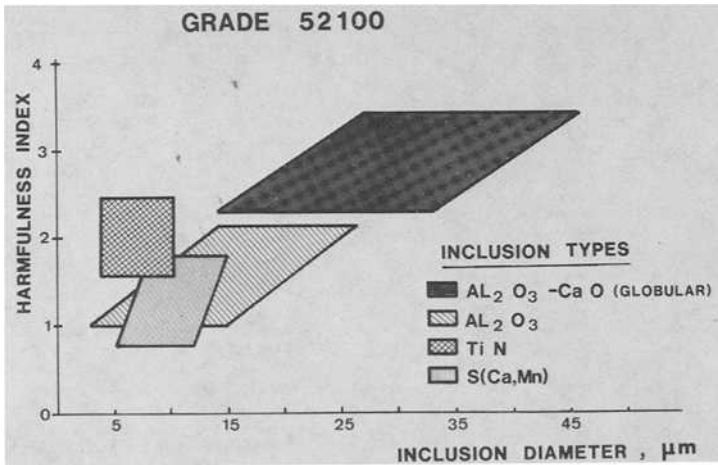


FIG. 8—Relative harmful effect of the various types of inclusions.

Cleanliness and Fatigue Properties

For a long time now, apart from basic chemical analysis, micrographic cleanliness has been considered one of the most important parameters in evaluating properties of bearing steel. This seems logical following the work carried out on inclusions. It has been observed in practice, however, that most of the existing methods of inclusion evaluation, as well as the limits used, have no scientific or technical basis.

The B thin rating of the ASTM Practice for Determining the Inclusion Content of Steel (E 45 Method A) is an example of a cleanliness parameter which, in characterizing the length of small oxide groups, assigns them a specific harmful factor. This standard has been used on a number of steelmaking practices from various manufacturers, and thus a relationship has been sought between the average B thin rating and fatigue properties. The results of this research are shown in Fig. 9. It would be possible, by selecting the steelmaking practices studied, to find a general relationship between the B thin cleanliness parameter and fatigue properties; when all the samples are considered, however, this relationship does not exist. For example, there is no difficulty in finding extremely clean steel with mediocre fatigue limits, as well as steel with good fatigue limits and poor micrographic cleanliness. In summary, there is therefore no absolute relationship between the B thin measurement and fatigue properties, and it does not represent the harmful effect of oxide inclusions.

In Figure 10 we have taken another approach, with RBF test specimens representing two steelmaking processes, A and B. On the upper part is a comparison of the distributions of oxide inclusion diameters, observed micrographically at cross sections of the test specimens. The logarithmic scale shows that there are ten times as many inclusions $\leq 5 \mu\text{m}$ in Process A than in Process B. The B thin rating in Process A is of the order of 2.5, whereas Process B has a rating of the order of 1. On the lower part of the figure is a comparison of the distributions of oxide inclusion diameter at the origin of fracture. In both cases, diameter means largest transverse dimension. Process A has a fatigue limit on the order of 830 MPa, whereas Process B has a limit on the order of 750 MPa. This demonstrates that the distribution of inclusions observed micrographically does not represent the distribution of inclusions at the origin of fracture, and it explains why a relationship cannot be drawn between the ASTM B thin measurement and fatigue properties.

The SAM method (ASTM Specification for High-Carbon Ball and Roller Bearing Steel (A 295-84) which attempts to take into consideration the size of inclusions using the

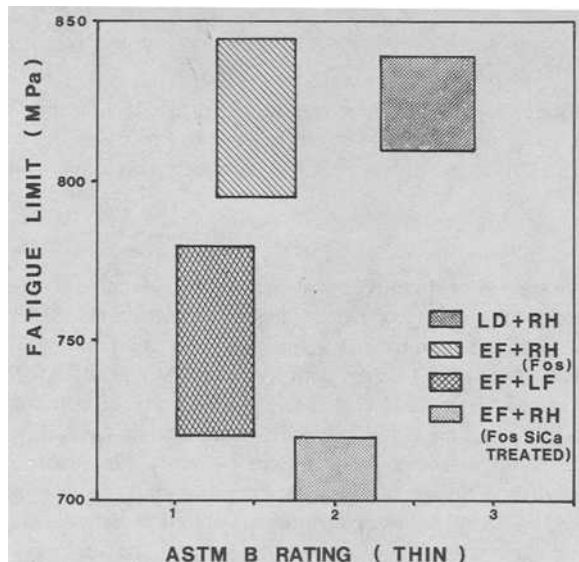


FIG. 9—ASTM B (thin) rating and RBF limit for various steelmaking processes.

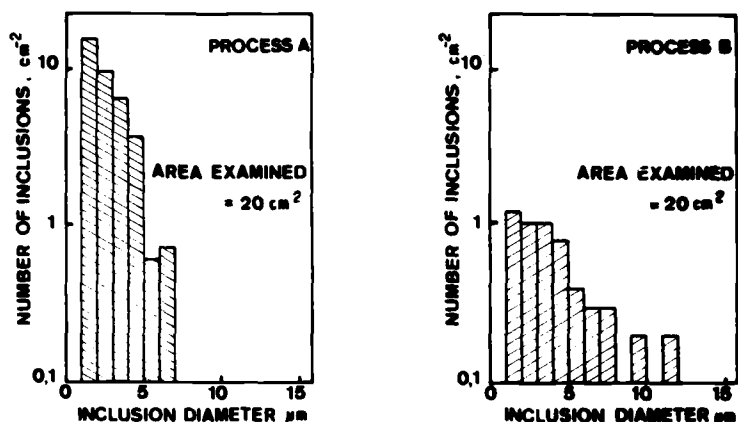
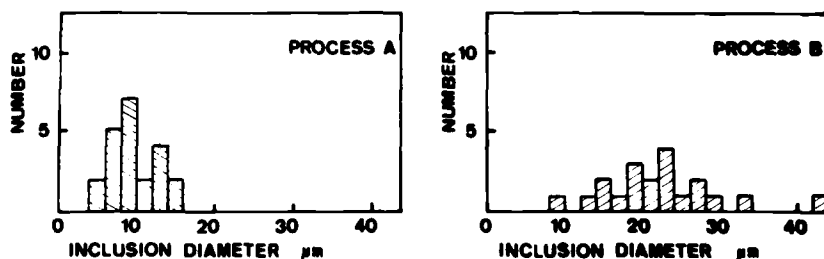
OXIDE INCLUSIONS ON POLISHED TRANSVERSE SECTIONS**OXIDE INCLUSIONS ON BROKEN RBF SPECIMENS**

FIG. 10—Comparison between the oxide inclusion size distribution on polished specimens and on fractured RBF specimens.

“thickness” qualification, should give better results. Unfortunately, in this case as well, it was found that Process B has a very low SAM count relative to Process A. Because Process B has a lower fatigue life, the largest inclusions are not sufficiently taken into consideration.

Oxygen Content

Oxygen is present in steel almost entirely in the form of oxide inclusions. The oxygen level, which represents the overall quantity of oxides, has now become mandatory in a number of customer specifications and has been used as an indicator of steel quality [19]. Figure 11 shows a number of different oxygen content ranges with respect to fatigue limits for various steelmaking practices. It can be seen that steel (LD and RH) with an oxygen content of approximately 20 ppm has very good fatigue limits. Furthermore, using a silico-calcium treatment makes it possible to make steel with a low oxygen content, which could give poor fatigue limits. However, it may be observed that within a steelmaking practice (EF and RH/Fos) and with numerous results, it is possible to determine a correlation between oxygen content and fatigue properties. The correlation, therefore, is not a general one, and oxygen content can be considered only as a relative parameter of fatigue limits in bearing steel within a particular process.

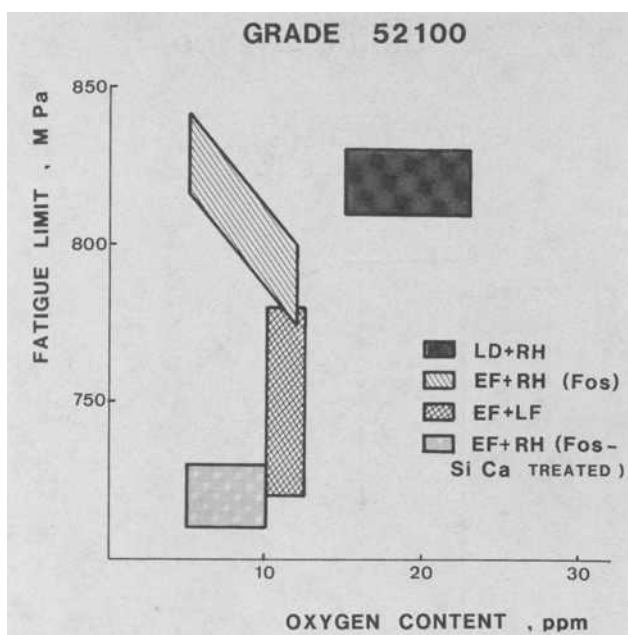


FIG. 11—Total oxygen content and RBF limit for various steelmaking processes.

These results are not surprising: oxygen content alone cannot represent the distribution of oxide inclusions. For example, as in Fig. 10, a high oxygen content (Process A) may represent numerous small inclusions and a low oxygen content (Process B) may represent a flatter distribution, where large inclusions are more frequent. In any case, most of the precipitated oxygen remains in the smaller less harmful oxide inclusions.

Furthermore, it is easy to verify, for example, that ten inclusions of $40 \mu\text{m}/\text{cm}^3$ yielding very poor fatigue properties would only contribute 0.1 ppm to total oxygen content.

Finally, it should also be noted that we have examined a number of vacuum arc remelted (VAR) or electro slag remelted (ESR) steels; in these cases as well, there is no relationship between oxygen content and fatigue properties. In fact, the results showed that remelted metal could have a higher oxygen content and cleanliness rating and retain better fatigue properties than nonremelted metal.

Titanium Content

The detrimental influence of titanium nitrides is well-known [16–20], and the amount of residual titanium has emerged in the last few years as a steel quality indicator [21] in a manner analogous to oxygen. Indeed, titanium is the element which governs the presence of these nitrides because the other elements concerned (carbon and nitrogen) are present in excess. Like that of oxygen, the distribution of titanium appears relatively constant in the cross section of a product; however, it appears that titanium nitrides segregate more than oxides, while both types of inclusions are fragmented and dispersed by hot-working reduction. The determination of many titanium nitride size distributions carried out on steels of various origins and a wide range of titanium contents has also clearly shown that titanium content increases as the number and average size of titanium nitrides increase.

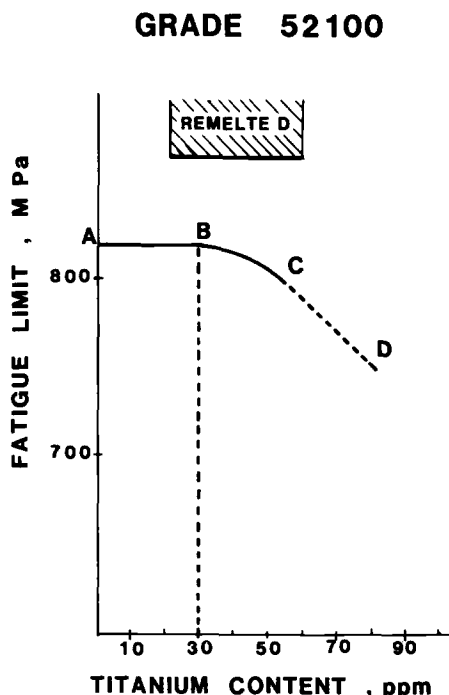


FIG. 12—Schematic representation of the role of titanium content on fatigue life.

Generally, for a given steelmaking practice, the proportion of RBF fractures initiated on titanium nitride inclusions also rises as titanium content increases, and a schematic representation of the role of titanium on fatigue life is proposed in Fig. 12. In the AB portion of the curve, the fatigue limit is controlled by oxides or calcium-enriched sulfides and therefore remains insensitive to titanium content. Conversely, in the CD portion of the line, most ruptures occur on titanium nitride inclusions. The break point depends on the steelmaking process, and the 30-ppm level appears to be the value for the Fos sur Mer process. Other processes seem to have about the same limit [21]. But it also appears that remelted metal has a higher tolerance for titanium, since we found limits greater than 850 MPa with a titanium content of 50 ppm. Research in progress on titanium nitrides reveals other phenomena that it is yet too soon to discuss, but that research will be presented in a future communication. In summary, the deterioration of fatigue life caused by titanium nitrides appears influenced by other factors than size and number. Titanium, as oxygen, therefore can not be considered as an absolute parameter.

Calcium Content

For a number of years we have been aware of the harmful effect of calcium when it is involved in the formation of calcium aluminates or calcium silico-aluminates, this effect being explained essentially by their size. They are often globular and, unlike other inclusions, are neither deformed nor broken by hot-working reduction. It has even been observed that beyond a certain point, increasing hot-working reduction reduces fatigue properties.

The negative effect of these inclusions is now known in other uses and in other hardened steels. It is equally known that they are not completely destroyed by VAR or ESR remelting, and therefore the quality of a remelted metal depends largely on the quality of the first melting. As we have mentioned above, they do not relate to oxygen content, and their distribution practically prevents their detection, even under severe micrographic testing. Their detection requires testing of a much larger volume of metal than that amenable to micrographic methods.

Calcium is now routinely used in order to enhance machinability, but it is also used to modify the shape and the distributions of nonmetallic inclusions. This is, in fact, how the B thin rating, easily seen micrographically but relatively harmless, can be changed into other inclusions that are larger, more widely dispersed, and therefore more difficult to detect. This is also why some steel producers prefer the use of calcium in order to more easily meet cleanliness specifications.

In bearing steels with low oxygen, calcium can partially replace manganese in manganese sulfides. Calcium-enriched sulfides are much less malleable and act as crack-initiation sites, in a way comparable to oxides, as seen in Fig. 8. Not added intentionally, residual calcium arises from the slag and the linings in contact with the metal. In this respect, the RH vacuum degassing system proves particularly well-suited, since it performs the metallurgical operations with a minimum of slag-metal interactions.

Thus a number of conclusions can be drawn that raise questions concerning the influence of calcium on fatigue properties. However, as the results obtained with oxygen and titanium reveal, it is obvious that caution must be exercised before considering calcium as an absolute parameter. Metallurgical phenomena being more complex than meets the eye, it is possible that calcium could be found in a nonharmful form.

Discussion

Observations that may be made at this stage of development are clear but pessimistic. It has been demonstrated that the major metallurgical characteristics currently used to predict or verify the behavior of bearing steel are not dependable. From the observations made, it appears necessary to specify how these various parameters should be used and what procedure can be developed to guarantee a reliable assessment of the intrinsic quality of a bearing steel, keeping in mind that segregation and hot-working reduction should be carefully checked before evaluating steelmaking parameters.

Cleanliness Ratings

Fatigue fractures occur on the most harmful inclusions present in the stressed zone. These inclusions may be oxides, titanium nitrides, or sulfides; for this discussion let us consider only oxides which form the most important type observed in cleanliness ratings. Figure 13 shows an overall view on this type and gives a schematic representation of the two size distributions A and B of Fig. 10. The graph is presented using a logarithmic scale: the x -axis shows the cross-sectional dimension of oxide inclusions, and the y -axis the number of inclusions per cubic centimetre of metal (density).

As it is explained by Chone [22], it is possible to have a three-dimension estimation of an inclusions distribution starting from a two-dimension examination like micrographical cleanliness rating. Thus, the micrographical examination of six ASTM Specification for High Carbon and Roller Bearing Steel (A 295-84) test specimens with a surface area of less than 20 cm² only represents an effective volume of 10 to 20 mm³, and therefore cannot represent a nonmetallic

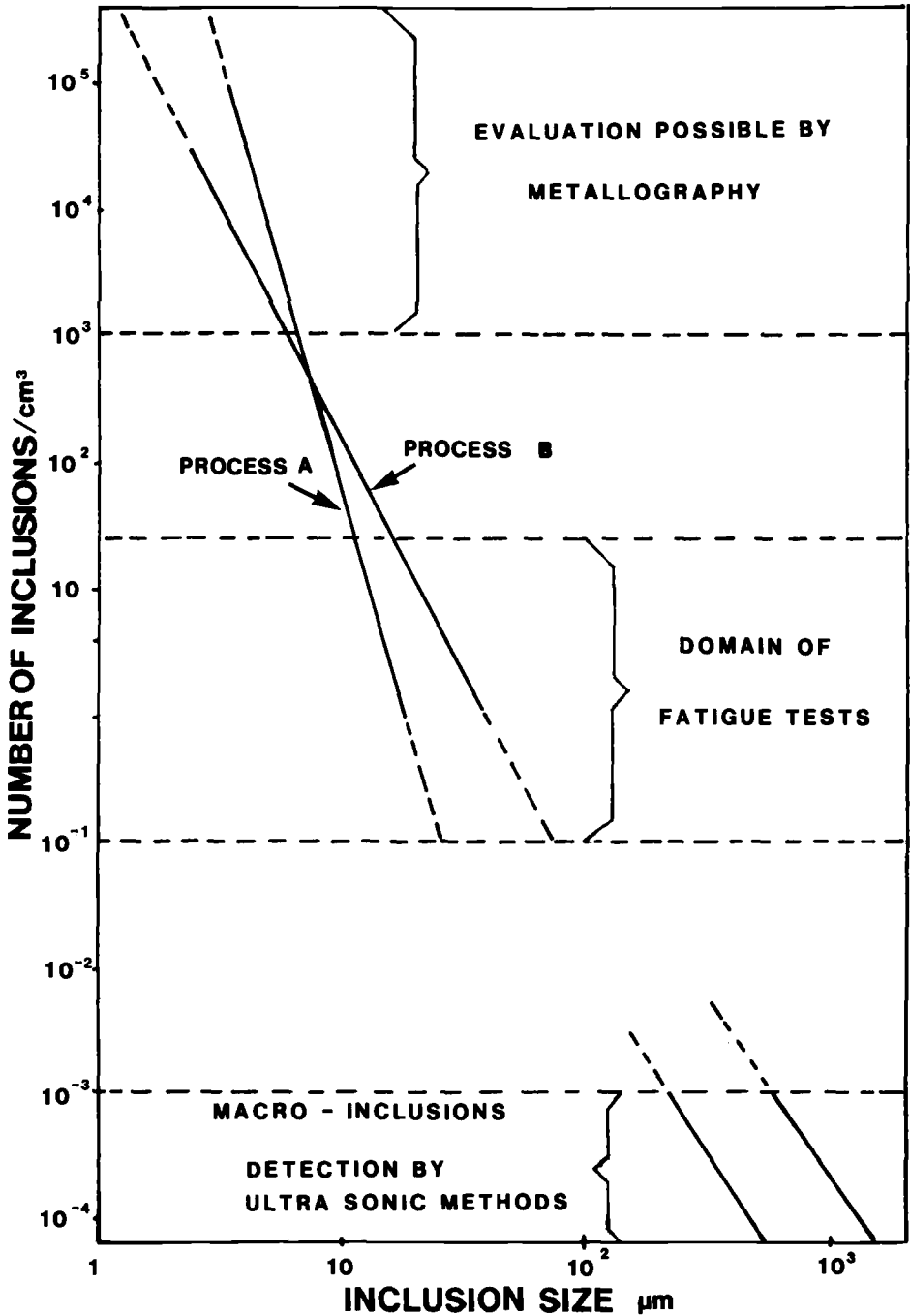


FIG. 13—Representation of different oxide inclusion size distributions.

inclusion density of less than $1000/\text{cm}^3$ [22]. This corresponds to the upper part of Fig. 13. Indeed, it is well-known that the greater the number of inclusions involved in a micrographic examination, the more reliable and significant are its results. Testing a series of fatigue test specimens or bearings takes into consideration much larger volumes of metal. It is very sensitive to densities to the order of a few inclusions per cubic centimetre, and this, therefore, corresponds to another part of Fig. 13 that can be located roughly between 0.1 inclusion/ cm^3 and 50 inclusion/ cm^3 . For the two Processes A and B in Fig. 13, the two distributions cross each other and illustrate the absence of correlation between cleanliness ratings and fatigue life. In certain processes, it is quite possible that modifications in steelmaking may result in a shift of the whole distribution. In such cases, a relation between cleanliness and fatigue will be found [23]. In the same figure, for information, is also shown an approximate representation of the distribution of macrographic inclusions; the area corresponding to these inclusions can be roughly located under 10^{-3} inclusions/ cm^3 . These exogenous inclusions depend on parameters other than those at the origin of the main distribution and thus cannot be expected to be related in number or size to the main trend. Progress in the casting process has greatly reduced these inclusions.

Given these conditions, it could be stated that micrographic examination should be considered only as a relative parameter of fatigue properties. Improved by means of quantitative metallography, it can be assumed that new rating methods will contribute better information to nonmetallic inclusion distributions.

Other Characteristics

By the same methods, we have demonstrated that the analytical parameters such as oxygen, titanium, and calcium content are incapable of providing a correct rating for the different processes. Therefore, they should be also considered as relative parameters and used only to indicate the consistency of a process.

With all the classical measurements failing to provide an absolute parameter for fatigue life, it clearly appears that only the fatigue test should serve as a reference for the performance of bearing steels. A standardization of the testing methods would indeed be welcome; even though it may not always produce absolute results, it could be used to provide comparisons between different steelmaking processes.

Finally, quality qualification and process control methods must evolve to keep up with today's user requirements in order to guarantee quality. Qualification of a new steel supplier or a variation in the manufacturing process should be made using fatigue tests under a quality assurance program. A statistical process control incorporating the critical metallurgical parameters would then be helpful to guarantee the reliability of the process.

Conclusion

The Phoenix Symposium on the Effect of Steel Manufacturing on the Quality of Bearing Steels is particularly welcome, since the advances in steel technology in the past few years have led to a large variety of steelmaking processes and an evolution of the classical metallurgical parameters. It is important to take a critical look at the influence of these parameters. The effects of inclusion type and size on fatigue life have been clearly demonstrated. The criteria consisting of oxygen content, titanium content, and the cleanliness rating do not relate directly to fatigue properties. For a specific process qualified by fatigue tests under a quality assurance program, these parameters can be useful as indicators for the statistical process control procedure.

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DISCUSSION

Erwin V. Zaretsky¹ (written discussion)—The technical content of the author's paper is excellent. However, the discussor has reservations with regard to the authors' results as related to rolling-element bearings. While it may be possible to study the relationship of inclusions and fatigue life using rotating beam specimens, it is not possible to quantify these effects to rolling-element bearings. Results published by ASTM in 1959 by Sachs, Sell, and Brown² and in 1960 by William J. Anderson and the discussor³ failed to show a relation between rolling-element fatigue life (Type II failure mode) with standard testing methods used to evaluate standard (nonrolling) fatigue life (Type I fatigue mode). This lack of correlation can be explained by the fact that classical rolling-element fatigue is subsurface originated while, as example, rotating beam (R. R. Moore) is of surface origin. Specimen preparation, surface condition, and environment significantly affect the resultant life. Specifically high-hardness brittle materials such as bearing steels will generally be more sensitive to surface imperfections than the softer more ductile materials. As a result, there is no reasonable substitute for rolling-element fatigue bench testing. Furthermore, it is the discussor's experience that most researchers using rotating beam specimens do not obtain a sufficient number of failures to reach statistically significant results, much less relate the results to rolling-element fatigue.

Jacques Monnot, Bernard Heritier, and Jean Y. Cogne (authors' closure)—The reason for the correlation and its limits between RCF (rolling contact fatigue on bearings) and RBF (rotating bending fatigue on Moore Specimens) are explained in our first publication.⁴

Concerning crack initiation on the RBF Moore Specimen, we used a special surface preparation with diamond polishing which gave us practically no surface crack initiation sites. On the contrary, we observed that these crack initiation sites are subsurface nonmetallic discontinuities and more often subsurface nonmetallic inclusions.

Rolling-contact fatigue is a very complex phenomenon where numerous parameters are involved. Many of these parameters are related to the bearing manufacturing, heat treatment, geometrical parameters, surface finish, lubrication, etc. Others are related to usage conditions such as stress conditions, speed, and temperature. Our research and tests on finished bearings show that under certain usage or test conditions, the steel manufacturing parameters appear more clearly to have an influence on the bearing life. As a result, it can be concluded that the correlation between rolling contact fatigue and rotating bending fatigue does exist, but only for the steel manufacturing parameters and under specific conditions.

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²Sachs, G., Sell, R., and Baum, W. F., Jr., "Tension, Compression, and Fatigue Properties of Several Steels for Aircraft Bearing Applications," *Transactions, American Society for Testing and Materials*, Philadelphia, Vol. 59, 1959, pp. 635-657.

³Zaretsky, E. V. and Anderson, W. J., "Relation Between Rolling-Contact Fatigue and Mechanical Properties for Several Aircraft Bearing Steels," *Transactions, American Society for Testing and Materials*, Philadelphia, Vol. 60, 1960, pp. 627-649.

⁴Monnot, J., Tricot, R., and Gueussier, A., *Revue de Métallurgie*, Vol. 67, July 1970, pp. 619-638.

The Distribution and Quantitative Relationship of Oxygen and Inclusions in High-Carbon Ball Bearing Steel

REFERENCE: Bombardelli, B., Pacchiani, G., Holzner, H., and Hoo, J. J. C., "The Distribution and Quantitative Relationship of Oxygen and Inclusions in High-Carbon Ball Bearing Steel," *Effect of Steel Manufacturing Processes on the Quality of Bearing Steels*, ASTM STP 987, J. J. C. Hoo, Ed., American Society for Testing and Materials, Philadelphia, 1988, pp. 166–188.

ABSTRACT: This paper summarizes the results of a series of investigations conducted in various time periods, in conjunction with each innovative change of equipment and process at Acciaierie di Bolzano S.p.A. in Bolzano, Italy. The results discussed are limited to three major areas of qualitative improvement for high-carbon ball bearing steel equivalent to AISI E 52100 grade: macroinclusions, microinclusions, and oxygen content. Based on the latest results obtained from the most up-to-date technology, a simplified method is proposed for the quality assurance sampling and testing in the production of AISI E 52100 steel.

KEY WORDS: macroinclusions, microinclusions, oxygen content, air melting, vacuum degassing, RH degassing, ASEA installation, ladle refining, top pouring, bottom pouring, shrouding, reoxidation, distribution of inclusions in ingots, homogeneity.

The Melt Shop of Acciaierie di Bolzano S.p.A. is equipped with two 45-metric ton basic electric furnaces, one Rukestahl-Heraeus (RH) vacuum degassing unit, and one ASEA ladle refining installation with degassing. The Allmanna Svenska Electuska Aktiebolaget (ASEA) ladle installation is equipped with facilities for vacuum oxygen decarburization treatment and powder injection. All heats produced in both production lines are bottom-poured into ingots with gas shrouding. Each heat is poured onto two plates, each plate contains six ingots of 3.6 metric tons each. Throughout the paper, quality comparisons are made with top-poured ingots produced in the past, when each heat was poured into 14 ingots of 3.1 metric tons each.

This paper presents the results and analyses of four methods of production of American Iron and Steel Institute (AISI) E 52100 steel for ball bearings at various time periods:

1. The early 1960s: air melt, not vacuum degassed, top-poured.
2. From 1967 to early 1970s: air melt, RH vacuum degassed, top-poured.
3. From 1975 to 1983: air melt, RH vacuum degassed, bottom-poured without shrouding.
4. Since 1984: air melt, ladle refining in ASEA installation, bottom-poured with shrouding.

Sampling Method

When the bottom-pouring method is employed, ingots have dimensions of 540 mm across the top, 455 mm across the bottom, and 2210 mm long. Ingots were rolled into 127 by 127-mm billets after cropping. Seven specimens were taken from each ingot for quality evaluation in

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accordance with the scheme shown in Fig. 1. Four ingots per heat, the first, middle, next to last, and last ingots, were targeted for quality investigation. They are referred to as ingots Nos. 1, 7, 13, and 14.

When bottom-pouring was adopted, only two ingots per heat were sampled. One ingot each was taken from the first and the second plate. Experience and quality control evidence indicated clearly that all ingots produced from the same plate possessed amazingly identical homogeneity of quality. Samples were taken from each billet for the examination of macroinclusion, microinclusion, and oxygen determination. The seven ingot specimens represented the top and bottom of the usable portion of the ingot after cropping and five other locations evenly distributed throughout the ingot.

Billet samples for macroinclusion examination were machined to rounds with diameters representing 90, 75, and 60% of the billet's cross section. These samples had diameters of 114, 95, and 76 mm, respectively, and were 200 mm long.

For microinclusion examinations, three specimens were cut from each billet sample, as shown in Fig. 2. Each specimen was 21.2 mm wide and 15.8 mm thick, representing the entire half cross section from the edge to the center of the billet. The specimens were designated as S, $\frac{1}{2}$ R, and C, representing surface, half-radius, and center of the billet, respectively. Since seven samples were taken from each ingot and three specimens were taken from each billet, 21 specimens of each ingot were examined for microinclusions with a total area of 7000 mm². For each billet, an area of 1000 mm² was examined under the microscope.

For oxygen determination, a LECO R016 apparatus was used in the early years and a more modern LECO TC136 apparatus in recent years. Three analyses were made for each sample, and the average readings were reported. These represented 63 analyses of oxygen for each ingot.

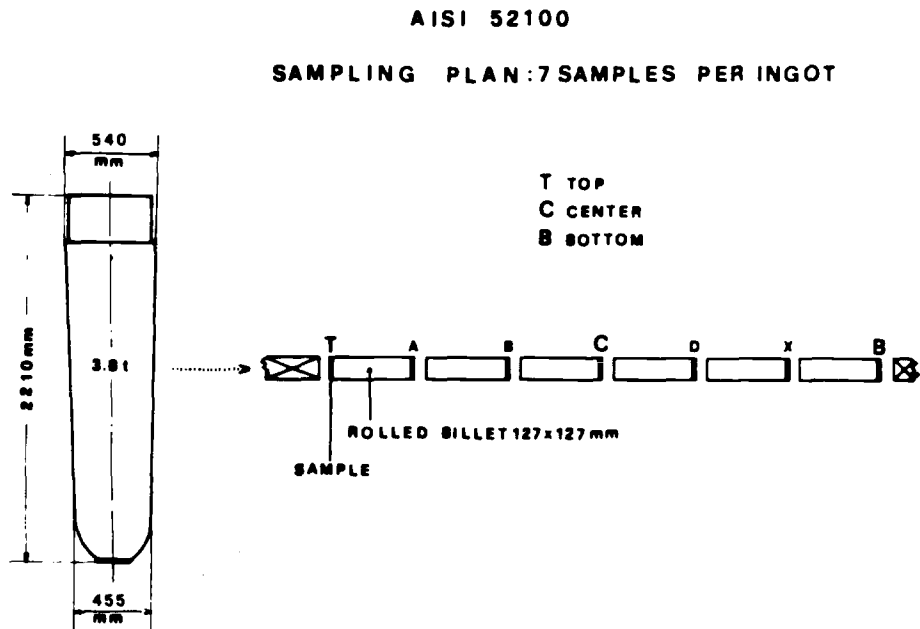


FIG. 1—Sketch for sampling.

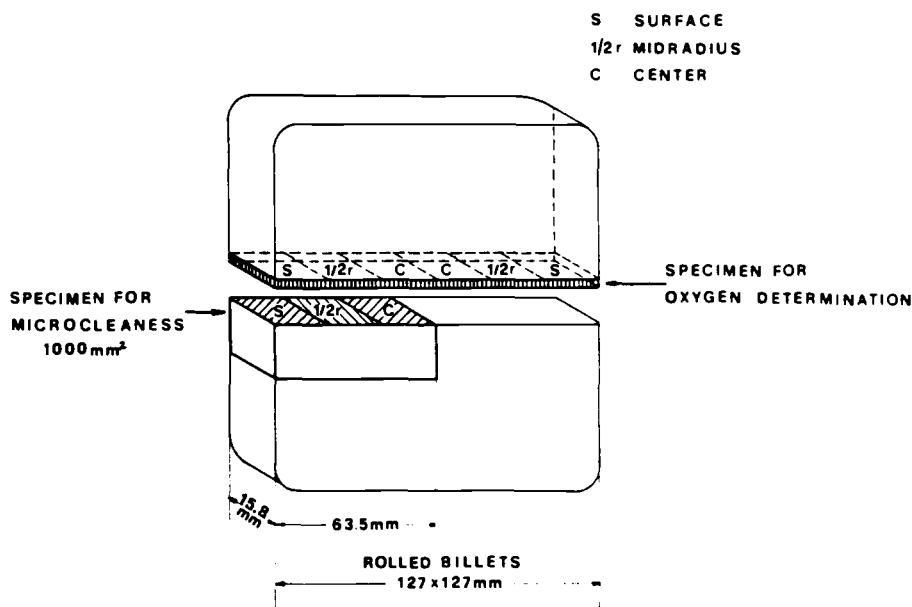


FIG. 2—Sketch for specimen locations, AISI E 52100.

It can be seen that samples were taken from the top to the bottom and throughout the body of the ingot for oxygen determination, and macro- and microinclusion ratings. Samples also included areas near the surface, from the mid-radius section, and the center of each ingot. This extensive sampling plan was rigorously maintained throughout the four time periods described in the introduction. The comparison of results can, therefore, be accepted with confidence.

Macroinclusions

Top-Poured Ingots from Vacuum-Degassed and Nonvacuum-Degassed Heats

All specimens were examined according to the ASTM Standard Practice for Determining the Inclusion Content of Steel (E 45-85, Section 3.1.3, Step-Down Method). It should be pointed out that this method requires that the test sample be machined to specified diameters below the surface and surveyed for inclusions under good illumination with the unaided eye or with low magnification. This test is essentially used to determine the presence of inclusions 3.18 mm ($\frac{1}{8}$ in.) in length and longer. Although this method has never been used as extensively as ASTM E 45, Section 3.1.4, Magnetic Particle Method in the United States, it was used extensively with high confidence in continental Europe for AISI E 52100 bearing steel.

The results of macroinclusion ratings, expressed in length of mm per square meter area examined, are shown in Fig. 3. The results were compiled from 48 vacuum-degassed heats and 33 nonvacuum-degassed heats produced in 1972. Figure 3 shows that there is an inhomogeneity between the first and the last ingot of the heat, with higher macroinclusion counts in the next to last and last ingots.

This trend of increasing macroinclusions in ingots as the top-pouring proceeded was perhaps an experience common to all producers employing top-pouring. The trend was related to the increased number of openings and shuttings of the slide gate on the bottom of the ladle. More

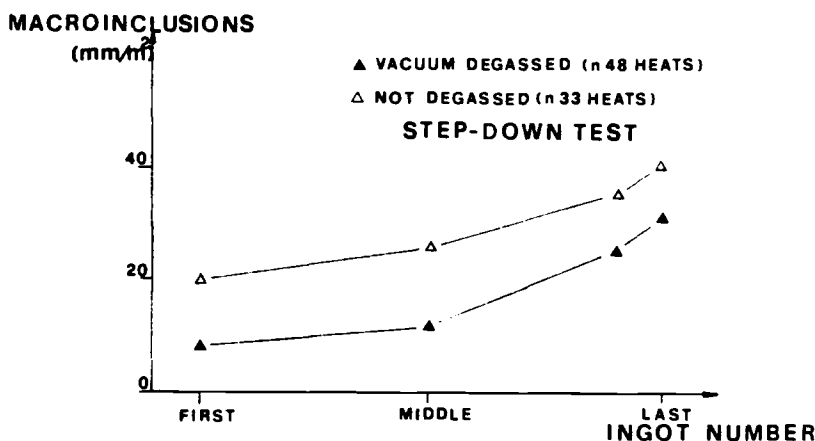


FIG. 3—Macroinclusions of top-poured AISI E 52100 steel ingots, vacuum degassed versus nonvacuum degassed, 1972.

reoxidation products would form in areas near the nozzle, and they would fall into the ingot as the teeming proceeded. Furthermore, nonmetallic inclusions would coagulate in the ladle and form aggregates of oxides that could not float completely and were, therefore, concentrated in the last portions of the liquid steel. The third cause was that as the sequence teeming proceeded, the reoxidation of the stream was easier because the stream was broken up, thus allowing more air to be entrapped. Furthermore, more refractory fragments would be eroded out due to the vehement turbulence of the liquid steel and fall into the ingot mold. More detailed discussion of this can be found in articles of Schwerdtfeger [1] and Schaefer [2].

The vacuum-degassed heats presented a decisive improvement as shown in Fig. 3, especially in the first ingots. The magnitude of improvement was almost 2:1, in spite of the fact that it was generally believed that the lower teeming temperatures of vacuum-degassed heats would produce more inclusions. The improvement of macroinclusion counts in the vacuum-degassed heats was the result of the oxygen removal from the liquid steel by vacuum before teeming. It should be noted that the trend for higher macroinclusions found in the last ingots remained the same as the nonvacuum-degassed heats.

The macroinclusion counts also showed an increase within a given ingot from the top to the bottom. Figure 4 shows this trend and represents the results of middle ingot (No. 7) of vacuum-degassed heats. The reason for this trend was that since the bottom part of the ingot solidified first, the solidification impeded the flotation of oxide aggregates; all oxidized drops of molten steel on the mold walls derived from the initial splashing were thus entrapped in this region.

Top-Poured and Bottom-Poured Ingots from Vacuum-Degassed Heats

During the second half of 1970s, bottom-pouring was introduced at Acciaierie di Bolzano for the teeming of American Iron and Steel Institute (AISI) E 52100 bearing steel. Studies were conducted to compare the results of top-pouring and bottom-pouring. When bottom-pouring was first introduced, shrouding of liquid steel stream had not yet been adopted. An ingot was top-poured from the same heat of steel between the bottom pouring of two plates and the macroinclusion counts compared.

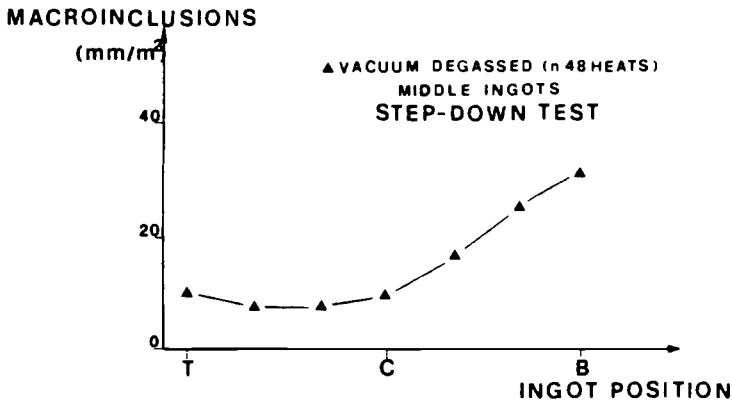


FIG. 4—Macroinclusions of top-poured AISI E 52100 steel, vacuum degassed, middle ingot, 1972.

The difference of macroinclusion counts between the two types of teeming methods was significant, especially near the bottom of the ingot. The average values obtained from a bottom-poured ingot from plate No. 1 and from the top-poured ingot from the same heat for 16 vacuum-degassed heats are given in Fig. 5.

Another significant advantage of bottom-pouring was the improved homogeneity from ingot to ingot of the same heat. There was virtually no difference in quality between the ingots from the first plate and those from the second plate. This experience coincided with those of other specialty steel producers [3,4].

Bottom-Poured Ingots from Heats Produced by ASEA Ladle Refinement

Macroinclusions in bearing steels, produced from electric furnace, ladle refined in ASEA installation, and bottom-poured, can practically be eliminated in the ladle. The macroinclusions found in bottom-poured AISI E 52100 bearing steel with low oxygen content is the result of the reoxidation of the liquid steel stream, the erosion of refractions, and the entrapping of mold powder.

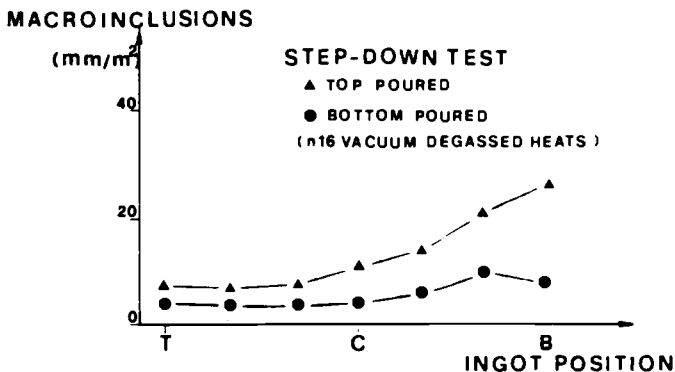


FIG. 5—Macroinclusions of AISI E 52100 steel heats, vacuum degassed, top-poured versus bottom-poured.

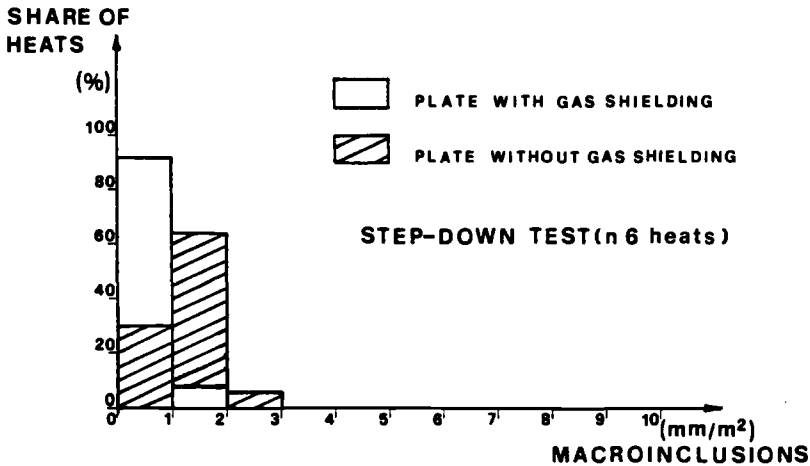


FIG. 6—Macroinclusions of AISI E 52100 steel heats, bottom-poured, with gas shrouding versus without gas shrouding, 1985.

The use of a shrouding gas to prevent reoxidation became very important. In Acciaierie di Bolzano, a strong flow mixture of argon and Tetrene gas (a mixture of methyl acetylene (C_3H_5) and propane (C_3H_8) gases) was directed to a ring of porous brick installed around the slide gate of the ladle. During the combustion of the Tetrene gas, all oxygen near the slide gate would be consumed. There was less than 0.2% oxygen found around the stream. The red flame that developed was very visible, and it prevented the cooling of the liquid steel stream. The operator could easily observe the process and make adjustments if any irregularities developed during teeming.

After the ladle was positioned in line with the trumpet, a large volume of shrouding gas was immediately introduced to purge all air in the runners and molds. Only when teeming began was the pressure of the shrouding gas reduced to a normal level.

Figure 6 shows the distribution of the average macroinclusion counts of seven samples in bottom-poured ingots. Protective gas shrouding was used in pouring Plate No. 1, while no protective gas shrouding was used for ingots in Plate No. 2 of the same heat.

It is evident from the results that for all practical purposes there was no measurable macroinclusion count, especially with a shrouding gas. Over 90% of the heats had a zero count, even when the method of magnetic particle inspection according to Aerospace Material Specification (AMS) 2301 was used.

The comparison of heats with gas shrouding versus those without also indicated an improvement on the average of 1.3 ppm of oxygen and 11 ppm of nitrogen in the steels produced with gas shrouding.

In conclusion, the adoption of a good teeming practice, the use of the right type and quantity of mold powder and flux [5], and the choice of high-quality refractories can preserve the high quality of steel already obtained in ladle refining. It is important that the possibility of reoxidation be eliminated so that macroinclusions do not form.

Microinclusions

The rating of microinclusions in AISI E 52100 bearing steel has been extensively discussed and well-documented [6]. Today, the general consensus is that when microinclusions are rated

according to ASTM E 45-85, Plate I and II, Type C silicate inclusions can no longer be found if silicon is not used as a deoxidation agent in the steelmaking process. Type A sulfide inclusions are related to sulfur content in the steel. The ASEA ladle refining process can effectively reduce sulfur content in the bearing steel to less than 0.005%, but to improve the machinability, the sulfur is intentionally controlled to slightly above 0.010%. The sulfide inclusions formed are not considered detrimental to the bearing fatigue life. Type B aluminates and Type D globular oxides became the target of all recent studies, and are hence the only types of inclusions specified to be reported by SAM method per the ASTM Specification for High-Carbon Ball and Rolling Bearing Steel (A 295-84). In recent years, the presence of titanium carbonitride inclusions attracted the attention of bearing metallurgists, but no great adverse effect on bearing rolling contact fatigue has been found if the titanium content in bearing steel is limited to 30 ppm maximum.

In the microcleanliness comparison study of AISI E 52100 bearing steels produced in various time periods, using various production methods discussed in the introduction of this presentation, quantitative ratings per German method Stahl und Eisen Pruefblatt 1570 (SEP 1570) was used. This method can very effectively evaluate the quantitative quality level of bearing steel. It was perhaps the most widely used quantitative method before the SAM method was adopted by ASTM Committee A01.28 in 1984. In the studies conducted at Acciaierie di Bolzano, all inclusions in sizes equal to or greater than picture 1 of the Diergarten Chart ($\geq DG 1$), were recorded to increase the sensitivity of the results. It should be noted that in SEP 1570-71, this number designated as characteristic value K (for "Kennwort") [7]. In the early 1960s, before risk factors of each inclusion type were adopted by German standard SEP 1570-71, Acciaierie di Bolzano internally adopted a system of risk factors so that a numerical value can be obtained by multiplying the number of inclusions, the size number of the inclusion per the Diergarten Chart and the risk factors (fg). The risk factors used by Acciaierie di Bolzano were

Diergarten Picture No. 1	Risk Factor fg
2	3.0
3	7.5
4	15.0
5	30.0
6	60.0
	99.0

After SEP 1570-71 adopted standardized risk factors, the company used that standard. Although Acciaierie di Bolzano used different risk factors in the 1960s, the essential ingenuity of the SEP 1570-71 method was already there.

Top-Poured Ingots from Vacuum-Degassed and Nonvacuum-Degassed Heats

Billet specimens cut per Fig. 2 were taken from first (No. 1), middle (No. 7), next to last (No. 13), and last (No. 14) ingots. Three specimens representing the surface (S), mid-radius ($\frac{1}{2}R$), and center (C) sections of the billet (and also ingot) were polished for microinclusion ratings. A preliminary study established that results of the first and middle ingots (Nos. 1 and 7) could be grouped together. Likewise, the results of next to last (No. 13) and last ingots (No. 14) could be grouped together. No significant difference was found according to this grouping.

Figures 7 to 10 show the distribution of microinclusions from top to bottom of the ingot and in positions uniformly scattered throughout. Total microinclusion counts, in Σni (number of inclusions) $\times$ fg (risk factor) were plotted in ordinate. Microinclusions found in the surface area, mid-radius, and center portions of the ingots were plotted separately.

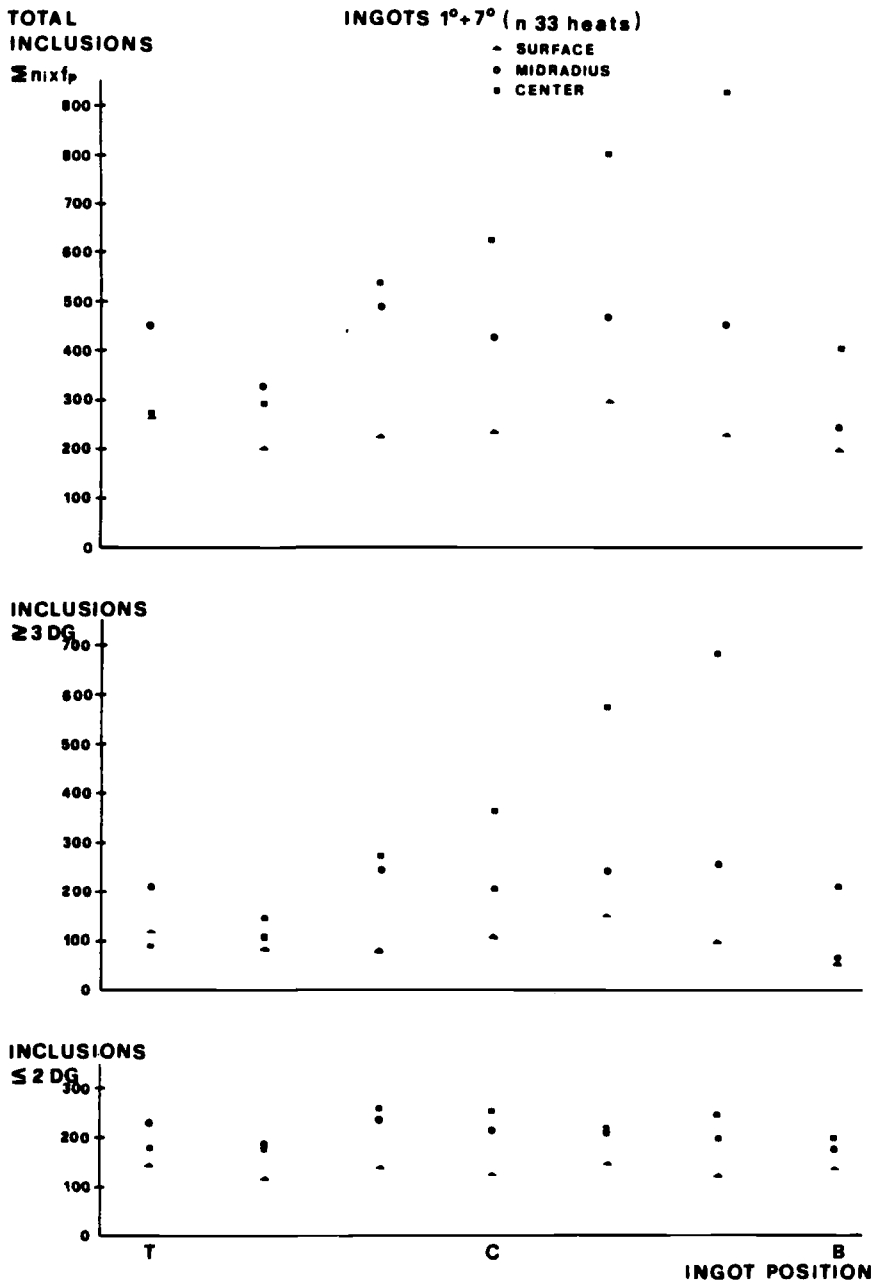


FIG. 7—Distribution of microinclusions, AISI E 52100 steel heats, nonvacuum degassed, top-poured, ingots Nos. 1 and 7.

The figures revealed a marked difference of microinclusion counts between the top portion (positions T, a, b) and the lower half of the ingot (positions C, d, x, B). In the lower half of the ingot there was in fact a progressive deterioration from the surface to the center of the ingot. The areas of higher microinclusion concentrations were in the regions of the central axis of the

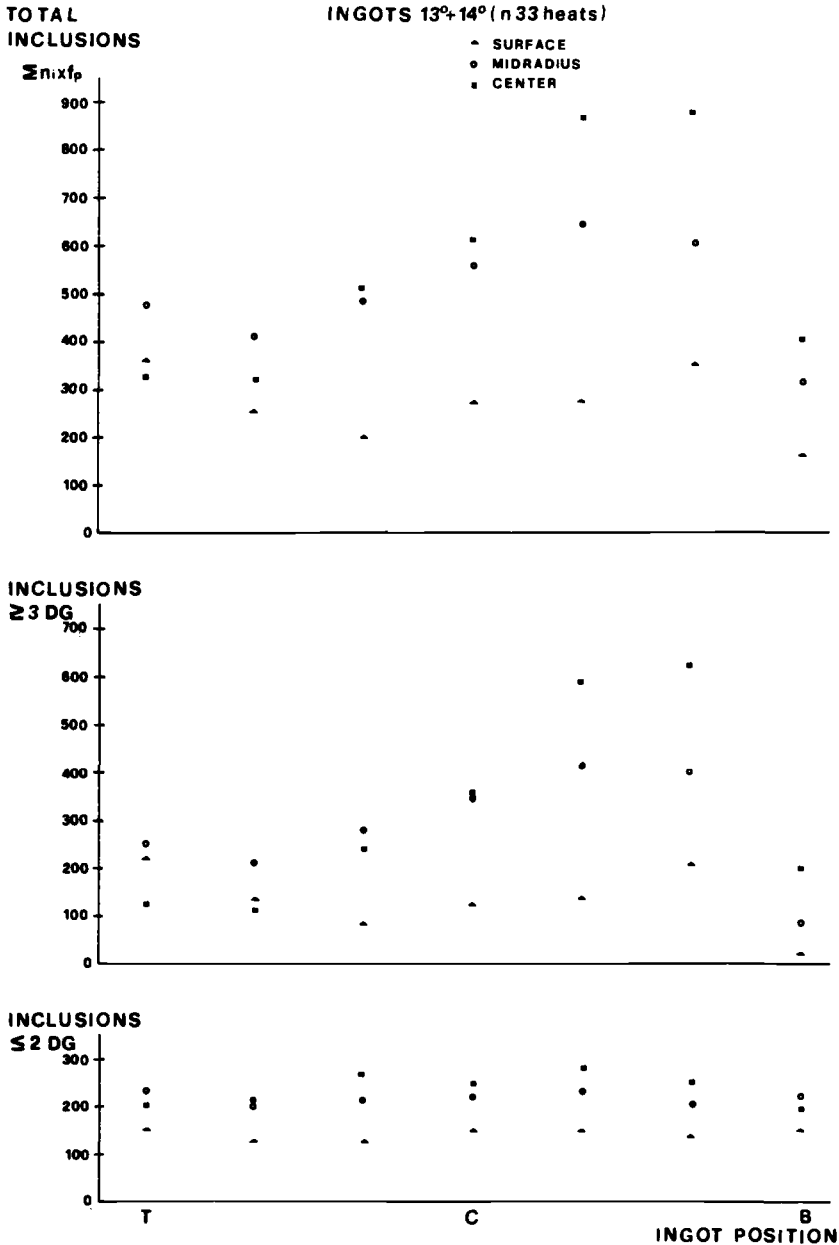


FIG. 8—Distribution of microinclusions, AISI E 52100 steel heats, nonvacuum degassed, top-poured, ingots Nos. 13 and 14.

ingot. The bottom position of each ingot again had fewer microinclusion counts than the next to last x position. Note that while more microinclusions were found in the lower $\frac{2}{3}$ of the ingot near the central axis, in the top $\frac{1}{3}$ of the ingot more microinclusions were found in mid-radius positions, thus forming a Y shape for microinclusion concentration in the ingot. This phenomenon was no doubt developed from the pattern of solidification.

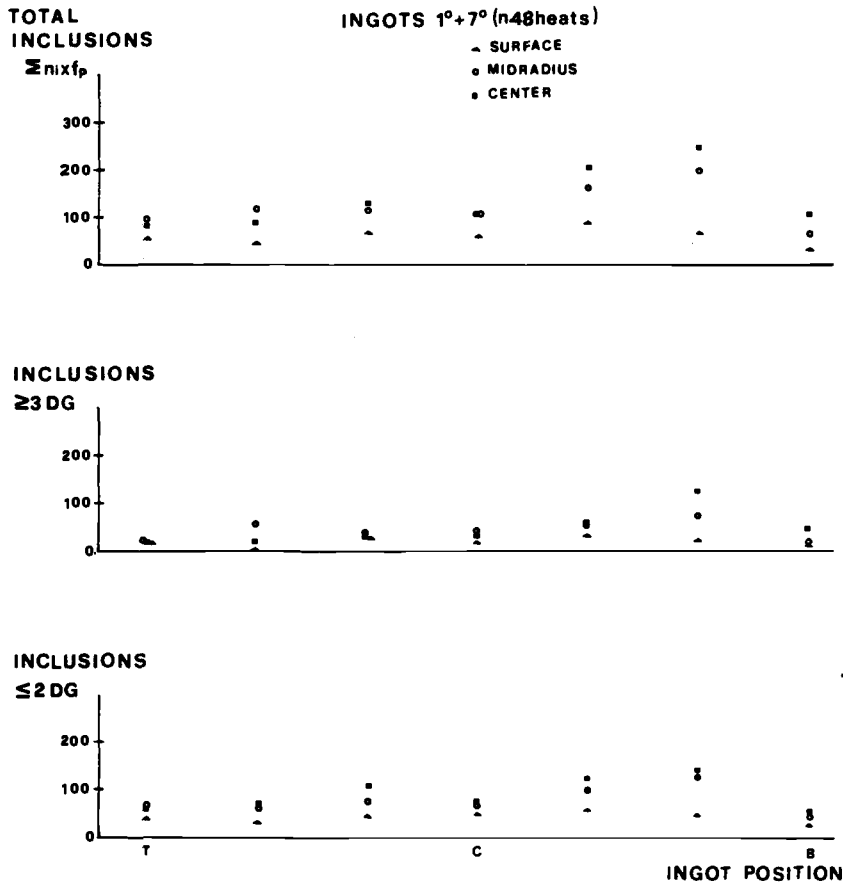


FIG. 9—Distribution of microinclusions of AISI E 52100 steel heats, vacuum degassed, top-poured, ingots Nos. 1 and 7.

Figures 7 and 8 are the results of nonvacuum-degassed heats, whereas Figs. 9 and 10 are the results of vacuum-degassed heats, all top-poured. The cleanliness improvement by vacuum degassing is evident by comparing the value on the ordinates. While Figs. 7 and 8 of the nonvacuum-degassed heat have microinclusion counts as high as 900, Figs. 9 and 10 of the vacuum-degassed heats have values of less than 300 for ingots Nos. 1 and 7, and 300 to 600 for ingots Nos. 13 and 14. The range of improvement was 3:1 to 2:1.

In Figs. 7 to 10, microinclusions in sizes equal to or less than Diergarten No. 2 ($\leq DG 2$) were plotted separately from the microinclusions with sizes equal to or greater than Diergarten No. 3 ($\geq DG 3$). These diagrams show that most differences in total microinclusion counts between the top and lower halves of the ingots were the result of the difference exclusively of sizes $\geq DG 3$. The small differences between the surface and the center region of the ingot, on the other hand, were caused by small microinclusions of $\leq DG 2$. Similar to the distribution of macroinclusions, there was an increase of microinclusions from the first to the last ingots, especially in the lower half of the ingot. The deterioration was more pronounced for vacuum-degassed heats (Figs. 11 and 12), which normally were teemed at lower temperatures.

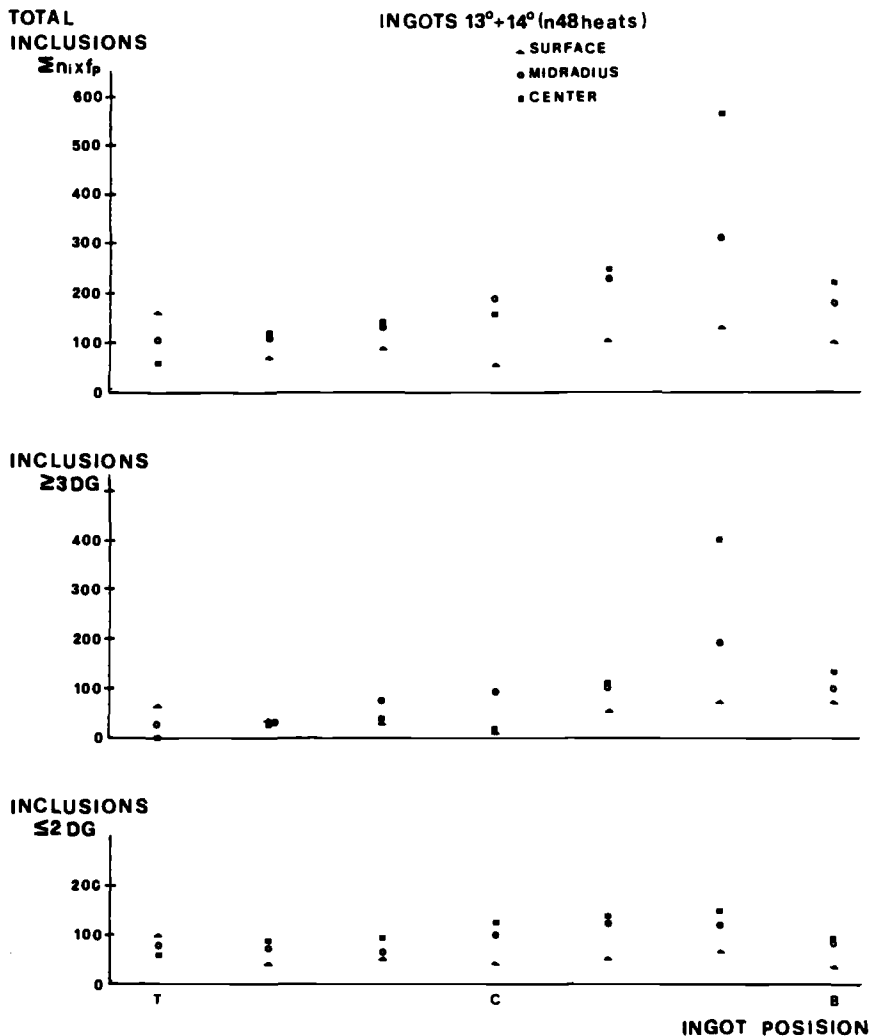


FIG. 10—Distribution of microinclusions of AISI E 52100 steel heats, vacuum degassed, top-poured, ingots Nos. 13 and 14.

In all cases, whether from heats vacuum-degassed or not, the difference of microinclusion counts could all be attributed to the difference in microinclusions of sizes $\geq DG 3$. The results of sizes $\leq DG 2$ were practically identical. This fact indicated strongly that the large-size microinclusions were developed due to the difficulty of the flotation of the inclusions in residual liquid steel in the ladle due to the lowering of the temperature.

The difference in the microcleanliness between vacuum-degassed and nonvacuum-degassed heats was found in all positions of the ingot (from top to bottom) and from the first to the last ingot of the heat. Only in the regions of maximum inclusion concentration, such as the center of the ingot in position next to the bottom (position x), was the ratio of improvement somewhat diminished.

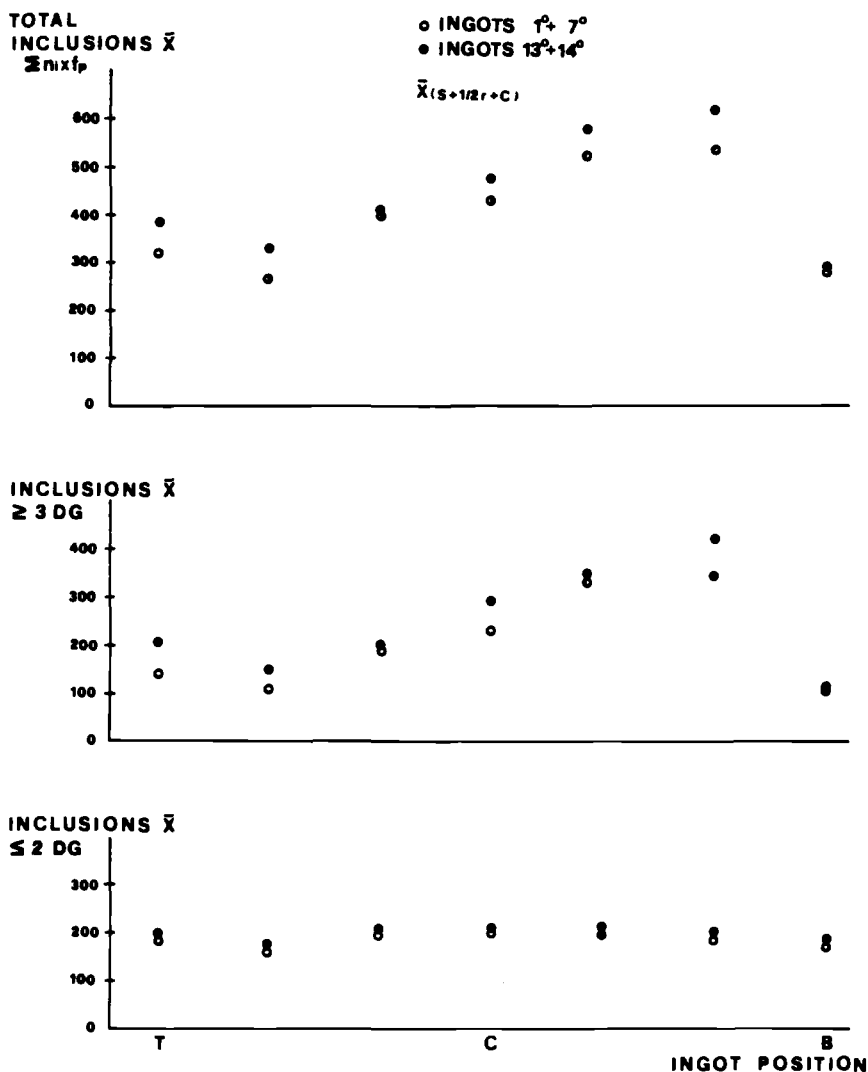


FIG. 11—Distribution of microinclusions of AISI E 52100 steel heats, nonvacuum degassed, top-poured, ingots Nos. 1 and 7 versus ingots Nos. 13 and 14.

Top-Poured and Bottom-Poured Ingots from Vacuum-Degassed Heats

With the introduction of bottom-pouring, studies were conducted to evaluate the difference of microinclusion counts between top-pouring and bottom-pouring in a method similar to that used for macroinclusion evaluation. The same heats were used to produce top-poured and bottom-poured ingots as described in the previous paragraph. The results also included all ratings of inclusions equal to or greater than Diergarten picture No. 1 ($\geq \text{DG } 1$) in order to increase sensitivity.

The improvement with bottom-pouring immediately became evident when data of summation obtained in the 1000 mm² area in the middle position of the ingot by multiplying the average number of microinclusions n_i and the risk factor f_g for the two types of teeming method.

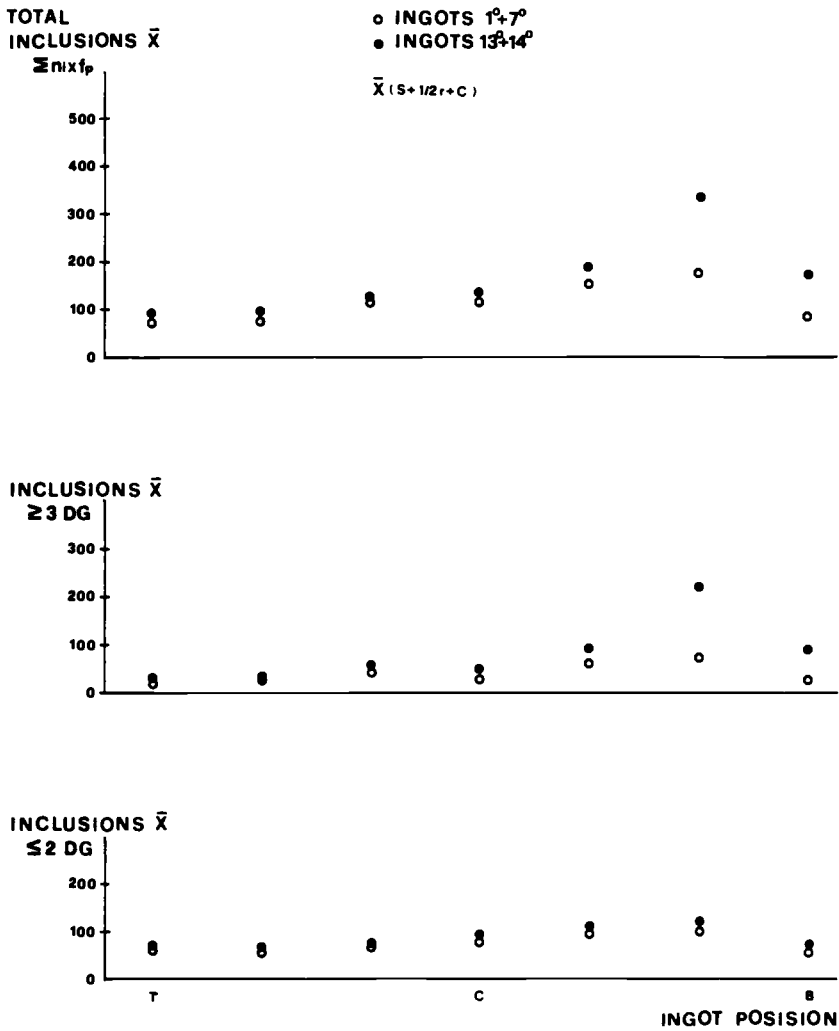


FIG. 12—Distribution of microinclusions of AISI E 52100 steel heats, vacuum degassed, top-poured, ingots Nos. 1 and 7 versus ingots Nos. 13 and 14.

Top pouring

$$\left[\sum n_i(AO) \times fg \right]_{\text{DG 1 to 6}} = 98.3$$

$$\left[\sum n_i(AO) \times fg \right]_{\text{DG 4 to 6}} = 25.8$$

Bottom pouring

$$\left[\sum n_i(AO) \times fg \right]_{\text{DG 1 to 6}} = 56.3$$

$$\left[\sum n_i(AO) \times fg \right]_{\text{DG 4 to 6}} = 8.6$$

For the center position of the ingot (position C), the improvement of counts of microinclusions of sizes $\leq$ DG 4 was less than the large size microinclusion $\geq$ DG 4.

The improvement was more evident when the differences of microcleanliness were calculated for position x (next to bottom) of the ingot where maximum concentration of inclusions was expected. The following summation values were obtained:

Top pouring

$$\left[\sum ni(AO) \times fg \right]_{DG 1 \text{ to } 6} = 242.4$$

$$\left[\sum ni(AO) \times fg \right]_{DG 4 \text{ to } 6} = 84.0$$

Bottom pouring

$$\left[\sum ni(AO) \times fg \right]_{DG 1 \text{ to } 6} = 102.5$$

$$\left[\sum ni(AO) \times fg \right]_{DG 4 \text{ to } 6} = 15.3$$

The difference was attributed to large microinclusion sizes $\geq$ DG 4 to 6. Throughout the heat, the distribution of microinclusions from bottom-pouring was much more uniform, and the total counts were sensibly better when compared with top-poured heats. The general trend of average values are plotted in Fig. 13.

Bottom-Poured Heats Ladle Refined in ASEA Installation

Studies were conducted on 20 heats produced in 1986 with and without vacuum degassing from an ASEA ladle refining installation. Contents of nonmetallic inclusions were rated in accordance with the German method SEP 1570-71, counting all inclusions $\geq$ 1 SEP (K1). Although the SAM method described in ASTM A 295-84 is easier to use and faster to rate, SEP 1570-71 provides more sensitive results as a research tool.

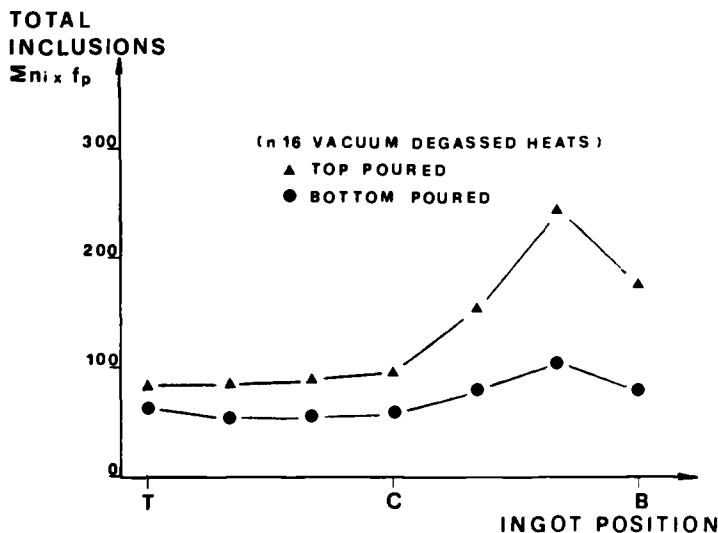


FIG. 13—Microinclusions of AISI E 52100 steel heats, top-poured versus bottom-poured.

It became evident immediately that there was no difference between ingots bottom-poured to the first plate and those to the second plate, when rated to K1. This implied an excellent degree of homogeneity from all heats. The results were thus grouped. Figures 14 and 15 show the distribution of oxidic microinclusions from the top to the bottom of the ingots. Again, in order for us to find the size distribution of microinclusions, sizes ≥ 3 SEP (K3), 1 + 2 SEP and 1 SEP were plotted separately.

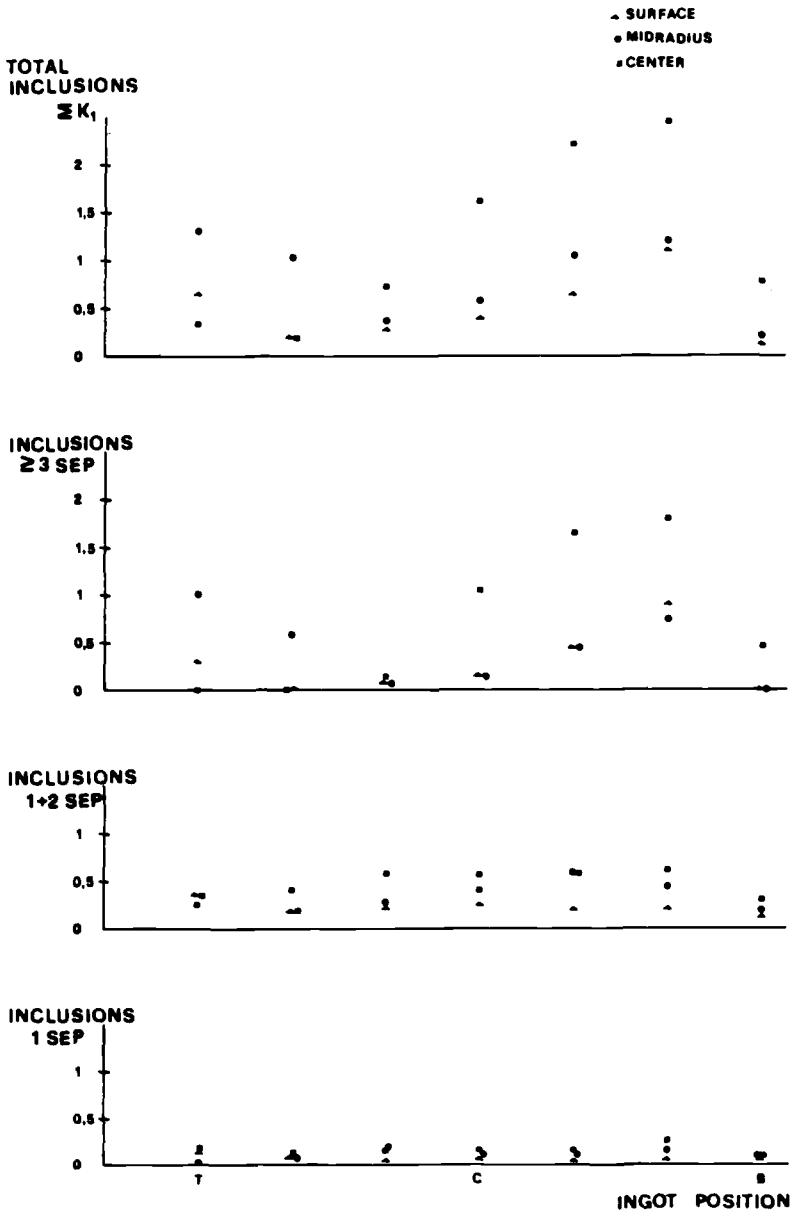


FIG. 14—Distribution of microinclusions of AISI E 52100 steel heats with ASEA ladle refinement, bottom-poured.

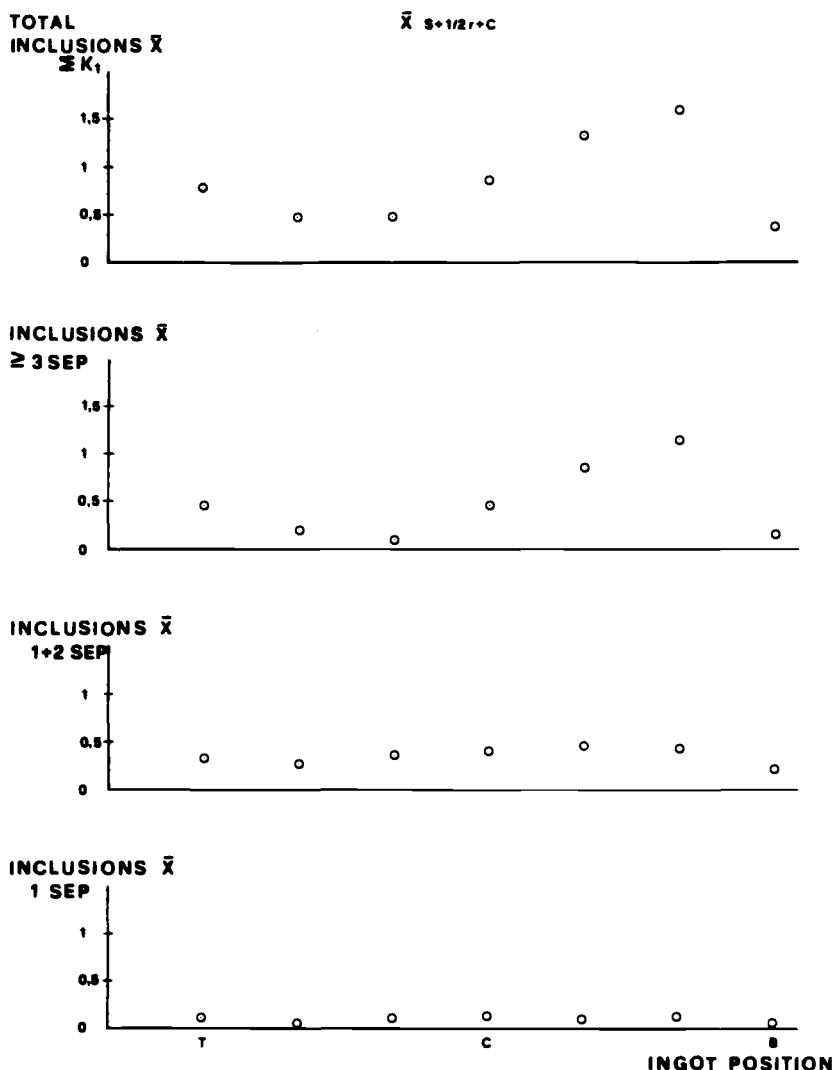


FIG. 15—Total microinclusions of AISI E 52100 steel heats with ASEA ladle refinement, bottom-poured.

Microinclusions found near surface, mid-radius, and center regions of the ingots were also plotted separately. These efforts were intended to divide the microinclusions by size and location within the ingot to ascertain that the new, more restricted microcleanliness levels were met in all parts of the ingot, both by frequency and in maximum size.

The microinclusions in size 1 SEP were practically distributed homogeneously, either lengthwise or in cross sections of the ingot. Microinclusions of size 2 SEP started to show a difference within a fraction of 0.25 of K1 indice. The difference of inclusion counts found between the surface and center regions of the ingots was caused by the few inclusions of sizes 3 SEP, although the total microinclusion count was greatly reduced. The trend of distribution remained the same as in the past ten years since the first heat was vacuum-degassed and the first heat was bottom-poured.

It should not be forgotten that both the frequency and the severity of microinclusions had

been greatly improved. The scale used in the past for microinclusion rating was easily ten times greater than the scale used today for vacuum-degassed heats and twenty times for nonvacuum-degassed heats. Heats produced by ASEA ladle refining equipment today, if properly deoxidized and adequately stirred to allow total flotation of oxides, show no quality difference whether vacuum-degassed or not.

The area with the highest probability of nonmetallic inclusions in a heat of steel produced with the most up-to-date technology remains to be the center axis region of the lower half of an ingot, and the mid-radius regions in the upper half of an ingot.

When a sample of 1000 mm², representative of the area from the surface to the center of the billet, is taken from the central position *C* of an ingot from the second bottom-poured plate, the microinclusions found under a microscope from this specimen should constitute a good representation of the overall quality level of the entire heat.

Oxygen Content

Top-Poured Ingot from Vacuum-Degassed and Nonvacuum-Degassed Heats

Oxygen contents were determined from specimens of each ingot (Figs. 16 and 17) to present the results of oxygen distribution within each ingot in the position described earlier in the paper. Such positions cover all parts of an ingot lengthwise from top to bottom, and in cross section from the surface to the center of an ingot. Figure 16 contains results of a vacuum-degassed heat, and Fig. 17 a nonvacuum-degassed heat. Samples were taken from ingots Nos. 1, 7, 13, and 14 and the results plotted separately. The two heats selected were representative heats.

As a general rule, lower values were obtained in the top portion of the ingot, and higher values were obtained from locations in the center, but near the bottom of the ingot. The difference of average value between the first and the last ingot was 1.6 ppm for a vacuum-degassed heat and 2.4 ppm for a nonvacuum-degassed heat.

The dispersion of values among various positions in an ingot and between the surface and center was higher when the total oxygen content of the steel was higher. Substantial consistency was found in the relationship of macroinclusion presence and oxygen content along the height of the top-poured ingot from positions *T* through *B*, even though the center portion of the ingot was not evaluated by the step-down test. On the other hand, the correlation of oxygen content with the distribution of microinclusions was not very close. This series of studies could not establish this correlation. The difference of oxygen contents between vacuum-degassed and nonvacuum-degassed heats produced in 1972 are shown in Fig. 18.

Bottom-Poured and Top-Poured Ingots from Vacuum-Degassed Heats

On the average, top-poured ingots contain more oxygen than bottom-poured ingots by 4.5 ppm. It is theorized that top-pouring causes more reoxidation because the liquid steel stream is longer and more turbulent. Figure 19 shows the results.

The difference is even more pronounced for normal alloy steel heats with a higher oxygen content than AISI E 52100 ball bearing steel heats. The bottom-pouring method was then extended to all steel production. It should be noted that bottom-pouring also produces a superior surface as well as better internal quality.

Bottom-Poured Ingots from ASEA Ladle Refined Heats

Twenty heats, produced by ASEA ladle refining installation and then bottom-poured, were

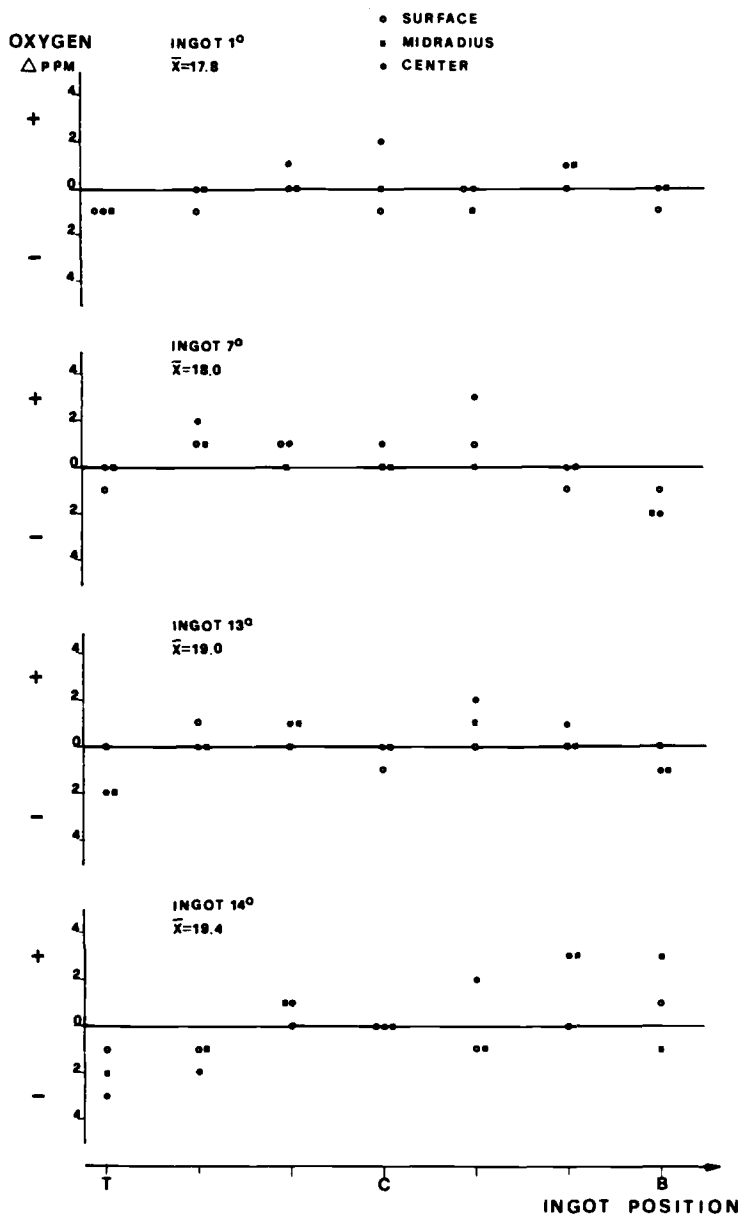


FIG. 16—Oxygen distribution of AISI E 52100 steel heats, vacuum degassed, top-poured.

used for the study. Results of oxygen determinations are shown in Fig. 20. Similar to the results of microcleanliness, there was no noticeable difference in oxygen content between ingots from the first and the second plate. Figure 21 shows the results of the oxygen determination from various positions of the ingot for three representative heats with oxygen contents of 5, 8, and 10 ppm, respectively. There was no noticeable difference in the dispersion of values between heats with oxygen contents from 5 to 10 ppm.

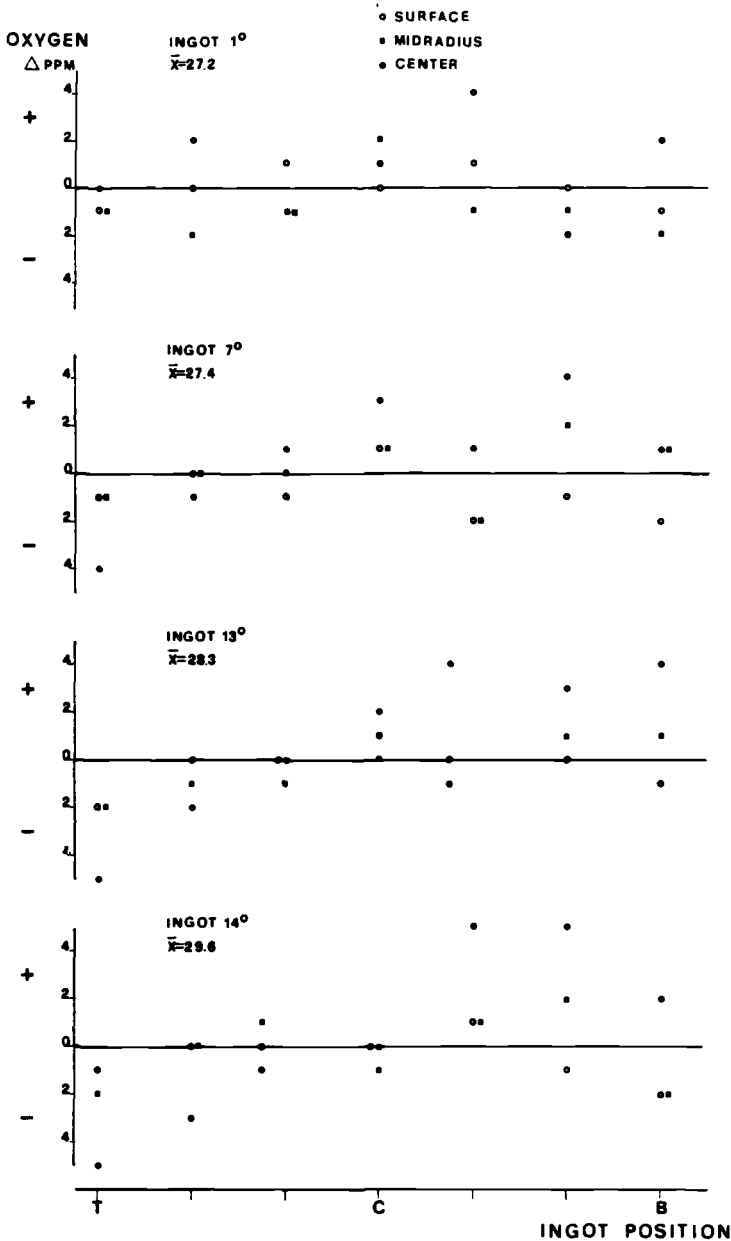


FIG. 17—Oxygen distribution of AISI E 52100 steel heat, not vacuum degassed, top-poured.

The average value of oxygen content in various ingot locations followed the count of oxidic inclusion of sizes ≥ 2 SEP (K2). The correlation was good, but not perfect, because the oxygen content represents the total content of oxidic inclusions in the steel, whereas the optical rating of microcleanliness is possible only when inclusions reach a certain minimum size.

Samples taken from wire products produced from these 20 heats were again tested to determine their oxygen content. The results were very comparable to results obtained from the billet samples.

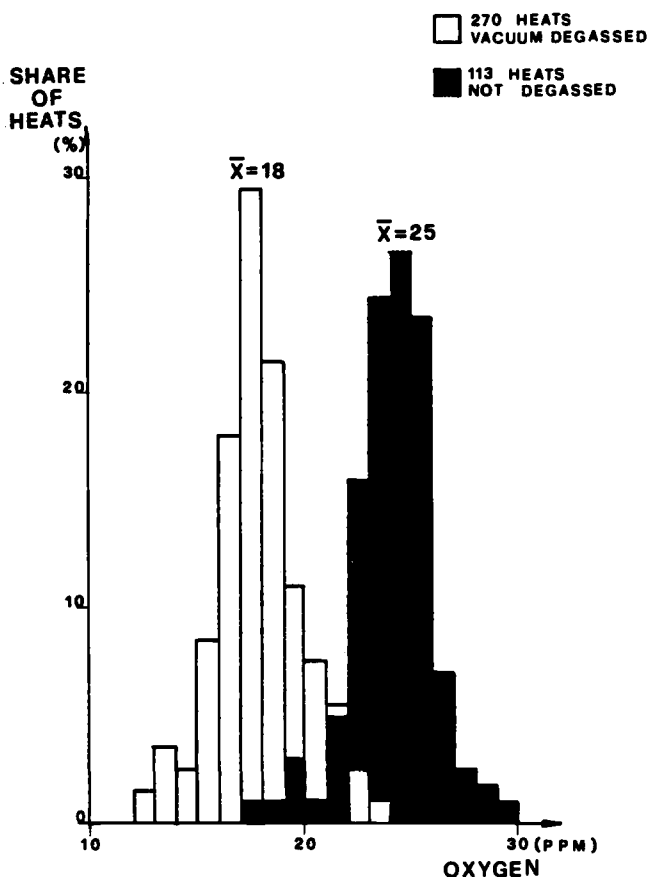


FIG. 18—Oxygen content of top-poured AISI E 52100 steel heats in 1972, vacuum degassed versus nonvacuum degassed.

Simplified Method for Heat Control

Heats of AISI E 52100 ball bearing steel, produced from the ASEA ladle refining installation and bottom-poured with protective gas shrouding, were found to be very homogeneous. It is felt that a microcleanliness specimen of 1000 mm² taken from the center position of an ingot can represent the general quality of the entire heat very well. This finding confirms the validity of our internal quality control procedure, which is used only in the absence of specific requirements from customers, to perform the following examinations from specimens taken from the center position of the ingot:

1. One specimen for step-down test for macroinclusions.
2. One specimen of 1000 mm² covering the surface to the center of the cross section of the billet for the rating of microcleanliness.
3. One specimen each from the surface, mid-radius, and center of the ingot for oxygen determination.

One ingot from each plate shall be examined.

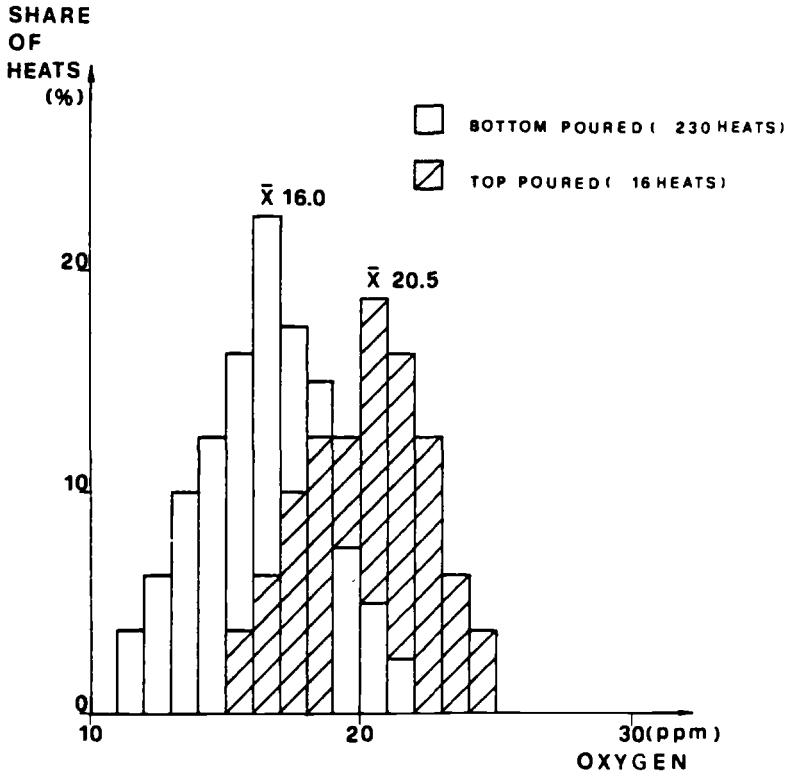


FIG. 19—Oxygen content of vacuum degassed AISI E 52100 steel heats in 1976, top-poured heats versus bottom-poured heats.

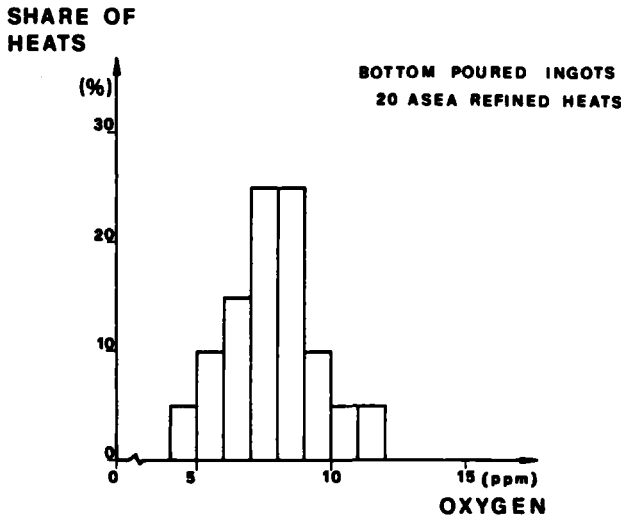


FIG. 20—Oxygen content of AISI E 52100 steel heat with ASEA ladle refinement, bottom-poured.

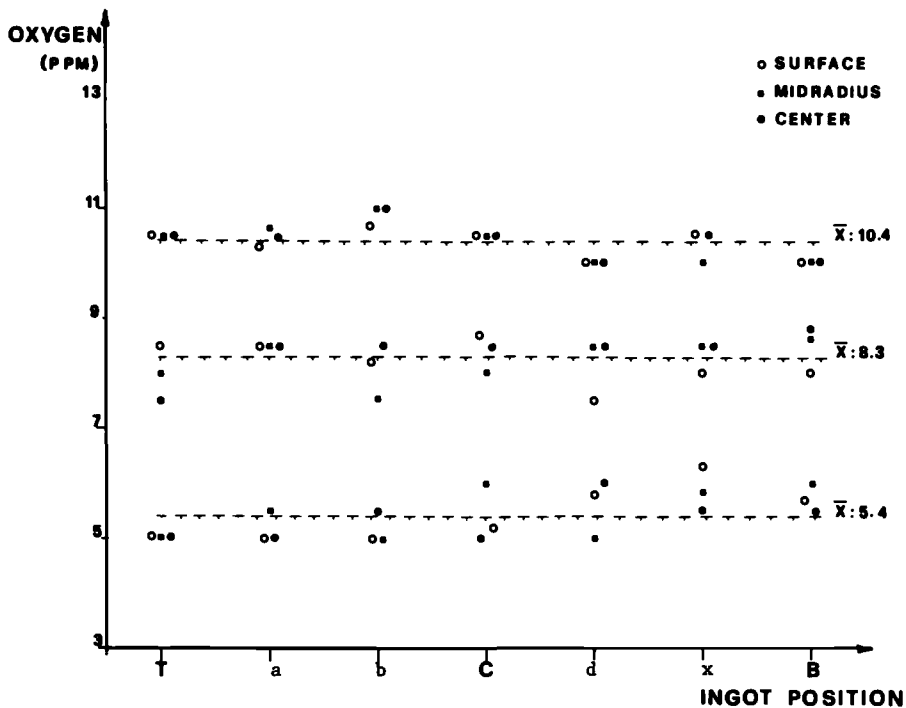


FIG. 21—Oxygen distribution in ingots of AISI E 52100 steel heats with ASEA ladle refinement, bottom-poured.

Conclusions

Today, the continued progress in steelmaking technology has satisfied the users' requirements in regards to macroinclusions, microinclusions, and oxygen content. In the past, such tight requirements could only be met by processes of remelting, such as electro slag remelting (ESR) and vacuum arc remelting (VAR). This paper presents processes which evolved over a period of time. We have almost achieved our goals of the total elimination of macroinclusion, drastic reduction of oxidic microinclusions by a factor as high as 20, and the lowering of the oxygen content to near or below 10 ppm.

There is no significant difference, either qualitatively or quantitatively, in microcleanliness when optically rated by all standard methods, among heats with between 6 and 10 ppm oxygen. The difference in oxygen content is reflected only in the amount of nonratable, extremely small oxidic inclusions, either not visible under an optical microscope at $\times 100$ magnification, or those that cannot be preserved by metallographic polishing. The bearing industry must establish that these very small inclusions are indeed detrimental to the bearing fatigue life. Further reduction of oxygen content in bearing steel is very costly. We may have reached the point of diminishing returns as far as bearing steel quality is concerned.

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New Methods to Evaluate Quality of Better Bearing Steels

The Development of an ASTM Standard Analytical Method for the Determination of Oxygen in Steel

REFERENCE: Diamondstone, B. I. and Flinchbaugh, D. A., "The Development of an ASTM Standard Analytical Method for the Determination of Oxygen in Steel," *Effect of Steel Manufacturing Processes on the Quality of Bearing Steels*, ASTM STP 987, J. J. C. Hoo, Ed., American Society for Testing and Materials, Philadelphia, 1988, pp. 191–197.

ABSTRACT: Efforts towards the development of a standard procedure for the determination of 5 to 50 ppm of oxygen in steel are described. This work is being carried out by ASTM committee E03 on Chemical Analysis of Metals, in cooperation with committee A-1 on Steel, Stainless Steel and Related Alloys. Data from three interlaboratory studies that have been carried out over the last six years are compared and improvement in the overall uncertainty of the proposed standard procedure is demonstrated. The various sources contributing to the uncertainty of oxygen measurements, including sample inhomogeneity and sample preparation procedures, are discussed, and suggestions for minimizing uncertainty are given. A short history of progress in Subcommittee E03.01 and suggestions for possible future courses of action are given.

KEY WORDS: inert gas fusion, vacuum fusion, oxygen, E 1019, steel, gases in metals, homogeneity

History

ASTM has been aware of the need for analytical methods for the determination of oxygen in steels for many years. A review of old committee reports of E03 indicates that plans for the formation of Division I were initiated in late 1962 and the first organizational meeting was held in March 1963. Division I was a subcommittee of E03 that was given the responsibility for the analysis of gases in metals. A symposium on Recent Advances in the Determination of Gases in Metals was held in 1966 in Atlantic City, New Jersey. By 1969, the members of Division I were discussing standard formats for two important analytical procedures: (1) the determination of oxygen in metals by the vacuum fusion method; and (2) the determination of oxygen in metals by the inert-gas fusion method. The first interlaboratory study to evaluate these two methods was also conducted in 1969.

Around this time, a joint project was being carried out between U.S. Steel and the National Bureau of Standards (NBS). This project involved the construction of an apparatus that combined the techniques of inert gas fusion and gas chromatography for the determination of oxygen and nitrogen in steels [1]. The goal of this project was to develop a rapid instrumental technique for the simultaneous determination of trace amounts (<50 ppm) of oxygen and nitrogen in all grades of steel. This apparatus was used to carry out analyses on a series of

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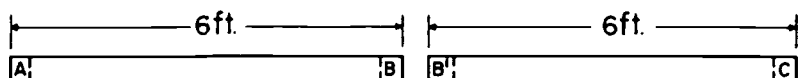
²Supervisor, Analytical Chemistry Research Department, Bethlehem Steel Corporation, Bethlehem, PA 18016.

maraging steel samples which were in the process of being certified by existing vacuum fusion techniques. A direct result of this program was the issuance of Standard Reference Material 1094 (Oxygen in Maraging Steel), along with several additional Standard Reference Materials (SRMs) certified for higher oxygen contents [2]. To this day, SRM 1094 remains the SRM with the lowest certified concentration of oxygen in steel (4.5 ppm).

The E03 committee report from 1970 indicates that Division I became inactive soon after this time and by 1971 had taken two important steps: (1) it had changed its name to Subcommittee E03.03, and (2) it had been disbanded because of insufficient participation. Since that time, the work on the determination of gases in ferrous materials has been the responsibility of Subcommittee E03.01. During the intervening years, this subcommittee has published standard methods for the determination of carbon, sulfur, and nitrogen, for example, ASTM Methods for Determination of Carbon, Sulfur, Nitrogen, Oxygen, and Hydrogen in Steel and in Iron, Nickel, and Cobalt Alloys (E 1019). This paper, and the following one by Green, Diamondstone, and Hoo which describes the current status of round robin testing of bearing steels, detail the most successful attempt to date to develop a standard procedure for the determination of oxygen in steel.

Present Needs

The need to determine accurately oxygen in metals can be traced to several critical areas in the production and use of steel. Both the producers and users of high-quality steel have considerable interest in being able to make accurate and precise measurements of low levels of oxygen in steel. The producers of steel monitor oxygen content for specific applications in processes such as continuous casting. It is known that soundness and homogeneity vary inversely with oxygen content at the time of solidification. Manufacturers of various steel products have to be concerned with the presence of oxide inclusions and blowholes, which can lead to premature failure of critical components.



MATERIAL	RODS ANALYZED	NUMBER OF DETERMINATIONS	MEAN ppm	SD ppm	R _{SD} %
NBS - SRM 1090 SAMPLES - B & B'	17	34	494.7	8.7	1.8
NBS - SRM 1090 SAMPLES - A, B, B', & C	17	68	494.9	15.1	3.1

FIG. 1—Homogeneity study on NBS-SRM 1090, ingot iron [5].

It is probably fair to say that, when properly maintained and calibrated, existing instrumentation for the analysis of oxygen in metals is capable of producing values which are sufficiently precise for the needs of both the producers and users of steel. Over the years, a considerable amount of research has been done using variations of the vacuum fusion and inert-gas fusion techniques [3,4]. The critical factor involved here is the overall question of accuracy. The instrumentation existing today is totally dependent on material standards, and the uncertainty of the resulting measurements must take into consideration the errors associated with the materials, the sample preparation, and the instrumentation itself.

Over the last 20 years, a considerable amount of work has gone into verifying that existing material standards, of adequate homogeneity, are satisfactory for evaluating proposed standard procedures. In the example illustrated in Fig. 1, duplicates of adjacent samples were taken for analysis from the center, as well as from the ends of a number of steel rods intended for use as SRMs. The total oxygen content was determined using the vacuum fusion technique with infrared detection. The somewhat lower imprecision of the values obtained for the adjacent samples from the center of the rod could lead an analyst to make an inaccurate estimate of the homogeneity of the entire rod. When measurements are made at both ends and the middle of the rod, we see that the original estimate of the mean value was good, but that the overall uncertainty of that value was underestimated. In addition, consecutive samples are often taken from individual rods over a period of 10 to 15 days to check the reproducibility of these measurements with time. The data shown in Fig. 2, along with the information compiled by users over the past 20 years, indicate that satisfactory primary reference materials can be made and certified. With the availability of precise, sensitive instrumentation and an adequate capability to produce certified reference materials for calibration, current work is aimed at obtaining consensus on sample preparation and operating procedures and then at conducting sufficient interlaboratory test programs to demonstrate the general utility of the method. This is an important part of the mission of Subcommittee E03.01.

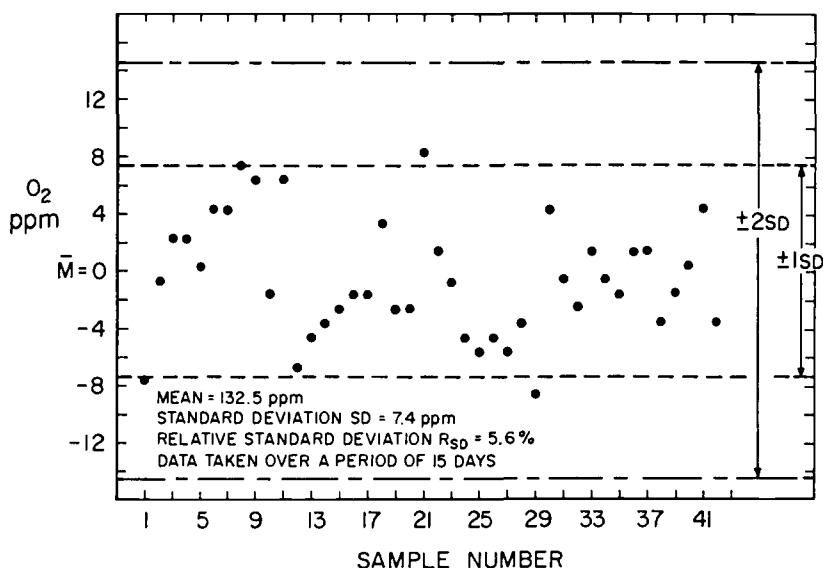


FIG. 2—Reproducibility of measurements on a single rod of NBS-SRM 1091 [5].

Role of E03

ASTM Committee E03 exists for the purpose of supplying test methodologies which allow producers and consumers to establish whether compositional specifications have been met. The standards that are developed by this committee should represent a common viewpoint of all interested parties, such as producers, users, consumers, and general interest groups.

One question facing committee E03 at this time is whether a standard method for the determination of oxygen in steel can be developed which meets the previously described needs of the steel community. Anything less than a procedure which meets these needs is unacceptable, as was shown during the last ballot of E03, held in 1985, where certain members of the committee felt that the precision of oxygen values obtained by the cooperating laboratories was not good enough to meet the needs of the industry. A successful method must be both precise and accurate to be used for monitoring commerce. In Committee E03, Recommended Practices for Conducting Interlaboratory Studies of Methods for Chemical Analysis of Metals (E 173) is used to evaluate the information provided by cooperating laboratories. This practice is precision driven and relies on primary reference materials to provide reliable information regarding the repeatability, reproducibility, and accuracy of a specific chemical method for the analysis of metals.

One important role that committees such as E03 have is to convince participants in interlaboratory studies of the importance of assuring that potential standard procedures are rigorously tested. Laboratories willing to cooperate in interlaboratory tests are the key to developing useful analytical procedures such as the one discussed here for the determination of oxygen in steels. The requisite precision can only be obtained if the cooperators are willing to take the steps necessary to ensure that all aspects of the analysis are being done in an identical fashion by all laboratories. Specifically, they must all carry out the procedures exactly as specified in the draft method. In order for cooperating laboratories to obtain the high precision required to meet the commercial needs for oxygen determinations, they must clean all surfaces of the specimen to avoid artificially high and inconsistent test results. This is by far the most labor-intensive part of the entire measurement scheme. Most cooperating laboratories must analyze samples as quickly as possible because of the high costs associated with analytical delays. As a result, many laboratories have good intentions when they volunteer to cooperate in these studies, but actually slow down the process of developing new procedures by failing to follow the directions provided with the proposed standards. For example, many laboratories find the

TABLE 1—*Effect of sample treatment on oxygen content.*

Sample No.	Sample Treatment ^a	Oxygen, ppm		
		Mean	Standard Deviation	Relative Standard Deviation, %
431-6	A	171.3	0.96	0.6
	B	136.2	1.9	1.4
431-9	A	170.5	5.26	3.1
	B	131.2	1.3	1.0
431-12	A	142.0	9.7	6.8
	B	134.8	1.9	1.4

^aSample treatment: A = Cut with water coated slitting wheel and grind surfaces on a Buehler paper disc grinder with a No. 120 grit silicon carbide paper. B = Hand hacksaw used to cut specimen and surface cleaned by filing with a clean sharp file.

process of filing analytical samples objectionable and, in the interest of saving time, resort to belt or wheel grinders to prepare the samples. The potential sources of error associated with these time-saving steps are numerous. Solvents, heat, and bonding agents are some of the factors that may contribute to errors in the determined oxygen contents relative to the "true" values. When laboratories deviate from the written procedures, they make it extremely difficult for the statisticians to isolate and identify potential sources of imprecision and uncertainty error.

In the example illustrated in Table 1, a test has been carried out to illustrate the effect of different procedures for preparing the surfaces of the test samples. When a method was used in which the sample was subjected to drastic treatment with an abrasion wheel, higher results were produced and, in most cases, the uncertainty for the hand-filed samples was significantly lower. It appears that automated surface preparation techniques often contaminate the surface with oxygen. This work was carried out in 1966 and illustrates that we must not take steps backward in the development of standard procedures.

Progress

The work on the oxygen standard which is in progress today was initiated in 1979 by D. A. Flinchbaugh of Bethlehem Steel. At that time, there were five separate standards for the analysis of steels, irons, and alloys which would theoretically cover the types of metals that would require oxygen procedures in the future. Those standards were designated E 350 through E 354.³ In order to avoid repetition, it was decided that the new standard, designated E 1019, would cover the analytical procedures for the analysis of carbon, sulfur, hydrogen, nitrogen, and oxygen that would be applicable to all of the metals described in Methods E 350 through E 354. A preliminary interlaboratory study, involving eight laboratories, was run in early 1980. Since that time, 12 revisions have been made to the original draft standard to arrive at the procedure which has been used during the latest study.

In 1984, a second interlaboratory study was conducted, and the results were presented to Committee E03 with a recommendation that the procedure be balloted for inclusion in E 1019. Materials ranging from a nickel-base alloy (about 20 ppm oxygen) to a cast iron sample (about 300 ppm oxygen) were used as unknown samples. The precision of the results obtained in this study was judged to be insufficient to meet the needs of today's producers and users of steel. As a result of this decision, it was decided that the procedures would be published in the ASTM Gray Pages, for information purposes, as a proposed Test Method for Determination of Oxygen in Steel and in Iron, Nickel, and Cobalt Alloys (P 187). This would give interested parties an opportunity to comment on the proposed method and provide suggestions on how a more useful method could be developed. A procedure can only reside in the Gray Pages for two years after which it must be elevated to a Standard or removed from the book of standards.

Immediately, plans were formulated for producing a new standard that would be acceptable as an E03 Standard Method in E 1019. It was decided to reduce the scope of the modified procedure and to focus on the critical aspects of sample preparation and material

³Method for Chemical Analysis of Carbon Steel, Low-Alloy Steel, Silicon Electrical Steel, Ingot Iron, and Wrought Iron (E 350), Method for Chemical Analysis of Cast Iron—All Types (E 351), Method for Chemical Analysis of Tool Steels and Other Similar Medium- and High-Alloy Steels (E 352), Method for Chemical Analysis of Stainless, Heat-Resisting, Maraging, and Other Similar Chromium-Nickel-Iron Alloys (E 353), and Method for Chemical Analysis of High-Temperature, Electrical, Magnetic, and Other Similar Iron, Nickel and Cobalt Alloys (E 354).

homogeneity. At this time representatives of ASTM Committee A01 approached E03 and indicated that they had heard about the ongoing work in E03 and were interested in participating in the next round of testing. It was then agreed that E03 would expand the next round robin to include international users and producers of steel. Committee A01 provided the additional cooperating labs and made arrangements to provide the material used for this study. The material, obtained from Torrington Steel, was believed to be sufficiently homogeneous for the purpose of evaluating the ability to determine oxygen at the 3 to 50 ppm level. Considerable emphasis was placed on preparing the samples for analysis. A procedure for removing surface oxides was described and each laboratory was asked to refrain from deviating from this recommended procedure.

A summary of a limited number of results obtained during the three different interlaboratory studies is shown in Table 2. Data for three materials at each of three different concentration ranges are shown. For the sake of discussion, comparisons are drawn between relative concentrations only, since the matrices are not identical. However, for similar concentrations it is seen that progress in reducing the within-laboratory and the between-laboratory variations has been made over the last five years. In addition, since the samples used for the first study were all SRMs, a considerable amount of information was available pertaining to the homogeneity of the samples. That meant that the main source of uncertainty was a result of the measurement process rather than of a combination of measurement process and sample inhomogeneity. For the next two studies, much less was known about the samples, and therefore, the total uncertainty must be a combination of the measurement uncertainty and sample inhomogeneity. If these latest materials were somewhat less homogeneous than the SRMs, then the improvement to the precision of the method may be even greater than is apparent in Table 2.

TABLE 2—Selected samples from Tests I, II, and III.^a

Sample	$\bar{x}^b$	S.E. ^c	s_w^d	s_b^e	R_1^f	R_2^g	%CV _{R1} ^h	%CV _{R2} ⁱ
I NBS 1096	11.9	1.2	1.8	3.1	2.9	9.2	24.5	76.9
II Inconel 625	17.4	1.2	1.5	3.4	2.4	9.8	13.8	56.3
III 3310	13.8	0.4	1.4	1.5	2.3	4.8	16.3	34.4
I NBS 1092	32.9	1.9	3.1	4.8	5.0	14.3	15.2	43.6
II L605	30.3	2.1	1.7	6.3	2.7	17.8	8.9	59.0
III 1026	35.7	0.8	1.4	3.2	2.3	9.2	6.2	25.7
I NBS 1093	44.4	5.8	6.1	15.7	9.9	45.1	22.2	101.5
II 15%Cr/25%Mo	41.3	2.2	3.5	6.3	5.7	18.5	13.8	44.8
III 440c	43.7	1.0	2.0	3.7	3.2	10.8	7.3	24.8

^aI = Preliminary interlaboratory study carried out in 1979 (81 laboratories and 12 materials). II = Interlaboratory study carried out in 1984 (20 laboratories and 8 materials). III = Latest interlaboratory study carried out in 1986 (16 laboratories and 12 materials).

^b $\bar{x}$ = overall mean.

^cS.E. = standard error.

^d s_w = standard deviation/within lab.

^e s_b = standard deviation/between labs.

^f R_1 = repeatability.

^g R_2 = reproducibility.

^h%CV_{R1} = percent coefficient of variation for repeatability.

ⁱ%CV_{R2} = percent coefficient of variation for reproducibility.

Conclusions

If this procedure is accepted by the E03 Committee, it will be very useful to the steel community. However, the concentration range in the scope has been severely reduced from the 5 to 500 ppm in the original draft to a more realistic 5 to 50 ppm. Therefore, the next logical step would be to find homogeneous standards for ferrous materials of different compositions and higher concentrations of oxygen and to begin to carry out studies that would allow E03 to gradually increase the scope of this latest practice. Expanding the scope back to its original range would most likely make it more useful to the user community.

In conclusion, considering that E03's role is to develop, test, and standardize analytical procedures for the chemical analysis of methods, it is obvious that Subcommittee E03.01 should do everything possible to see that this work on oxygen in ferrous materials is carried to completion. In this paper we have described the procedures used and special technical and operating problems associated with the development of a planned standard method to determine oxygen in steel. The work has been enhanced by active participation and support from members of A01. It is imperative that members of the metallurgical community communicate their analytical needs to the analytical chemists. Because developing a successful standard procedure requires a considerable amount of time, chemists must have sufficient lead time to generate methods that are suitable for monitoring the compositional characteristics of specific steel products.

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Current Status of Round-Robin Testing of Oxygen Content in Bearing Steels

REFERENCE: Green, W. B., Jr., Diamondstone, B. I., and Hoo, J. J. C., "Current Status of Round-Robin Testing of Oxygen Content in Bearing Steels," *Effect of Steel Manufacturing Processes on the Quality of Bearing Steels, ASTM STP 987*, J. J. C. Hoo, Ed., American Society for Testing and Materials, Philadelphia, 1988, pp. 198–210.

ABSTRACT: The results of round-robin testing for oxygen content are presented. These results represent 3 standards and 12 unknowns, analyzed by 16 international laboratories and institutions. The procedures for analysis methods were detailed by ASTM E03, Subcommittee E03.01, Proposed Method for the Determination of Oxygen in Steel and in Iron, Nickel and Cobalt Alloys, Draft No. 12, 29 April 1986. Statistical analysis of data is presented. The results show that reliable and reproducible oxygen analyses can be obtained from bearing steels by using proper procedures and commercially available analysis equipment.

KEY WORDS: oxygen, bearing steels, inclusions, chemical analysis

Background

Analytical methods to measure the oxygen content of steels have come under close scrutiny in recent years and, in fact, an agreed-upon ASTM method does not exist. The responsibility for the development of a method lies with the E03.01 Subcommittee on Ferrous Metals under the jurisdiction of ASTM Committee E03 on Chemical Analysis of Metals. With the recent changes in steelmaking technology aimed at improved quality, new methods for assessing quality have been sought. Oxygen content, which is related to steel oxide inclusion content, is one of the items which has been proposed as a measure of steel quality. This, of course, is of great interest to the bearing industry where "clean" steel technology is of utmost importance to product performance. The steelmakers have begun to measure the oxygen content of their product on a routine basis and are using these values as a measure of steel quality. Some have proposed replacing the JK inclusion rating systems now being used with oxygen content values. The latest revisions of the bearing steel specifications under the jurisdiction of the ASTM A01.28, Subcommittee on Bearing Steels have acknowledged this trend by including oxygen determinations by request in the supplementary sections of the specifications. The requirement for inclusion ratings has not been altered, however. Due to the increased use of oxygen determinations in bearing steels, especially at low levels of oxygen, the accuracy of the equipment and analysis methods have been questioned. As mentioned, ASTM E03 has been working on a standard analysis method for several years, but has been unable to resolve problem areas with the method. This International Symposium on Effect of Steel Manufacturing Processes on the Quality of Bearing Steels provided an excellent opportunity for a

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collaborative effort between the E03 and A01 committees of ASTM to work on a common problem. Because of the experience of the bearing steelmakers and users in measuring low-oxygen contents, A01.28 was able to provide the additional laboratories necessary for a comprehensive round-robin test as well as samples for testing, most of which are bearing steels. The National Bureau of Standards and E03.01 provided additional samples, coordinated the analyses, provided the writeup on the methods to be used, and performed the statistical analysis of the data generated.

Procedures

The following laboratories volunteered to participate in the oxygen determination round-robin:

Timken Steel
SKF Steel
ARMCO Steel
Accaierie di Bolzano
Altech
Stockbridge Engineering Steel
Leco
Bethlehem Steel
LTV Steel
Mercury Marine
Sanyo Special Steel
Thyssen Steel
Ugine F&S-Ascometal
National Bureau of Standards
Torrington Company
Lukens Steel

Each of these laboratories was provided with samples, detailed sampling procedures, and the Proposed Method for the Determination of Oxygen in Steel and in Iron, Nickel and Cobalt Alloys (Appendix II). Three standards, representing oxygen levels of 4.5, 21, and 61 ppm and 12 unknowns were sent to each laboratory. The 12 unknowns used and their materials were as follows:

1. 1026 steel
2. 8620
3. 8622
4. 1570
5. 440C
6. M-50
7. 1070
8. 52100
9. 52100
10. 3310
11. 4330
12. Iron

Each of the laboratories reported their results on the form provided as requested. Fourteen laboratories used Leco TC-136 units, one used as Leco TC-30, and one used an Horiba EMGA 2200. The analysis method used by these machines is the same as is the size of the sample analyzed.

Results and Statistical Analysis

The results are presented in tabular form in Table 1. Statistical analysis of the results was performed by National Bureau of Standards personnel. The symbols and formulae used for these analyses are shown in Table 2. These are consistent throughout. Table 3 presents the statistical analysis of all the data in Table 1. It is arranged by increasing mean oxygen content (X). The unknown sample numbers are listed for information and identification. The top portion of the table shows the results using all of the individual samples ($m = 3$) while the lower portion uses the laboratory mean value ($m = 1$). Outlier values for individual unknowns from Laboratories 3 and 9 were not included in the analysis due to their wide variance from the mean. Unknown No. 2 showed great inconsistency, as can be seen in Table 3. It was eliminated from any further consideration in the statistics relating to the accuracy and repeatability of the method.

Figure 1 shows a plot of the percent coefficient of variation (CV) versus mean for both r_1 and r_2 (see Table 2). This shows that in both cases the percent CV asymptotically decreases with increasing oxygen content, the variation being considerably higher in the r_2 between laboratory analysis. Analysis of the r_1 and r_2 data from Table 3 does not show a strong trend, although there is some evidence that the repeatability and reproducibility are somewhat better at the low oxygen levels and level off as the oxygen content increases.

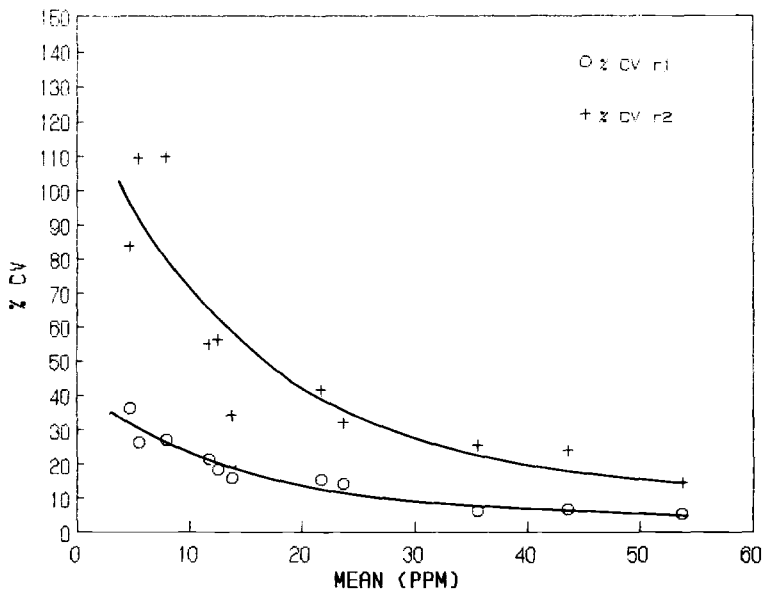


FIG. 1.—Plot of percent CV for both r_1 and r_2 .

TABLE 1—Oxygen by inert gas fusion, ppm.

Laboratory No. Sample No.	1	2	3	4	5	6	7	8	9	10	11	12	13	14	15	16	Mean	1s
Blank (mean)	0.2	-0.3	0.0	0.1	0.0	0.4	0.0	1.3	0.4	0.7	-0.5	0.0	1.0	2.0	0.2	1.0	0.4	0.8
Blank (CAL-BIO)	0.0	-0.3	1.5	0.2	-0.4	0.7	-0.2	0.8	0.5	0.2	-0.5	-0.3	-0.3	3.3	0.4	1.0	0.4	
Blank Values Determined from Std A values below	-0.3 -0.4 0.8 0.0	-0.4 -0.2 -0.2 -0.2	2.1 2.7 -0.2	0.2 -0.1 0.3 0.4	-0.3 -1.0 0.2 -0.6	0.4 0.7 0.7 0.8	0.0 -0.9 0.7 0.4	3.9 -1.0 -2.3 2.6	0.2 0.3 1.1 0.2	0.2 1.0 0.5 -1.1	0.2 -0.7 -1.0 -1.1	0.5 0.5 -1.0 -1.0	-1.5 -0.5 -1.5 2.5	3.5 2.5 3.5 3.5	-0.5 0.9 1.3 0.0	0.5 1.5 0.5 1.5		
Std A (mean) 1s	4.5 0.5	4.3 0.1	6.0 1.5	4.7 0.2	4.1 0.5	5.2 0.2	4.3 0.6	5.3 2.9	5.0 0.4	4.7 0.9	4.0 0.4	4.5 0.6	4.3 1.9	7.8 0.5	4.9 0.8	5.5 0.6	4.9	1.0
Std B (mean) 1s	24.0 0.6	22.0 1.0	29.9 3.2	22.5 0.5	29.6 1.6	21.0 2.0	21.5 1.0	23.8 1.1	21.5 1.3	21.9 2.1	20.0 0.6	21.3 2.0	20.8 2.0	24.5 0.6	25.7 1.7	19.8 1.0	23.1	3.1
Std C (mean) 1s	61.0 0.8	59.1 0.7	71.8 6.6	60.4 0.6	69.9 0.7	61.0 1.3	61.5 1.2	77.4 11.5	62.0 1.8	66.7 0.1	57.8 1.5	61.0 0.1	67.3 0.2	59.3 0.1	61.7 0.2	62.8 0.1	63.8	5.5
Unk 1 (mean) 1s	35.8 0.5	34.9 0.1	44.0 2.3	34.4 0.4	37.9 1.3	33.8 0.8	38.3 0.2	37.1 2.7	32.0 0.1	39.2 1.3	35.1 0.5	36.0 1.0	35.7 1.0	32.7 1.0	34.9 1.9	29.7 1.5	35.7	3.4
	1	35.4	34.6	41.7	36.7	34.6	41.0	39.7	32.0	37.8	34.5	37.0	35.0	34.0	34.5	28.0		
	2	35.7	34.6	46.3	34.5	37.6	33.9	37.0	32.0	39.3	35.5	35.0	37.0	32.0	36.9	30.0		
	3	36.3	35.6	44.0	34.7	39.3	33.0	34.3	32.0	40.4	35.3	36.0	35.0	32.0	33.2	31.0		

TABLE I—Continued.

Laboratory No. Sample No.	1	2	3	4	5	6	7	8	9	10	11	12	13	14	15	16	Mean	1s
Unk 2 (mean)	26.0	35.4	60.4	24.9	52.7	37.4	37.0	31.3	31.0	22.2	26.5	50.7	34.7	49.7	37.7	29.0	36.7	11.6
1s	1.2	3.3	4.5	0.8	0.6	3.2	1.2	1.6	4.0	3.0	1.0	9.0	2.3	2.0	5.0	1.0		
1	27.4	38.8	65.5	17.7	52.5	39.0	50.0	49.3	26.0	19.4	26.0	50.0	36.0	50.0	43.4	30.0		
2	25.8	32.2	57.4	24.0	46.8	33.7	27.0	18.3	32.0	21.9	28.2	42.0	36.0	52.0	37.0	29.0		
3	25.1	35.2	58.2	33.0	58.8	39.5	34.0	26.4	34.0	25.3	25.4	60.0	32.0	47.0	32.7	28.0		
Unk 3 (mean)	25.0	20.1	28.2	26.3	20.2	22.7	22.7	20.9	43.3	28.8	25.2	21.3	23.7	25.0	23.8	21.7	24.9	5.7
1s	1.4	2.1	1.7	0.2	1.0	2.2	1.1	3.0	1.0	3.5	0.9	2.0	1.5	3.0	1.0	2.5		
1	24.1	20.4	28.6	26.4	21.8	25.2	22.0	17.0	44.0	32.9	24.6	20.0	22.0	24.0	25.5	19.0		
2	26.6	17.8	29.6	26.1	18.7	21.8	24.0	22.3	44.0	26.8	24.8	24.0	24.0	28.0	23.2	22.0		
3	24.4	22.0	26.3	26.5	20.1	21.1	22.0	23.4	42.0	26.8	26.3	20.0	25.0	23.0	22.6	24.0		
Unk 4 (mean)	13.1	11.5	14.6	14.1	11.7	9.4	11.3	11.1	13.3	17.9	12.6	12.3	7.3	16.7	11.2	12.7	12.6	2.2
1s	0.1	0.3	0.5	0.6	0.9	1.2	0.6	1.0	0.6	4.6	0.5	0.6	1.5	2.1	0.4	1.5		
1	13.1	11.2	14.8	13.8	12.7	8.7	11.0	10.1	13.0	17.4	13.0	12.0	6.0	19.0	10.8	13.0		
2	13.2	11.4	15.0	14.8	10.8	10.8	12.0	11.2	14.0	13.6	12.0	13.0	7.0	15.0	11.6	14.0		
3	13.1	11.8	14.1	13.6	11.7	8.6	11.0	12.1	13.0	22.7	12.7	12.0	9.0	16.0	11.2	11.0		
Unk 5 (mean)	48.2	44.1	57.1	45.2	47.3	42.1	43.0	43.5	21.0	47.3	39.6	47.0	45.0	41.7	44.5	33.3	43.1	7.9
1s	1.6	0.7	1.0	0.8	2.9	1.3	2.6	3.2	1.0	1.5	3.0	1.0	2.0	1.0	0.6	2.1		
1	50.0	44.0	56.0	45.3	46.2	41.9	46.0	45.4	21.0	47.0	39.2	47.0	43.0	43.0	44.9	35.0		
2	47.5	43.4	57.8	44.4	50.6	41.0	42.0	45.2	21.0	48.9	43.1	48.0	47.0	41.0	44.9	31.0		
3	47.1	48.8	57.5	45.9	42.2	43.5	41.0	39.8	21.0	46.0	36.4	46.0	45.0	41.0	43.8	34.0		
Unk 6 (mean)	7.1	5.3	6.8	5.6	5.7	3.7	4.1	1.5	5.3	6.2	9.3	4.0	5.0	10.0	2.9	6.7	5.6	2.3
1s	0.3	0.4	0.3	0.5	0.8	0.6	0.4	1.3	1.1	0.7	0.6	1.0	1.0	0.0	0.3	2.3		
1	7.2	4.8	7.2	5.0	6.0	3.2	4.6	2.3	4.0	7.0	8.8	5.0	4.0	10.0	2.6	4.0		
2	6.8	5.6	6.6	5.6	6.3	3.6	3.8	2.2	6.0	5.6	9.1	4.0	5.0	10.0	3.2	8.0		
3	7.4	5.4	6.7	6.1	4.8	4.4	3.9	0.0	6.0	6.1	9.9	3.0	6.0	10.0	2.8	8.0		

Unk 7 (mean)	11.0	10.7	15.5	10.3	11.2	8.9	13.3	8.6	14.0	14.5	12.9	10.7	9.7	16.0	9.6	11.3	11.8	2.3
1s	0.6	0.9	0.6	0.6	0.6	0.7	2.3	1.3	2.0	3.3	1.2	1.1	0.6	2.0	2.0	2.3		
1	11.7	10.8	14.8	10.8	11.8	8.4	16.0	8.0	14.0	11.5	14.3	12.0	10.0	15.0	8.2	14.0		
2	10.6	11.6	16.0	10.5	11.2	9.7	12.0	10.1	12.0	18.1	12.2	10.0	9.0	15.0	8.6	10.0		
3	11.0	9.8	15.7	9.6	10.5	8.7	12.0	7.6	16.0	13.9	12.3	10.0	10.0	18.0	12.0	10.0		
Unk 8 (mean)	7.6	6.7	13.2	5.9	5.2	5.1	6.8	6.5	8.7	7.8	7.2	6.7	5.7	15.7	6.4	13.0	8.0	3.2
1s	0.5	0.4	1.7	0.1	1.4	0.7	0.5	2.4	1.5	1.2	1.2	0.6	2.1	2.5	1.1	0.1		
1	7.3	6.4	14.1	5.8	6.8	4.9	7.3	5.1	9.0	8.6	5.9	7.0	4.0	13.0	7.6	14.0		
2	8.2	7.2	11.2	6.0	4.5	4.5	6.5	5.1	7.0	6.4	7.7	6.0	5.0	16.0	5.7	13.0		
3	7.4	6.6	14.3	5.9	4.4	5.9	6.5	9.3	10.0	8.4	8.1	7.0	8.0	18.0	5.8	12.0		
Unk 9 (mean)	22.2	20.8	30.0	21.4	20.5	20.0	16.5	21.0	17.0	24.6	25.3	20.7	21.0	25.0	19.8	22.3	21.8	3.4
1s	0.4	1.6	1.1	0.3	0.7	0.9	3.5	0.2	0.2	3.5	5.6	0.1	0.0	0.2	0.5	1.5		
1	21.8	21.6	28.8	21.3	21.3	19.0	14.0	18.9	17.0	26.8	22.7	20.0	21.0	27.0	19.5	21.0		
2	22.6	21.8	30.0	21.7	20.5	20.8	19.0	21.4	19.0	20.6	21.4	21.0	21.0	24.0	19.5	24.0		
3	22.1	19.0	31.1	21.1	19.8	20.3	xx	22.8	15.0	26.5	31.7	21.0	21.0	24.0	20.3	22.0		
Unk 10 (mean)	14.1	14.0	17.2	12.3	15.2	12.4	14.0	11.6	13.0	14.8	15.1	12.3	11.0	16.3	14.6	13.3	13.8	1.6
1s	0.5	0.3	1.3	0.2	0.6	1.2	1.7	0.1	1.7	1.6	1.4	0.6	2.6	0.6	1.7	2.5		
1	13.8	14.2	15.8	12.3	15.4	13.2	26.0	11.8	12.0	13.7	13.9	13.0	8.0	16.0	13.1	16.0		
2	14.7	13.6	17.5	12.6	14.6	13.1	13.0	11.5	12.0	16.7	14.8	12.0	12.0	16.0	14.4	11.0		
3	13.8	14.2	18.4	12.1	15.7	11.0	13.0	11.6	5.0	14.1	16.7	12.0	13.0	17.0	16.4	13.0		
Unk 11 (mean)	4.5	3.8	5.1	4.8	5.7	3.9	6.2	3.3	6.0	4.9	5.8	3.3	4.3	8.3	2.2	5.0	4.8	1.5
1s	0.2	0.4	0.7	0.5	0.4	0.8	1.5	1.9	1.0	1.6	0.9	0.6	1.5	1.0	0.7	1.7		
1	4.5	4.2	4.3	5.1	5.2	3.0	6.6	3.6	6.0	3.2	6.9	3.0	4.0	7.0	15.0	4.0		
2	4.8	3.8	5.4	5.1	5.9	4.6	7.5	5.0	5.0	5.3	5.4	3.0	3.0	9.0	3.0	4.0		
3	4.3	3.4	5.5	4.2	5.9	4.0	4.6	1.3	7.0	6.3	5.2	4.0	6.0	9.0	2.1	7.0		
Unk 12 (mean)	54.6	52.6	74.1	50.8	55.4	49.6	54.0	52.0	51.7	56.8	55.4	59.0	57.0	48.7	55.3	53.7	55.0	6.0
1s	0.9	1.8	1.9	0.9	0.6	1.2	2.6	1.6	1.5	2.8	1.0	2.0	3.5	2.0	0.6	3.2		
1	55.4	51.2	72.2	51.6	56.0	48.6	51.0	50.6	52.0	56.3	55.3	57.0	53.0	47.0	54.9	56.0		
2	53.6	52.0	74.0	51.0	55.3	49.4	55.0	51.6	50.0	59.9	56.4	61.0	59.0	50.0	55.1	55.0		
3	54.7	54.6	76.1	49.8	54.9	50.9	56.0	53.7	53.0	54.3	54.4	59.0	59.0	49.0	56.0	50.0		

TABLE 2—*Symbols and formulae used in analyses.*

Overall mean	$\bar{X}$
Standard deviation	S_w = within lab S_b = between labs
Number of readings	m
Standard error	$S.E. = [1/16(S_b^2 + S_w^2/3)]^{1/2}$
Repeatability	$r_1 = 2.8 S_w/(m)^{1/2}$ $r_2 = 2.8(S_w^2/m + S_b^2)^{1/2}$
% coefficient of variation	$\%CVr_1 = (r_1/\bar{X})100$ $\%CVr_2 = (r_2/\bar{X})100$

Figures 2 and 3 illustrate the effect of doing multiple measurements on the percent CVs for the within- and between-laboratory conditions. The effect is far greater in Fig. 2, the within-laboratory data, than in the between-laboratory condition, Fig. 3. This leads to the observation that multiple measurements may improve precision, but overall accuracy is not helped if the largest source of error is between laboratories, as is the case here.

Data from individual laboratories can be plotted against the mean values from all laboratories. From these, laboratory trends can be seen. For example, if all data lie above the perfect correlation line, a high bias is indicated; if data lie below, a low bias is indicated; and if data cross the line at an angle, a calibration problem is indicated. This exercise is left to the reader.

TABLE 3—*Statistical analyses of the results from Table 1.*

Unknown	$\bar{X}$	S.E.	S_w	S_b	$r_1(m=3)$	$\%CVr_1$	$r_2(m=3)$	$\%CVr_2$	Outliers
11	4.83	0.36	1.10	1.30	1.78	36.8	4.05	83.9	
6	5.58	0.55	0.91	2.12	1.47	26.4	6.12	109.6	
8	8.01	0.78	1.37	3.04	2.21	27.6	8.80	109.8	
7	11.77	0.58	1.60	2.13	2.59	22.0	6.50	55.2	
4	12.55	0.64	1.49	2.40	2.41	19.2	7.14	56.9	
10	13.85	0.43	1.40	1.50	2.26	16.3	4.77	34.4	
9	21.78	0.82	2.11	3.02	3.41	15.7	9.12	41.9	
3	23.70	0.70	2.10	2.44	3.39	14.3	7.63	32.2	(9)43.3
1	35.71	0.81	1.43	3.17	2.31	6.5	9.17	25.7	
# 2	36.64	2.80	6.55	10.53	10.59	28.9	31.33	85.5	
5	43.70	1.03	1.98	3.69	3.20	7.3	10.82	24.8	(3)57.1(9)21.0
12	53.77	0.74	1.94	2.66	3.14	5.8	8.08	15.0	(3)74.1
					$r_1(m=1)$	$\%CVr_1$	$r_2(m=1)$	$\%CVr_2$	
11					3.08	63.8	4.77	98.7	
6					2.55	45.7	6.46	115.8	
8					3.84	47.9	9.34	116.6	
7					4.48	38.1	7.46	63.4	
4					4.17	33.2	7.91	63.0	
10					3.92	28.3	5.75	41.5	
9					5.91	27.1	10.32	47.4	
3					5.88	24.8	9.01	38.0	(9)43.3
1					4.00	11.2	9.74	27.1	
# 2					18.34	50.1	34.72	94.8	
5					5.54	12.7	11.73	26.8	(3)57.1(9)21.0
12					5.43	10.1	9.22	17.1	(3)74.1

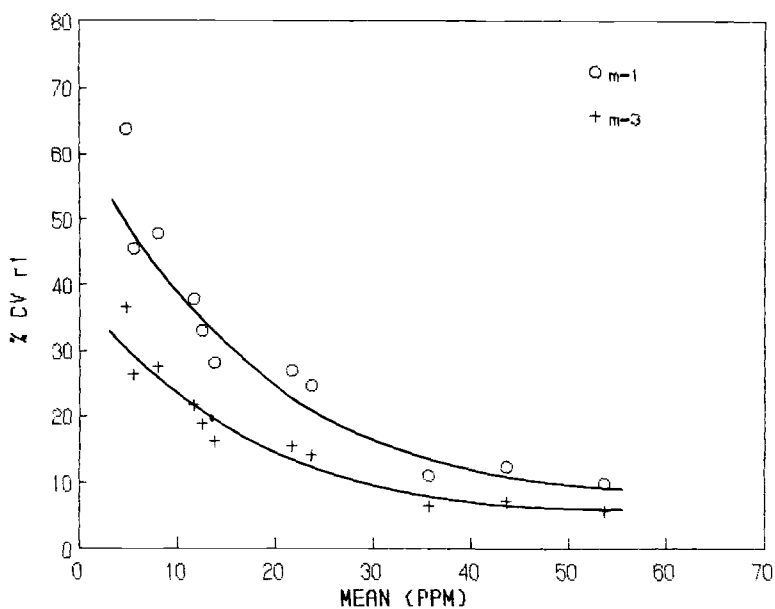


FIG. 2—Percent CV versus mean, within-laboratory data.

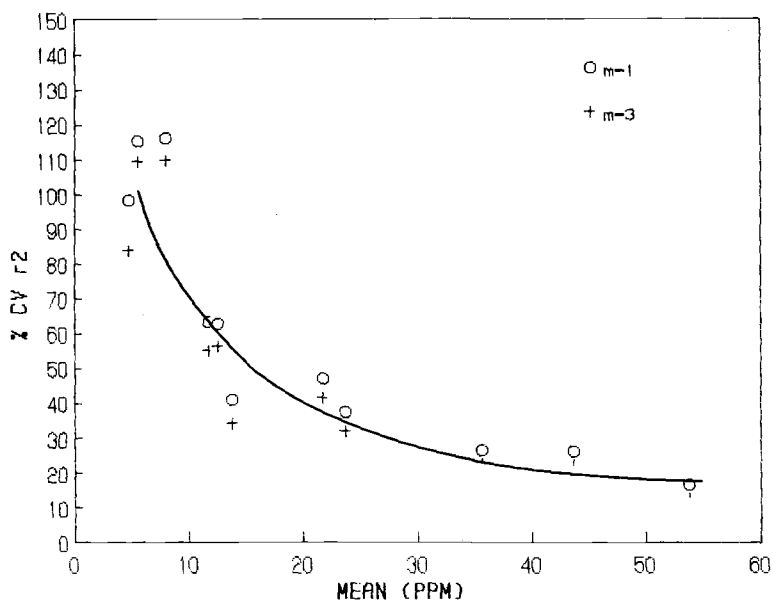


FIG. 3—Percent CV versus mean, between-laboratory data.

Conclusions

1. The reproducibility of data within a given laboratory is considerably better than among laboratories.
2. The percent CV improves substantially at higher oxygen contents both within and among laboratories.
3. Sample preparation, calibration, and operating procedures must be rigidly followed to produce acceptable results.
4. The proposed method specification for oxygen determination and the analysis equipment commercially available appear to give adequate results for bearing industry use.

APPENDIX I

Instructions for Round-Robin Testing for Oxygen Content

Thank you for agreeing to participate in the ASTM E03 Round Robin for the determination of oxygen in steels and in iron, nickel and cobalt alloys. The purpose of this study is to evaluate a proposed method for the determination of oxygen at concentrations from 0.0003 to 0.005%.

I realize that studies such as these require a considerable amount of effort on the part of the cooperating laboratories and therefore I must emphasize the importance of doing everything possible, *one time*, to evaluate the provided technique. This means that the number of variables between labs must be kept to a minimum. The best way to assure that this is done is to follow the procedure as closely as possible and to avoid making any changes to the procedure. Below is a set of important notes which should be carefully read before the analyses are carried out.

1. If there are specific questions concerning the prescribed procedures, please don't hesitate to contact me for clarification before you carry out the analyses.

2. The samples provided in this study include 12 unknowns and 3 standards. The samples vary in size from $\frac{1}{4}$ -in. rods to segments from 3.5-in. disks. On samples 2, 4, and 10, 0.125 in. of material per side should be machined off before any samples are cut. In order to assure sample homogeneity, *please cut your final samples from as close to the center of the original diameter as possible*. In the case of the pie-shaped segments, this will mean at the narrow end of the piece. The 2-in.-diameter disks can be quartered before the samples are taken, and the small rods can simply be filed to a suitable diameter.

NOTE: Any additional measurements made on the larger samples at the surface and at the mid-radius points would be useful in evaluating the homogeneity of the samples. *This information would be optional only and is not required for this study.*

3. One of the most critical factors in this procedure is the sample preparation step. Once the samples have been cut to an appropriate size please file all surfaces as completely as possible. *Avoid using sample preparation procedures other than those described in the method.*

4. Complete all information on the analysis information sheet which is attached to the procedure. The information on this sheet is necessary for a complete analysis of the round robin study.

5. Once the analyses have been completed, please save any sample fragments that you have left over until the round robin has been completed (around the first week in November). This may allow us to resolve any problems associated with particular samples, that arise after the data has been compiled.

6. *Report the original measurements for the unknown samples. I do not want averaged results.*

7. Each laboratory will be assigned an identifying code which will be provided with the procedure. The results will be kept confidential, and the code will be the only reference to individual labs when the data are presented.

8. ASTM is very interested in any suggestions you may have for improving this procedure, and all recommendations will be considered. *However, it is critical that the analyses be carried*

out in accordance with the procedure provided. Then attach your recommendations to the final report.

9. Our goal is to have all samples mailed to cooperating laboratories by 1 June 1986 and to have all results returned to me by 1 Aug. 1986. Hopefully, this will allow me to evaluate all the data in time to present the results in early November at the symposium on the "Effect of Steel Manufacturing Processes on the Quality of Bearing Steels" in Phoenix, Arizona. If this presents serious problems for you, please let me know as early as possible.

Once again, on behalf of ASTM E03, I thank you for your cooperation and hope that your considerable effort will help to produce a useful standard in the near future.

APPENDIX II

Proposed Method for the Determination of Oxygen in Steel and in Iron, Nickel and Cobalt Alloys⁴

1. Scope

1.1 This method covers the determination of oxygen in steel and in iron, nickel and cobalt alloys having chemical compositions within the limits shown in Methods E 1019, Section 1 in concentrations from 0.0003 to 0.005.

1.2 This standard may involve hazardous materials, operations, and equipment. This standard does not purport to address all of the safety problems associated with its use. It is the responsibilities of whoever uses this standard to consult and establish appropriate safety and health practices and determine the applicability of regulatory limitations prior to use. Specific precautionary statements are given in Section 6.

2. Applicable Documents

2.1 ASTM Standards:

E 29 Standard Recommended Practice for Indicating Which Places of Figures Are To Be Considered Significant in Specified Limiting Values

E 50 Practices for Apparatus, Reagents, and Safety Precautions for Chemical Analysis of Metals

E 59 Standard Method of Sampling Steel and Iron for Determination of Chemical Composition

E 173 Standard Practice for Conducting Interlaboratory Studies of Methods for Chemical Analysis of Metals

E 1019 Methods for the Determination of Oxygen, Nitrogen, Carbon, and Sulfur in Steel and in Iron, Nickel and Cobalt Alloys.

3. Significance and Use

3.1 These methods for the chemical analysis of metals and alloys are primarily intended to test such materials for compliance with compositional specifications. It is assumed that all who use these methods will be trained analysts capable of performing common laboratory procedures skillfully and safely. It is expected that work will be performed in a properly equipped laboratory.

⁴This proposed method is under the jurisdiction of ASTM Committee E03 on Chemical Analysis of Metals and is the direct responsibility of Subcommittee E03.01 on Ferrous Metals. This is a working document for committee use only, and is presented here only as information. Reproduction is prohibited without the express approval of the Chairman of Committee E03 or the President of the Society.

4. Summary of Method

4.1 This method is written for use with automated, commercially available analyzers which are based on the inert gas fusion principle and use a variety of gas conditioning and measuring techniques. All use calibration methods traceable to primary Standard Reference Materials.

4.2 The sample, contained in a small, single-use graphite crucible, is fused under a flowing inert gas stream at a temperature sufficient to release oxygen, nitrogen, and hydrogen. The oxygen combines with carbon from the crucible to form carbon monoxide (CO) and is carried by the flowing inert gas stream to either thermal conductivity or infrared detectors. The detector output is compared to that obtained from Standard Reference Materials and is displayed as oxygen content of the steel.

4.3 In a typical instrument based on thermal conductivity detection (Fig. 1)⁵, the sample is fused in a stream of helium which carries the released gases over copper oxide which oxidizes CO to CO₂ and hydrogen to water. The water is then absorbed by magnesium perchlorate and the remaining nitrogen and carbon dioxide are separated chromatographically. The nitrogen is eluted from the chromatographic column within seconds and is either measured or disregarded. The oxygen as (CO₂) then elutes from the column and enters the measuring cell of the thermistor bridge. Bridge output is integrated and the data are processed and displayed directly as percent oxygen.

4.4 In a typical instrument based on infrared detection (Fig. 2), the sample is fused in a stream of nitrogen or argon and passed directly into an infrared cell through which infrared energy is transmitted. The absorption of energy by the CO in the stream causes a reduction in energy received at the detector. The absorption of IR energy by the CO is measured by an infrared detector. The data are processed and displayed directly as percent oxygen.

5. Interferences

5.1 The elements usually present in these alloys do not interfere.

6. Safety Precautions

6.1 For precautions to be observed in the use of certain reagents in this method, refer to Practices E50.

6.2 Use care when handling hot crucibles and operating furnaces to avoid personal injury by either burn or electrical shock.

7. Apparatus

7.1 Instrument—The general features of the instruments used in developing the test data accompanying this method are shown schematically in Figures 1 and 2. These units are commercially available.

7.2 Graphite Crucibles—The crucibles must be made of high purity graphite and should be those recommended by the manufacturer or their equivalent.

8. Reagents

8.1 Acetone—Residue after evaporation must be <0.0005%.

8.2 Ascarite II (Sodium hydroxide on clay)—Used in some instruments to absorb CO₂.

8.3 Inert Gas (helium or argon, as specified for the instrument being used)—Use the purity specified by the manufacturer.

8.4 Magnesium Perchlorate—Used in most instruments as a moisture trap. Use the purity specified by the manufacturer.

⁵The figures mentioned in this Appendix are not included here.

9. Preparation of Apparatus

9.1 Assemble the apparatus as recommended by the manufacturer. Make the required power, gas and water connections. Turn on the instrument and allow sufficient warm up time to stabilize the system.

9.2 Change the chemical traps and filters as required. Test the furnace and the analyzer to ensure the absence of leaks. Make a minimum of two determinations using a sample as directed in 11 to condition the instrument before attempting to calibrate the system or to determine the value of the blank.

10. Sample Preparation

10.1 Use only solid samples to minimize the potential for errors due to surface oxidation. Samples must be of the proper size to permit free introduction into the sample loading device, if required, and to fit in the graphite crucible.

10.2 Cut the sample to an appropriate size using a silicon carbide, water-cooled, cut-off wheel or by other means that will avoid overheating. Avoid oxide cutting or oxide abrading materials. When appropriate, flat samples may be stamped using a punch and die. File all surfaces of the sample to remove all traces of surface oxidation oxides or other impurities, rinse in acetone, air-dry, and weigh. Be sure that all pores in the surface of the sample are free of acetone.

10.3 Do not touch the sample with fingers during and following the final stages of cleaning. Store the prepared samples in a desiccator. If they are not analyzed within four hours of preparation, repeat the filing step prior to analysis. For some material, such as plain carbon sheet steel, a rimmed or semi-killed deoxidation practice is used and the oxygen may not be uniformly distributed throughout the thickness of the sample. In these cases, the removal of too much surface material may cause erroneous results.

NOTE 1—Careful adherence to the sample preparation procedures described above is critical to obtaining accurate and precise results. The use of small and irregular shaped samples requires a diligent effort to ensure that all surface contamination has been removed.

11. Calibration

11.1 Calibration Standards—Select those standards containing approximately 0.0005, 0.002, and 0.005% oxygen and designate them as standards A, B, and C, respectively.

NOTE 2—The accuracy of the method is dependent to a large extent upon the accuracy of the methods used to certify the oxygen concentration in the calibration standards as well as upon their homogeneity.

11.2 Gas Dosing—Automated and manual gas dosing (recommended by some instrument manufacturers) can be used to set up the instrument, but instrument response must be confirmed as described below using the standards described in 11.1

11.3 Adjustment of Response of Measurement System—Using standard B as the sample, proceed as directed in 12.2–12.4. Repeat 12.2–12.4 until the absence of drift is indicated; continue running a series of samples until the last four readings have a maximum range of 0.0005% oxygen, making appropriate adjustments to the calibration potentiometer, if applicable. Record potentiometer settings. Adjust the instrument calibration to provide a reading within $\pm 0.0003\%$ absolute of the certified value for the standard using the average of the last four readings. Refer to the manufacturer's instructions.

11.4 Determination of Blank Reading—Using standard A as the sample, proceed as directed in 12.2–12.4. Repeat a sufficient number of times to establish that a low average (less than 0.0002% of oxygen) and consistent ($\pm 0.0001\%$ oxygen) individual blank values are obtained. Blank values are equal to the total result of the crucible and standard A minus the standard value of A. Record the average value of four successive blank determinations which meet the above requirements for maximum and consistent values. If the blank readings are too high or inconsistent, determine the cause, correct it, and repeat the steps as directed in 12.2–12.4. Enter the average blank value in the appropriate mechanism of the analyzer (Note 3);

refer to manufacturer's instructions. This mechanism should electronically compensate for the blank value.

NOTE 3—If the unit does not have this function, the blank value must be subtracted from the total result prior to any calculation.

11.5 Calibration—Weigh an appropriate sample size of standard C, (refer to manufacturer's instructions) to the nearest 1 mg, and place in the instrument sample loading device. Follow the calibration procedure recommended by the manufacturer using standard C as the primary standard, analyzing samples of standard C to determine the calibration slope. Run samples until the maximum range of four successive samples does not exceed 0.0005% oxygen. Treat each standard as directed in 12.2–12.4 before proceeding to the next one. Confirm the calibration by analyzing an additional standard C after the calibration procedure is completed. The value should be within the allowable limits of the standard value. If not, repeat the calibration procedure. Next, weigh at least two appropriate-sized samples of standard B, weighed to the nearest 1 mg, and transfer to the instrument sample loading device. Treat each standard as directed in 12.3 and 12.4 before proceeding to the next one. Record the results and compare them to the actual certified oxygen value of standard B; they should be within the allowable limit of the standard. If not, refer to the manufacturer's instructions for checking linearity of the system.

NOTE 4—Repeat the calibration when (1) a different lot of crucibles is used, (2) the system has not been in use for one hour, or (3) the carrier gas supply has been changed.

12. Procedure

12.1 Assemble the apparatus, calibrate, set the blank, and test the performance as directed in 9 and 11.

12.2 Transfer an appropriate sized sample, weighed to the nearest 1 mg to the instrument sample loading device.

12.3 Place a crucible on the furnace pedestal and raise the pedestal into position.

12.4 Start the crucible degassing cycle (Note 5); refer to the manufacturer's recommended procedure regarding entry of sample weight.

NOTE 5—For some instruments this procedure automatically precedes the analysis cycle.

12.5 Transfer the sample to the crucible and start the analysis cycle.

13. Calculations

13.1 Follow the manufacturer's directions to ensure that all essential variables in the calculation of analysis results have been accounted for, including the blank, weight of sample, and calibration data. Since the output of most modern instruments is given directly in percent concentration, post-analysis calculations may not be required.

Analysis of Microinclusions in Through-Hardening Bearing Steels

REFERENCE: Lane, S., "Analysis of Microinclusions in Through-Hardening Bearing Steels," *Effect of Steel Manufacturing Processes on the Quality of Bearing Steels, ASTM STP 987*, J. J. C. Hoo, Ed., American Society for Testing and Materials, Philadelphia, 1988, pp. 211–225.

ABSTRACT: An automatic optical image analyzer, the Quantimet 800 Steelscan, has been applied for microinclusion assessment in bearing steels according to the Stahl Eisen Prüfblatt (SEP) 1570/71 method. A brief description of the system is given, including the use of electronic image amendment to distinguish reliably between different types of oxides.

Although the SEP comparison chart is well-formulated and should be sensitive enough for inclusion assessment of present-day bearing steels, results are presented showing that manual assessments with this method should suffer from very poor reproducibility. The automatic method has provided a more reliable basis for discriminating different degrees of inclusion content and, as a consequence, has highlighted the difficulty in obtaining a representative measurement of microinclusion content, especially in worked material such as bar or tube. The effects of deformation by rolling, sampling errors due to material inhomogeneity, and small variations in the orientation of the plane of sectioning on the microinclusion *K* ratings are discussed.

The limitations of comparison chart methods and automatic image analysis systems are reviewed, with emphasis on the required sensitivity for assessing bearing steels, as compared with an accurate chemical analysis of impurity elements for the prediction of microinclusion content.

KEY WORDS: microinclusion content, microinclusion rating methods, image analysis, bearing steels

For many years, the bearing industry and its steel suppliers have given a lot of attention to lowering the nonmetallic inclusion content of bearing steels. The use of light optical microscopy coupled with manual comparative rating charts, for example, the ASTM Practice for Determining the Inclusion Content of Steel (E 45-85, JK charts) and SEP 1570/71, has been the traditional approach for the assessment of microinclusion contents in steels. Although the charts have provided a common basis to conveniently describe microinclusion populations, ratings by these methods are often inaccurate due to the variable subjective judgment of the metallographer and, in particular, to the incompatibility of the picture gradings and the microinclusions being assessed. This situation has been improved to some extent by the introduction of automatic optical image analysis systems, which are able to perform comparison chart ratings with a much improved reproducibility. However, as a result of the continuous development in modern steel refining technology, the sensitivity of these microinclusion rating methods for discriminating subtle discrepancies in impurity levels of "clean" (that is, <15 ppm oxygen) bearing steels is becoming increasingly inadequate.

Most recently, improved reliability of chemical methods has led to the consideration that oxygen, titanium, and sulfur analyses, for example, give a satisfactory indication of microinclusion content. This paper reviews the applicability of the different methods to assess "clean" bearing steels, including errors introduced by sampling, especially on intermediate raw material products.

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Experimental Procedure

The microinclusion content of rolling bearing rings, tubes, and bars of SAE 52100 steel, representing numerous heats from various sources, have been investigated using the SEP 1570/71 and ASTM E 45-85 JK comparison chart methods and compared with the results of chemical analysis. The assessments by the SEP method were performed automatically using an optical image analyzer, the Quantimet 800 Steelscan (Fig. 1). This system, which has been described in previous publications [1,2,3], utilizes a conventional grey level discrimination technique for detection of oxides and sulfides. As illustrated in Fig. 2, the three main oxide types are distinguished by morphology. For each field of view, measurements on detected features, such as length/width ratios and counts combined with image amendment routines, are used initially to identify possible silicates and, if necessary, subsequently to distinguish between alumina or globular type oxides. The total measured area fraction of each inclusion type per field is used for conversion to a size level rating depicted on the chart.

Results and Discussion

Accuracy of Manual/Automatic Comparison Chart Ratings

Standard comparison chart methods attempt to provide a rating which gives some indication of the size distribution and quantity of microinclusions present, according to type, in a steel. Ideally, for rolling bearing applications, they should be able to predict the probability of certain types of microinclusion occurring above a critical size at some critical position in the Hertzian zone during rolling contact loading. Other than the heat-treated steel matrix

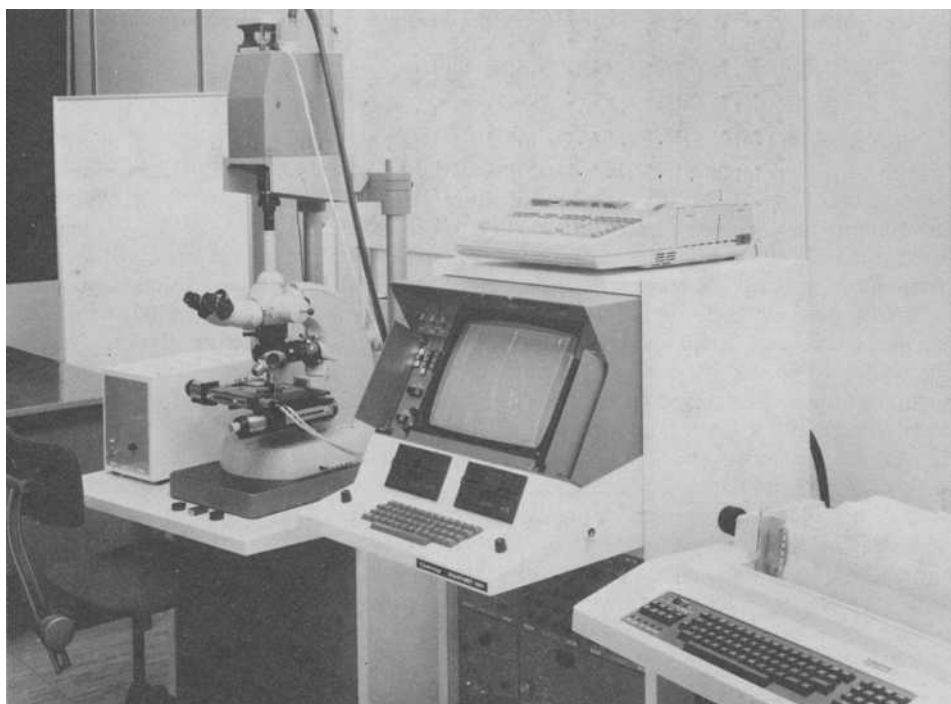


FIG. 1—Quantimet 800 Steelscan.

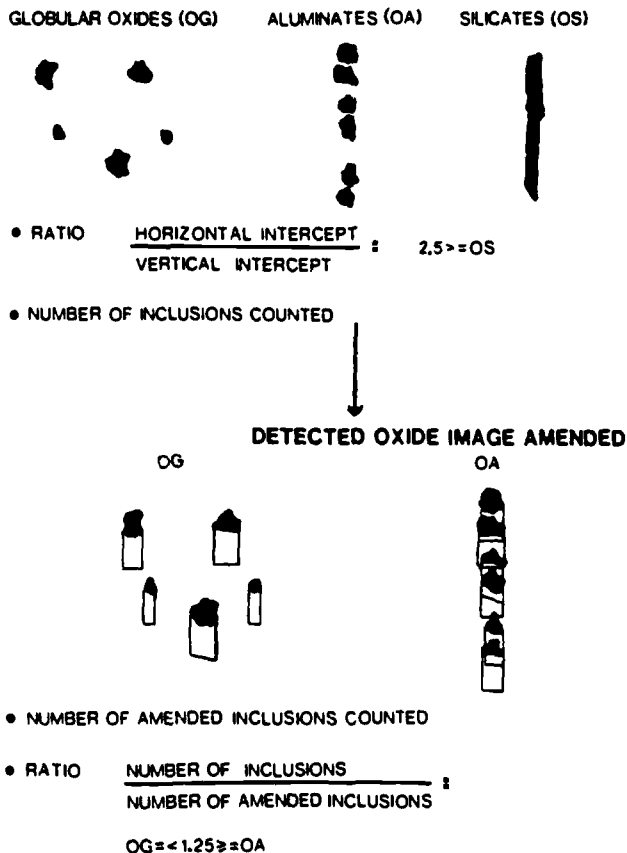


FIG. 2—Field-specific measurements for oxide identification.

properties and the associated residual stresses, it is generally accepted that microinclusion characteristics such as quantity, size, shape, chemical composition, location, and orientation are of significant importance for the initiation of rolling contact fatigue. As a consequence, it is not surprising that there is a general low success rate for trying to establish a relationship between microinclusions when using worst-field, average worst-field or index-type (SEP K-value) ratings, and the fatigue life of bearings manufactured from various grades of bearing steels. Additional factors mitigating against the success of a correlation between steel inclusion content and performance are the accuracy of the microinclusion assessment method, the effects of material variation, and reliability of sectioning.

A disadvantage of comparison charts is the limited number of picture gradings which can be used to describe the microinclusions present in a given field of view. Further, it is common that errors arise in personal judgment when comparisons are performed between the investigated sample and the chart. An example of the poor reproducibility of manual assessment is shown in Fig. 3a. Sixteen different metallographers were asked to determine a *K3* rating (inclusions with a maximum size of ~ 2 , A- and B-type thin series, according to the ASTM-JK chart), according to the SEP method, for twelve different specimens representing a range of steel qualities. All samples were polished automatically such that specimen quality in no way influenced the assessment. A consistent rating was obtained by all metallographers on the

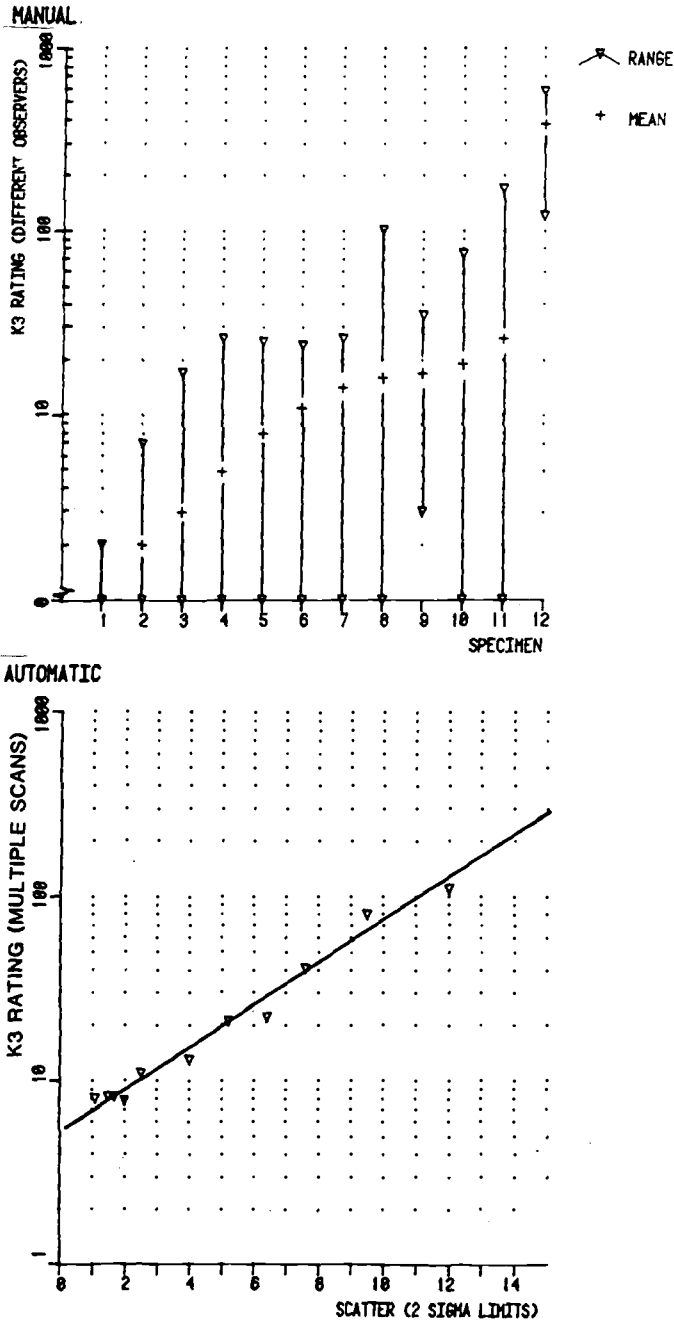


FIG. 3—Reproducibility of manual and automatic microinclusion ratings using the SEP 1570-71 method.

cleaner steel (Number 1); however, scatter in results increased substantially with decreasing steel quality. For example, in Steel Number 8, the $K3$ ratings were in the range of 0 to 100, with a resultant mean of 16. Similar degrees of variation have been reported elsewhere [4,5] when using the SEP and JK/ASTM methods, although perhaps for the latter this problem is

somewhat less pronounced since the chart is more coarsely graded. By comparison, Fig. 3b gives the automatic ratings of similar steels. The accuracy at low impurity content is similar, but in the increasingly "dirty" steels, a significant improvement in reliability is observed. For example, a steel with a $K3$ rating of 50 would have a 95% chance of being rated between 42 and 58 for repeated measurements using the Quantimet, while for manual assessment, a rating on such a steel could be in the range of approximately 0 to 150. Accordingly, for comparative purposes, an automatic assessment method is to be preferred, especially when considering the added advantages of speed of assessment ($300 \text{ mm}^2/6 \text{ min}$) and reliability over a period of time.

Some limitations of automatic methods in general are, however, the necessity of automatic polishing techniques and a very clean working environment in order to suppress artefacts and dust. Further, co-detection is unavoidable of porosity and oxides, titanium carbonitrides and sulfides, very fine oxides (less than $\sim 3\text{-}\mu\text{m}$ diameter, depending on the resolution of the system) and sulfides. The field specific type measurements with this system enable only one oxide type per field to be identified and do not account for inclusions which overlap a field of view, in terms of actual size. In addition, complex or fine agglomerated type microinclusions, or microinclusions which have an unusual morphology (perhaps as a result of being continuously cast), would be problematic for correct identification.

Sampling Errors in Microinclusion Assessment

Regardless of the type of microinclusion assessment method, there are factors which can seriously introduce sampling errors. From a steel consumer point of view, the inhomogeneous distribution of inclusions retained from the ingot are a particular problem. Steelmaking parameters, such as methods of deoxidation and teeming, ingot solidification rates and sizes, subsequent rolling temperatures, and degree of reduction (per pass and final), can all influence the size and distribution of microinclusions in steel products. Consequently, the variation of practice between different steel producers is also a variant which can promote errors when attempting to obtain representative samples or measure inclusion contents.

Figure 4 shows the variation of microinclusion content, measured automatically, on one bar sample, repolished between ten successive examinations, with some 0.1 to 0.2 mm being removed between planes of section. In both steel qualities, the percentage dispersion ($2S$ limits for all ratings) of the K -ratings is approximately 50% of the average microinclusion level reached. Figure 5 shows for a single section plane with the same steels the influence of using increasingly investigated areas (same start point) on the total K -rating. Although fairly consistent values were obtained with the good quality steel for areas of $\geq 200 \text{ mm}^2$, the bad quality steel clearly shows the effect of segregation, and thus there is no optimum area below 1600 mm^2 with which an absolutely reliable result is reached.

Usually, particular care is taken to ensure that the plane of sectioning is parallel to the original direction of metal working to avoid the foreshortening effects with sectioning in other planes [6]. The improved reproducibility of the automatic assessment has highlighted the magnitude of this problem on samples which had been sectioned with a specific angle of deviation from the true fiber-flow orientation in bar material. Figure 6a illustrates that this effect is particularly marked on alumina oxide (OA) stringers. The $K2$ value (inclusions with a maximum size of ~ 1.5 , A- and B-type thin series, according to the ASTM-JK chart) obtained on sections 10 deg to the rolling axis is less than 30% of the values obtained parallel to the rolling axis. Globular oxides (OG), as might be expected from their geometry, are relatively unaffected. Somewhat surprisingly, sulfides are not seriously influenced, and this can be explained by their higher frequency of occurrence and more homogeneous distribution throughout the steel such that more sulfides are cut by the angled plane which compensates for

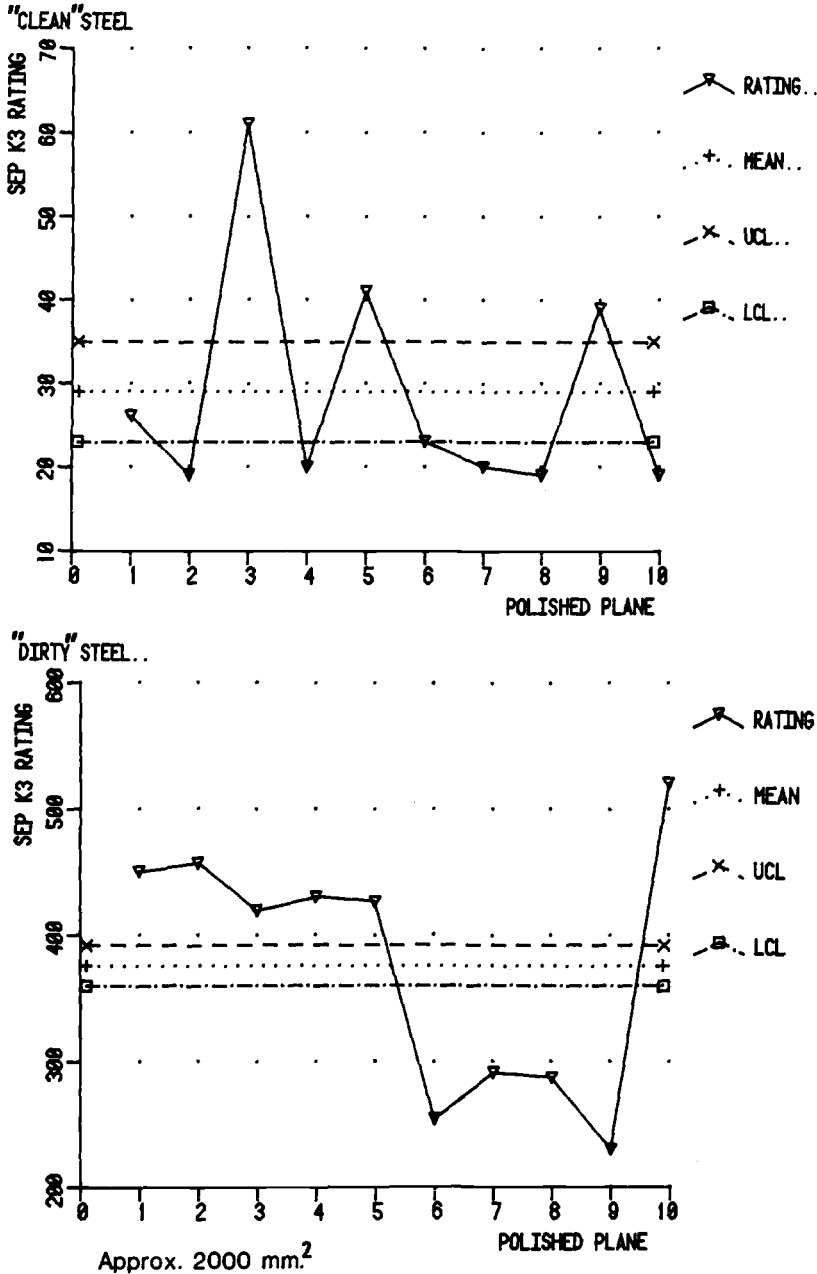


FIG. 4—Influence of different polishing planes on microinclusion ratings.

the inevitable reduction in inclusion projected length. In "clean" steels, these effects are less serious, as shown in Fig. 6b, although errors may arise when attempting to find large alumina oxides in rings or tubes where the spirally orientated material flow deviates significantly from normal forms of sections (Fig. 7). This problem and also the extra dispersion of inclusions caused by rolling deformation are considered to be the main reasons for exaggerated reduction

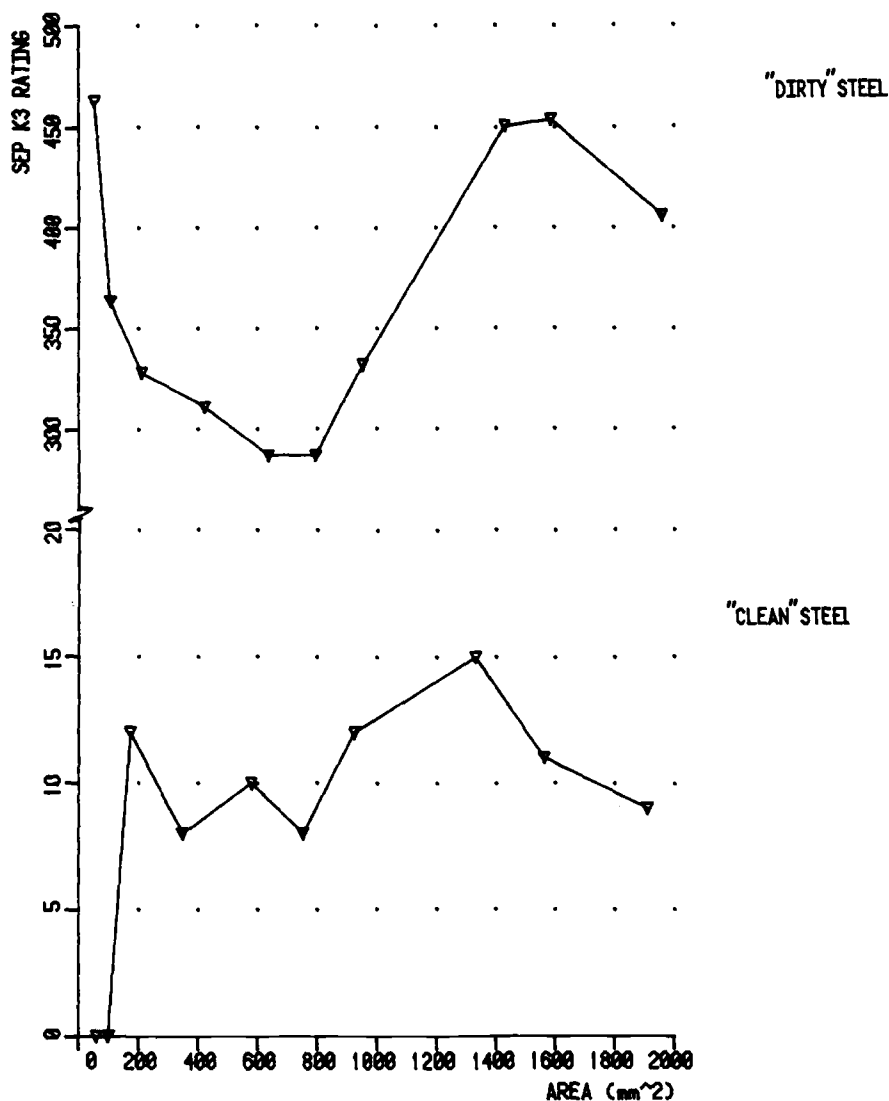


FIG. 5—Influence of area investigated on microinclusion rating.

in microinclusion contents in tubes if compared with ratings determined on original billet/bar sections. Table 1 shows this effect for three experimental steels.

Further demonstration of the poor accuracy for microinclusion assessment on intermediate product forms is given in Figs. 8 and 9. In these figures, the relationship between sulfur/oxygen contents and sulfide/oxide microinclusion contents determined on random tube and bar samples representing 73 and 57 heats, respectively, is shown. Although the degree of scatter is appreciable, it can be seen that the correlation between these parameters is much better when investigating bar material. Because the SEP method is based on gradings according to area

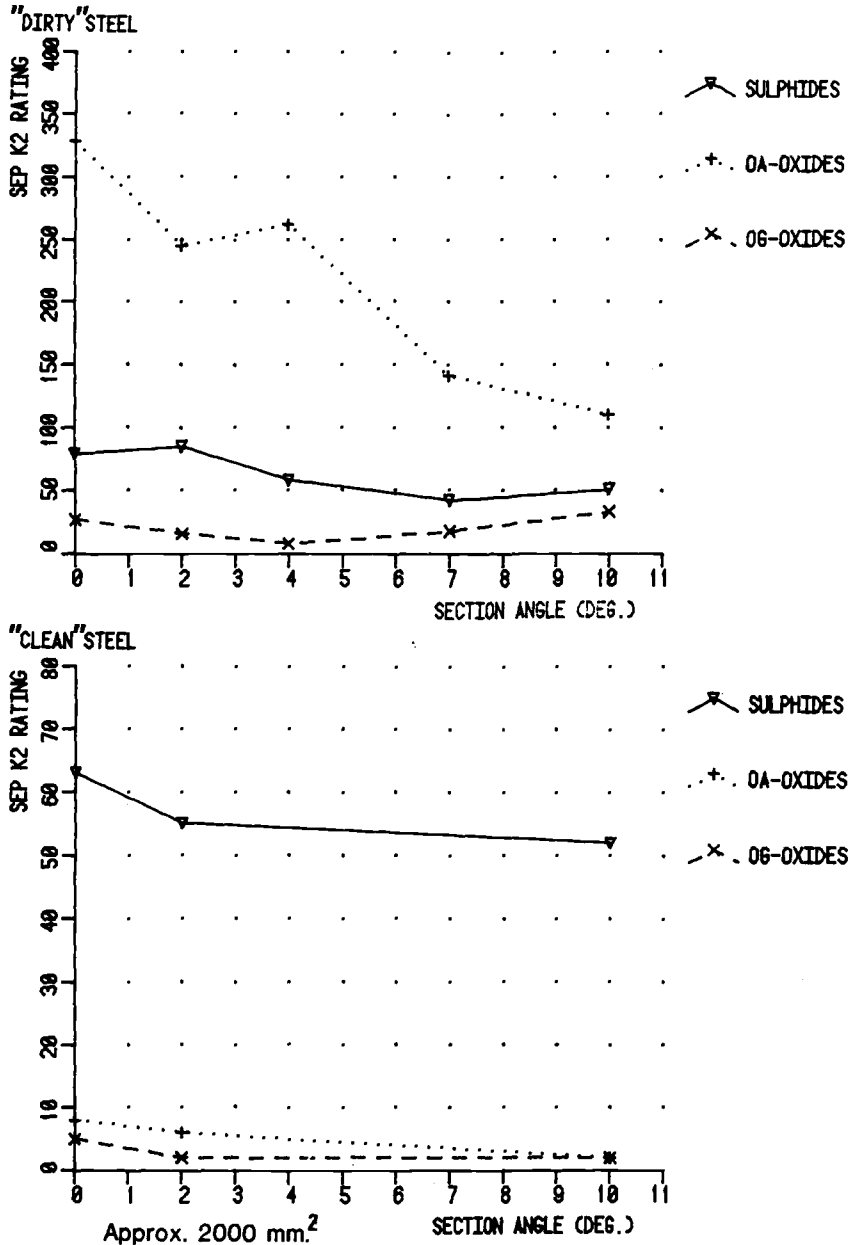


FIG. 6—Influence of angle of section plane on microinclusion rating.

fraction, which in turn should be related to oxygen and sulfur contents, it is anticipated that the correlation would be further improved if the automatic inclusion ratings are performed on billet slices originating from the same ingot positions for a given steel source.

The ASTM JK worst-field rating method is not completely related to the oxygen and sulfur contents, but it is more a qualitative measure of segregation or contamination in a heat. Because it is only the largest microinclusions which are being recorded, this could well

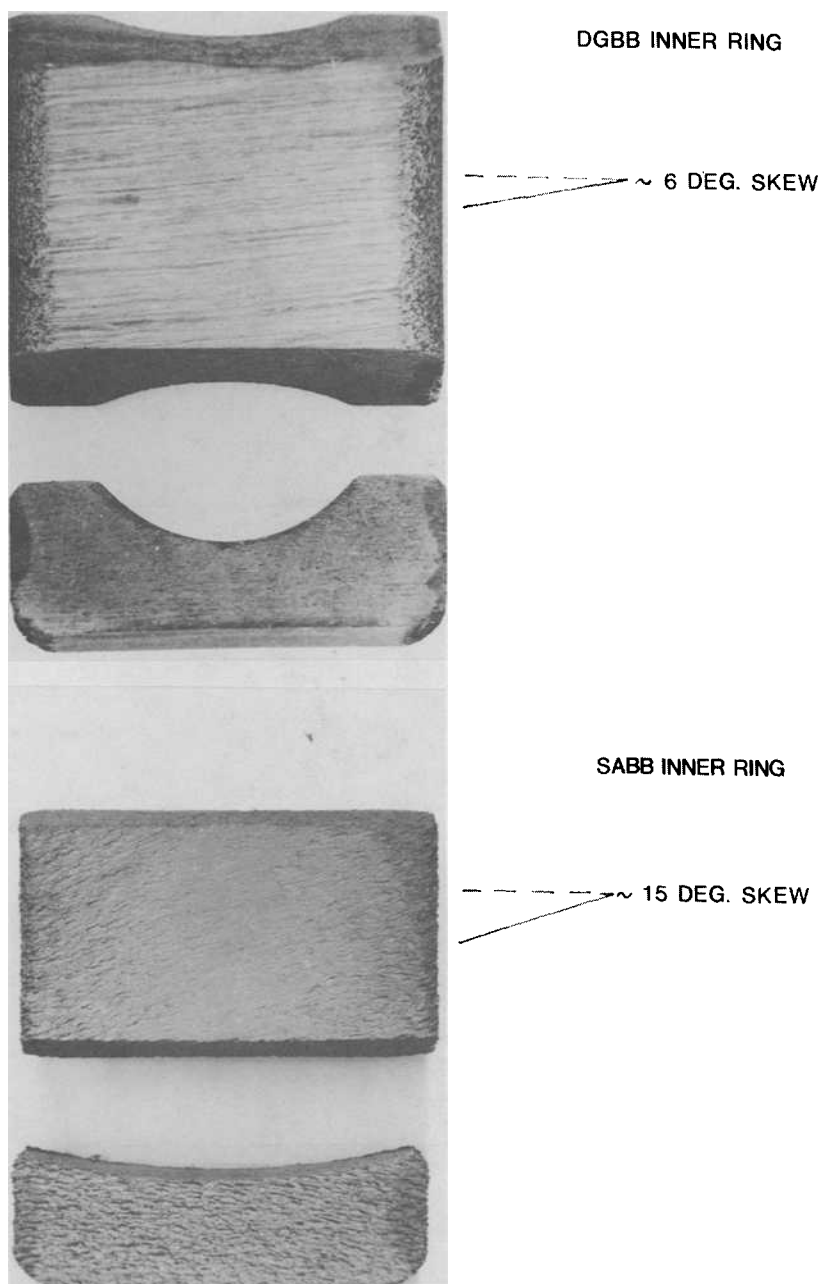
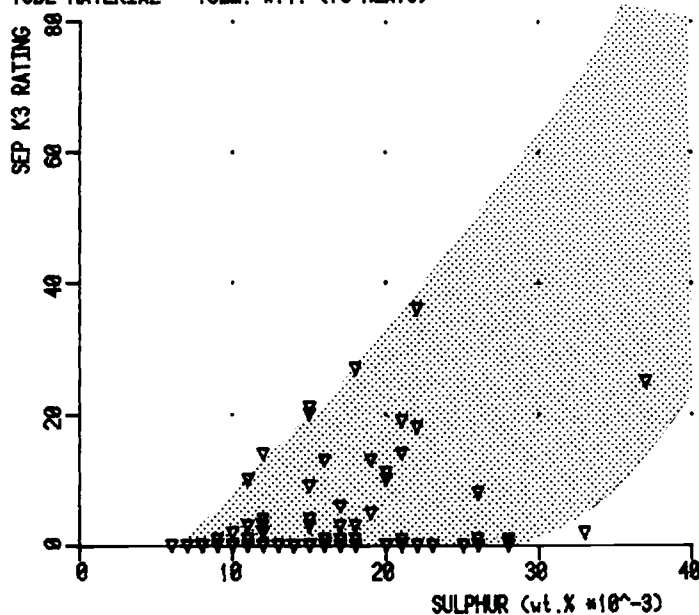


FIG. 7—Material flow lines on deep-etched ring sections.

TUBE MATERIAL ~ 10mm. W.T. (73 HEATS)



BAR MATERIAL, 80-120mm. DIA. (57 HEATS)

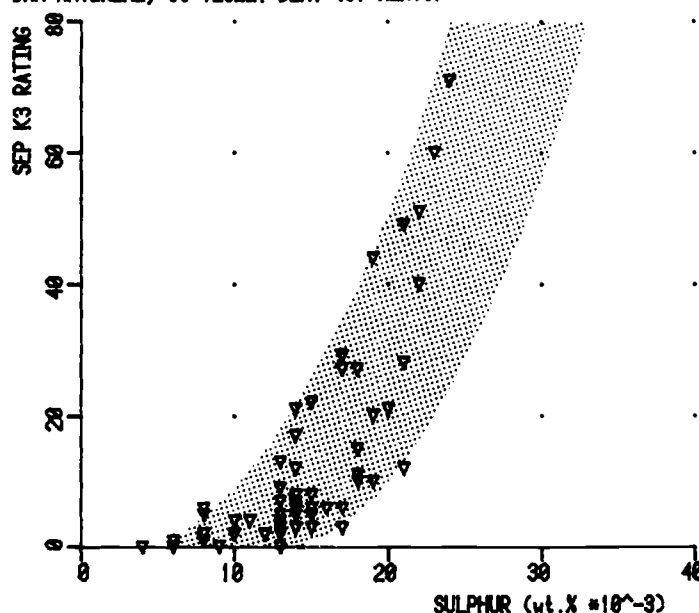
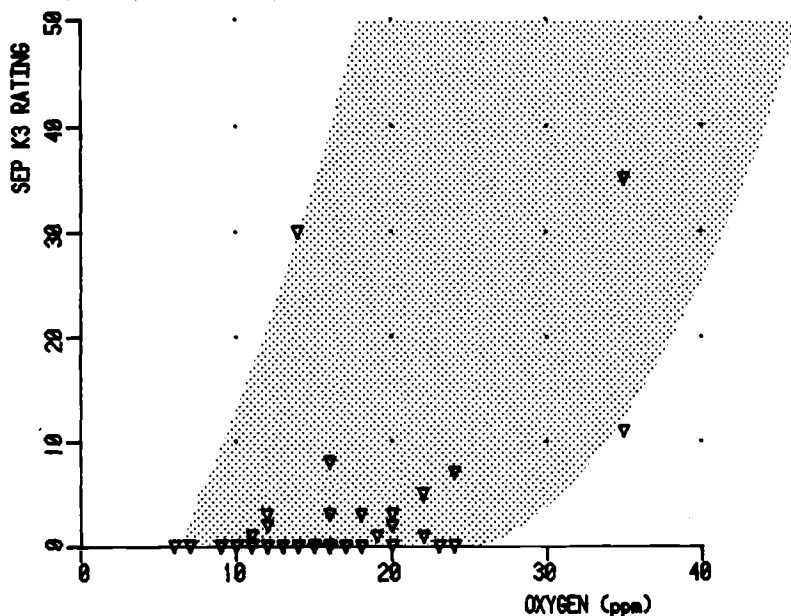


FIG. 8—Sulfide microinclusion content versus sulfur content.

correspond to a localized region where a concentration of microinclusions exists or could be due to the fact that the microinclusions are exogeneous in nature and therefore not really representative of the general distribution of the inclusions. As would be expected, the relationship between oxygen and sulfur contents and manual ASTM JK ratings for oxides and sulfides on identical bar specimens is much less apparent (Figs. 10 and 11) in relation to equivalent SEP

TUBE MATERIAL ~ 10mm. V.T. (73 HEATS)



BAR MATERIAL, 80-120 mm. DIA. (57 HEATS)

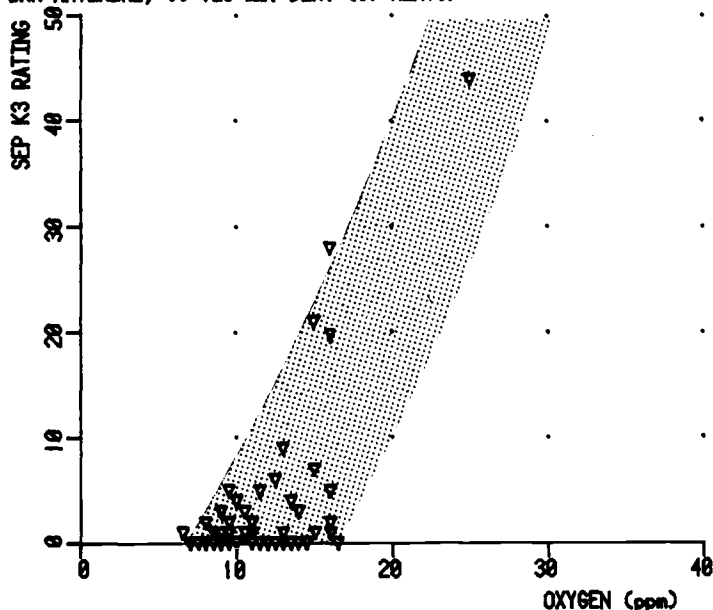


FIG. 9—Oxide microinclusion content versus oxygen content.

ratings. A thin series worst-field rating of 1.5 for sulfides and oxides could be possible for a steel containing sulfur levels between 0.003 and 0.028 weight %, and oxygen levels between 10 and 32 ppm, respectively.

Clearly, in practice none of these inclusion rating techniques appear to give a satisfactory measure of steel inclusion content, especially when investigating intermediate raw material

TABLE 1—Inclusion ratings on different product forms manufactured from three experimental steel qualities.

Steel Oxygen, ppm Sulfur, weight %	1 39 0.013		2 24 0.020		3 11 0.012	
SEP-K3 Rating	Sulfides	Oxides	Sulfides	Oxides	Sulfides	Oxides
Billet 115 mm ϕ	3	191	6	11	2	3
Bar 80 mm ϕ	7	145	14	9	3	3
Tube 100/83 mm ϕ	0	15	0	0	0	0
Tube 62/45 mm ϕ	7	1	0	0	0	0

BAR MATERIAL 80-120 mm. DIA. (57 HEATS)

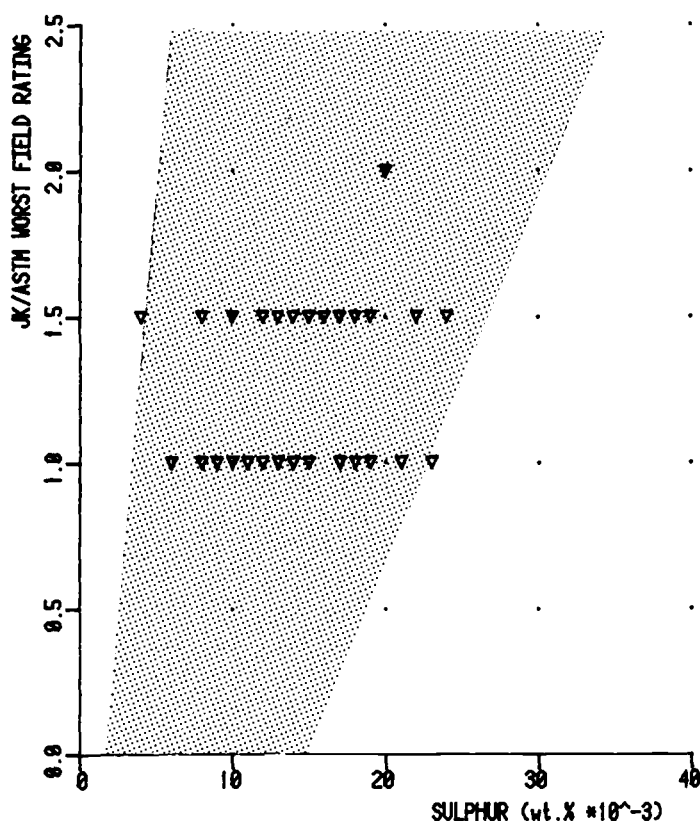


FIG. 10—Sulfide (A thin) microinclusion content versus sulfur content.

product forms such as tubes, or modern bearing steels containing less than 15 ppm oxygen. This situation may improve somewhat if the specified ASTM A295 sampling procedure, that is, taking a minimum of six specimens representing top and bottom of the first, middle, and last cast ingot, is stringently applied to check a given heat of steel.

BAR MATERIAL, 80-120 mm. DIA. (57 HEATS)

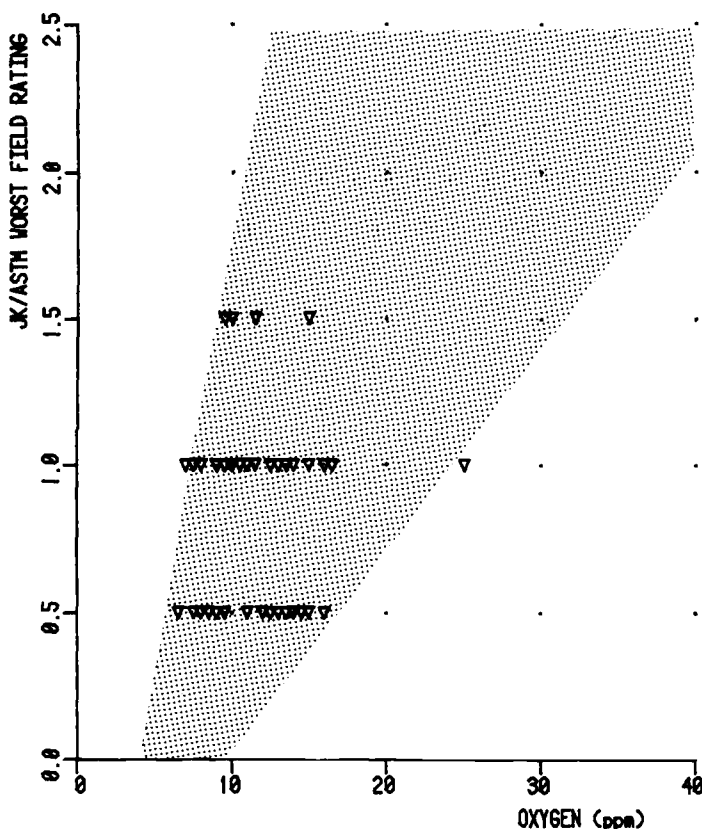


FIG. 11—Oxide (B-thin) microinclusion content versus oxygen content.

Inclusion Content Determined by Chemical Analysis

The contribution of microinclusions towards bearing performance is nowadays becoming more difficult to assess, due to the extremely long lives of steels containing low oxygen levels and the lack of evidence of microinclusions being associated with spall-type failures. In our search for a relationship between steel "cleanness" and fatigue life, the one parameter which seems the most efficient, in terms of quantification of microinclusions, is simply the oxygen level determined by accurate chemical analysis [7]. Assuming that the oxygen analysis technique is well-calibrated with suitable standards and the precautionary measures are taken to prevent surface oxidation of specimens, this procedure appears attractive from a reproducibility point of view. Figure 12 compares oxygen levels measured by three different laboratories on the same steels from different heats. The degree of scatter approaches a maximum of 5-ppm oxygen, which is governed to some degree by the inhomogeneity of the steels. When we fit a linear regressed curve on the two populations of data, a fairly constant difference of 2 ppm is evident between SKF-ERC and the other laboratories, and 4 ppm between Lab A and Lab B. Other advantages of chemical analysis are that it is insensitive to product form and that the effective sampling volume is substantially larger than typical size specimens used for microscopic examination. For example, a 1-g oxygen analysis specimen is approximately equivalent to a 15-cm² metallographic specimen containing oxide inclusions with diameters in the range

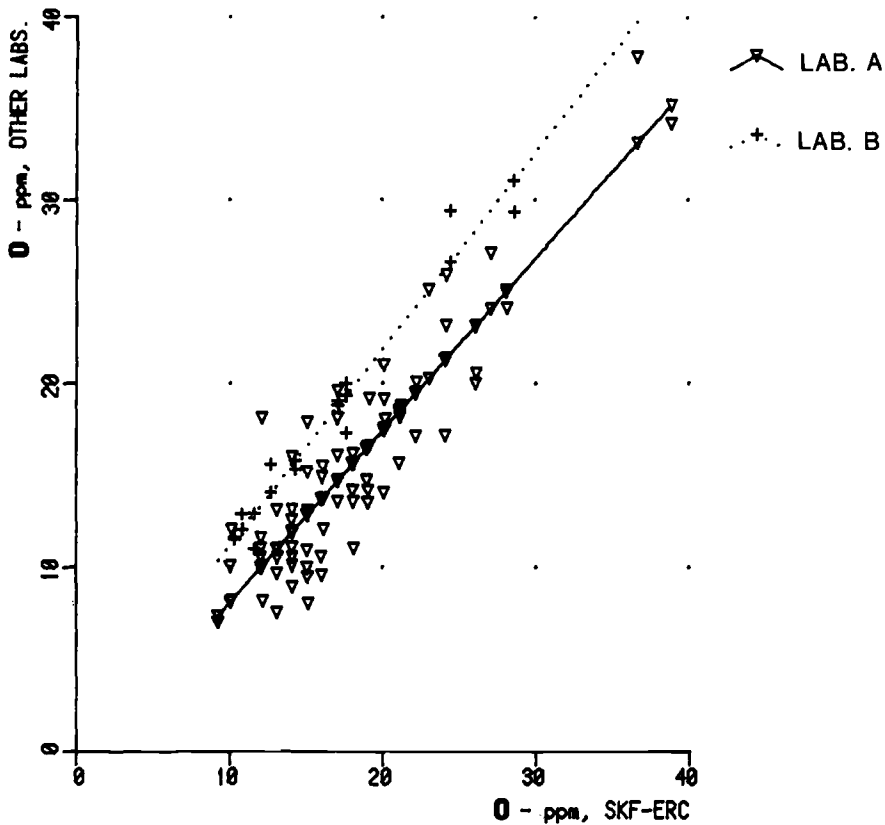


FIG. 12—Oxygen analyses by different laboratories on material from same heats.

3 to 10 μm [8]. The main shortcoming of the method, however, is that no information is obtained regarding type or size distribution of the oxides. As is well-known, large inclusions are more deleterious to fatigue than small ones; globular calcium aluminates are more damaging than broken stringer aluminates, although in good practice, the latter have to be maintained also at a very low level. Accordingly, some microscopic evaluation is necessary to supplement the chemical analysis, whether it be a quick scan for general identification of microinclusion types or an accurate, automatic quantification of microinclusion size distributions on extremely large areas, preferably prepared from billet sections. However, it must be reflected if such exorbitant experimental effort is economically viable.

Conclusions

1. Manual assessments of microinclusions using international comparison chart methods suffer from poor reproducibility.
2. An automatic image analysis technique combined with an international comparison chart method is more reliable than manual assessment, although it is very dependent upon accuracy of sampling.
3. International comparison chart methods performed on randomly selected samples from intermediate products provide only limited information on the general inclusion content.

The accuracy of these methods when assessing steels of oxygen contents below 15 ppm is questionable.

4. A recommended procedure for inclusion assessment is an accurate chemical analysis, complemented with a microscopic examination in order to identify types of inclusions and, if possible, size distributions with a high resolution automatic technique. To minimize sampling errors, billet slices of substantial size which represent top/bottom ingot positions should be used.

Acknowledgments

The author wishes to thank the Managing Director of SKF Engineering and Research Centre B.V., Dr. I. K. Leadbetter, for his kind permission to publish this paper. In addition, the author would like to acknowledge Dr. G. E. Hollox and Mr. J. M. Beswick for their useful assistance in writing this paper, and colleagues of the Physical Metallurgy Department who contributed to the work.

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Measurement of Extremely Low Inclusion Contents by Image Analysis

REFERENCE: Vander Voort, G. F., "Measurement of Extremely Low Inclusion Contents by Image Analysis," *Effect of Steel Manufacturing Processes on the Quality of Bearing Steels*, ASTM STP 987, J. J. C. Hoo, Ed., American Society for Testing and Materials, Philadelphia, 1988, pp. 226-249.

ABSTRACT: Image analysis measurements using a stereological approach have been performed on the inclusions in VIM/EBR, VIM/VAR, and VIM/ESR material from the same parent VIM heat of M-50 bearing steel. Tests made on billet samples produced excessively high, inaccurate results due to very fine microshrinkage cavities associated with the carbide due to insufficient hot reduction. Subsequent tests on bar specimens revealed a closer correlation with the calculated volume fractions based on sulfur and oxygen chemical analysis. However, the measured area fractions were consistently lower than the calculated values. Details of the metallographic preparation problems are provided.

KEY WORDS: inclusions, image analysis, M-50 bearing steel, sampling, stereology, oxides, manganese sulfides

Traditionally, inclusions in steels have been assessed using standard chart methods, most commonly by the ASTM Practice for Determining the Inclusion Content of Steel (E 45-85) and by German Standard Microscopical Examination of Special Steels for Non-Metallic Inclusions Using Standard Micrograph Charts (VDEh 1570-71) [1], although other chart methods have been developed as reviewed in Ref 2. The limitations of standard chart methods for assessing the inclusion content has been recognized [2-7] for many years.

Several approaches have been used to improve measurement reliability using comparison charts. The quality of the chart picture grading and the depiction of inclusion types relative to currently produced steels can be improved, as has been done with VDEh 1570 [1,2]. Worst-field measurement results can be better described by producing frequency distributions of the worst-field ratings, as recommended by Wojcik and Walter [8]. Chart ratings can also be supplemented by data on the length and thickness of the inclusions, as suggested by Spies [9]. The chart ratings may also be combined to produce a single index that reflects the overall inclusion content [10,11] such as the SAM index, discussed elsewhere by Tayeh and Woehrle [12]. Another approach, proposed by Johansson [13-16], employs a new chart and a programmed calculator to analyze field-by-field results quantitatively. The development of sophisticated image analyzers has prompted development of image analysis quantitative chart rating procedures for ASTM E 45-85 ratings [17-20], which has led to the new ASTM Standard Practice for Obtaining JK Inclusion Ratings Using Automatic Image Analysis (E 1122), and for VDEh 1570 [19-26].

The introduction of automatic image analyzers has also prompted development of inclusion assessment procedures using principles of quantitative stereology [27-38] in order to overcome

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chart rating deficiencies. Subcommittee E 0.04 of ASTM is presently balloting a stereologically based image analysis test method for assessment of inclusion content, tentatively titled Method for Determining the Inclusion Content of Steel and Other Metals by Image Analysis.² The ISO has also been working on a similar document.

In this paper, results are presented for the measurement of inclusions in three ingots of M-50 bearing steel from the same parent vacuum induction melted (VIM) heat that were remelted by electron beam (EBR), vacuum arc (VAR), and electroslag (ESR) processes. Measurements were made using the proposed ASTM method on both billet and bar specimens from the top, middle, and bottom locations with respect to the original ingot. Because the inclusion contents of these steels were extremely low, no effort was made to produce separate measurements for both oxides and sulfides. Chart rating methods, such as ASTM E 45, are not useful for these steels because nearly all of the fields are below the lowest severity limits depicted by the charts.

Proposed ASTM Measurement Method

This proposed ASTM method for measuring inclusion content utilizes field measurements only. Feature-specific measurement procedures are not included because performance of such measurements on several hundred fields would be time-consuming, particularly for quality control purposes. In addition, not all machines are capable of such measurements. Instead, only simple counting type, stereologically-based measurements that are independent of any assumptions regarding shape are employed. These measurements are made on each field, and after the desired number of fields have been measured, the means and standard deviations of the measurements are computed. Next, the 95% confidence interval and the percent relative accuracy are calculated in the same manner as described in the Standard Practice for Determining Volume Fraction by Systematic Manual Point Count (E 562-83). Oxides are separated from sulfides, if desired, by gray level differences. Sampling and specimen orientation are chosen in the manner described by ASTM E 45-85. Besides the usual mid-thickness location, the proposed method also permits testing at other cross-sectional locations as well, such as subsurface and center locations. All specimens are longitudinally oriented.

Four measurements are made on each field, for the oxides and sulfides, either separately or collectively: the area of the detected inclusions; the number of detected inclusions; the number of interceptions of inclusions with the scan lines; and the projected height of the inclusions. Each of these measurements is normalized by the appropriate parameter:

1. The area of inclusions per field is divided by the mask area to obtain the area fraction A_A in percent.
2. The number of inclusions per field is divided by the true mask area to obtain the number per unit area N_A .
3. The number of inclusion interceptions with the scan lines is divided by the total true length of the scan lines in the mask area to obtain the number of interceptions per unit length N_L .
4. The total projected height of the inclusions is divided by the number of inclusions in the mask area to obtain an average inclusion length $\bar{L}$.

Before the measurements are taken, the sample is placed on the microscope stage so that the hot working direction, that is, the direction of inclusion elongation, is vertical on the screen. Thus, the simple projection measurement, when divided by the number of particles in the mask area, yields an average inclusion length for each field.

²Subsequently passed sub-, main-, and society ballots and given the designation E 1245-88.

These measurements are used to calculate two additional parameters for each field. The area fraction (not in percent) is divided by the N_A value to obtain the average inclusion area $\bar{A}$, per field. The N_L and A_A values (not in percent) are used to calculate the mean free path perpendicular to the hot working axis $\bar{\lambda}$ according to

$$\bar{\lambda} = \frac{1 - A_A}{N_L} \quad (1)$$

This provides an estimate of the mean edge-to-edge spacing between inclusions in the direction perpendicular to the hot working direction. Thus, for each field, measurements are made of the area fraction A_A (equal to the volume fraction V_V), the number per unit area N_A , the average length $\bar{L}$, the average area $\bar{A}$, and the mean free path $\bar{\lambda}$. A_A is in percent, N_A is in mm^{-2} , $\bar{L}$ is in μm , $\bar{A}$ is in μm^2 , and $\bar{\lambda}$ is in μm .

After n fields have been measured, mean values for A_A , N_A , $\bar{L}$, $\bar{A}$, and $\bar{\lambda}$ are calculated and the field-to-field variability is used to compute a standard deviation s . Next, the 95% confidence interval (95CI) is calculated for each measurement

$$95\text{CI} = \pm \frac{ts}{(n)^{1/2}} \quad (2)$$

For simplicity purposes, t is set equal to 2.0, which is satisfactory for $n \geq 100$. Then, the percent relative accuracy, %RA, is calculated for each measurement:

$$\% \text{RA} = \frac{100(95\text{CI})}{\bar{X}} \quad (3)$$

where $\bar{X}$ represents the mean values. In general, a 10% relative accuracy is satisfactory for most measurements. However, when dealing with measurements of inclusions, except for resulfurized steels, it is difficult to obtain this level of confidence, and one must generally accept a higher %RA value. As the inclusion content decreases, the %RA increases for a constant value of n . From a practical viewpoint, n values between 100 and 300 are employed. To obtain a reasonably good assessment of the inclusion content of a heat of steel, it is better to measure a greater number of locations using about 200 measurements per specimen than measure 500 or more fields per specimen on a smaller number of specimens.

Test Materials

Table 1 lists the chemical analysis of the VIM electrodes before remelting by the three procedures. The chromium level for the electrode to be remelted by EBR was intentionally raised in anticipation of expected losses during electron beam remelting. After EBR, the chromium content was reduced by about 0.3%. The manganese content was also reduced by the same amount to a level below the specification limits. The molybdenum and vanadium levels were not affected.

A few samples were obtained from a disk cut from the top of one of the VIM electrodes. The average oxygen content of these samples was 31 ppm, and the average sulfur content was 0.0040% (analyzed by the experimental method described next). Although this is inadequate sampling, it does give us an idea of the relative amount of oxygen present before remelting, at least at this location. Oxygen and sulfur were measured after remelting. The oxygen content was determined by the inert-gas fusion method using the Leco TC-136 according to the Proposed Method for Determination of Oxygen in Steel and in Iron, Nickel, and Cobalt

TABLE 1—*Chemical analysis of M-50 VIM electrodes, weight percent.*

Remelt Type	C	Mn	Si	P	S	Cr	Mo	V
EBR	0.82	0.34	0.20	0.011	0.003	4.88	4.10	1.04
VAR	0.84	0.34	0.19	0.011	0.003	4.17	4.14	1.04
ESR	0.84	0.34	0.19	0.011	0.003	4.17	4.14	1.04

Alloys,³ to be added to the ASTM Methods for Determining of Carbon, Sulfur, Nitrogen, Oxygen, and Hydrogen in Steel and in Iron, Nickel, and Cobalt Alloys (E 1019-85). Measurements were made at the top, middle, and bottom of bars forged from each remelted billet. The average values were 6.4, 6.7, and 20.1 ppm, respectively, for the EBR, VAR, and ESR material. The sulfur content was initially determined with the Leco CS-344 device by the ASTM combustion-infrared absorption method (E 1019-85) at the same test locations. The average sulfur contents were 0.0015, 0.002, and 0.0012%, respectively, for the EBR, VAR, and ESR material. Because this test method is not ideal for such low sulfur levels, the specimens were reanalyzed using an experimental combustion/pararosanilin spectrophotometric method, a test method originally developed by Allegheny Ludlum and refined by T. R. Dulski.⁴ The average sulfur levels were 0.00207, 0.0033, and 0.00177% for the VIM/EBR, VIM/VAR, and VIM/ESR bars, respectively. These values are slightly higher than those obtained with the CS-344, but the same relative trends are observed by each method.

The remelted ingot sizes were 305 mm (12 in.) diameter for the EBR ingot, 406 mm (16 in.) in diameter for the VAR ingot, and 381 mm (15 in.) in diameter for the ESR ingot. Each ingot was rotary forged to a 152-mm (6 in.) square billet size. Billet disks were cut from the top, middle, and bottom after inspection and cropping. Measurements were made at subsurface, mid-radius, and center locations on each billet disk using longitudinally oriented specimens. Sections were also cut from each billet at the top, middle, and bottom locations and press forged to 25 mm (1 in.) square. Six specimens were cut from these bars, at each location, and split longitudinally. Measurements were spaced over the cross-section of each bar specimen.

Specimen Preparation

M-50 bearing steel is a very difficult grade to polish to the level necessary for accurate image analysis, due primarily to two problems: cracking or holes in the carbides or at the carbide/matrix interface and very fine pitting. All specimen preparation has been conducted using the Struers Abramatic system and unmounted specimens.

We have found that the best results are obtained when the specimens in the same holder exhibit the same approximate hardness. In this work, all billet test specimens were quenched and tempered to about 63 HRC. The as-cast specimens were hot isostatically pressed (the microshrinkage cavities present in the as-cast condition made accurate measurement impossible) and slow cooled but not hardened and the bar specimens were slow cooled after forging but not hardened. Hence, these specimens were hardness tested and grouped in 5 HRC unit increments before polishing. The specimens are placed in a holder and ground for about 40 s at 10 N pressure using a 150-grit alumina stone to achieve a planar condition for all of the specimens in the holder.

³Draft Number 10, 21 Aug. 1985.

⁴Carpenter Technology Corporation, Reading, PA 19612-4662.

TABLE 2—Polishing procedure for M-50.^a

Step	Procedure	Time and Pressure		
		Preliminary	Main	Final
1	6- μ m diamond on Petrodisc-M	5 s/50 N	230 s/200 N	5 s/100 N
2	3- μ m diamond on red felt cloth	5 s/50 N	60 s/50 N	5 s/50 N
3	1- μ m diamond on Microcloth	5 s/50 N	30 s/50 N	5 s/50 N
4	0.05- μ m γ -alumina on Microcloth	5 s/50 N	5 s/50 N	5 s/50 N

^a150 rpm used for all steps.

Our experience with M-50 has shown that grinding with silicon carbide papers through the usual sequence does not produce results as good as achieved with the Struers Petrodisc-M. Table 2 presents our current procedure for polishing M-50. The times shown are varied depending on the hardness of the specimens (240 to 480 s on the Petrodisc-M, 70 to 140 s with 3- μ m diamond, 40 to 80 s with 1- μ m diamond, and 15 to 30 s with 0.05- μ m alumina). The degree of lubrication is important because excessively moist cloths seem to promote pitting. The cloths are just kept damp using Struers "blue" lubricant. Extended times in an ultrasonic cleaner should also be avoided. As soon as each polishing cycle has been completed, it is imperative that the samples be washed and dried immediately. After polishing, the specimens are placed in a desiccator; they should be measured within 24 h, otherwise pitting and other surface oxidation effects can occur.

At these very low inclusion levels, even a minor amount of surface deterioration cannot be tolerated. Scratches are relatively easy to control. When the specimen is to be measured, we hand-polish it for about 10 s using γ -alumina (deagglomerated Meller γ -alumina). The specimen is inspected before measurement. One additional 10-s hand polish can usually be performed without producing excessive relief at the carbides. The ASTM Practice for Preparing and Evaluating Specimens for Automatic Inclusion Assessment of Steel [E 768-80 (1985)] is used to assess specimens. In general, relief is not a problem. Colloidal silica has not been used for final polishing for image analysis measurements of M50 due to a greater tendency for promoting relief than alumina. We do use colloidal silica extensively for other steels.

Results

As-Cast Specimens

A disk had been cut from the top of one of the VIM electrodes, and three specimens were obtained from the disk. Although this degree of sampling is inadequate to characterize the initial inclusion content before remelting, the results give an idea of the initial condition, at least for this ingot location. Sampling of as-cast ingots or electrodes is rarely done due to the difficulty of such sampling.

The specimens were polished but could not be measured with accuracy due to the fine microshrinkage cavities present in the as-solidified condition. Hence, the specimens were hot-isostatically pressed to heal the microshrinkage sufficiently to permit analysis. Measurements (Table 3) were made using a Leitz TAS-Plus image analyzer using a $\times 50$ objective. Some measurements were also made with $\times 32$ and $\times 80$ objectives. The $\times 32$ objective did not

TABLE 3—Inclusion measurements of as-cast ingot.^a

A_A , %	N_A , mm^{-2}	$\bar{L}$, μm	$\bar{A}$, μm^2	$\bar{\lambda}$, μm
0.013 ± 0.008	131.2 ± 64.9	0.90 ± 0.09	1.02 ± 0.15	8960 ± 5115

^aThree samples from the top of one VIM ingot were measured after hot isostatic pressing to heal the microshrinkage cavities.

provide adequate resolution for detection of the finer inclusions. Resolution was adequate with the $\times 50$ and $\times 80$ objectives, but the former produced better %RA values because of the larger field size which produced less field-to-field variability than the $\times 80$ objective.

Billet Specimens

The remelted ingots were rotary forged to produce 152-mm (6-in.) square billets. Disks were cut from the top, middle, and bottom, and specimens were cut for measurements at the subsurface (S), mid-radius (MR), and center (C) locations on a longitudinal plane. These specimens were quenched and tempered to about 63 HRC before polishing. Two hundred fields were measured on each specimen using the Leitz TAS-Plus. Table 4 presents the mean values for each specimen and location. Table 5 summarizes the values by billet location by computing the averages and 95% CI from the S, MR, and C data at each location. Note that the A_A and N_A values are equal to or greater than those of the as-cast specimens. This was caused by the presence of cracks and holes in the carbides and holes at the carbide/matrix interface. These very fine defects have the same black appearance on the monitor as oxides, and they cannot be separated from the oxides with the threshold setting. Table 6 presents the average values of all of the specimens from each billet. Statistically, there is no difference between the values at a 95% confidence level, and because of the presence of the artifacts, no conclusions can be drawn from this data.

Because of the holes present in carbides and at carbide/matrix interfaces, several samples were selected for metallographic examination using high power light microscopy and scanning electron microscopy. Figure 1 shows a typical example of holes observed at the end of elongated carbides in a middle location specimen from the VIM/EBR billet specimen at the mid-radius. The sample was lightly etched to reveal the relationship of the holes to the carbides. Figure 2 shows another example from the same specimen illustrating holes within the carbides. Figure 3 shows another example of a large hole in a massive carbide, the angular dark spots (in the scanning electron microscope [SEM] micrograph) within the carbide are vanadium-rich MC regions in the larger M_6C type carbide. EDS analysis within the large hole did not reveal any evidence of inclusions.

Figure 4 shows a manganese sulfide inclusion in a middle location, mid-radius billet specimen from the VIM/VAR ingot. Again, this specimen was lightly etched so that the relationship of the sulfide to the carbides can be observed. This sulfide is typical of the largest observed in this specimen.

Figure 5 shows an example of holes associated with carbides in a middle location, mid-radius specimen from the VIM/ESR billet specimen. Again, the specimen has been etched lightly so that the carbides can be observed. Figure 6 shows an example of a duplex oxide (aluminate)/manganese sulfide inclusion in this specimen. Note the small hole also shown in the micrographs. Figure 7 shows a similar complex aluminate/sulfide inclusion in this specimen where the oxide is angular rather than globular, as in Fig. 6.

TABLE 4—Inclusion measurements of M-50 billets (average values).

Type	Location ^a	A_A , %	N_A , mm^{-2}	$\bar{L}$, μm	$\bar{A}$, μm^2	$\bar{\lambda}$, μm	
VIM/EBR	top	S	0.030	312	1.02	0.96	2940
		MR	0.068	504	1.30	1.36	1491
		C	0.055	422	1.36	1.31	1797
	middle	S	0.006	117	0.71	0.48	13684
		MR	0.015	104	1.12	1.44	8533
		C	0.020	118	1.08	1.67	6811
	bottom	S	0.012	203	0.78	0.59	5991
		MR	0.021	217	1.19	0.97	4434
		C	0.034	264	1.18	1.27	3001
VIM/VAR	top	S	0.006	176	0.62	0.37	7800
		MR	0.014	176	1.00	0.78	5486
		C	0.018	155	1.02	1.19	5060
	middle	S	0.013	147	1.05	0.90	5817
		MR	0.031	192	1.34	1.61	3268
		C	0.026	234	1.14	1.10	3922
	bottom	S	0.022	257	0.87	0.84	3538
		MR	0.017	164	1.08	1.06	5334
		C	0.011	190	0.71	0.60	6655
VIM/ESR	top	S	0.014	248	0.62	0.55	5788
		MR	0.009	110	0.92	0.83	7947
		C	0.021	212	0.92	1.01	4322
	middle	S	0.011	169	0.75	0.66	7645
		MR	0.009	122	0.83	0.74	9104
		C	0.025	130	1.21	1.94	5054
	bottom	S	0.021	468	0.60	0.46	3420
		MR	0.028	484	0.60	0.58	2896
		C	0.044	474	0.81	0.93	2137

^aS = subsurface, MR = mid-radius, C = center.TABLE 5—Inclusion measurements of billets by ingot location (averages $\pm$ 95% CI).

Type	Location	A_A , %	N_A , mm^{-2}	$\bar{L}$, μm	$\bar{A}$, μm^2	$\bar{\lambda}$, μm
VIM/EBR	top	0.051 ± 0.048	413 ± 239	1.23 ± 0.45	1.21 ± 0.54	2076 ± 1898
	middle	0.014 ± 0.018	113 ± 20	0.97 ± 0.56	1.20 ± 1.57	9676 ± 8884
	bottom	0.022 ± 0.027	228 ± 79	1.05 ± 0.58	0.94 ± 0.84	4475 ± 3714
VIM/VAR	top	0.013 ± 0.015	169 ± 30	0.88 ± 0.56	0.78 ± 1.02	6115 ± 3662
	middle	0.023 ± 0.022	191 ± 109	1.18 ± 0.37	1.20 ± 0.91	4336 ± 3290
	bottom	0.017 ± 0.014	204 ± 119	0.89 ± 0.47	0.83 ± 0.57	5176 ± 3888
VIM/ESR	top	0.015 ± 0.015	190 ± 178	0.82 ± 0.43	0.80 ± 0.58	6019 ± 4532
	middle	0.015 ± 0.022	140 ± 62	0.93 ± 0.61	1.11 ± 1.79	7268 ± 5096
	bottom	0.031 ± 0.030	475 ± 20	0.67 ± 0.30	0.66 ± 0.60	2818 ± 1603

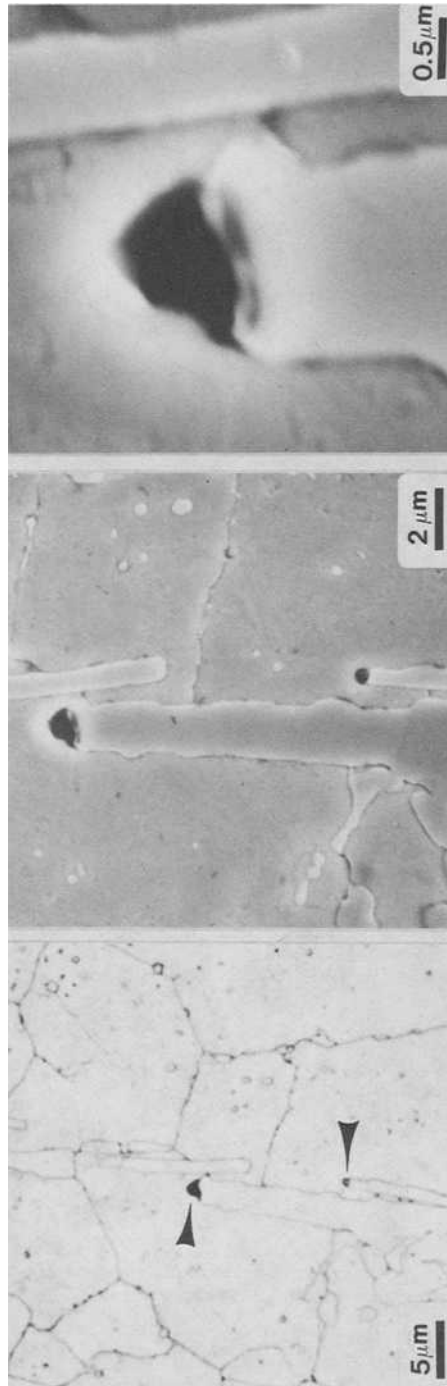


FIG. 1—Examples of holes (arrows) at carbide/matrix interface observed in middle location, mid-radius VIM/EBR billet specimen as observed by light microscopy (left, $\times 1640$) and by scanning electron microscopy (center, $\times 4100$ and right, $\times 16400$, both secondary electron images). All micrographs taken after lightly etching with Vilella's reagent.

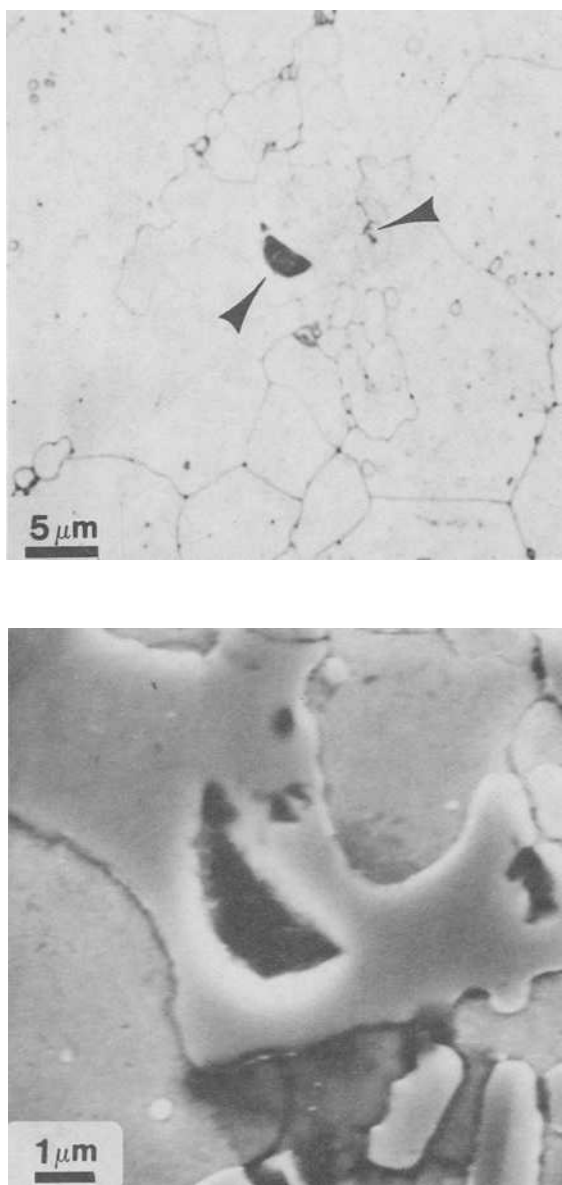


FIG. 2—Examples of holes within carbides (arrows) observed in middle location, mid-radius VIM/EBR billet specimen as examined by light microscopy (top, $\times 2000$) and scanning electron microscopy (bottom, $\times 8000$, secondary electron image).

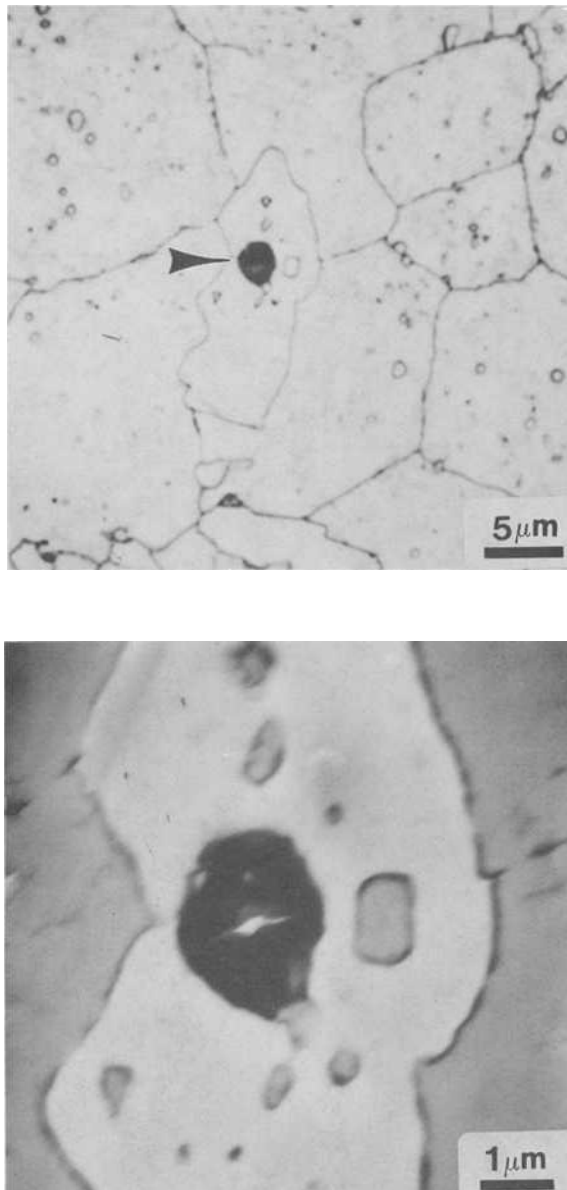


FIG. 3—Example of hole (arrow) in large carbide observed in middle location, mid-radius VIM/EBR billet specimen as examined by light microscopy (top, $\times 2000$) and scanning electron microscopy (bottom, $\times 10\,000$, secondary electron image). The angular particles within the massive carbide that appear white in the LOM image and gray in the SEM image are V-rich MC-type carbides.

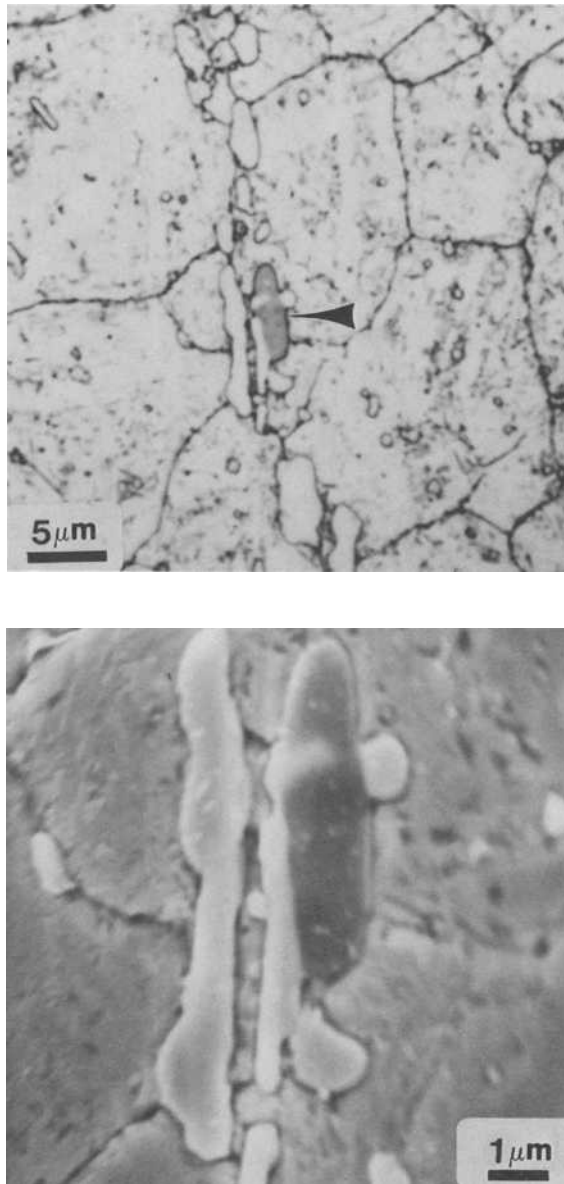


FIG. 4—Example of manganese sulfide inclusion (arrow) observed in middle location, mid-radius VIM/VAR billet specimen as examined by light microscopy (top, $\times 2000$) and scanning electron microscopy (bottom, $\times 8000$, secondary electron image).

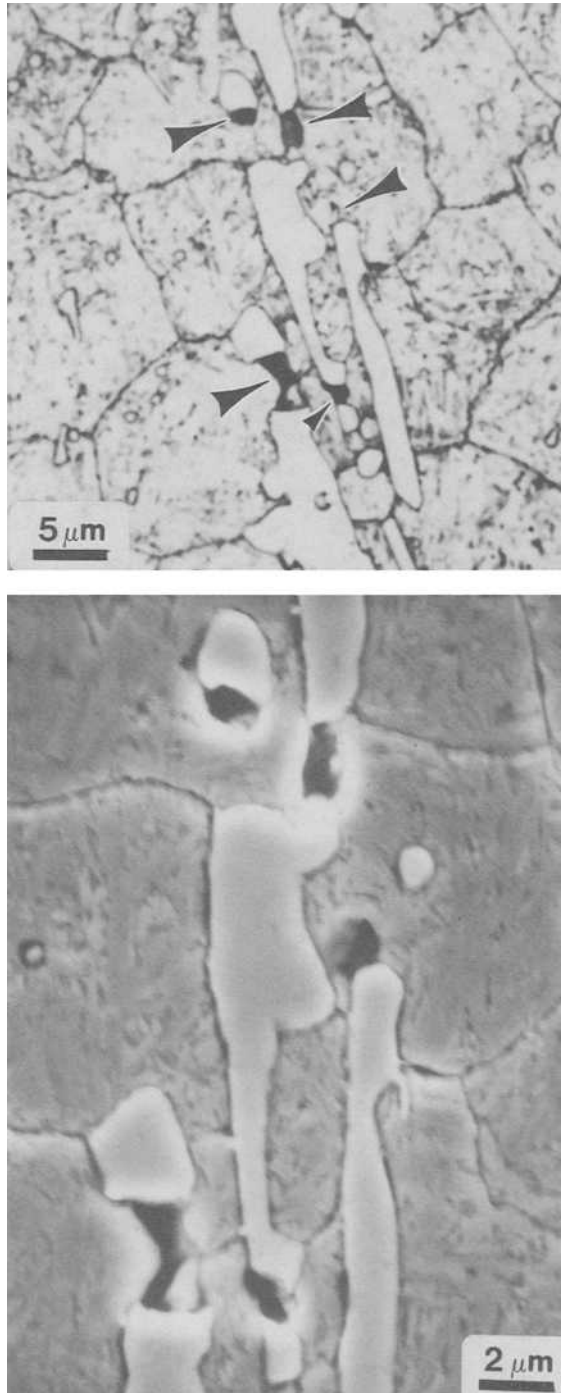


FIG. 5—Examples of holes (arrows) at carbide/matrix interfaces in middle location, mid-radius VIM/ESR billet specimen as examined by light microscopy (top, $\times 2000$) and by scanning electron microscopy (bottom, $\times 5000$, secondary electron image).

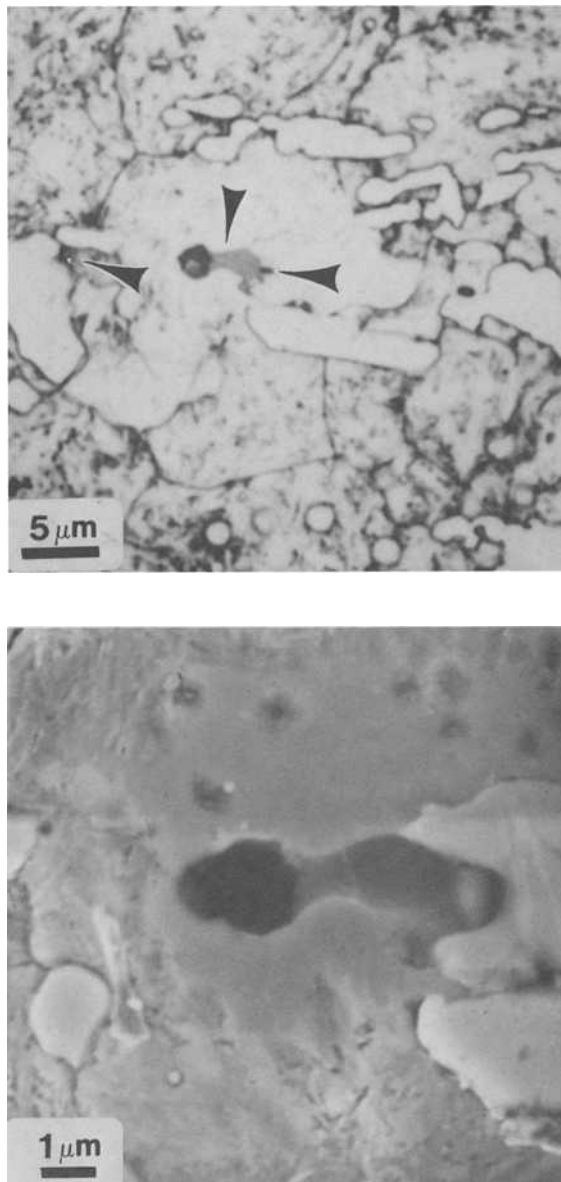


FIG. 6—Example of duplex oxide (black)-sulfide (gray) inclusion found in the middle location, mid-radius VIM/ESR billet specimen as examined by light microscopy (top, $\times 2000$) and by scanning electron microscopy (bottom, $\times 7000$, secondary electron image). Vertical arrow points to sulfide; aluminate is the dark adhering spot. Horizontal arrows indicate holes.

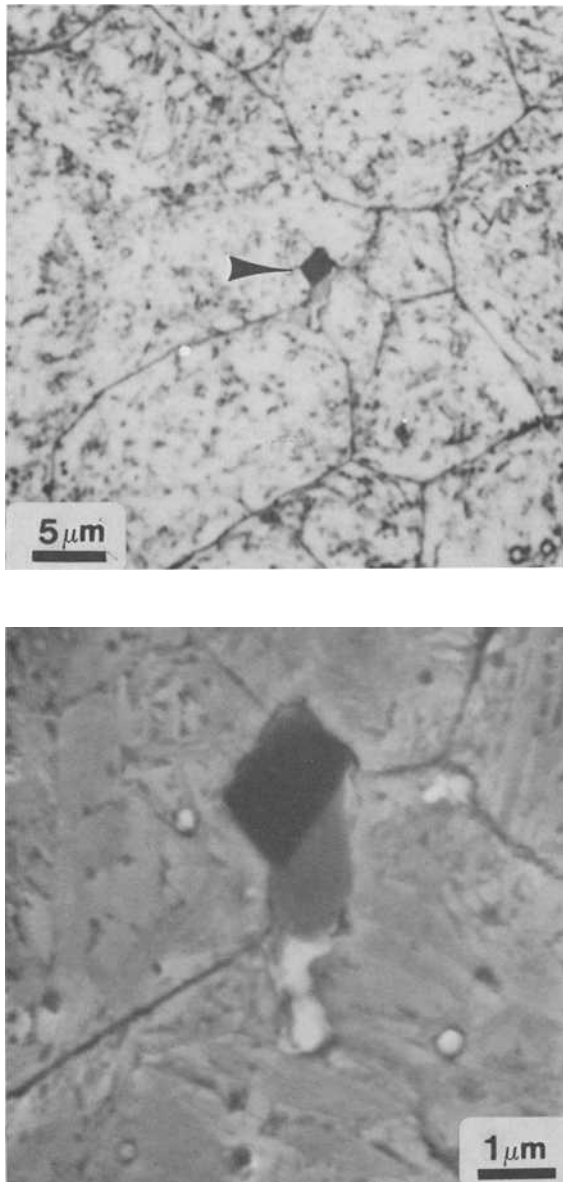


FIG. 7—Example of duplex oxide (black) -sulfide (gray) inclusion found in middle location, mid-radius VIM/ESR billet specimen as examined by light microscopy (top, $\times 2000$) and scanning electron microscopy (bottom, $\times 10000$, secondary electron image). Arrow points to angular aluminate (black); below it is gray sulfide.

TABLE 6—Inclusion measurements of billets by melt type.

Type	A_A , %	N_A , mm^{-2}	$\bar{L}$, μm	$\bar{A}$, μm^2	$\bar{L}$, μm
VIM/EBR	0.029 ± 0.016	251 ± 108	1.08 ± 0.17	1.12 ± 0.31	5409 ± 2999
VIM/VAR	0.018 ± 0.006	188 ± 28	0.98 ± 0.17	0.94 ± 0.28	5209 ± 1137
VIM/ESR	0.020 ± 0.009	269 ± 124	0.81 ± 0.15	0.86 ± 0.34	5368 ± 1873

Bar Specimens

Because of the interfering holes in and adjacent to the carbides, sections were cut from the top, middle, and bottom billet locations and press forged to 25-mm (1-in.) square bars. Six specimens were cut from each bar with the six spaced along the bar lengths. These specimens were split in half longitudinally, polished, and the inclusions were measured collectively using 200 fields per specimen with the Leitz TAS-Plus.

Tables 7 to 9 list the results for the 18 test specimens from the VIM/EBR, VIM/VAR, and VIM/ESR bars, respectively. Table 10 summarizes the averages and 95% confidence intervals for each remelt type bar as a function of billet location. The *t*-test was used to assess the statistical significances of the differences as a function of location. Table 11 summarizes the results of the *t*-test. This analysis reveals that the inclusion content in the VIM/EBR and VIM/VAR material was lowest at the top and the differences between the middle and bottom were not significant. The inclusion content of the VIM/ESR material was lowest in the bars from the middle of the ingot.

All of the bar measurements for each remelt type were averaged, and the results are presented in Table 12. These data were analyzed using the *t*-test to determine statistically

TABLE 7—Inclusion measurements of M-50 VIM/EBR bars (average values).

Location	Sample Number	A_A , %	N_A , mm^{-2}	$\bar{L}$, μm	$\bar{A}$, μm^2	$\bar{L}$, μm
Top	1	0.0051	54.2	1.01	0.94	21924
	2	0.0033	53.4	0.81	0.61	26058
	3	0.0023	16.8	1.37	1.38	53165
	4	0.0006	6.9	0.86	0.83	220261
	5	0.0015	12.4	1.14	1.23	86912
	6	0.0014	9.6	1.17	1.40	105124
Middle	1	0.0066	162.5	0.66	0.41	10255
	2	0.0075	117.5	0.79	0.64	11916
	3	0.0031	79.6	0.58	0.39	23479
	4	0.0018	49.0	0.57	0.36	42630
	5	0.0054	175.2	0.57	0.31	11254
	6	0.0037	31.7	0.86	1.16	35040
Bottom	1	0.0036	106.1	0.58	0.34	18139
	2	0.0051	88.1	0.72	0.58	18103
	3	0.0086	73.5	0.82	1.16	14209
	4	0.0059	163.3	0.59	0.36	11680
	5	0.0010	28.4	0.59	0.37	60071
	6	0.0027	64.5	0.66	0.42	27614

TABLE 8—*Inclusion measurements of M-50 VIM/VAR bars (average values).*

Location	Sample Number	A_A , %	N_A , mm^{-2}	L , μm	$\bar{A}$, μm^2	$\bar{L}$, μm
Top	1	0.0075	87.0	0.98	0.86	12706
	2	0.0030	41.0	0.85	0.73	27614
	3	0.0029	20.1	1.30	1.46	34389
	4	0.0026	34.4	1.06	0.77	23791
	5	0.0039	43.5	0.98	0.91	20603
	6	0.0039	38.8	1.13	1.00	20835
Middle	1	0.0090	166.4	0.81	0.54	7872
	2	0.0054	83.2	0.92	0.64	11226
	3	0.0128	189.0	0.83	0.68	6267
	4	0.0159	270.0	0.83	0.59	4863
	5	0.0066	64.2	1.11	1.04	12236
	6	0.0161	57.0	0.98	2.83	8142
Bottom	1	0.0157	323.1	0.74	0.49	4547
	2	0.0179	393.4	0.72	0.45	3911
	3	0.0056	45.5	1.12	1.24	19808
	4	0.0061	91.2	0.90	0.67	12848
	5	0.0071	62.0	0.99	1.14	14499
	6	0.0091	105.2	0.98	0.86	9401

TABLE 9—*Inclusion measurements of M-50 VIM/ESR bars (average values).*

Location	Sample Number	A_A , %	N_A , mm^{-2}	L , μm	$\bar{A}$, μm^2	$\bar{L}$, μm
Top	1	0.0064	26.7	1.50	2.40	21265
	2	0.0078	47.1	1.32	1.65	15678
	3	0.0077	39.1	1.28	1.98	16314
	4	0.0068	41.0	1.27	1.66	16577
	5	0.0121	34.2	1.82	3.55	15090
	6	0.0069	48.5	1.17	1.43	15976
Middle	1	0.0067	71.9	0.91	0.93	9532
	2	0.0028	20.9	1.26	1.34	35856
	3	0.0032	20.9	1.37	1.52	33396
	4	0.0062	44.0	1.22	1.41	15575
	5	0.0075	73.4	1.10	1.01	9337
	6	0.0038	21.5	1.30	1.75	31899
Bottom	1	0.0140	123.1	1.10	1.14	7940
	2	0.0097	46.0	1.31	2.10	14058
	3	0.0137	171.3	0.97	0.80	6747
	4	0.0085	74.4	1.02	1.14	12268
	5	0.0089	43.0	1.51	2.07	13524
	6	0.0058	36.1	1.26	1.62	20110

TABLE 10—Summary of bar inclusion measurements by location (average $\pm$ 95 CI).

Type	Location	A_A , %	N_A , mm^{-2}	$\bar{L}$, μm	$\bar{A}$, μm^2	$\bar{\lambda}$, μm
VIM/EBR	top	0.0024 ± 0.0017	25.6 ± 23.2	1.06 ± 0.22	1.07 ± 0.34	85574 ± 77371
	middle	0.0047 ± 0.0023	102.6 ± 62.1	0.67 ± 0.13	0.55 ± 0.34	22429 ± 14481
	bottom	0.0045 ± 0.0028	87.3 ± 47.7	0.66 ± 0.10	0.54 ± 0.33	24969 ± 18923
VIM/VAR	top	0.0040 ± 0.0019	44.1 ± 23.7	1.05 ± 0.16	0.96 ± 0.28	23323 ± 7676
	middle	0.0110 ± 0.0049	138.3 ± 88.9	0.91 ± 0.12	1.05 ± 0.93	8434 ± 2972
	bottom	0.0103 ± 0.0055	170.1 ± 156.3	0.91 ± 0.16	0.81 ± 0.35	10836 ± 6428
VIM/ESR	top	0.0080 ± 0.0022	39.4 ± 8.6	1.39 ± 0.25	2.11 ± 0.82	16817 ± 2351
	middle	0.0050 ± 0.0021	42.1 ± 26.5	1.19 ± 0.17	1.33 ± 0.33	22599 ± 13066
	bottom	0.0101 ± 0.0033	82.3 ± 56.8	1.20 ± 0.21	1.48 ± 0.56	12441 ± 5040

TABLE 11—Summary of statistically significant differences between test locations—bar measurements.

Remelt	Location Comparison	Level of Significance (<i>t</i> -Test)				
		A_A	N_A	$\bar{L}$	$\bar{A}$	$\bar{\lambda}$
VIM/EBR	top versus middle	90%	98%	99%	98%	90%
	top versus bottom	80%	98%	99%	98%	90%
	middle versus bottom	...	...	...	...	...
VIM/VAR	top versus middle	99%	95%	80%	...	99%
	top versus bottom	98%	90%	...	...	99%
	middle versus bottom	...	...	...	...	...
VIM/ESR	top versus middle	95%	...	80%	95%	...
	top versus bottom	...	90%	80%	80%	90%
	middle versus bottom	99%	80%	...	...	90%

TABLE 12—Bar inclusion measurements by remelt type (average $\pm$ 95 CI).

Type	$\bar{A}_A$, %	$\bar{N}_A$, mm^{-2}	$\bar{L}$, μm	$\bar{A}$, μm^2	$\bar{\lambda}$, μm
VIM/EBR	0.0038 ± 0.0012	71.8 ± 27.0	0.80 ± 0.12	0.72 ± 0.20	44324 ± 25610
VIM/VAR	0.0084 ± 0.0025	117.5 ± 54.1	0.96 ± 0.07	0.94 ± 0.27	14198 ± 4285
VIM/ESR	0.0077 ± 0.0016	54.6 ± 19.1	1.26 ± 0.11	1.64 ± 0.32	17286 ± 4225

TABLE 13—Summary of statistically significant differences between remelt types—bar measurements.

Comparison	Level of Significance (<i>t</i> -Test)				
	$\bar{A}_A$	$\bar{N}_A$	$\bar{L}$	$\bar{A}$	$\bar{\lambda}$
EBR versus VAR	99%	80%	95%	80%	95%
EBR versus ESR	99%	...	99%	99%	95%
VAR versus ESR	...	95%	99%	99%	...

TABLE 14—Bar inclusion measurements (top locations only) made on transverse specimens (average $\pm$ 95 CI).

Type	A_A , %	N_A , mm^{-2}	$\bar{L}$, μm	$\bar{A}$, μm^2	$\bar{\lambda}$, μm
VIM/EBR	0.0017 ± 0.0016	25.1 ± 30.0	0.85 ± 0.18	0.70 ± 0.46	58967 ± 55306
VIM/VAR	0.0050 ± 0.0120	104.3 ± 288.4	0.76 ± 0.30	0.62 ± 0.83	31621 ± 61733
VIM/ESR	0.0073 ± 0.0033	56.9 ± 41.7	1.16 ± 0.28	1.33 ± 0.52	17688 ± 11232

TOP LOCATION MEASUREMENTS (LONGITUDINAL PLANE) FROM TABLE 10					
VIM/EBR	0.0024 ± 0.0017	25.6 ± 23.2	1.06 ± 0.22	1.07 ± 0.34	85574 ± 77371
VIM/VAR	0.0040 ± 0.0019	44.1 ± 23.7	1.05 ± 0.16	0.96 ± 0.28	23323 ± 7676
VIM/ESR	0.0080 ± 0.0022	39.4 ± 8.6	1.39 ± 0.25	2.11 ± 0.82	16817 ± 2351

significant differences. Table 13 presents this information. This shows that the total inclusion area fraction of the VIM/EBR material is significantly lower than that of the VIM/VAR and VIM/ESR material. This, of course, agrees with the chemical analysis results which showed that the VIM/EBR material had the lowest combined oxygen and sulfur contents. The inclusions in the VIM/EBR bars are also smaller and more widely spaced. The area fractions of the inclusions in the VIM/ESR and VIM/VAR are not significantly different, but those in the VIM/ESR ingot are fewer in number but larger in size.

Because the inclusions in these materials were not highly elongated, a few measurements were made on transversely oriented bar samples to see how the measurements would compare to that from the longitudinally oriented specimens. Three transverse samples from the top locations of each remelt type were prepared and measured. Table 14 summarizes the results from the three transverse samples which are compared to the data from the six longitudinal specimens from the top locations. The 95% confidence intervals for the transverse data are larger than for the longitudinal data due to using only three specimens rather than six. However, we can see that the data are quite similar, particularly for the area fraction data. Because of the different sample orientation, we should expect higher N_A values and lower $\bar{L}$, $\bar{A}$, and $\bar{\lambda}$ values for the transverse specimens, and indeed, this is observed, although the confidence intervals for the $\bar{\lambda}$ values for the transverse planes are very high.

A comparison of the measurement of the inclusions in the as-cast specimen from the top of one of the electrodes (Table 3) with the longitudinal bar measurements from the top locations (Table 14) demonstrates the degree of refinement provided by these remelt procedures. The A_A and N_A values are reduced substantially by each process. The average inclusion sizes, except for the VIM/ESR material, are about the same.

Comparison of Calculated and Measured A_A Values

It has been shown [2,28,30,34,35,37,38] that chemical analysis results can be converted to area fractions which can be compared to measured area fraction values. In general, the correlation between measured and calculated A_A values for sulfides produces better agreement than for oxides. The relationship between the weight percent oxygen and the area fraction (%) of oxides is

$$\text{Wt\% O} = A_A \left[\frac{d_{\text{inclusion}}}{d_{\text{matrix}}} \cdot \frac{M_{\text{O}_2}}{M_{\text{inclusion}}} \right] \quad (4)$$

where

$d_{\text{inclusion}}$ = density of the inclusion,

d_{matrix} = density of the matrix,

M_{O_2} = weight of oxygen in a mole of inclusion, and

$M_{\text{inclusion}}$ = molar weight of the inclusion.

Values for the four defined parameters are known and can be grouped into a constant k . Hence, to calculate the area fraction in percent, we simply divide the weight percent of oxygen by k . If the density of the matrix is assumed to be 7.86 g/cm³ (iron) and the density of the inclusion (alumina, for example, is 4.0 g/cm³), the k value for Al₂O₃ will be 0.235. In like manner, we can calculate a k value for manganese sulfide (0.187). Because the density of M50 is 7.97 g/cm³, the k values have been recalculated yielding 0.236 for alumina and 0.185 for manganese sulfides. In this manner, the A_A values have been calculated for sulfides and oxides from the chemical analysis data and combined for comparison with our total measured A_A values. This information is presented in Table 15.

Inspection of the data reveals that the measured data are consistently lower than the calculated values (Figure 8). Several factors can account for these differences. There is undoubtedly some minor amount of solubility for sulfur and oxygen, and the calculations assume none at all. Metallographic examination revealed only a few sulfides in two of the 18 VIM/EBR specimens, but small quantities were visible in all of the VIM/ESR specimens. However, both analytical results indicated a slightly lower sulfur content in the VIM/ESR specimens. This could be due to a difference in sulfide size. According to Malm [37], A_A measurements are generally lower than the theoretical values because some of the inclusions are smaller than the resolution limit of the light microscope. Test measurements by Bergh [39] have shown that inclusions may be as small as 0.03 to 0.04 μm , a size that cannot be detected by light microscopy. The VIM/VAR samples did contain the most sulfides, when viewed optically, which agrees with the analysis results. This does show that such a correlation at extremely low inclusion levels is less precise than for higher oxygen and sulfur levels.

TABLE 15—Comparison of calculated and measured area fractions.

Type	Location	S, %	O, ppm	Calculated A_A , %			Measured A_A , %
				Sulfides	Oxides	Total	
VIM ^a	top	0.004	31.0	0.0216	0.0131	0.0347	0.0130
VIM/EBR	top	0.0022	8.5	0.0119	0.0036	0.0155	0.0024
	middle	0.0018	4.1	0.0097	0.0017	0.0114	0.0047
	bottom	0.0022	6.5	0.0119	0.0028	0.0147	0.0045
VIM/VAR	top	0.0035	4.6	0.0189	0.0019	0.0208	0.0040
	middle	0.0032	8.6	0.0173	0.0036	0.0209	0.0110
	bottom	0.0032	7.0	0.0173	0.0030	0.0203	0.0103
VIM/ESR	top	0.0018	19.2	0.0097	0.0081	0.0178	0.0080
	middle	0.0018	16.7	0.0097	0.0071	0.0168	0.0050
	bottom	0.0017	24.5	0.0092	0.0104	0.0196	0.0101
VIM/EBR	all	0.00207	6.35	0.0112	0.0027	0.0139	0.0038
VIM/VAR	all	0.0033	6.72	0.0178	0.0028	0.0206	0.0084
VIM/ESR	all	0.00177	20.13	0.0095	0.0085	0.0180	0.0077

^aAverage of three specimens from the top of one as-cast electrode.

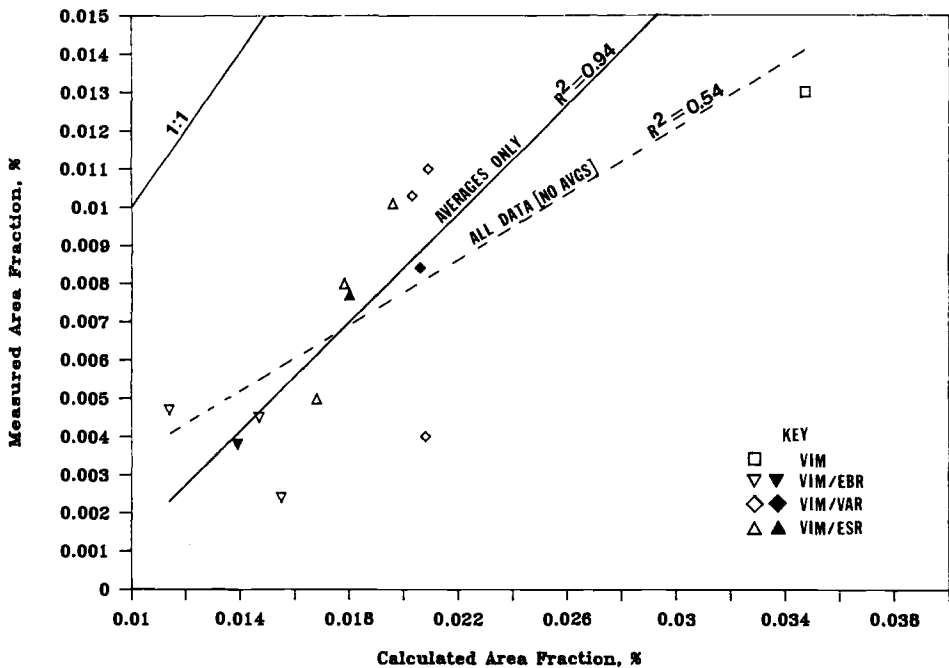


FIG. 8—Comparison of calculated and measured area fractions. Solid symbols are for averages of 18 specimens per remelt type (solid regression line) while open symbols are averages of six specimens per location by remelt type (dotted regression line) and as-cast VIM location (three specimens).

Influence of Measurement Area on %RA

Repetitive measurements were made on a bar specimen from the top of the VIM/VAR ingot (Specimen 1, $A_A = 0.0075\%$) to assess the influence of the number of fields measured, that is, the measurement area, on the relative accuracy of the area fraction measurement. Accordingly, 58 measurements were made with the $\times 50$ objective, and 52 measurements were made with the $\times 80$ objective with from 5 to 500 fields measured. Rather than using a constant value of 2.0 for t in the calculation procedure, we used the exact t value for each n field value. Figure 9 shows a plot of the measurement area (mm^2) as a function of the relative accuracy. The trend lines shown are for each set of data based on regression analysis (see below). Note that the data using the two different objectives, which are most appropriate for such an inclusion content, fall along a common path.

The $\times 50$ and $\times 80$ objective data were analyzed by linear regression analysis with the following results:

$\times 50$

$$\text{Log}(\%RA) = -0.431 \text{ Log}(\text{Area}) + 1.75 \quad (R^2 = 0.917)$$

$\times 80$

$$\text{Log}(\%RA) = -0.445 \text{ Log}(\text{Area}) + 1.66 \quad (R^2 = 0.890)$$

All data

$$\text{Log}(\%RA) = -0.415 \text{ Log}(\text{Area}) + 1.712 \quad (R^2 = 0.883)$$

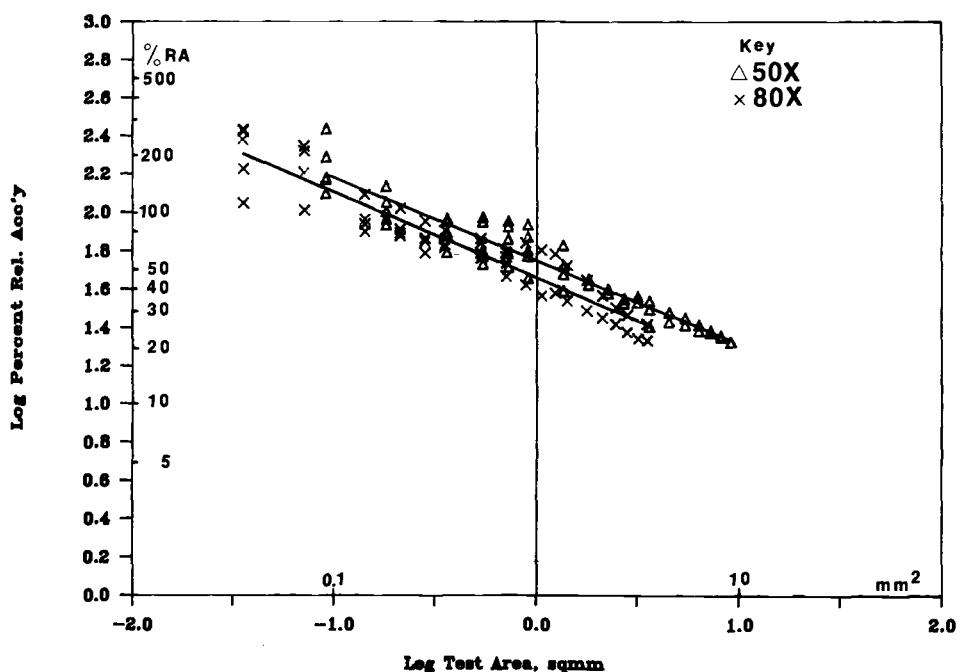


FIG. 9—Empirical relationship between total measurement test area and percent relative accuracy for a specimen with an area fraction of 0.0075% using $\times 50$ and $\times 80$ objectives.

TABLE 16—Estimates of number of fields required to obtain different relative accuracies.^a

Objective	50%	40%	30%	20%	10%	5%
$\times 50$	72	121	237	606	3029	15124
$\times 80$	115	190	363	904	4290	32449

^aArea fraction of test specimen was 0.0075%. Field area with $\times 50$ objective was 0.01815 mm². Field area with $\times 80$ objective was 0.00709 mm².

The $\times 50$ and $\times 80$ regression equations were used to provide estimates of the number of fields, based on the required total measurement area, to obtain various percent relative accuracies, as shown in Table 16.

Conclusions

This work has demonstrated application of the proposed ASTM stereological image analysis inclusion measurement procedure for characterizing the inclusion content, size, and spacing in M-50 bearing steel made by VIM/EBR, VIM/VAR, and VIM/ESR processes. The work has revealed the difficulties that are encountered when working with extremely low inclusion contents. The very low oxygen and sulfur contents in these steels clearly push image analysis and chemical analysis techniques to the limits of today's technology.

Very careful mechanical polishing procedures are required when characterizing such low inclusion levels. In addition, the polishing and measurement sequence must be closely sequenced to avoid problems with surface oxidation artifacts. A relatively high magnification

must be used to permit resolution of the small inclusions in these steels. When the $\times 50$ objective and 200 measurement fields per specimen were used, the relative accuracies for the A_A measurements were typically around 35%. The analytical work has demonstrated that achievement of a 10% relative accuracy, a typical target for stereological analysis, cannot be obtained realistically. Hence, to improve the statistical characterization of the inclusion content, it is better to accept a higher %RA value per specimen and measure a greater number of specimens.

The billet measurements revealed that the degree of hot reduction (80.1% for the VAR billet, 76.9% for the ESR billet, and 68.2% for the EBR billet) is very important. M-50, and other highly alloyed steels containing substantial quantities of carbides, must receive a greater amount of hot reduction to close the microshrinkage cavities present within the carbides and at the carbide/matrix interfaces. The bar specimens, which received from 99.5 to 99.1% reduction, were free of such problems. Only a few tight crack-like defects were observed. Improper polishing practices, however, can exaggerate these defects and create others. A polishing practice, which will usually produce acceptable quality surfaces, was described.

The chemical analysis results showed that the VIM/EBR process achieved the lowest combined oxygen and sulfur contents. The VIM/VAR process achieved the same oxygen level but a slightly higher sulfur level. VAR, of course, is noted for reducing the gas contents of steels substantially but not the sulfur content. The VIM/ESR process achieved the lowest sulfur level, but did not significantly reduce the oxygen level, as would be expected. The inclusion measurements agreed with the chemical analysis trends.

The measured inclusion contents were lower than the inclusion contents calculated from the chemical analyses. There are several possible reasons for this. First, there is undoubtedly some minor solubility of sulfur, and perhaps for oxygen, in M-50 at room temperature, although the calculation assumes there is none. Second, some of the inclusions are certainly in the submicroscopic size range that cannot be detected by light microscopy. The data suggest that the sulfides in the VIM/EBR material are smaller than those in the other steels. The average size of the inclusions was smallest in the VIM/EBR ingot, and visible sulfides were observed in only two of the eighteen specimens compared to all of the VIM/ESR specimens, even though the sulfur content of the VIM/ESR material was lower. Electron beam remelting has the potential to become an important process for producing high quality bearing steels. The ingot in this study was the first of this grade remelted by the EB process. An excessive carbide segregation condition and the unexpected loss of nearly all of the manganese precluded further processing of this material. Additional work is anticipated to exploit properly the potential of this process.

Acknowledgments

The author acknowledges the contributions of J. A. Leshner, who prepared the metallographic specimens, and D. G. Kauffman and W. A. Walthier, who performed the image analysis measurements. He is also grateful to the Carpenter Technology Corporation for permission to publish the results for the VIM/VAR and VIM/ESR material and to Pratt & Whitney for permission to publish the results for the VIM/EBR material.

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Inclusion Assessment in Steel Using the New Jernkontoret Inclusion Chart II for Quantitative Measurements

REFERENCE: Johansson, S., "Inclusion Assessment in Steel Using the New Jernkontoret Inclusion Chart II for Quantitative Measurements," *Effect of Steel Manufacturing Processes on the Quality of Bearing Steels*, ASTM STP 987, J. J. C. Hoo, Ed., American Society for Testing and Materials, Philadelphia, 1988, pp. 250-259.

ABSTRACT: The new Swedish standard SS 11116 for assessing the content of nonmetallic inclusions in steel is a manual image analysis method that uses a comparative picture chart for describing the content of inclusions in a microscopic field of view. With the aid of this new picture chart, it is possible to obtain the same results as with optical image analysis regarding basic parameters such as area percent, length and numbers of inclusions per unit area. The frequency distribution of the inclusion sizes can also be obtained. SS 11116 is a tool for inclusion assessment that retains the advantages of both the old comparative methods (ASTM E 45-85 and SEP 1570) and optical image analysis, but reduces the disadvantages to a minimum.

KEY WORDS: inclusion assessment, clean steel, bearing steel, image analysis, comparative method, size distribution, frequency distribution, inclusion size

To determine the inclusion content of a clean steel is a difficult problem. When the oxygen content is low, the frequency of oxides is also very low. Still, we do have nonmetallic inclusions which can be critical for the product. An inclusion assessment using the traditional comparative charts does not give information about the presence of critical inclusions. The low level of nonmetallic inclusions also means that sample preparation defects will have a proportionally greater influence on the results compared with a case when the amount of inclusions is higher.

In optical image analysis, the sample preparation defects and other artifacts cannot be separated from inclusions. This problem can only be obviated by very careful specimen preparation; automatic preparation methods are a necessity. Of course, it is possible to use the editor on each field of view to remove artifacts, but this will slow down the analysis so much that it will no longer be a practical method.

At Sandvik Steel, we have been working with optical image analysis since the middle of the 1960s and have gained considerable knowledge from this technique. From this knowledge, we went further and created an inclusion assessment method that combines the best from the traditional comparative methods and the optical image analysis, Fig. 1. This method is an inclusion counting method using comparative pictures, Fig. 2, based on a new principle for describing the inclusion content in the steel. The principle for this description has been reported elsewhere [1], and the new method has become a Swedish standard, SS 11116.² It has proven to be a very reliable method especially for evaluating clean steels, Sandvik Electronics has used this new method in the specially designed instrument, SANTRON Morphologic Inclusion Classifier (MIC).

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²MNC, Box 4089, S-102 61 Stockholm, Sweden.

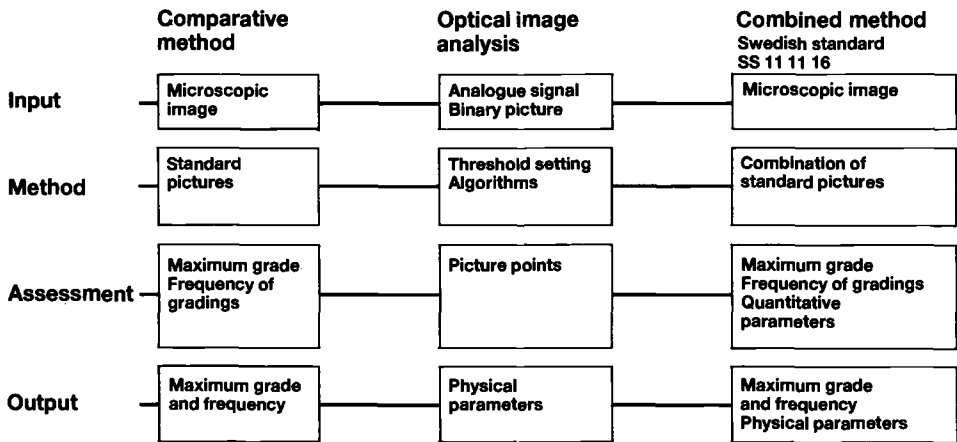


FIG. 1—Method structure for comparative method, optical image analysis, and SS 111116.

Critical Inclusions

What is Important to Know About Steel Inclusions?

For every steel application, there is a critical inclusion size. This inclusion can be of different composition, shape, and size, dependent on the type of application. Figure 3 shows some of the most common inclusion shapes found in modern steels. It must be possible to recognize and record the different morphological types by the method of determination in order to separate harmful inclusions from others. The high quality of bearing steel is very dependent on the absence of some types of inclusions, while other products may not suffer from these specific types. More than one type of inclusion is harmful to bearings; for example, highly elongated manganese silicates, broken stringers of aluminates, and large globular oxides are known to be detrimental to bearings. If the harmful types are present, what is the probability that the size of any individual inclusion is big enough to be a critical inclusion? The word *probability* does indicate that the critical inclusion is a very rare object.

If it were easy to find a critical inclusion, the steel producer would not bother to sell that product because of future customer complaints. The inclusions that normally are observed in the analysis are not the critical inclusions. However, the distribution of sizes can be used in order to make a qualified guess about the critical ones. Figure 4 compares two products. One product has a very high number of inclusions and a rapid decrease in number with increasing size. The other product seems to have a better inclusion level with a much lower total number, but the decrease in number with increasing size indicates that the probability of larger inclusions is higher than for the first product.

Bearing-Quality Inclusion Assessment

Normally, a bearing-quality steel is assessed using the ASTM Practice for Determining the Inclusion Content of Steel (E 45-85) Method D or Mikroskopische Prüfung von Edelstählen auf nichtmetallische Einschlüsse mit Bildreihen (DIN 50602, Stahl Eisen Prüfblatt [SEP] 1570). The result can be presented either as a maximum rating number or as a calculated value from fields with a certain amount of inclusions. The SAM rating system from the ASTM Specification for High-Carbon Ball and Roller Bearing Steel (A 295-79) is one such calculation

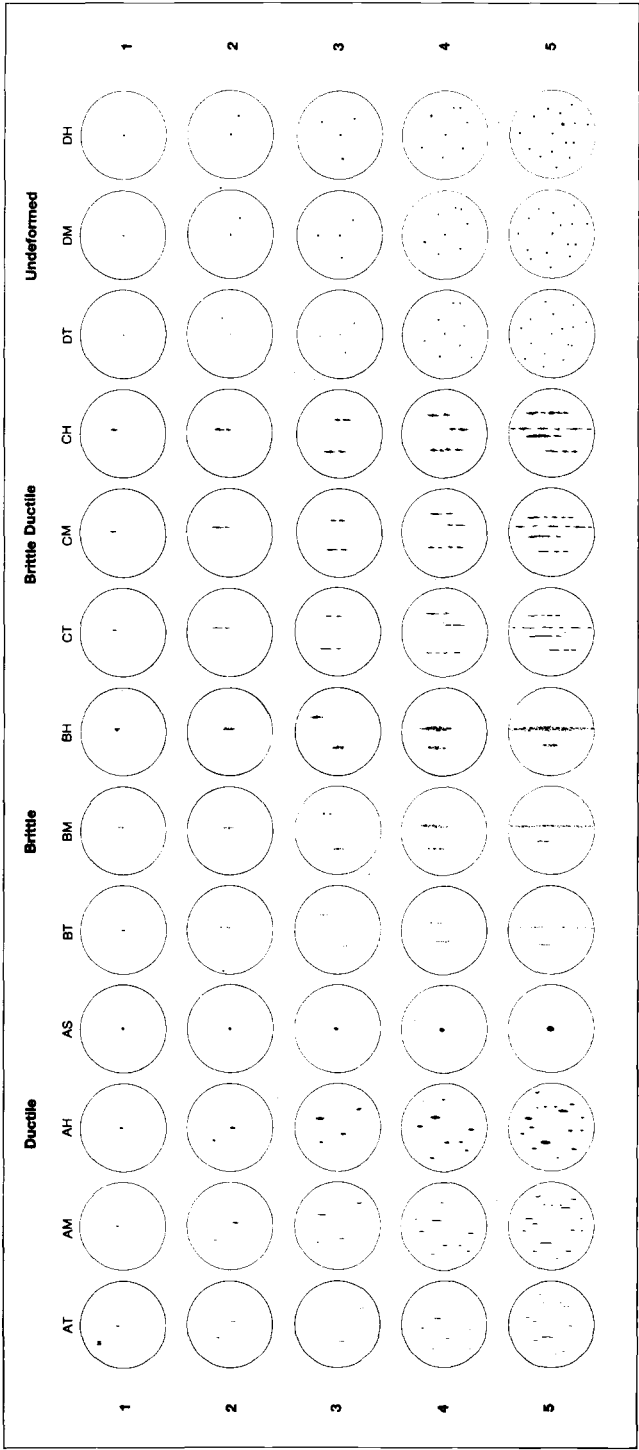


FIG. 2—The new Jernkontoret inclusion chart II.

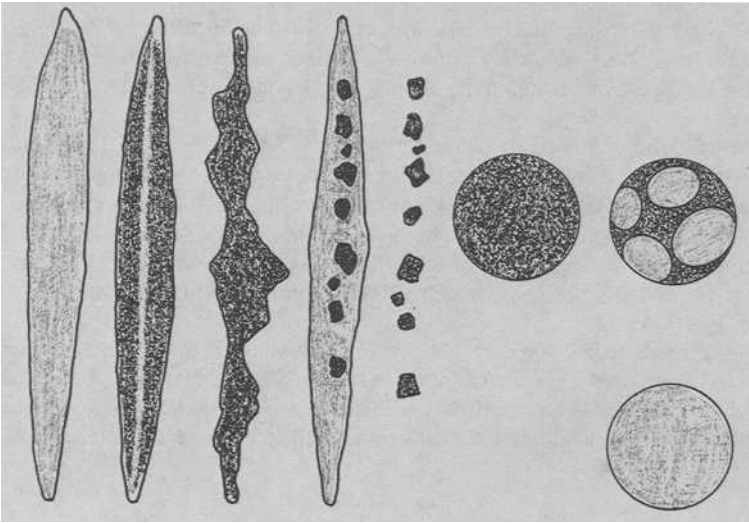


FIG. 3—Common inclusion shapes. Which one is harmful?

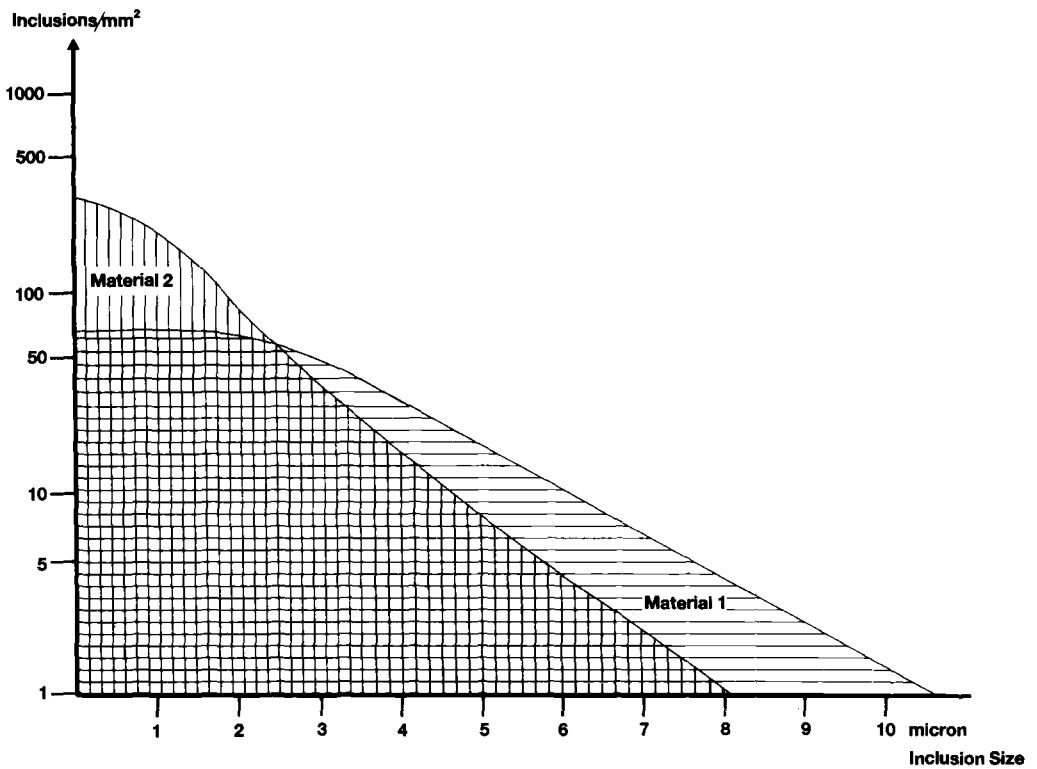


FIG. 4—Different inclusion size distributions. What is the probability of having a critical one?

using the ASTM E 45-85 Method D inclusion chart for classification of B-type and D-type inclusions. In DIN 50602, a $K1$ or $K4$ value can be calculated from the frequency of fields with inclusion content equal or greater than picture row 1 and 4, respectively, on the SEP 1570 picture chart.

Figures 5 and 6 show the results of an ASTM E 45-85 and a SEP 1570 assessment of two specimens with different contents of nonmetallic inclusions. If a SAM rating value is calculated from the ASTM E 45-85 table and a $K1$ value from the SEP 1570 table, we can establish that there is no difference between the two samples concerning B-type inclusions, Fig. 5, that there is a small increase in Sample 2 of D-type inclusions calculated as SEP 1570 $K1$ value, and that the difference for D-type inclusions is much more pronounced in the calculation of a SAM rating value, Fig. 6.

If we use the new Swedish standard SS 111116 to establish if there is a difference between Specimen 1 and Specimen 2, we can see also from Figs. 5 and 6 that there is such a difference in both the distribution of B-type and D-type inclusions. The difference in area percent distribution is only one of many parameters that can be used to indicate a difference using the new Swedish method.

Different Production of Carbon Steel

This is another case where the new method proved its usefulness. One-percent carbon steel was produced in three different ways by ladle refining (LR), by electroslag remelting (ESR), and by a converter argon oxygen decarburization (AOD) process with special treatment for modification of the inclusion size distribution. Table 1 shows the total recorded amount of inclusions in 50 fields from the edge and 50 fields from the middle position in each of the three different samples. These inclusions were recorded at a very high microscopic magnification, and the result was calculated using the new Swedish standard. As can be seen from this table, LR gave a high amount of small inclusions, whereas both AOD and ESR gave a smaller number totally.

The cumulative size distribution of inclusions in Fig. 7, however, shows that there are many more large inclusions in ESR and AOD steels compared with LR. Even if the total for the LR sample in Fig. 6 looked worse than AOD and ESR, after studying the size distribution it is no longer quite as obvious that it is worse.

The new Swedish standard is a tool for assessment of the inclusion content in clean steels, and it extends into the range of inclusion contents where no other comparative method can give information and where sampling preparation defects become detrimental for the result obtained by optical image analysis. If the level of inclusion contents, on the other hand, is in the range where traditional methods can be used, the Swedish standard can also be used in order to obtain much more information about the inclusions.

TABLE 1—Amount of oxide inclusions in the same grade of steel processed in different ways. $\times 950$ magnification.

Process	Area, %	Length of Inclusions, mm/mm ²	Number of Inclusions, mm ²
LR	0.031	0.437	587
AOD	0.012	0.143	170
ESR	0.011	0.106	121

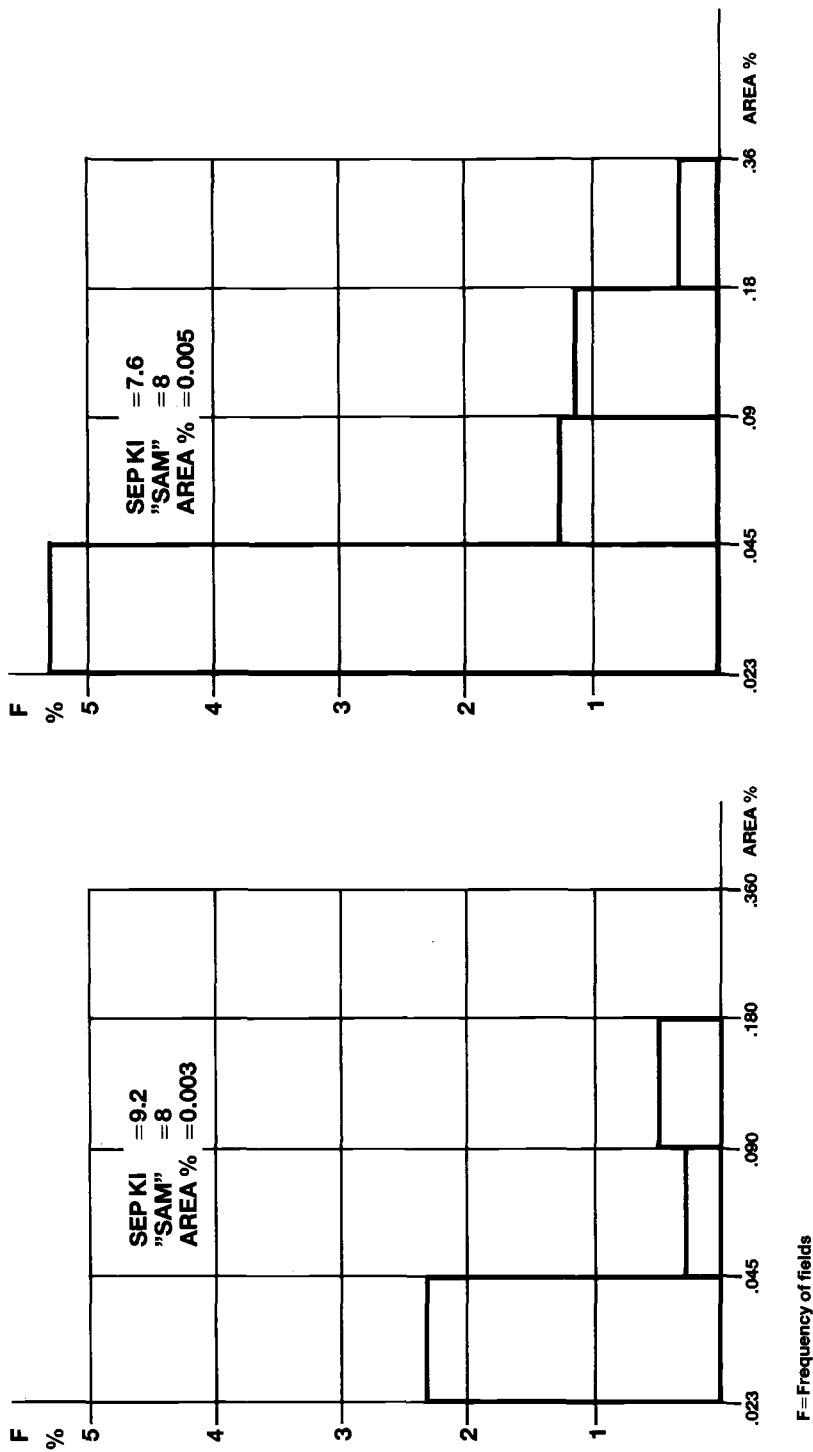


FIG. 5—Two steel samples with different amount of B-type stringers.

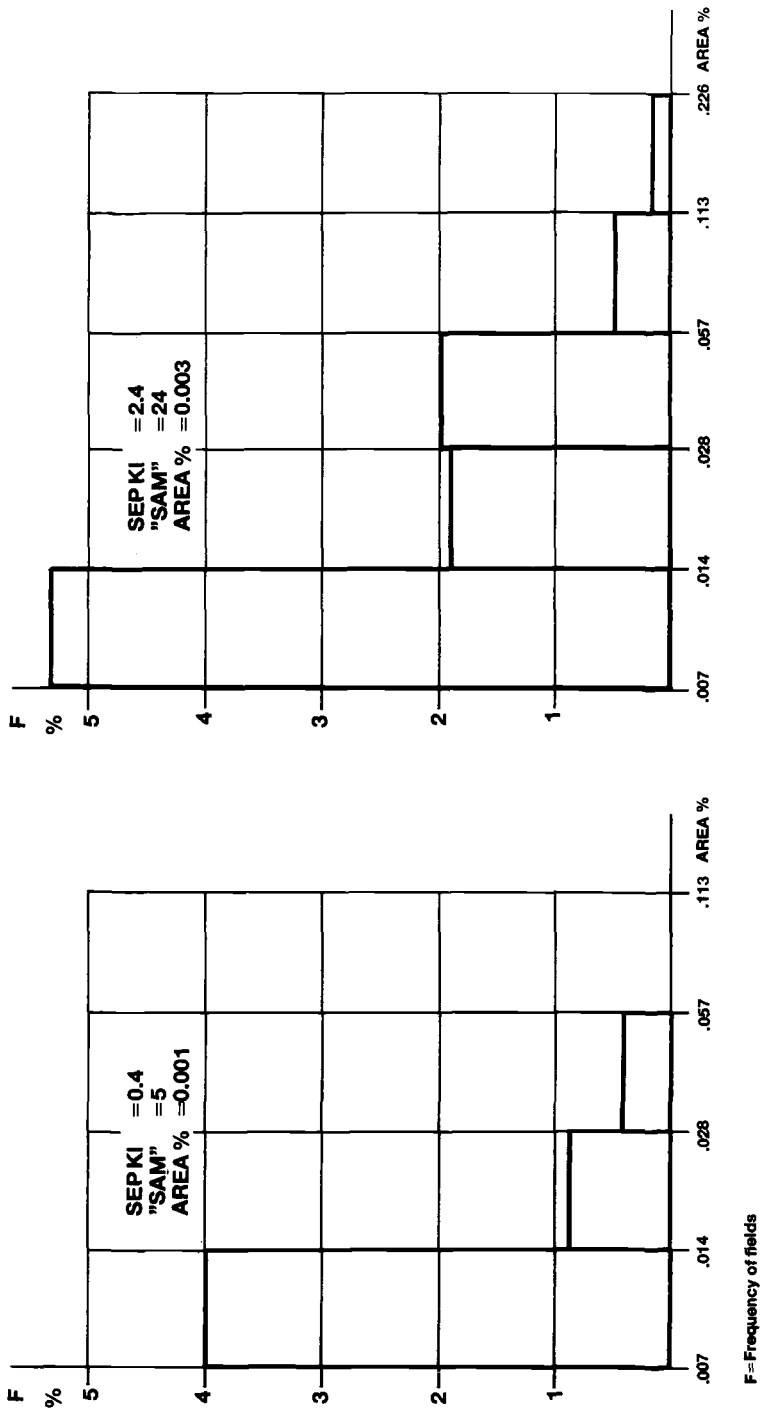


FIG. 6—The same steel samples as in Fig. 5, showing different amount of randomly distributed D-type inclusions.

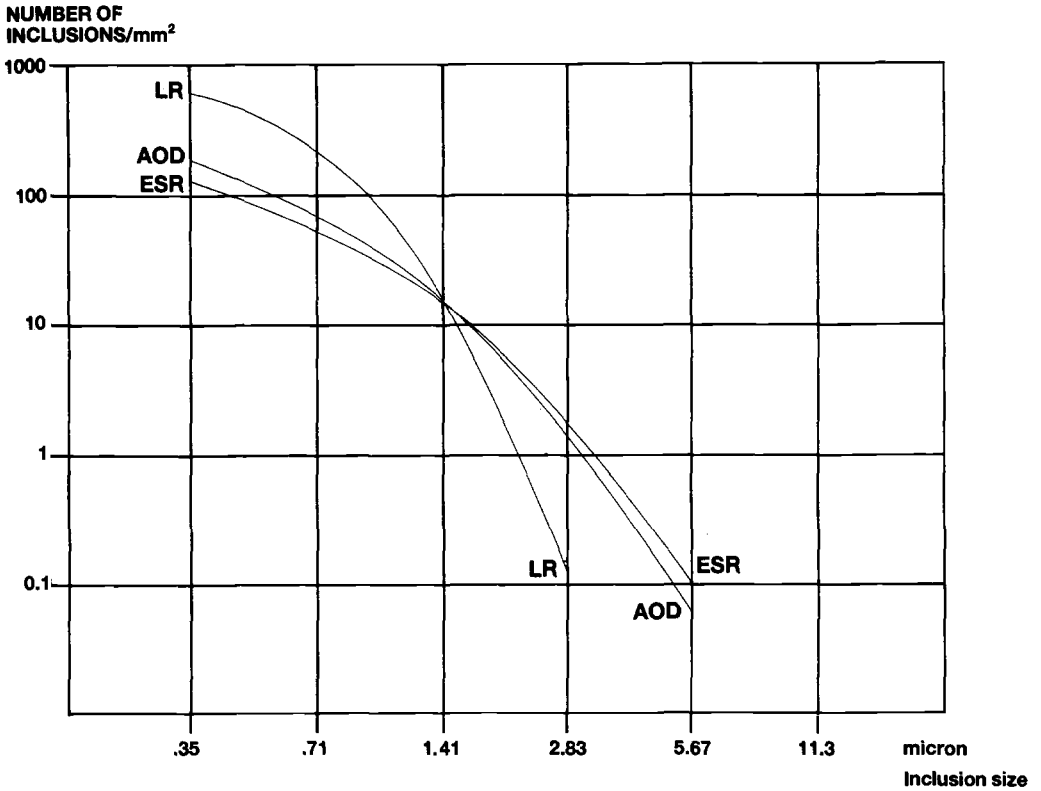


FIG. 7—Cumulative size distribution for oxide inclusions in steels processed differently.

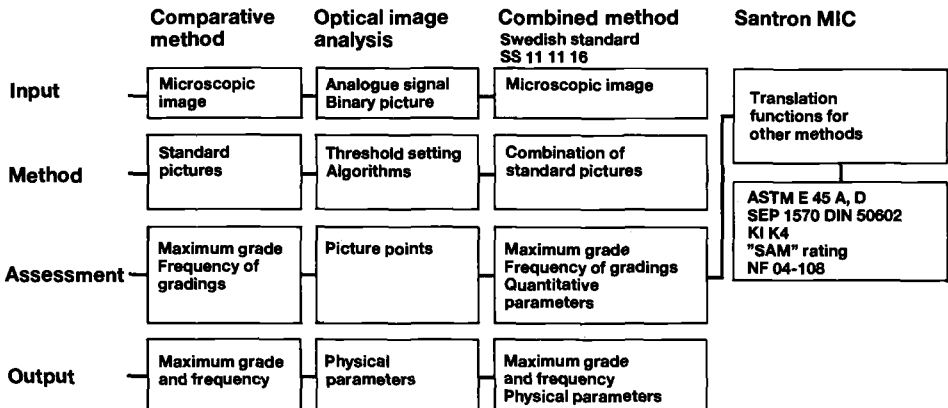


FIG. 8—Method structure for SANTRON MIC.

In order to make the changeover from traditional methods to this new one easier, we have made extensive measures to sort out the traditional rating numbers for ASTM E 45-85 and SEP 1570 from all the information; and, if customers still want an ASTM E 45-85 frequency table, they can get it. The translation to ASTM E 45-85 is made according to the recommendation for obtaining a JK count from optical image analysis [2]. Figure 8 shows which translations into other standard methods are made from the combinations of standard pictures in the new SANTRON MIC.

Productivity Effects

Quality control has to be carried out quickly; it is not possible to spend hours with each sample. In order to reach the stage where the quality control can be carried out quickly and safely, you must gain some knowledge about your process or your product or both.

Figure 9 shows the relationship between the return in knowledge and the time for the assessment. The traditional comparative chart can be used and gives a low level of knowledge in a reasonably short time. If optical image analysis is used, the level of knowledge can be much higher, but the time needed to reach this level may be much longer. The Swedish standard SS 11116, in combination with a powerful software program, can give a higher level of knowledge than optical image analysis, and the analyzing time is still not more than for the traditional methods.

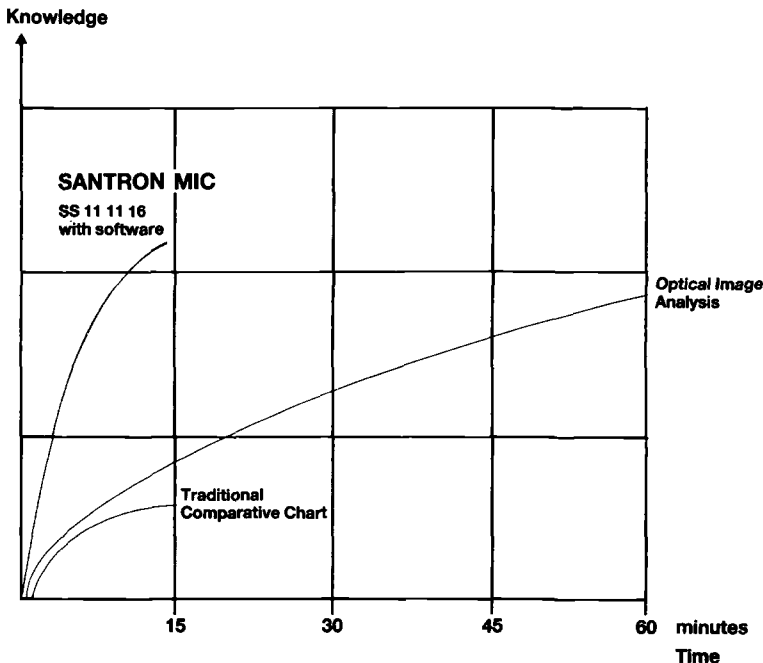


FIG. 9—Output in knowledge related to analyzing time.

Conclusion

The advantages in using SS 111116 are most significant when assessing clean steel. The traditional comparative charts have no pictures for differentiating between a clean and a very clean steel. In most cases, the result will be a zero. Artifacts from a poor specimen preparation will increase the value for oxide and sulfide or both when using optical image analysis. For a clean steel, the contribution from artifacts can be more than from inclusions.

The new Swedish standard does not have these drawbacks; it is a comparative method that has pictures for a single small particle of any size. It is also an image analysis which gives quantitative parameters not influenced by artifacts. The nature of the inclusion size distribution can easily be determined and the probability for critical inclusion sizes predicted. Results can be presented as a corresponding ASTM E 45-85 or a SEP 1570 rating system. It can also be converted into a frequency of pictures according to the AFNOR Méthode micrographique de caractérisation de la forme des sulfures à l'aide d'image-types (A 04-108) for modified sulfide shape. These translations make it possible to compare results collected over the years using traditional comparative methods with results obtained by using this new method.

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New Processes to Produce Better Bearing Steels

The Ladle Refining Process for Bearing-Quality Steels

REFERENCE: Odar, J. A. and Fecich, D. J., "The Ladle Refining Process for Bearing-Quality Steels," *Effect of Steel Manufacturing Processes on the Quality of Bearing Steels*, ASTM STP 987, J. J. C. Hoo, Ed., American Society for Testing and Materials, Philadelphia, 1988, pp. 263–277.

ABSTRACT: Bearing-quality steels that have been previously refined by the vacuum degas carbon deoxidation process for microcleanliness considerations are now able to achieve lower levels of oxygen and inclusion content by ladle refining. Through aluminum deoxidation in the presence of strong stirring energy, inclusions are removed by mechanical rather than chemical means. Such refining techniques are provided by ladle reheat capability, slag control, de-slagging, stirring, alloy and deoxidizer injection, and vacuum degassing, further enhanced with bottom pouring, shrouding, and casting.

Several ladle refining options were studied to evaluate the relative cleanliness levels that the processes were capable of producing. Superior levels of steel quality and chemistry control were achieved by these out-of-the furnace techniques in terms of microcleanliness as measured by the JK-worst field and SAM ratings, and of macrocleanliness as measured by magnetic particle inspection procedures of AMS 2301.

KEY WORDS: inclusions, vacuum degassing, ratings, slags, deoxidation, aluminum oxides, sulfides, refining

The bearing industry has long required steels of the utmost "clean" quality levels. These levels were always the best that steelmakers could provide on a production basis. The steel was frequently melted by the basic open-hearth (single-slag) or electric-furnace (double-slag) refining practices. With vacuum degassing-carbon deoxidation (VDCD), lower oxygen and cleaner steels could be achieved. Then the industry required stricter cleanliness limits of non-metallic inclusions as measured by the ASTM Practice for Determining the Inclusion Content of Steel (E 45), JK-worst field rating system (Table 1). These demands were necessitated by higher performance engines and lighter weight vehicle components. With the vacuum degas processes, such as ladle, stream, RH, or DH, a large step was taken toward ladle-refined clean steel as we know it today. This form of ladle metallurgy provided low oxygen, 10 to 25 ppm, and low inclusion contents via the carbon-oxygen reaction.

For 20 years VDCD was, indeed, the basis for clean steel. Aircraft and bearing manufacturers came to rely on the process, and many specifications were written around it. However, the best clean steel practice soon became obsolete. During the last ten years improvements and technology have dramatically changed steel refining with out-of-the-furnace ladle metallurgy treatments.

The chief factors to address in the clean steel issue are oxygen and sulfur in the form of oxides and sulfides—aluminum oxide (Al_2O_3), silicon dioxide (SiO_2), calcium oxide, manganese sulfide, etc. The size, shape, and frequency of these inclusions are very significant. All steels contain these inclusions in varying degrees. However, steel is generally considered

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TABLE 1—Standard microcleanliness limits, JK E 45 worst field.

Steel	A		B		C		D	
	T	H	T	H	T	H	T	H
ASTM A 534								
OH	3.5	2.5	3.5	3.0	3.0	2.0	2.5	2.0
EF	3.0	2.0	3.0	2.5	2.5	1.5	2.5	1.5
A 295-79	2.5	1.5	2.0	1.5	2.0	1.5	1.5	1.5
A 295-84	2.5	1.5	2.0	1.0	0.5	0.5	1.5	1.0
A 535-79	1.5	1.0	1.5	1.0	1.5	1.0	1.5	1.0
CLEAN STEEL MICROCLEANLINESS LIMITS								
LRS LIMIT	2.5	1.5	2.0	1.0	0	0	1.0	0.5

“clean” with probably less than 20 ppm oxygen and 0.015% sulfur levels. Data seem to indicate that bearing fatigue life is better at 10 ppm oxygen than at 20 ppm [1,2]. However, can oxygen really be accurately and reproducibly measured within less than 5 or 10 ppm?

Hence, it must be our goal to optimize properties and quality—fitness for use—not necessarily to shoot for elemental levels which are so low as to be prohibitively expensive to attain, analytically questionable, and technically unjustifiable.

Another feature of ladle refining in the processing of bearing steels is chemistry control. Consistency of chemistry and hardenability are of great importance in the highly automated production lines of the automotive industry.

With the presence of ladle refining throughout the world, the traditional and more costly techniques of refining become less attractive. Economic and quality benefits are realized by these new processes.

At the LTV-Canton Bar Company, clean steel can be produced by ladle refining and pouring techniques with such elements as:

1. Ladle reheat capability.
2. Slag control.
3. De-slagging.
4. Porous plug-argon bubble.
5. Desulfurization.
6. Wire feed—powder injection.
7. Ladle load cells—chemistry control.
8. Oxygen control ability.
9. Vacuum degassing.
10. High-purity ladle refractories.
11. Bottom-pour technology.
12. Ladle stream shrouding.
13. Advanced caster technology.

With the use of these refining techniques, as well as bottom pouring, shrouding, and casting, world-class clean bearing-quality steel with precision hardenability can be achieved.

In order for us to examine the ladle refining process, each of these features will be described and their contribution will be discussed. The quality results that can be achieved through the use of these options will then be examined.

Ladle Refining Technology Features

The Ladle Refining Station (LRS) at the LTV Canton, Ohio, Works began operation on 7 March 1986. The station was constructed by dismantling the shell of a 220-tonne (240 ton) electric arc furnace and installing a ladle stand and cover in its place. Electrical power to reheat the steel in the ladle came from existing transformers that were used to power the furnace.

The ladle refining station is designed to allow the remaining three 220-tonne (240 ton) electric arc furnaces to do the job that they are designed to do best—melt down scrap. All alloy additions and refining of the steel now take place in the ladle, rather than in the furnace. But while the steel is being refined in the ladle, the electric furnace is busy melting down its next heat. With furnace tap-to-tap times from 3 to 3½ hours and LRS refining times under 1 hour, this duplex operation can dramatically improve a melt shop's productivity and has the added advantage of making better-quality steels at the same time.

There are a number of features at the LRS station that enable us to produce an *engineered, clean, bearing-quality steel* uniquely suited to a customer's specific application. First of all, the ladle sits on a stand that is equipped with load cells. These load cells enable us to determine the actual hot metal weight. Knowing the true hot metal weight enables the melter to calculate accurately alloy additions in order to work to tighter chemistry ranges.

A second major feature at the LRS is the use of two porous plugs in the bottom of the ladle to inject argon into the steel. The argon is used to continuously stir the steel in the ladle to improve temperature and chemistry uniformity. Not only does this tandem, two-plug operation stir in alloy additions, but it plays a major role in producing steels with improved cleanliness.

Also included at the LRS is a three-strand wire-feeding operation. Currently the wire feeder is used to add pure aluminum wire to the steel and cored wires containing calcium-silicon alloy and pure sulfur. Wire feeding of these alloys creates a "deep injection" system whereby the wire dissolves deep in the ladle and is stirred in by the continuous argon bubble. This "deep injection" of alloys prevents any slag-alloy reactions and allows virtually 100% recovery of these reactive or volatile elements. By counting on 100% alloy recovery of these wires, and by knowing the true heat weight of the steel in the ladle, the melter can maintain very close control of aluminum and sulfur contents.

Now, let us return to the beginning of the steelmaking operation for an electric furnace shop.

Chemistry control at the LTV Canton Works begins with proper scrap selection. Steel scrap is the principal source of minor "residual" elements found in any given heat of steel, and control of these residuals is by scrap charge design.

Once the scrap has been completely melted down and the carbon content of the steel adjusted to the desired level, the steel is tapped out of the furnace and into the ladle. At this point, major alloy additions of manganese, chromium, silicon, nickel, or molybdenum are added to the ladle to take advantage of the mixing action of the tap stream. A certain amount of furnace slag is also carried over into the ladle. Although this slag is very oxidizing, it is also very hot and rich in lime. It is to our advantage to use this slag as part of the LRS refining step.

When the heat is tapped and transferred to the LRS, the slag depth is measured using a "dip-stick" type of burn-off rod. To maximize heating efficiency, it is best to bury the arc in about 12 cm (5 in.) of slag. Excess furnace slag can be removed from the ladle by raking. If there is not enough slag, then bulk slag additions—lime, spar, dolomite, and aluminum—are added to create a slag. In any case, the slag is always treated with deoxidizers to turn it from an oxidizing slag to a reducing slag. This is a key step in producing "clean steel" with low-oxide inclusion contents.

Now that the slag has been turned into a cleansing slag, the stirring action of argon bubbling becomes extremely important to assist flotation of nonmetallic inclusions that are formed as a part of the oxygen-steelmaking process. Oxide nonmetallics, such as aluminates and silicates, are always less dense than liquid steel, and therefore, they will float out of the steel and eventually become a part of the slag. However, the relative buoyancy of a nonmetallic is dependent on the size of the particle; that is, larger particles will float to the surface faster than a "cloud" of small particles. Bubbling the steel with argon through two porous plugs adds a stirring energy to the steel that sweeps the tiny particles to the surface and into the slag. The refining slag dissolves these inclusions and helps to form a protective layer on top of the steel to prevent any reoxidation. Also, because the ladle is covered, the spent argon collects to form an inert atmosphere above the slag to further protect the heat of steel from air contamination.

In order for us to further refine the steel to meet bearing- and aircraft-quality standards, the steel can be sent on a transfer car into the teeming aisle, where the entire heat of steel can be placed into a Stokes vacuum chamber for degassing at pressures as low as 0.2 torr. During the vacuum cycle (20 to 30 min), the dual argon porous plugs are continuously stirring the steel, exposing the surface of the steel to the vacuum above in order to remove dissolved gases. Furthermore, argon stirring under vacuum has the effect of greatly multiplying the stirring energy of the argon bubbles. This high level of stirring energy improves steel cleanliness, while the vacuum of the chamber prevents reoxidation of the "boiling" steel in the ladle.

A further practice refinement for producing superior-quality bearing steels involves the use of an oxygen probe at the LRS to maintain the dissolved oxygen content at specified levels prior to vacuum treatment. While under vacuum, the dissolved oxygen reacts with carbon to form carbon monoxide gas, which is easily removed from the ladle. Again, argon bubbling assists this reaction until the dissolved carbon-oxygen content reaches equilibrium with the vacuum above the steel and the boiling ceases. At this time, silicon and aluminum can be added to scavenge the remaining dissolved oxygen, and the few resulting nonmetallics can be floated up into the slag.

During the last ten years, many significant advances have been made in steelmaking operations to improve greatly the overall cleanliness of steel. But these improvements are of no use at all if the steel is allowed to come into contact with air during either the teeming or casting operations. Technologies have developed superior shrouding systems for protection of pouring streams in top pouring, bottom pouring, and continuous casting of steels at the LTV Canton Works.

Top-pour shrouding is accomplished by a shield at the bottom of the ladle that is lowered to the top of the mold so that it completely covers the mold. The system is then flooded with argon gas at a positive pressure to prevent air infiltration.

Bottom-pour shrouding is much more easily accomplished. The ladle nozzle is pressed against a kaowool ring at the top of the downsprue trumpet. The entire trumpet system is then pressurized with a low flow rate of argon gas. The rising steel in the molds is further protected from reoxidation by a synthetic slag that is placed inside each mold.

Shrouding at the bloom caster is accomplished with ceramic tubes. From the ladle to the tundish, a preheated alumina-graphite tube is attached to protect the stream. A "can" type shroud is placed around the joint and flooded with argon. This completely seals the pouring stream against air contamination. The steel in the tundish is then covered with a synthetic slag that is both reducing and insulating. This slag protects the steel from air oxidation and helps to prevent temperature losses in the tundish. From the tundish to the mold, a fused silica shroud protects the steel. The shroud joint is sealed with mortar, and the steel in the mold is protected with a synthetic slag.

Because the physical description of the features of the LRS has now been provided, it should be noted that there are both quality and operating advantages that are achieved:

1. Internal quality of microinclusion: size and frequency.
2. Control of gases: hydrogen, oxygen, and nitrogen.
3. Low sulfur <0.005%.
4. Temperature control: casting, degassing, bottom pouring.
5. Ability to meet restricted chemistry and hardenability.
6. Control of deoxidation.
7. Increase productivity rates and reduced costs.

Evaluation Program

With the start-up of the LTV Canton LRS, it was desirable to evaluate all the practice options that were available in order to provide an engineered clean steel, designed to suit the specific needs of the customer and applications by the techniques described. The options that were evaluated included:

1. Ladle refined steel (LRS) with a minimum prescribed argon stir.
2. LRS with a vacuum treatment.
3. LRS-clean steel practice (clean steel practice—optimum hot top and ingot design, bottom pour, shrouding, or strandcast techniques).
4. LRS-VDCD combination-clean steel.
5. LRS-cast.

Each category represents steel grades with carbon levels of 0.20 to 0.70%. With the ladle refining mechanism of precipitation deoxidation, removal of oxygen is no longer driven by the chemical reaction of carbon and oxygen, but by mechanical stirring energy. For this reason distinctions between high and low-carbon grades were not made.

The quality of the heats was evaluated by the standard techniques of ASTM Practice for Determining the Inclusion Content of Steel (E 45) JK Method A worst-field, presenting the data for B-alumina type stringers and D-globular type oxides. The SAM technique as described in ASTM Specification for High-Carbon Ball and Roller Bearing Steel (A 295-84) was used to further define the quality levels attained. Magnetic particle testing, as described by AMS 2301, was also performed converting the inclusion ratings into units of mm/m². These qualification tests were performed at the usually prescribed top and bottom ingot positions of the first, middle, and last portions of the heat. In the case of strandcast material, tests were taken from the front, middle, and back positions of an inside and outside strand. Oxygen was analyzed from samples taken at the same positions using a LECO oxygen-nitrogen analyzer.

Results and Discussion

The data of Table 2 show the cleanliness results of historical traditional practices available at LTV Canton, with the open-hearth-single-slag practice (high and low carbon); vacuum degas carbon deoxidation practice-bearing quality/aircraft quality (high and low carbon); a VDCD practice with clean steel considerations of hot topping, ingot design, bottom pouring, and shrouding; and strandcast. Of these variations, the VDCD practice with clean steel provisions has always been very effective and thorough in providing a quality product.

The JK-worst field data show few differences in quality of B- and D-type inclusion ratings among the several traditional processes. Although the quality between open hearth and VDCD

TABLE 2—Standard steel quality, E 45-JK inclusion rating.

Practice	Worst Field				SAM B	SAM D	AMS 2301, mm/m ²	Oxygen, ppm
	B-alumina		D-globular					
	T	H	T	H				
OH $\bar{X}$,LoC	1.8	0.3	1.0	0.3	> 40	> 40	600	30
VDCD $\bar{X}$,LoC	1.6	0.3	1.0	0.4	12.4	8.9	350	18
OH $\bar{X}$,HiC	1.6	0.4	1.0	0.2	> 40	> 40	500	25
VDCD $\bar{X}$,HiC	1.6	0.2	1.0	0.2	4.0	4.5	240	13
VDCD $\bar{X}$ clean steel	1.2	0.1	1.0	0.4	3.5	2.5	10	12
Strand $\bar{X}$ cast LoC	1.0	0.2	1.1	0.2	3.9	2.0	350	18
Strand $\bar{X}$ cast HiC	0.9	0.2	1.0	0.1	3.9	2.5	145	16

are approximately equivalent as measured by the JK, it is apparent by the SAM B and SAM D (ASTM A 295-84) ratings that the VDCD process is superior. The open hearth quality evaluations are largely incalculable, with SAM B and D being estimated in excess of 40. The AMS 2301 macroinclusions ratings and oxygen levels show similar improvements with vacuum treatment. Further enhancements are realized by the use of clean steel pouring practices. Here great improvements are attained particularly in levels of the macroinclusions of AMS 2301. The strandcast product shows levels of microinclusions equivalent to clean steel quality as measured by the JK and SAM evaluations. However, with respect to macroinclusion ratings, comparative improvements are not fully realized. It should be pointed out that the casting operation, protected through elaborate shrouding systems, contains inherent sources of inclusions that are peculiar to that process and are not fully addressed by shroud protection from atmospheric reoxidation.

The data of Table 3 immediately reveal that bearing quality, once achieved by VDCD, can readily be produced without vacuum by the ladle refining process. With the use of the ladle metallurgy tools of re-heat, wire feeding, argon bubbling, and slag control, low-oxygen clean steels are produced at reduced levels of B- and D-type inclusions. Minimum stir cycles are specified that will produce these known levels of oxygen and cleanliness. If the ladle is treated through the refining station, further stirring energy can be imparted by placing the fully deoxidized heat in the vacuum chamber and stirring more intensely and efficiently to reduce hydrogen levels below 2.0 ppm. The clean steel practice to prevent reoxidation offers benefits primarily in reducing macroinclusions as expected through provision of bottom pouring and argon shrouding.

With the use of the ladle refining station as a tool coupled with a more precise control of the traditional VDCD process, further benefits can be achieved. By determination of oxygen, the aluminum deoxidizer can be wire-fed to provide the optimum carbon-oxygen reaction under vacuum. Thus, the best combination of micro- and macroinclusion levels are attained with low oxygen.

The ladle reheat capability also provides the opportunity to perform extensive alloying, stirring, and degassing and going to the strandcaster without the previous furnace super heat

TABLE 3—Clean steel quality, E 45-JK inclusion rating.

Practice	Worst Field				SAM B	SAM D	AMS 2301, mm/m ²	Oxygen, ppm
	B-alumina		D-globular					
	T	H	T	H				
LRS $\bar{X}$	1.3	0.4	1.0	0.3	4.0	3.5	250	15
LRS + V $\bar{X}$	1.0	0.2	1.0	0.3	3.0	3.0	130	16
LRS $\bar{X}$ clean steel	1.4	0.4	1.0	0.4	4.0	2.5	70	14
LRS + VDCD clean steel	1.0	0.1	1.0	0.2	1.8	0.6	8	10
LRS $\bar{X}$ cast	0.8	0.1	1.0	0.3	0.7	2.5	250 60 MLC ^{*a}	20
LRS $\bar{X}$ cal treat	0.5	0	0.5	0	0	0	0	16

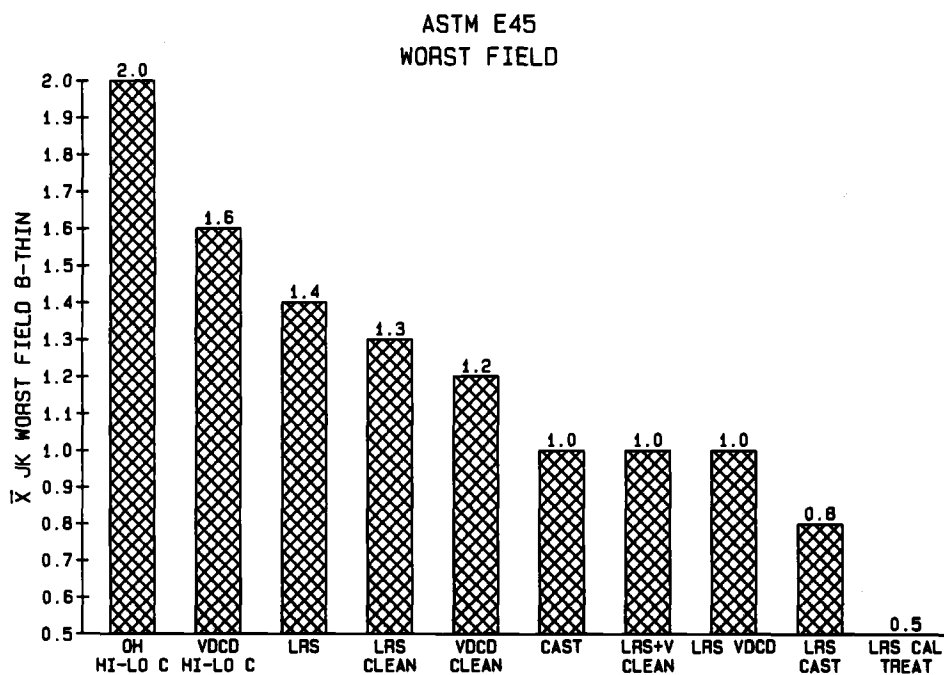
^{a*}MLC = mold level control.

FIG. 1—JK-B thin rating: process comparison.

penalty. Through the use of state-of-the-art strandcast technology such as mold level control, up-to-date tundish design, and ladle-tundish and tundish-mold shrouding, very favorable micro- and macroinclusion cleanliness results are provided in direct comparison with ladle-refined ingot cast and vacuum-treated ingot cast product. If we compare the quality levels of all the processes, it is clear that the ladle refining practices have replaced the traditional furnace refined steels in quality as measured by the JK (B-thin) worst-field, SAM-B, SAM-D, and magnetic particle (macroinclusion) ratings of AMS 2301 (Figs. 1 to 4).

One practice that has not been discussed is the use of calcium injection at low levels in the presence of extended stirring. The calcium transforms the alumina stringers into calcium aluminates, which are stirred to normal residual levels. Thus, with the removal of aluminum oxide, the steel is virtually inclusion free as depicted by the data of Table 3. Although calcium in bearing steel is not to be used for bearing fatigue life considerations, the potential of this ultra-clean steel should be considered.

Oxygen

Oxygen levels of the various processes certainly show no surprises (Fig. 5). The open-hearth practices show high levels: 25 to 30 ppm. Strandcast is at slightly lower levels with the use of protective shrouding to protect from reoxidation. The ladle-refining practices show consistently the lowest levels with the combination of LRS and VD CD at 8 to 12 ppm. Here, excellent bearing life should be achieved. However, the ladle-refined precipitation deoxidation process should achieve 5 ppm or lower oxygen with sufficiently long stirring times [3]. This is where engineering and fitness for use must be considered. Will infinitesimal decreases in oxygen

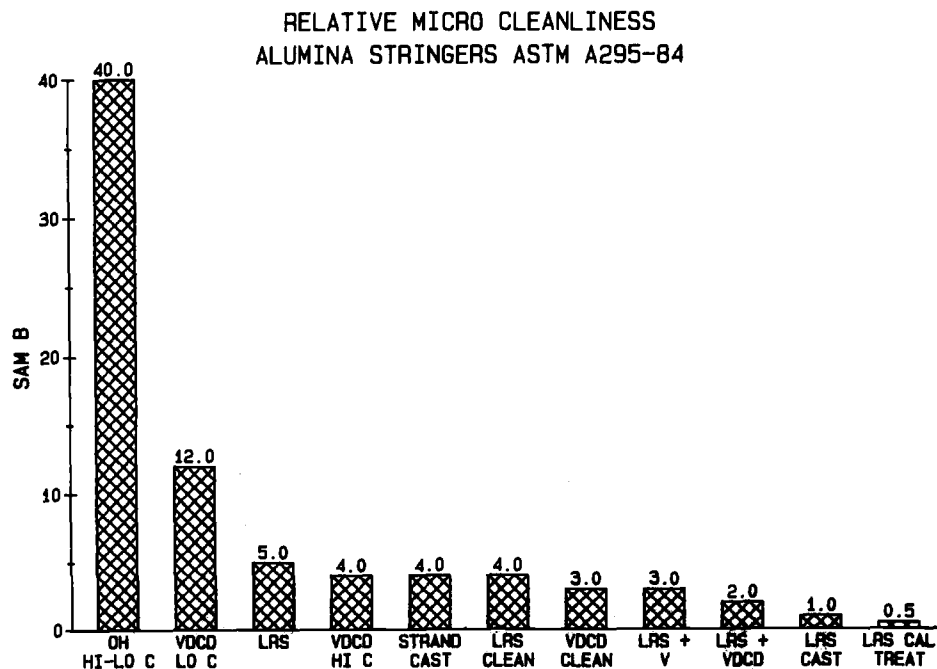


FIG. 2—SAM B rating: process comparison.

RELATIVE MICRO CLEANLINESS
GLOBULAR OXIDES - ASTM A295-84

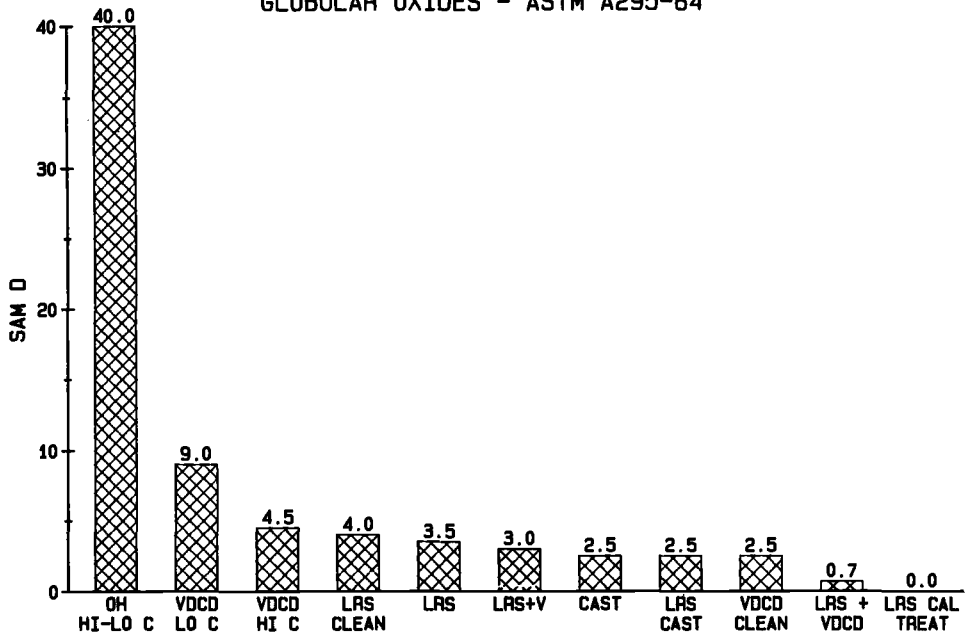


FIG. 3—SAM D rating: process comparison.

AMS 2301

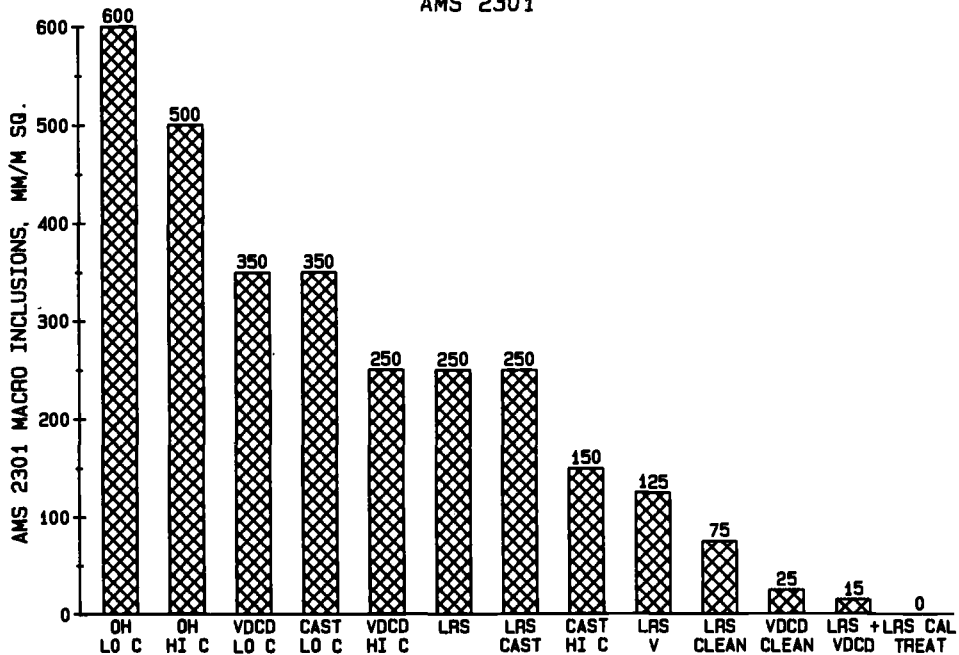


FIG. 4—Macroinclusion rating: process quality comparison.

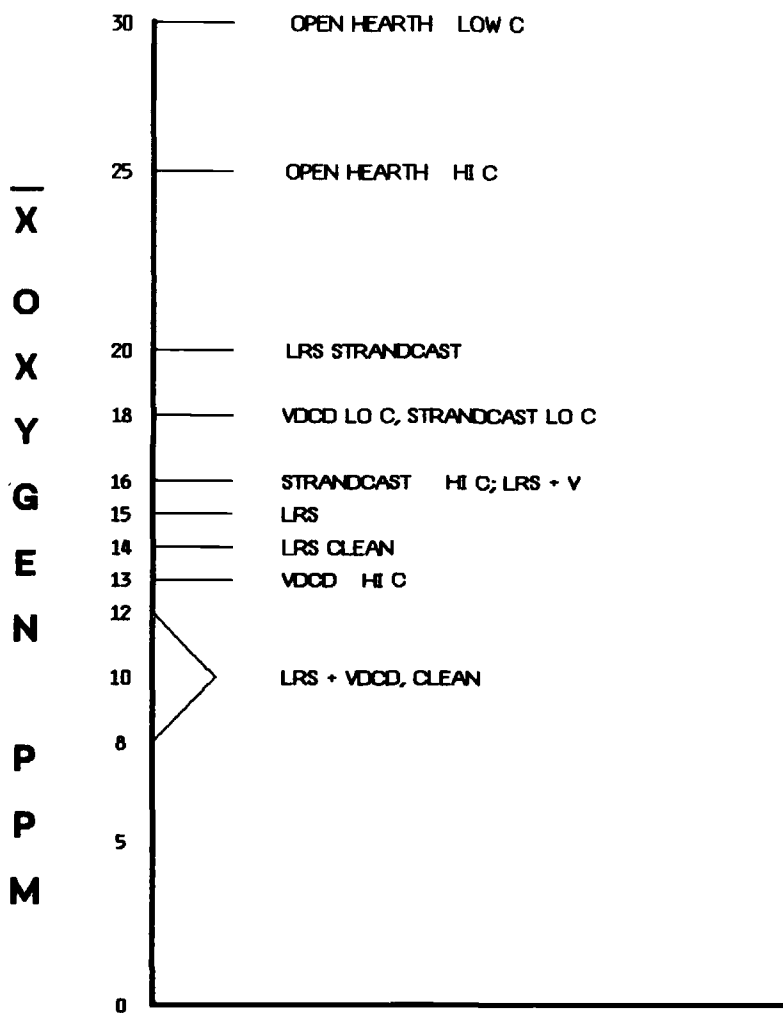


FIG. 5—Average levels of oxygen by process from capability study.

at this level provide economical returns in increased bearing life? It is doubted that oxygen can be even accurately measured at these levels. Round-robin oxygen testing (Fig. 6) by many labs shows a wide variation in results with the same samples [4]. Oxygen analysis is extremely sensitive to sample source within the ingot, ingot size, test equipment condition, and operator technique. Thus, interpretation of oxygen results should be made with caution, since repeatability and reproducibility can vary.

However, the objective of achieving low oxygen is theoretically to provide lower levels of oxides. With less oxygen to combine with aluminum, fewer inclusions form. The steel is cleaner. At a recent AIME meeting in Birmingham, Kolarik and Stover of The Timken Co. [5] presented data from a paper showing the lack of correlation between macroinclusions and

TOTAL OXYGEN ANALYSES-AISI ROUND ROBIN

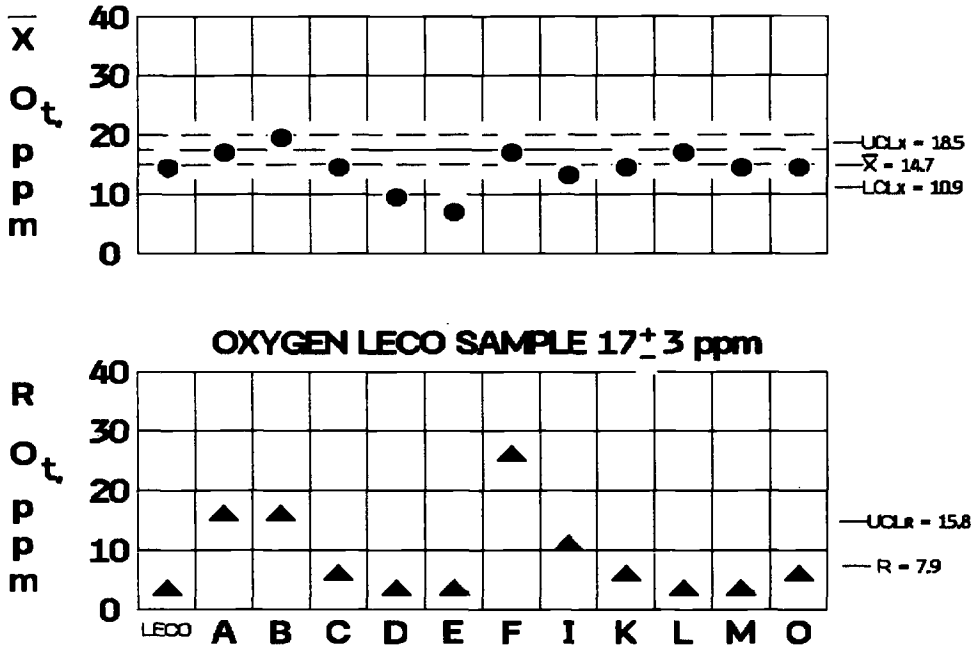
 $\bar{X}$, R FOR SIX SAMPLES PER LAB

FIG. 6—AISI round-robin laboratory test program for oxygen analysis.

oxygen content. Similar comparisons with LTV experiences could be made. With this in mind, all the data accumulated for this study—SAM-B, SAM-D, and AMS-2301 macroinclusions (mm/m^2)—were plotted against the corresponding oxygen levels taken at identical heat qualification locations within the heat. Absolutely no correlation could be found; cleanliness was random at all levels (Figs. 7, 8, and 9). Perhaps at very high and very low oxygen content, fatigue performance differences could be evident.

Chemistry Control

A feature of the ladle-refined steel, as indicated, has been the ability to control chemistry. With metered alloy additions, in the presence of a clean low oxide containing slag, and the use of precise ladle weights determined through the load cell devices, chemistry spot aims become very accurate. Uniformity throughout the heat is afforded by ladle stirring. Standard deviations from the melt aim before the implementation of ladle refining and after implementation are shown in Fig. 10. Carbon, manganese, chromium, and sulfur can be now controlled with pinpoint accuracy. Along with the chemistry control hardenability ranges can be restricted. Based on calculated hardenability, since actual test variability limitations can be prohibitive, half or quarter band (high, medium or low) would be common.

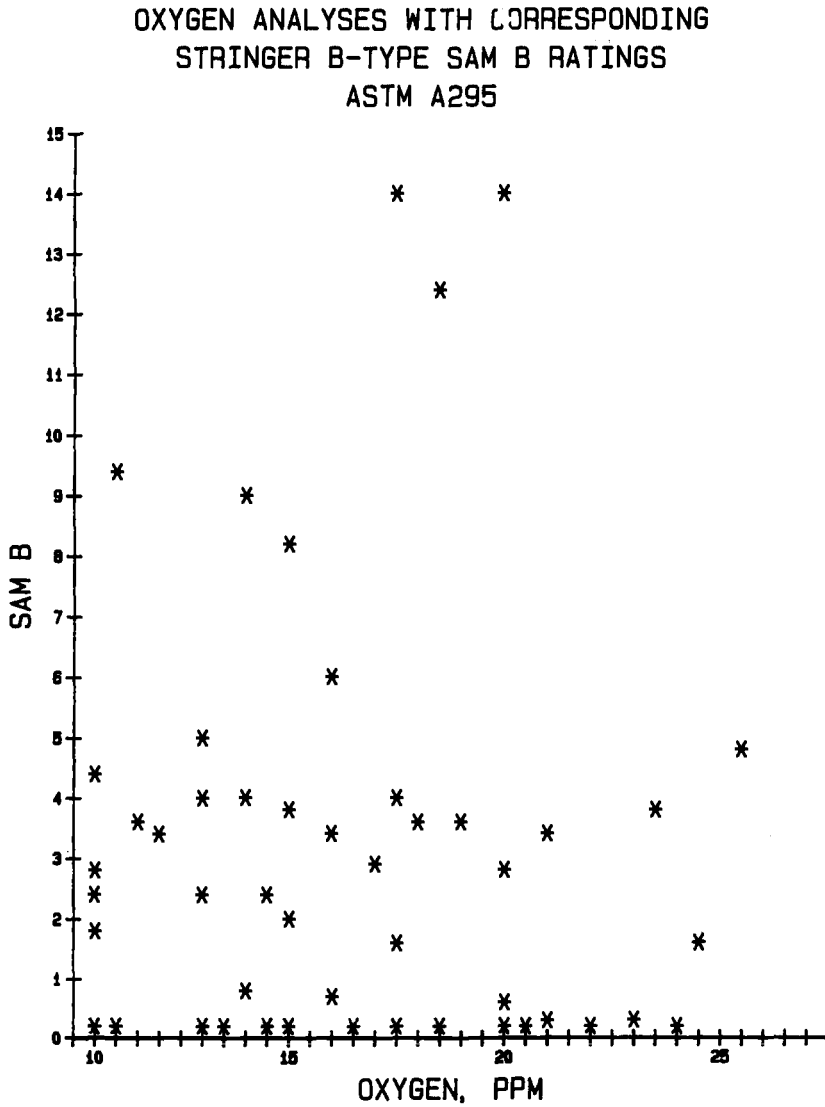


FIG. 7—Oxygen content versus SAM B microcleanliness.

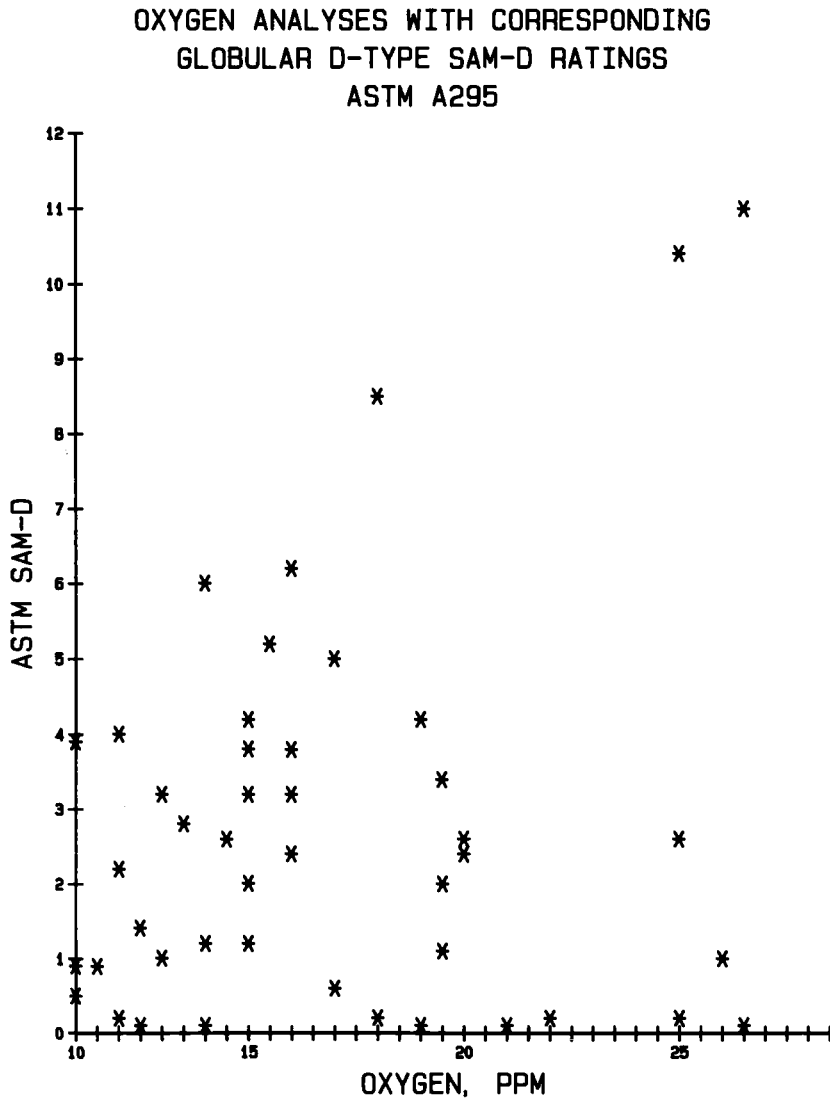


FIG. 8—Oxygen content versus SAM D microcleanliness.

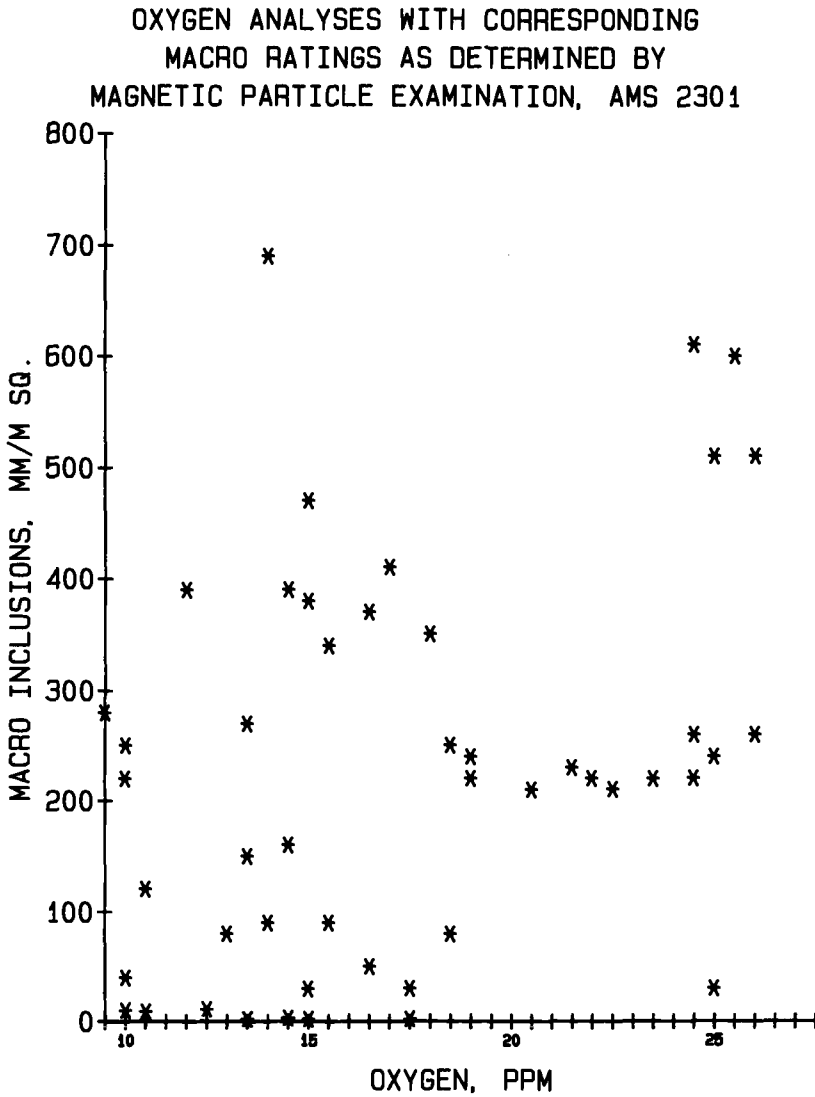


FIG. 9—Oxygen content versus AMS 2301 macroinclusions.

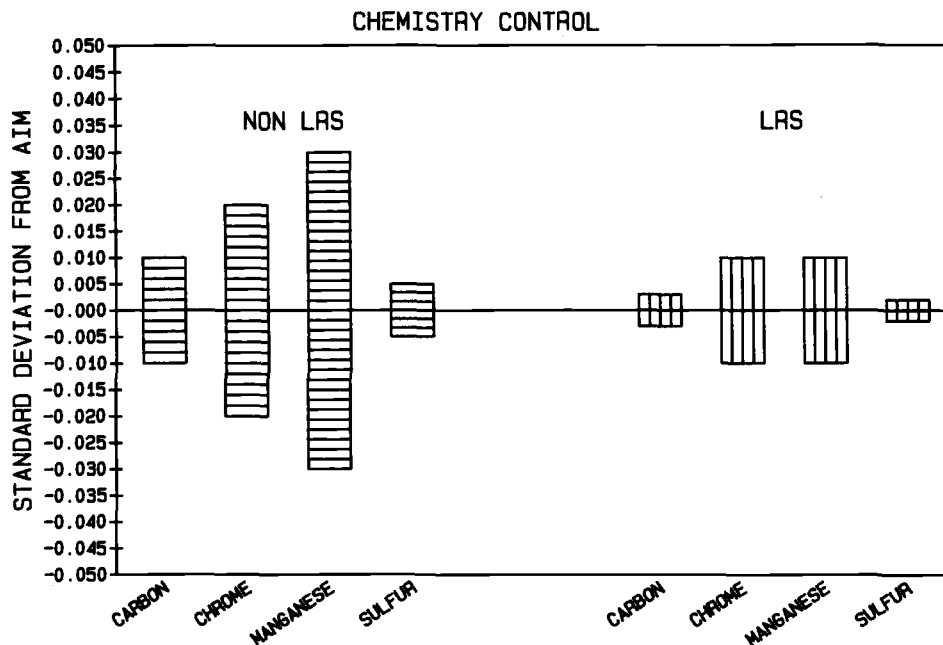


FIG. 10—Standard deviation from melt aims.

Conclusion

Thus, by the use of all the tools available at the ladle station, of the degasser, and of the teeming practices in the pits or at the caster, a clean, bearing-quality steel can be engineered to fit the needs and requirements of the customer. It may have low oxygen and low inclusion levels with specific levels of sulfur for machining or forming considerations. It may be engineered to known levels of inclusions that will perform with optimum cost considerations. However, clean steel can be achieved by various routes and methods. Products superior to the traditionally vacuum refined steels can now be provided by ladle refining techniques.

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Quality of High-Carbon Chromium Bearing Steel Produced in the Electric Arc Furnace—Ladle Furnace—RH Vacuum Degassing Vessel—Vertical Continuous Caster

REFERENCE: Uesugi, T. and Tsubota, K., "Quality of High-Carbon Chromium Bearing Steel Produced in the Electric Arc Furnace—Ladle Furnace—RH Vacuum Degassing Vessel—Vertical Continuous Caster," *Effect of Steel Manufacturing Processes on the Quality of Bearing Steels*, ASTM STP 987, J. J. C. Hoo, Ed., American Society for Testing and Materials, Philadelphia, 1988, pp. 278–292.

ABSTRACT: The quality of high-carbon chromium bearing steel, SUJ2 (SAE52100), produced by the electric arc furnace (EF)-ladle furnace (LF)-Ruhrstahl-Heraeus vacuum degassing vessel (RH)-vertical type continuous caster (CC) process, including the effects of impurities and cold working on its fatigue life, was investigated.

Bearing steel made in the experimental 100-kg induction furnace was also subjected to a fatigue life test in order to confirm the effect of impurities. The substantially low oxygen content of the CC bearing steel has resulted not only in high cleanliness which doubles the life of conventional ingot steel, but also in the reduction of buildup of nonmetallic inclusions on the inner surface of an immersion nozzle. Due to this reduction, approximately 10 000 metric tons (68 heats) of steel can be cast from one tundish-one nozzle without interruptions.

The fatigue life test was performed on specimens where the center segregation of bearing steel lay on their running surfaces with results that flaking sites did not tend to concentrate in the center portion. In other words, tests were done when center segregation did not affect the life of the specimen.

It has become clear that the fatigue life of the CC-bearing steel, despite reduced impurities, still depends mainly on the oxygen content of the steel. Also, the life test on the specimens made in the experimental furnace has confirmed that titanium and sulfur are hardly related to the life as long as their contents are kept at levels generally found in commercial bearing steel.

The life was further extended by cold working even when the oxygen content was extremely reduced. The bearing steel manufactured by the EF-LF-RH-CC-cold working process is accordingly considered to exhibit the longest life in commercially mass produced bearing steel.

KEY WORDS: high-carbon chromium bearing steel, vertical continuous caster, rolling contact fatigue, nonmetallic inclusion, oxygen content, impurity, center segregation, cold pilger

Rolling bearings have been used for critical parts which support the turning axles of automobiles, industrial machinery, etc. Improvement in rolling contact fatigue life of bearing steel has become important because smaller or heavier duty rolling bearings have recently been in increasing demand.

The life of rolling bearings, that is, the period during which the bearings can be used, is usually determined by the occurrence of flakings. Since it has long been known that the occurrence of flakings is remarkably affected by oxide nonmetallic inclusions [1], bearing steel manufacturers have been making every effort to reduce oxygen content in steel.

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The principal sources of oxygen are raw materials, oxygen injected at an oxidation stage, slag, air, and refractories. In order to reduce oxygen in steel, it is necessary to prevent molten steel from contacting air or refractories throughout the casting operation as well as to adopt vacuum degassing process. The less oxygen there is in steel prior to vacuum degassing, the less might be expected after the vacuum treatment.

With these factors taken into account, it was concluded that the melting process with oxidation reactions should be separated from the refining process involving reduction reactions, both of which are conventionally executed in the electric arc furnace. Also, continuous casting should be adopted to minimize contacts between molten steel and refractories. A vertical-type continuous caster is the most desirable with respect to the uniformity in distribution of nonmetallic inclusions.

The combined process going through an electric arc furnace (EF), a ladle furnace (LF), an Ruhrstahl-Heraeus vacuum degassing vessel (RH), and a vertical-type continuous caster (CC) has been thus established and has contributed to the mass production of bearing steel with lower oxygen as had been planned.

Not only can a lower oxygen content be achieved compared with that obtained through the conventional bottom-pouring ingot casting (IC), but titanium and sulfur can be also reduced through the EF-LF-RH-CC process by removing old slag prior to LF operation and also by strong stirring of a melt under new high basicity slag in LF operation for the former and the latter, respectively. Sulfur has been reported to either improve the fatigue life or shorten it [2,3], while the effect of titanium on the fatigue life has scarcely been reported. The effects of these impurities on the fatigue life were examined, using both the EF-LF-RH-CC bearing steel with reduced impurities and the experimental induction furnace-made heats with various levels of impurities.

Since cold working has been known to extend the fatigue life [4], a combination of the EF-LF-RH-CC process and cold working was expected to achieve the outstanding life.

The quality of the EF-LF-RH-CC bearing steel, SUJ2, and the effects of impurities and cold working on the fatigue life are discussed in this paper. The effects of impurities on the fatigue life was examined using the experimental 100-kg vacuum induction melted (VIM) bearing steels are also mentioned.

Features of Manufacturing Facilities and Operations

Electric Arc Furnace (EF)

The maximum capacity is 150 metric tons. Water-cooled panels are embedded in 75% of the sidewalls and 70% of the roof. Molten steel is tapped in a barely deoxidized state and transferred to the LF station. Tap to tap time is approximately 75 min for SUJ2.

Ladle Furnace (LF)

The chemical compositions of the ladle refractories and the refining slag play an important role in the reduction of oxygen in steel. Examples of these compositions are shown in Table 1 and Table 2. Both the bricks and the slag contain little silicon oxide to reduce oxygen in steel. The LF operation for SUJ2 is described briefly as:

Volume of steel	150 metric tons maximum
Treating time	30 to 50 minutes
Stirring method	Inert gas stirring by double porous method
Basicity of slag	Calcium oxide/silicon oxide ≈ 4 to 5
Treating temperature	1793 to 1843 K (1520 to 1570°C)

Sulfur in steel is reduced down to 0.01% or less during the LF operation.

TABLE 1—*Chemical composition of ladle bricks, weight %.*

	MgO	Al ₂ O ₃	SiO ₂	Fe ₂ O ₃	C
Slag line	77.6	...	...	0.4	15.0
Sidewall and bottom	...	83.0	11.6	1.1	...

TABLE 2—*Chemical composition of LF slag in weight %.*

CaO	SiO ₂	Al ₂ O ₃	MnO	MgO	Total Fe	Cr ₂ O ₃	P ₂ O ₅	CaF ₂	S
57.8	13.3	15.8	<0.1	4.3	0.6	<0.1	<0.1	7.8	1.1

TABLE 3—*Major particulars of continuous caster.*

Type of machine	Vertical bloom caster
Number of strands	3
Bloom size	370 by 470 mm
Maximum withdrawal speed	0.65 m/min

RH Vacuum Degassing Vessel (RH)

Steel refined in the LF is subjected to the RH degassing treatment for 20 to 30 min before transfer to the CC. Aluminum deoxidation and final adjustment in chemical composition are carried out during the RH degassing.

Continuous Caster (CC)

Particulars for the CC are indicated in Table 3. The refractories of the tundish, to achieve low oxygen content, are made mainly of magnesium oxide, which prevents steel from being contaminated by the refractories. Inert gas shrouding, which keeps the oxygen concentration in atmosphere in the tundish under 0.1% throughout the casting operation, is applied to protect steel against atmospheric contamination.

Experimental Procedures*Specimens*

All materials used were SUJ2. The chemical composition of the specimens is shown in Table 4. The chemical composition of samples produced in the VIM to evaluate the effects of impurities on fatigue life is given in Table 5.

Oxygen Analysis

Oxide inclusions in steel exert a fatal influence on its life, but steelmaking technology has successfully responded by meeting lower oxygen requirements for which the authors have carried on a vigorous campaign. Not only oxygen levels themselves but also oxygen analysis procedures are now the focus of attention.

TABLE 4—Chemical composition of steel used. [EF-LR-RH-CC] weight %.

Heat	C	Si	Mn	P	S	Ni	Cr	Mo	Cu	SolAl	Ti	O	Remarks
A	1.00	0.22	0.41	0.008	0.011	0.05	1.42	0.01	0.05	0.018	0.0028	0.0003	Fig. 3
B	0.99	0.20	0.38	0.012	0.015	0.07	1.40	0.02	0.06	0.014	0.0033	0.0003	Fig. 3
C	0.97	0.21	0.40	0.011	0.011	0.04	1.41	0.01	0.07	0.014	0.0041	0.0004	Fig. 3
D	0.98	0.27	0.36	0.015	0.009	0.05	1.39	0.01	0.06	0.017	0.0035	0.0004	Fig. 3
E	0.99	0.19	0.38	0.017	0.007	0.07	1.44	0.01	0.05	0.015	0.0038	0.0004	Fig. 3
F	0.98	0.22	0.40	0.011	0.005	0.06	1.43	0.01	0.04	0.015	0.0030	0.0005	Fig. 3
G	0.99	0.23	0.42	0.014	0.013	0.06	1.40	0.01	0.03	0.022	0.0045	0.0005	Fig. 3
H	1.00	0.29	0.40	0.012	0.004	0.06	1.44	0.01	0.08	0.020	0.0033	0.0005	Fig. 3
I	0.98	0.20	0.40	0.013	0.006	0.07	1.42	0.02	0.04	0.018	0.0028	0.0006	Fig. 3
J	0.97	0.18	0.36	0.015	0.004	0.04	1.40	0.01	0.08	0.014	0.0025	0.0006	Fig. 3
K	1.01	0.20	0.35	0.007	0.008	0.06	1.35	0.01	0.09	0.017	0.0022	0.0007	Fig. 3
L	0.99	0.23	0.44	0.011	0.003	0.05	1.38	0.01	0.05	0.012	0.0031	0.0007	Fig. 3
M	0.98	0.22	0.35	0.007	0.001	0.04	1.42	0.01	0.02	0.011	0.0010	0.0007	Fig. 3
N	0.99	0.19	0.34	0.015	0.003	0.04	1.34	0.01	0.04	0.015	0.0014	0.0003	Fig. 3
P	0.98	0.21	0.38	0.007	0.007	0.06	1.37	0.02	0.10	0.009	0.0040	0.0007	Fig. 3
Q	0.98	0.29	0.45	0.019	0.015	0.04	1.46	0.02	0.09	0.015	0.0025	0.0008	Fig. 4
R	0.97	0.30	0.42	0.012	0.010	0.04	1.43	0.01	0.05	0.022	0.0030	0.0011	Fig. 4

TABLE 5—Chemical composition of steel used to test effect of impurity on the rolling contact fatigue life, weight %.

	TPNO	C	Si	Mn	P	S	Ni	Cr	Mo	Cu	SoAl	Ti	O
Sulfur	S-1	1.02	0.27	0.35	0.017	0.007	0.07	1.42	0.01	0.01	0.022	0.001	0.0005
	S-2	0.98	0.26	0.36	0.015	0.006	0.07	1.43	0.01	0.04	0.016	0.003	0.0008
	S-3	0.98	0.26	0.37	0.015	0.010	0.07	1.41	0.01	0.04	0.017	0.003	0.0007
	S-4	0.99	0.26	0.38	0.015	0.019	0.07	1.41	0.01	0.04	0.019	0.003	0.0007
	S-5	0.97	0.26	0.41	0.014	0.027	0.07	1.41	0.01	0.04	0.019	0.003	0.0009
	S-6	0.99	0.26	0.26	0.015	0.011	0.07	1.41	0.01	0.04	0.016	0.003	0.0008
	S-7	0.98	0.26	0.52	0.016	0.009	0.07	1.41	0.01	0.04	0.017	0.002	0.0008
Titanium	T-1	0.96	0.26	0.42	0.005	0.001	0.01	1.44	0.01	0.01	0.022	0.001	0.0003
	T-2	0.98	0.26	0.42	0.005	0.001	0.02	1.42	0.01	0.01	0.020	0.004	0.0003
	T-3	0.98	0.25	0.42	0.005	0.001	0.01	1.43	0.01	0.01	0.020	0.008	0.0004
	T-4	0.97	0.26	0.40	0.004	0.001	0.02	1.42	0.01	0.01	0.026	0.014	0.0003
	T-5	0.97	0.26	0.41	0.008	0.006	0.07	1.43	0.02	0.07	0.021	0.001	0.0008
	T-6	1.01	0.25	0.41	0.012	0.009	0.07	1.43	0.02	0.07	0.025	0.002	0.0006
	T-7	0.99	0.25	0.40	0.012	0.009	0.07	1.41	0.01	0.07	0.024	0.004	0.0007
	T-8	0.98	0.26	0.41	0.012	0.010	0.07	1.42	0.02	0.08	0.014	0.008	0.0006
	T-9	1.01	0.26	0.41	0.013	0.009	0.07	1.42	0.02	0.05	0.023	0.011	0.0008
	T-10	1.00	0.25	0.41	0.014	0.009	0.07	1.42	0.02	0.08	0.025	0.016	0.0006

TABLE 6—Standard calibration samples.

Oxygen Level	Sample
5 ppm	NBS1094 and in-house sample
15 ppm	JSS-GS 2b
20 ppm	in-house samples
45 ppm	JSS-GS 1c

The oxygen contents were measured in the new oxygen analyzer, equipped with a temperature stabilizer for oxide reduction-extraction by the "inert gas carrier melting-infrared absorptiometric method." The previous analyzer, based on the "inert gas carrier melting-thermal conductometric method," was replaced in December 1985.

The standard deviation of analysis results has been now almost halved. The standard calibration samples are listed in Table 6.

Oxygen contents lower than 4.5 ppm, the value of NBS1094, are determined by extrapolation.

Calibration is carried out with in-house standard samples after each operation sequence (that is, a certain number of samples are tested, in addition to the monthly calibration work using NBS and JSS samples), so that the accuracy of analysis, particularly for lower ends of oxygen which are prone to frequent disputes, may be maintained and improved.

Preparation of Fatigue Life Test Specimens

Ordinary specimens for the thrust-type fatigue tester were taken from the transverse sections perpendicular to the axis of rolled or forged 65-mm diameter bars. Specimens for the cylindrical roller-type fatigue tester were also taken from the mid-radius ($\frac{1}{4}$ diameter) of the 65-mm bars. The rolling contact surface is therefore parallel to the bar's axis. The outer surface of the specimens arranged from tubing which was heated to 650°C and flattened was subjected to the thrust-type fatigue test. The specimens used to examine the effect of center segregation on the

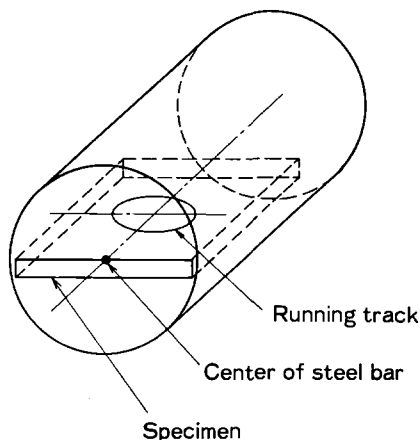


FIG. 1—Running track of rolling contact fatigue life test specimen.

TABLE 7—Life test conditions.

Item	Thrust Type	Cylindrical Type
Maximum nominal hertzian stress, $P_{\max}$, N/mm ²	4903	5884
Stress repetition rate, cpm	1800	46240
Lubricant	No. 60 spindle oil	No. 56 turbine oil
Surface finish of the specimen, $H_{\max}$, μm	0.35	0.25

fatigue life were prepared so that the axis of the 65-mm bar could lie on the running surface, as illustrated in Fig. 1. The specimens for fatigue life tests were heat treated as

835°C × 20 min. → oil quenched → 180°C × 90 min. → air cooled

Fatigue Life Tests

Principally the thrust-type rolling contact fatigue testers [5] were used for the life tests. Since the cylindrical roller-type fatigue tester [6] was used in the paper where the negative influence of sulfur on the fatigue life was reported, the same type of tester was also used to determine the effect of sulfur on the life. The test conditions are summarized in Table 7.

Results and Discussion

Oxygen Content

Histograms of oxygen contents are presented in Fig. 2 for SUJ2 currently produced by the EF-LF-RH-CC and the conventional ingot casting (IC) processes. On the average, CC steel is lower in oxygen content than IC steel by 2.6 ppm.

Because NBS 1094, the lowest oxygen standard sample, represents 4.5 ppm, it might be said that the reliability of analysis lower than 4.5 ppm is doubtful. Remember, however, that SUJ2

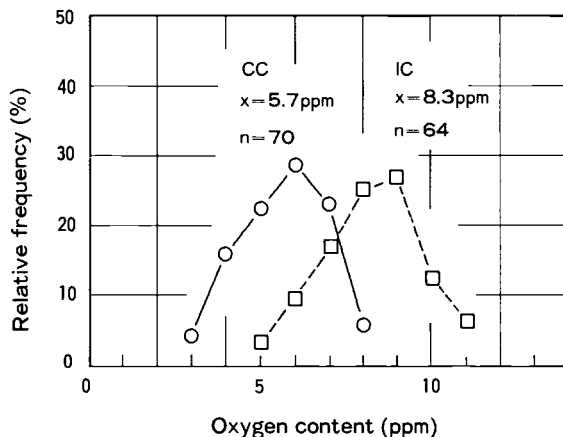


FIG. 2.—Oxygen distribution of SAE 52100 (n = number of heats).

TABLE 8—Nonmetallic inclusions rated to SAM method.

Inclusion	Process	
	CC ^a	IC ^b
Type B	0.06	0.4
Type D	0.06	0.8

^aContinuous cast, number of heats = 70.^bIngot cast, number of heats = 64.

with lower than 5 ppm oxygen has been produced. This level of oxygen in the EF-LF-RH-CC steel is almost on a par with that in the vacuum arc remelted steel.

Such low oxygen content has almost eliminated problems such as the buildup of nonmetallic inclusions on the inner surface of an immersion nozzle, and about 10 000 metric tons (68 heats) of bearing steel accordingly can be cast from one tundish without nozzle change.

Cleanliness

Cleanliness of the CC and IC steel currently produced was examined by the SAM method, specified in ASTM Specification for High-Carbon Ball and Roller Bearing Steel (A 295-84). Results are shown in Table 8.

According to the results, the ratings of the CC steel are reduced to $\frac{1}{7}$ for B-type inclusions and $\frac{1}{13}$ for D-types as compared with those of the IC steel. It is, as indicated in Fig. 2, the lower oxygen content that has improved the ratings of the CC steel.

Fatigue Life of CC Steel

As stated earlier, CC steel has such low oxygen content that longer fatigue life can be expected. The results of the thrust-type rolling contact fatigue life test on the CC steel currently produced are shown in Fig. 3. The fatigue life band previously obtained for a relationship between oxygen content and L_{10} life [4] is also shown in Fig. 3.

The L_{10} life of the CC steel with 5.7 ppm average oxygen content was $(12 \text{ to } 80) \times 10^6$, while that of the IC steel with 8.3 ppm average oxygen content was $(7 \text{ to } 35) \times 10^6$. Roughly speaking, the life of the CC steel has so far been doubled as compared with that of the IC steel.

Because oxide inclusions, which work as a medium to form a stress-concentration zone where carbon atoms diffused by rolling stress are trapped [7], initiate micro cracks, they should be reduced to as few as possible.

Effect of Center Segregation on Fatigue Life

Segregation of nonmetallic inclusions or carbon is found in the center of both the CC and the IC steel. The effect of center segregation on the fatigue life of the CC and the IC steel was investigated using the specimens prepared as indicated in Fig. 1.

The microstructure in the center of the CC steel is shown in Fig. 4, where the microscopic carbide segregation is observed.

The fatigue test results are demonstrated in Fig. 5. In addition, the flaking sites were located on the specimens because it had been expected that the flaking would have a tendency to occur on the center line of the specimen, that is, the axis of the steel bar, if center segregation indeed

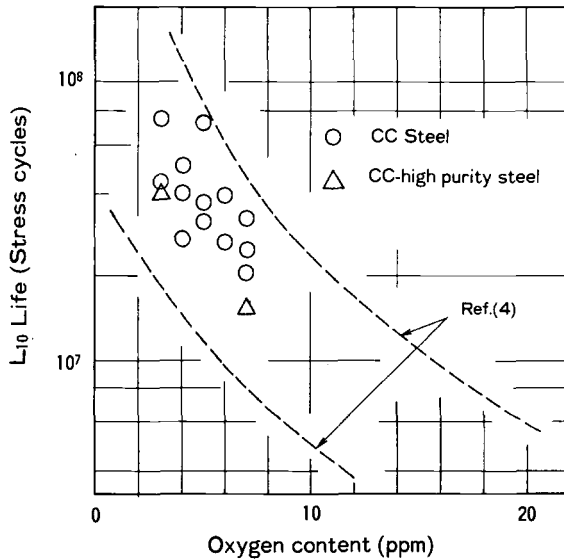


FIG. 3—Life test results of CC steel and CC-high purity steel.

promoted flakings. The relationship between life and flaking site is plotted in Fig. 6. It can be recognized that there is little difference in the number of flaking sites in each of three 30-deg zones in a quadrant between the CC and the IC steel. In addition, since the absolute values of the CC steel life are better than those of the IC steel, the flaking sites of the CC steel scatter away from the origin of the quadrant.

Two factors are thought to contribute to the results of the longer life with the CC steel and absence of tendency of flaking sites being segregated on the center portion in the CC steel as in the ingot cast steel:

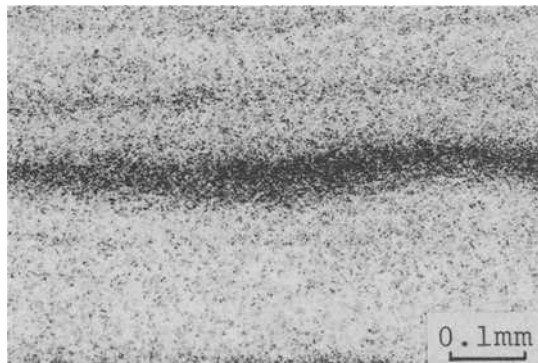


FIG. 4—Microstructure of center segregation in continuous cast SAE 52100 (65-mm ϕ , life test specimen).

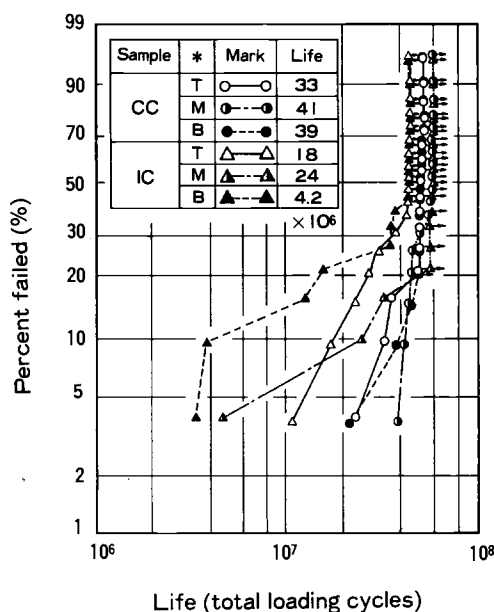


FIG. 5—Weibull distribution of rolling contact fatigue life test (longitudinal section). * = position of bloom or ingot. CC = continuous cast (Heat Q). IC = ingot cast (Heat R).

1. Oxygen content in the CC steel is lower than that in the IC steel, as shown in Fig. 2.
2. While the carbon segregation ratio, C/Co , where "C" is carbon at the center, "Co" is ladle analysis carbon, is approximately 1.15 in the CC steel, it is 0.9 to 1.15 in the IC steel depending on the location of an ingot. The carbon segregation is therefore not so severe as is generally expected of the CC steel.

Some similar tests on the relationship between center segregation and fatigue life were previously reported with almost the same results as above [8–10].

Fatigue Life of High-Purity CC Steel

Recently, a reduction of impurities such as sulfur or titanium has been attempted in order to improve the fatigue life [11,12]. The reduction of these impurities in addition to a reduction of oxygen can be achieved through the EF-LF-RH-CC process and several heats of bearing steel with the lowest possible impurities were made for the fatigue life tests.

According to the results shown in Fig. 3, the reduction of either titanium or sulfur did not result in any improvement in the life. The results still fall within the fatigue life band, previously determined for 0.002 to 0.005% titanium and 0.005 to 0.02% sulfur.

To further investigate the effect of impurities on the life, we made specimens in the VIM and subjected them to the fatigue life test as well as the rotating-bending fatigue test.

Effect of Impurities on Fatigue Life and Rotating-Bending Fatigue Strength of Experimental Steel

The results obtained by the fatigue life test and the rotating-bending fatigue test are shown

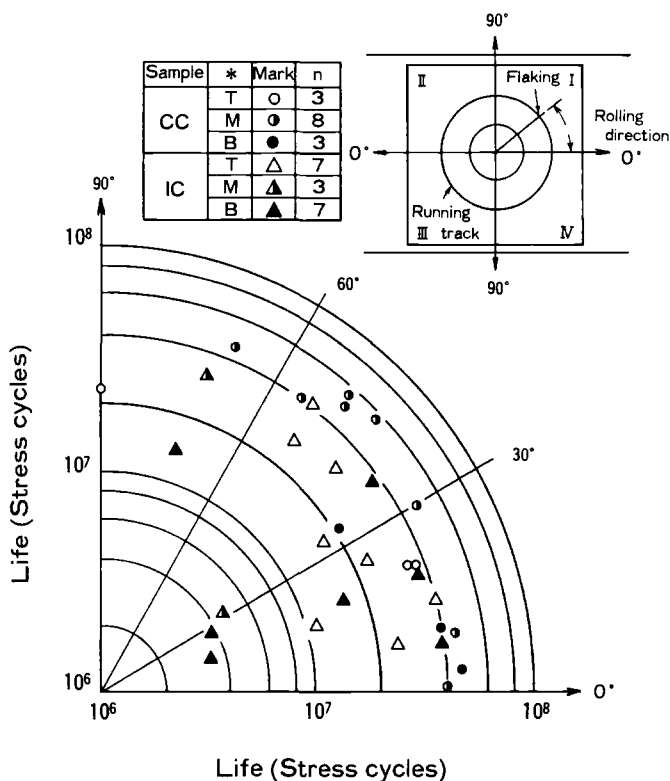


FIG. 6—Relationship between flaking site on the specimen and fatigue life. * = position of bloom or ingot. CC = continuous cast. IC = ingot cast.

in Figs. 7 and 8, respectively. The cylindrical roller-type fatigue test was also employed to examine the effect of sulfur (Fig. 7). It can be said from these figures that

1. 0.03% maximum of sulfur, the level in commercial steel, has little influence on the rolling contact fatigue life.
2. Titanium under 0.005%, the level in commercial steel, has little effect as well on the fatigue life, whereas higher titanium deteriorates it.
3. The rotating-bending fatigue limit, that is, 10^7 strength, is not affected by sulfur, but within the tested sulfur range, it tends to be improved by an increase in sulfur content.
4. The 10^7 strength is not influenced by titanium, either.

Although sulfur has been reported to either improve the fatigue life [2] or shorten it [3], the present result indicates that sulfur has nothing to do with the life. It seems reasonable to theorize that manganese sulfide relaxes the stress concentration by enveloping oxide inclusions to improve the life [2] when there is a significant amount of oxygen in steel. However, the fewer oxide inclusions there are, the less sulfur should be required to wrap them. The fact that there is always excessive sulfur in the commercial low oxygen steel to make up for adverse effects of oxygen may explain why there is no relationship between sulfur and life.

Although titanium has been reported to have a negative influence on the fatigue life [1] and the rotating-bending fatigue strength [12], the negative influence of titanium on the rolling

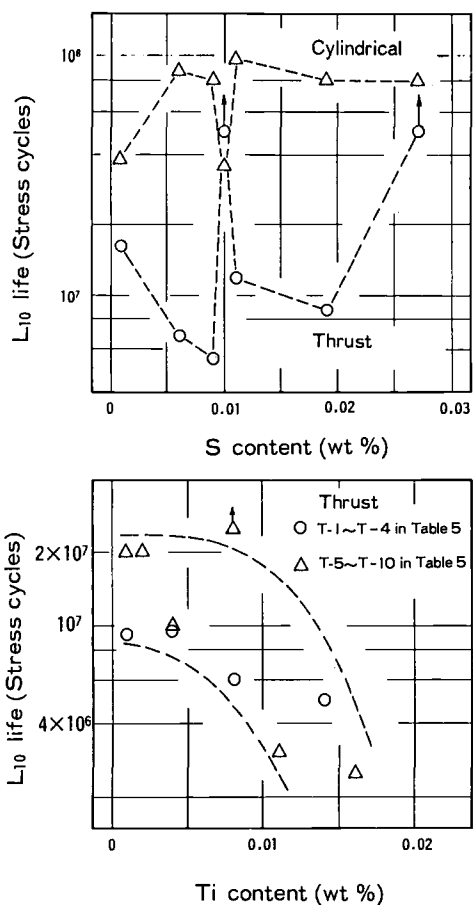


FIG. 7—Effect of impurities on rolling contact fatigue life.

contact fatigue life was observed only when it was present in the excess of 0.005%, and its effect on the rotating-bending fatigue strength was not seen in the present work.

Since these results are somewhat different from those in the previous reports, the effect of titanium nitride (TiN) on the fatigue strength was investigated further.

A "fish eye" [13] is often found on the fracture-surface of the smooth-type specimen after the rotating-bending fatigue test. At least one sample of fish eye was taken from each heat, and the nonmetallic inclusions in the center of eighteen fish eyes altogether were analyzed in the electron probe microanalyzer. TiN particles were found in only one fish eye out of the 18, while alumina-base oxide inclusions were identified in the rest. It is accordingly reasoned that TiN is less likely to cause the fatigue fracture.

The behavior of the TiN particles on the running track of the thrust-type fatigue test specimens was furthermore observed during the tests. Although oxide inclusions are known to break up and ultimately flow out from the specimens along with a lubricant [7], the TiN particles did not behave likewise, nor was the formation of microcracks seen around them until the running surface became too rough for the TiN to be microscopically identified.

It seems therefore that TiN has much smaller effect on the rolling contact fatigue than oxide

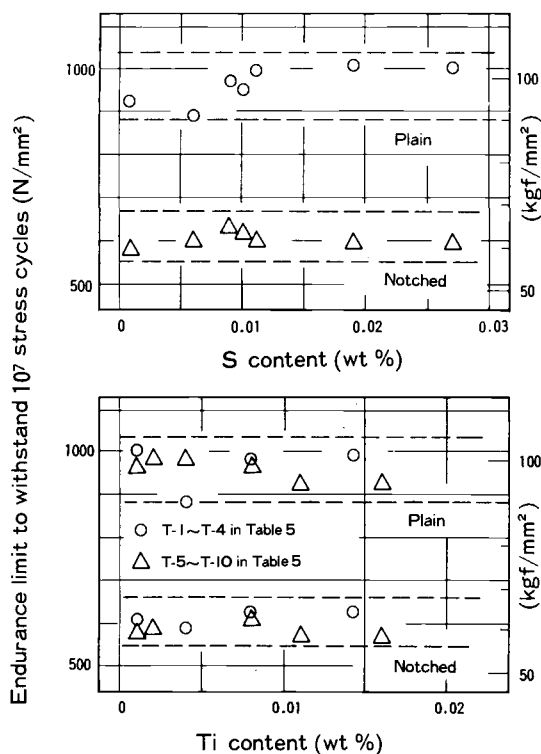


FIG. 8—Effect of impurities on rotating bending fatigue strength.

nonmetallic inclusions because, as stated above, the latter form voids while the former does not, apart from a similar degree of residual stress [14] produced around both of them by heat treatment. The shorter fatigue life to be caused by an increase in titanium over 0.005% is presumed to be due to increasing occurrences of microcracks induced by interactions between TiN and oxide inclusions, or between growing TiN clusters themselves.

Effect of Cold Working on Fatigue Life

Probably because of the destruction of oxide inclusions, the fatigue life of SUJ2 containing 6 to 14 ppm oxygen has been almost tripled by cold working in the cold pilger machine [4]. This extended life would be more conspicuous if more oxides were in steel. To confirm the effect of cold working on the fatigue life of such low oxygen steel as the CC steel, some tubes representing two bearing steel heats of 3 and 7 ppm oxygen were cold pilgered with an approximate 70% reduction in area prior to the fatigue tests.

The results shown in Fig. 9, where the fatigue life was doubled by cold working, indicate that the life is still significantly influenced by oxide inclusions even if the oxygen content is substantially reduced.

The tubing produced by the EF-LF-RH-CC cold-working process can, according to the results, be assumed to exhibit the longest life in current, mass produced bearing steels.

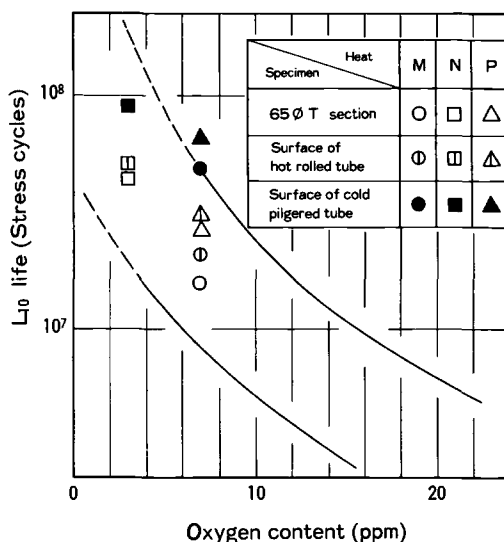


FIG. 9—Effect of cold rolling on rolling contact fatigue life.

Summary

The results of the present work are summarized below:

1. Due to the substantial reduction of oxide inclusions in SUJ2 produced by the EF-LF-RH-CC process, the clogging of the immersion nozzle during casting has been eliminated, and the continuous casting of approximately 10 000 metric tons of the steel from one tundish-one nozzle without interruptions has been made possible.
2. The fatigue life of the CC steel has been approximately doubled, as compared with that of the conventional IC steel, by a drop in its oxygen content.
3. The center segregation of the CC steel has shown little effect on the fatigue life because of the significantly reduced number of oxide inclusions.
4. The life has not been affected by 50 ppm maximum of titanium which is generally found in commercial bearing steel, whereas a further increase in titanium has detrimental effects.
5. The fatigue life has not been influenced by sulfur, either, as long as its content falls within the range generally found in commercial bearing steel.
6. Cold working has improved the fatigue life even when the oxygen content is extremely low.

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New Developments in the Production and Testing of Bearing Steels

REFERENCE: Dressel, P. G., Kremer, K.-J., Spitzer, H., Vöge, H., and Weber, L., "New Developments in the Production and Testing of Bearing Steels," *Effect of Steel Manufacturing Processes on the Quality of Bearing Steels*, ASTM STP 987, J. J. C. Hoo, Ed., American Society for Testing and Materials, Philadelphia, 1988, pp. 293–307.

ABSTRACT: The production routes for bearing steels, available in the works of Krupp Stahl AG, and developments to improve product quality and to lower production costs are described. The requirements for chemical composition, nonmetallic inclusions, and microstructure are discussed in respect to machinability and rolling contact fatigue life of bearings. Melting and refining of the bearing steel AISI 52100 (DIN 100 Cr 6) and its processing to billets, bars, and wire-rods are illustrated under the aspect of process technology developments and economical factors. The introduction of the continuous casting process and improvements in manufacturing technology will render the production of bearing steels less expensive at a high quality level. This, however, requires further research and development which should be assisted by appropriate requirements for bearing steel products.

KEY WORDS: production routes for bearing steel products, product requirements, chemical composition, nonmetallic inclusions, machinability, rolling contact fatigue, economical factors, concast, continuous casting, ingot casting

With an annual production of about 1.5 Mio-t, bearing steels share an important part of the special steel production of western industrialized countries. The production of these steels with a high-quality standard on a low-cost basis, therefore, is a particular challenge to the special steel producers.

Besides other factors, the rolling contact fatigue life of bearings are influenced by the fatigue behavior of the material [1–6]. The mutual efforts of steel producers and bearing manufacturers to ensure a high bearing life, based on material properties and considering economical factors, are established in the specifications for bearing steels. Table 1 gives a comprehensive survey of the requirements on chemical composition, nonmetallic inclusions, and microstructure, specified for bars and wire-rods of the bearing steel grade American Iron and Steel Institute (AISI) 52100 (DIN 100 Cr 6) according to DIN and ASTM standards. The bearing manufacturers, FAG, INA and SKF, partly specify more restricted properties.

The steel producers have to control their production routes carefully in order to meet the specified requirements accurately. Beyond that, they have to improve product quality and lower production costs, considering the results of investigations on the material behavior as well as developing their process technology.

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TABLE 1—Specifications (chemical requirements, inclusion rating, microstructure) for bars and wire-rods in high-carbon ball and roller bearing steel AISI 52100 (DIN 100 Cr 6), spheroidized.

	DIN 17230-80	FAG Specifications 1982/1984	INA Specifications 1985	ASTM A 295-84	SKF Specifications 1986
	CHEMICAL COMPOSITION (CONTENT, %)				
Carbon (C)	0.90 to 1.05	0.90 to 1.00	0.95 to 1.05	0.98 to 1.10	0.95 to 1.10
Manganese (Mn)	0.25 to 0.45	0.30 to 0.40	0.25 to 0.45	0.25 to 0.45	0.25 to 0.45
Phosphorus (P)	0.030 maximum	0.025 maximum	0.030 maximum	0.025 maximum	0.025 maximum
Sulfur (S)	0.025 maximum	0.020 maximum	0.025 maximum	0.025 maximum	0.010 to 0.025 (bars) 0.025 maximum (wire-rods)
Silicon (Si)	0.15 to 0.35	0.20 to 0.35	0.15 to 0.35	0.15 to 0.35	0.15 to 0.35
Chromium (Cr)	1.35 to 1.65	1.45 to 1.65	1.40 to 1.65	1.30 to 1.60	1.35 to 1.65
Nickel (Ni)	0.30 maximum	0.30 maximum	0.20 maximum	0.25 maximum	0.25 maximum
Copper (Cu)	0.30 maximum	0.25 maximum	0.20 maximum	0.35 maximum	0.30 maximum
Molybdenum (Mo)		0.08 maximum		0.10 maximum	0.10 maximum
Titanium (Ti)			0.0030 maximum	purchaser may specify	0.0030 maximum
Oxygen (O)		0.0020 maximum	0.0020 maximum	that analysis of	0.0015 maximum
Tin (Sn)			...	titanium, aluminum	0.03 maximum
Lead (Pb)			...	and/or oxygen be	0.002 maximum
Arsenic (As)			...	provided by agreement	0.04 maximum
Antimony (Sb)			...	with steel manufacturer	0.005 maximum
Aluminum (Al)			0.020 to 0.050		0.010 to 0.060

INCLUSION RATING (MICROSCOPICAL METHODS)
According to Stahl-Eisen-Prüfblatt 1570-71^b

$d = 200$ K4 = 22	$d = 150$ K4 = 15	From billets 102 by 102 mm two thirds of all specimens and at least one from each ingot tested as well as the average of all specimens shall not exceed ASTM E 45-85.			
140 = $d = 200$ K4 = 20	70 = $d = 150$ K4 = 10	$d = 200$ K4 = 22	$d = 200$ K4 = 20	thin series	heavy series
100 = $d = 140$ K4 = 18	30 = $d = 70$ K4 = 8	140 = $d = 200$ K4 = 20	100 = $d = 140$ K4 = 18	A 2 $\frac{1}{2}$	A 2 $\frac{1}{2}$
70 = $d = 100$ K4 = 15	8 = $d = 30$ K3 = 10	70 = $d = 100$ K4 = 15	35 = $d = 70$ K4 = 12	B 1 $\frac{1}{2}$	B 1 $\frac{1}{2}$
35 = $d = 70$ K4 = 12	$d = 8$ K2 = 10	35 = $d = 70$ K4 = 12	17 = $d = 35$ K3 = 15	C 0	C 0
17 = $d = 35$ K3 = 15	Backgroud: maximum B	17 = $d = 35$ K3 = 15	8 = $d = 17$ K3 = 10	D 1	D 1
8 = $d = 17$ K3 = 10	according to ASTM E 45-85	8 = $d = 17$ K3 = 10	$d = 8$ K2 = 12		
$d = 8$ K2 = 12		$d = 8$ K2 = 12	Titanium carbonitrides		
		K2 = 1			

MICROSTRUCTURE
According to Stahl-Eisen-Prüfblatt 1520-78^c

If specified by the purchaser: carbide size, amount of lamellar pearlite, carbide networks, carbide streaking	CG 2.1 to 2.3 CN ^d CZ ^e	wire-rods		Completely spheroidized, free from excessive carbide segregation	According to Stahl-Eisen-Prüfblatt 1520-78 ^c bars wire-rods
		bars	CG 2.1-2.3 2.0-2.2		
		PA 3.0 3.0	PA 3.0 3.0		CG 2.1-2.3 ^f PA 3.0 = 3.1 CN = 4.2
		CZ = 6.3 ^g = 7.3 ^g	CN = 5.3 ^f = 5.1		CN = 5.2
					CZ = 6.3 = 6.2 (= 15 mm) = 6.1 (= 15 mm) = 7.3 (= 15 mm) = 7.2 (= 15 mm)

^aSpecifications are in discussion.

^bSee also DIN 50 602.

^cSee also ISO 5949-1983.

^dCarbide size outside segregation 12 μ m max. diameter and 2 μ m by 35 μ m max.

^eOutside of the center: $\frac{1}{3}$ radius.

^f10 = $d = 30$; CN = 5.1, $d = 30$ CN 5.4 in few fields allowed.

^g $d = 30$ CZ 6.4, 7.4 allowed in the center.

Production Routes at Krupp Stahl AG

Figure 1 shows the production routes for billets, rounds, bars, and wire-rods, operated by Krupp Stahl AG in the Siegen, Rheinhausen, and Hagen works [7]. Basically, one has to distinguish between the melting with the 300 *t*-BOP-process at Rheinhausen and the 140 *t*-EAF-process at Siegen. Both melting facilities have complete, modern equipment for secondary metallurgy (for example, ladle furnaces and degassing units [Dortmund-Horder (DH)- or Ruhrstahl-Heraeus (RH)-process]) and for gas-stirring and alloy-trimming. To meet special requirements with respect to cleanliness and micro-structure, electro slag remelted (ESR) units are available.

The 300 *t*-heats are cast as 5 *t* ingots (bottom cast) or as continuous-cast blooms with cross sections of 260 by 600 mm and 300 by 400 mm, respectively. The 140 *t*-heats are cast as 3 *t* ingots (bottom cast) or as continuous-cast cross sections of 265, 150, and 135 mm². For special steels and small quantities, a horizontal caster is available.

Except for the concast billet sections, ingots and concast blooms are directly hot-charged into the reheating furnaces of the primary rolling mills. The bar and wire-rod mills at Siegen and Hagen are provided with rolled blooms or concast billets with a maximum size of 150 or 135 mm².

Careful control of reheating, rolling, and cooling ensures a uniform structure and low degree of decarburization. In case of the wire-rod rolling, a salt-bath facility enables controlled cooling from the rolling heat.

All requirements specified for structure and hardness can be obtained by modern heat-treatment facilities. Modern equipment for finishing, inspection, and further processing of the

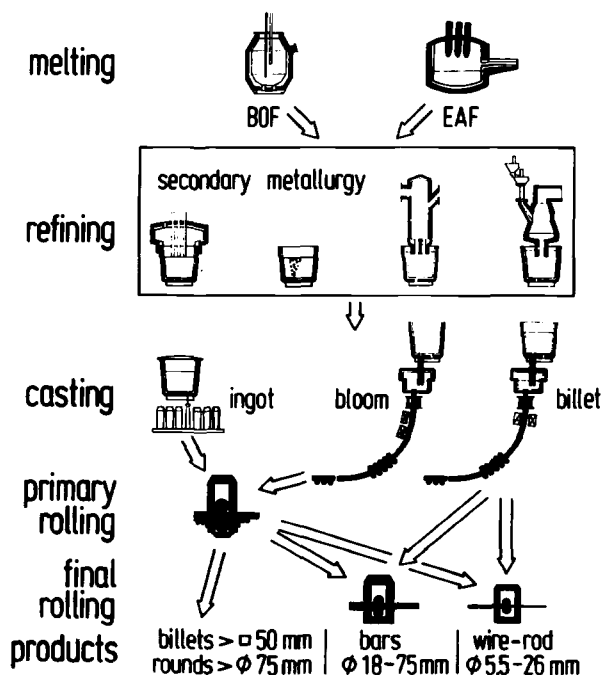


FIG. 1—Production routes for long products as billets, rounds, bars, and wire-rods in bearing steels at Krupp Stahl AG.

rolled products is available in all works. The quality-assurance system is adjusted to the demands of special steel production.

It should be noted that the available melting facilities offer the possibility for flexible adaption to variations in costs of raw materials and energy and to different requirements of processing.

Apart from the influence of tramp elements, which are mainly controlled by the raw material, the properties of steel products are achieved by the metallurgical practice during ladle treatment and casting, as well as by the further manufacturing processes. Today, the casting process is of particular interest [8]. Up until now, bearing steels have been cast as ingots and hot-charged to the primary rolling mill. To lower production costs, as well as to improve product quality, many efforts are concentrated on the continuous casting of bearing steels. Continuous casting of blooms was initially considered because this is a production route comparable to conventional ingot casting. The one-stage production of bars and wire-rods from continuous-cast small sections, however, offers particular economical advantages. As opposed to the two-stage production from rolled ingots or blooms, in this case, hot-charging and inspection usually are not achieved.

Developments on the Production of Bearing Steel DIN 100 Cr 6

Efforts to improve product quality and to lower production costs are helped by the information obtained from fatigue tests on bearings and specimens as well as by accompanying material investigations. Influence of the material on rolling contact fatigue life is determined by its structure, which should be uniform, fine-grained, and free of local inhomogeneities such as precipitations on phase boundaries or nonmetallic inclusions, by which the fatigue life can be decreased remarkably [9–16]. In particular, nonmetallic inclusions such as oxides or titanium-carbonitrides can initiate the formation of flaws under alternating stresses, and this affects the rolling contact fatigue life. Consequently, efforts of the steel producers concentrate on getting a uniform structure and avoiding such local inhomogeneities as far as possible.

Requirements on Chemical Composition

If we consider the rolling contact fatigue life of bearings, the melting process is of interest insofar as different raw materials are used. The steels melted with the electric arc furnace (EAF)-process generally show a remarkably higher level of tramp elements like molybdenum and nickel (Fig. 2), copper and tin (Fig. 3), and arsenic and antimony (Fig. 4) than steels melted with the basic oxygen furnace (BOF)-process. Because the contents of these elements practically cannot be influenced by the metallurgical practice, the accomplishment of restricted values requires the BOF-process or—in the case of the EAF-process—the availability of raw materials with low tramp element levels. The limits, specified in the standards, are aligned to the melting with the EAF-process, with which bearing steels are primarily melted today. Any restrictions, however, will increase the production costs due to the required raw materials.

It is also important to maintain low phosphorus contents. Figure 5 shows that the phosphorus content normally is higher with the BOF-process than with the EAF-process. By the use of adequate metallurgical practices, however, low phosphorus contents can be achieved with both melting processes, but this means greater expense in the melting process.

Results of rolling contact fatigue tests are not yet available which allow reliable quantitative correlations between the contents of tramp elements and the fatigue behavior of the steel. On the other hand, the advantageous effect of low tramp element contents on cold processing of the DIN 100 Cr 6 steel is well-established. For economical reasons, however, the limits for tramp elements should be considered carefully for any individual application, taking into account the knowledge about the material behavior and the requirements of the processing.

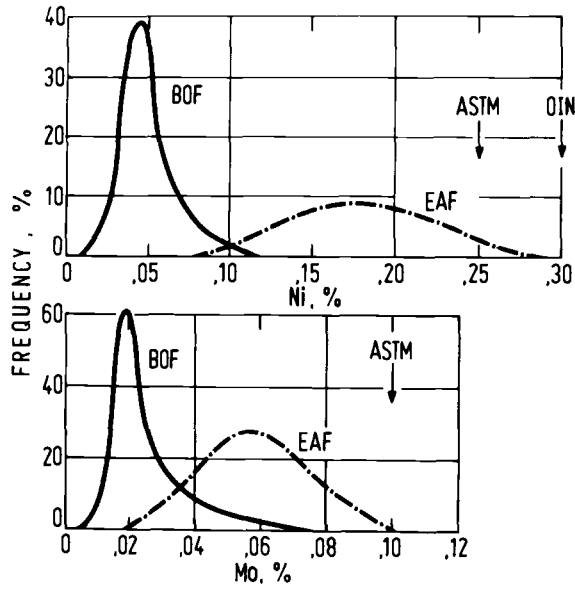


FIG. 2—Nickel and molybdenum contents of bearing steel DIN 100 Cr 6 with BOF- and EAF-melting.

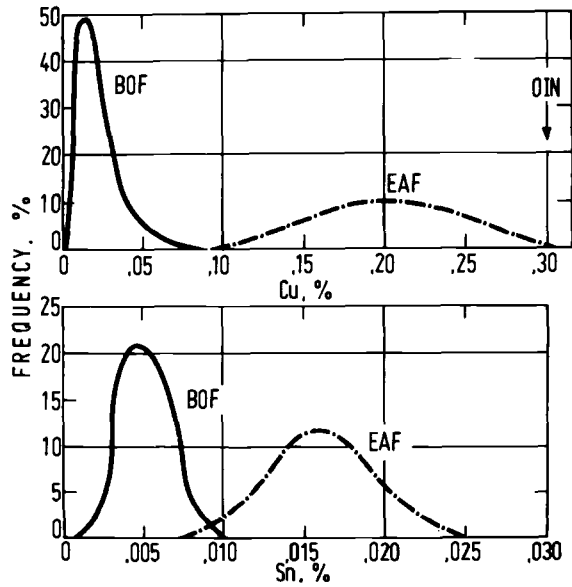


FIG. 3—Copper and tin contents of bearing steel DIN 100 Cr 6 with BOF- and EAF-melting.

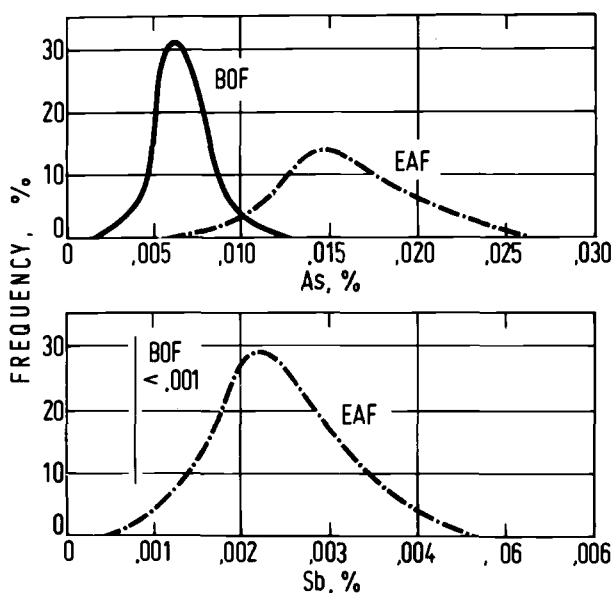


FIG. 4—Arsenic and antimony contents of bearing steel DIN 100 Cr 6 with BOF- and EAF-melting.

Requirements on Nonmetallic Inclusions

Type, size, amount, and distribution of nonmetallic inclusions in steels depend mainly on the metallurgical practice during ladle treatment and casting. Ladle treatment at Krupp Stahl is carried out in basic, lined ladles with facilities for heating up, degassing, gas-stirring, and alloy-trimming. Using these facilities, metallurgical practices have been developed to provide a homogeneous melt with narrow scatter bands for chemical composition and temperature, low total oxygen contents, and low hydrogen contents.

Today, bearing steels are mainly deoxidized with aluminium. Nevertheless, the specification of a minimum aluminium content does not seem to be very appropriate, because other deoxidation practices also may achieve a low amount of nonmetallic inclusions of a desired

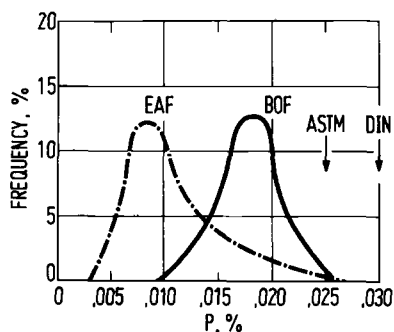


FIG. 5—Phosphorus contents of bearing steel DIN 100 Cr 6 with BOF- and EAF-melting.

composition [17]. Thus deoxidation with silicon or calcium and vacuum-carbon-deoxidation are preferred.

To characterize the metallurgical practice with aluminium, Figure 6 shows the distribution of the total oxygen contents after ladle treatment and in the as-rolled condition. The total oxygen contents of 15 or 20 ppm, respectively, required by some bearing manufacturers for rolled products, is met for ingot material as well as for concast material.

Low oxygen contents correspond with the microcleanliness assessments for oxides by metallographic examinations [18]. Ratings of oxides (Types B and D, according to the ASTM Practice for Determining the Inclusion Content of Steel [E 45-85]), show that generally the results comply with the requirements of the ASTM Specification for High-Carbon Ball and Roller Bearing Steel (A 295-84) (Fig. 7). Further restriction of these values, however, will increase the probability for rejections. Inclusion Type C is not found in aluminium-killed steels.

The uncertainty in metallographic determinations of microcleanliness by reference charts [19,20] raises the question of whether such restrictions are justified, especially in the case of low inclusion contents. The question is to what extent the specified metallographic methods are appropriate for the valuation of microscopic cleanliness, considering the low number of inclusions.

Figure 8 shows the results of an inclusion rating for bearing steel DIN 100 Cr 6, ingot and concast materials, using the KO-method according to DIN 50602-85 (Stahl-Eisen-Prüfblatt 1570-71) and quantitative image analysis (Quantimet 900). These methods [21,22] disclose small differences of inclusion amounts, but because they need considerable time for evaluation, they are not appropriate for routine applications. For this purpose, the analysis of the total oxygen content offers an alternative solution. Today, it can be carried out very accurately with the use of modern equipment, and it characterizes the microcleanliness adequately.

The determination of the total oxygen contents is not an adequate measure for macroinclusions, which are particularly harmful in respect to the fatigue properties. The amount of oxide macroinclusions depends on a careful separation of deoxidation products during the ladle treatment and on prevention of reoxidation during the casting process. During ingot casting as well as continuous casting, all efforts are made to prevent reoxidation of the melt and formation of oxide inclusions. The blue fracture test or the step-down test or both are used to check the content of macroinclusions. An ultrasonic inspection of the rolled product can disclose comparatively large inclusions only.

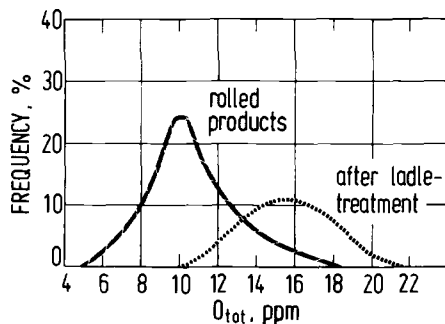


FIG. 6—Total oxygen contents of bearing steel DIN 100 Cr 6 after ladle treatment and in the rolled products (ingot and continuously cast material).

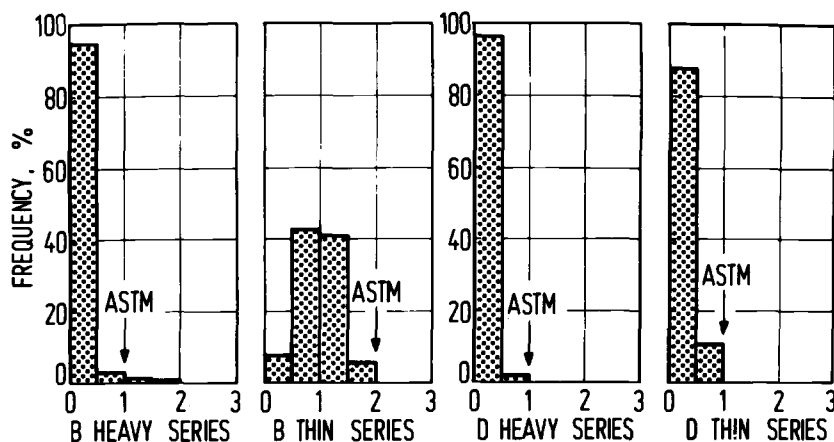


FIG. 7—Inclusion rating oxides according to ASTM E 45-84 for bearing steel DIN 100 Cr 6 (average values according to ASTM A 295-84).

Like alumina, titanium-carbonitrides in the metal matrix of bearing steels cause stress concentrations under alternating stresses. By this, formation of flaws can be initiated [11,12].

According to Fig. 9, the specified titanium content of 30 ppm maximum requires raw materials with low titanium impurities. Despite a special metallurgical refining practice, higher titanium contents only can be removed from the melt insufficiently (Practice A). The application of ferroalloys with low titanium contents can increase the production costs remarkably.

Comprehensive examinations have been made on the assessment of low amounts of titanium-carbonitrides. Figure 10a shows the relation between microscopical results and spectrometrically analyzed titanium contents. The comparatively high scattering is caused mainly

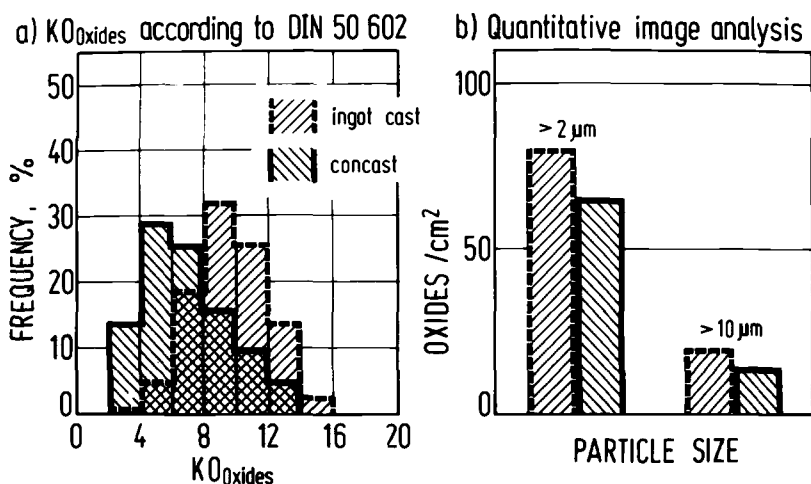


FIG. 8—Inclusion rating (oxides) for bearing steel DIN 100 Cr 6. (a) (Stahl-Eisen-Prüfblatt 1570-71). (b) Quantimet 900.

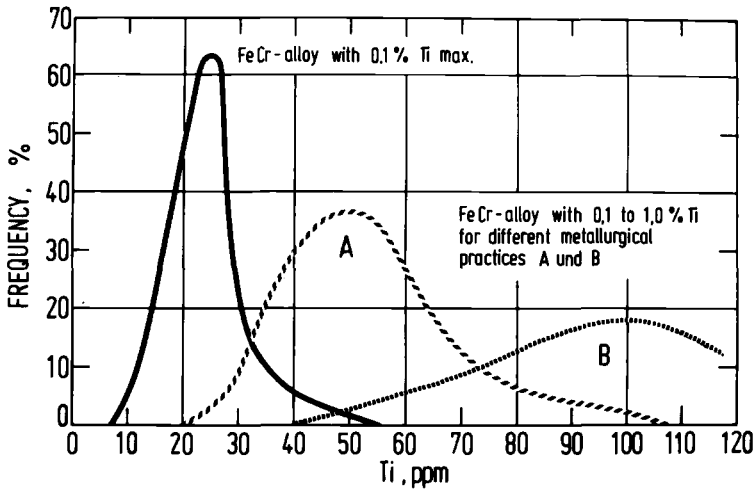


FIG. 9—Titanium contents of bearing steel DIN 100 Cr 6 for various alloying practices.

by the metallographic examination. This is due to the nonuniform distribution of carbonitrides in the solid state. On the other hand, spectralanalysis of low titanium contents can be achieved with high accuracy today [23].

It is also well-established that the scattering of metallographic results will increase with rising titanium contents (Fig. 10b). The KO-method characterizes the size and the number of particles and, therefore, the number of carbonitrides. As can be seen from Fig. 10b, this value varies only slightly at titanium contents below about 30 ppm.

The results of the investigations may be explained by the formation of the titanium-carbonitrides during the cooling of the melt. They suggest specifying an allowable titanium content and ensuring this by spectralanalysis. At low amounts of titanium-carbonitrides, a time-consuming metallographic examination is not appropriate.

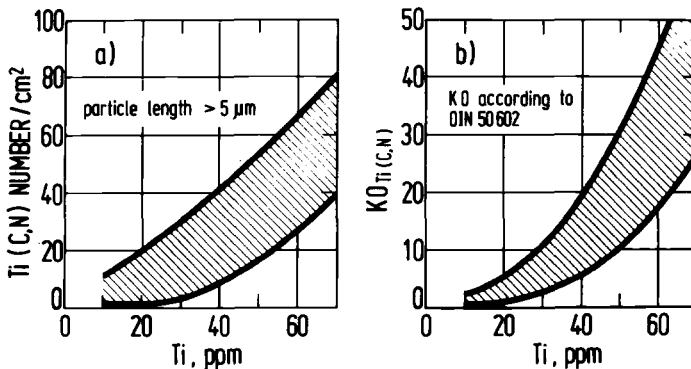


FIG. 10—Inclusion rating (titanium-carbonitrides) in relation to titanium contents of bearing steel DIN 100 Cr 6. (a) Metallographical rating of particles. (b) KO-method according to DIN 50602-85 (Stahl-Eisen-Prüfblatt 1570-71).

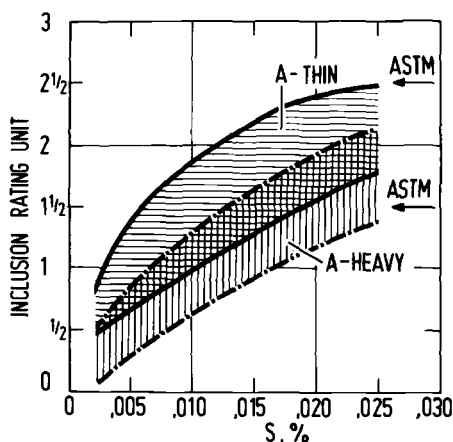


FIG. 11—Inclusion rating (sulfides) according to ASTM E 45-85 in relation to sulfur contents of bearing steel DIN 100 Cr 6 (average values according to ASTM A 295-84).

An evaluation of sulfide inclusions in relation to the sulfur contents (Fig. 11) indicates that the requirements for the type A-Heavy according to ASTM A 295-84 cannot be met with sulfur contents above 0.015%.

Inconsistent results are reported about the influence of sulfide inclusions on the rolling contact fatigue life. The investigations, however, mainly show only a slight effect of sulfur [10,14].

Requirements on Machinability

To improve machinability, some bearing manufacturers specify sulfur contents between 0.010 and 0.025%. Under shop conditions, the low total oxygen contents at the end of the ladle treatment, necessary to ensure a good castability in the continuous casting process and a small oxide inclusion amount, can be met easier at decreasing sulfur contents. Thus, low sulfur contents are advantageous for metallurgical reasons.

The influence of sulfur on the machinability is difficult to estimate because the influence of microstructure and oxide inclusions, as well as the machining practice, have to be considered. The tool material and the cutting speed applied are of particular influence. It is usually assumed that economical machining of bearing steel DIN 100 Cr 6 with high-speed steel tools as well as with carbide tools requires a minimum sulfur content. Due to the latest development of cutting tools and tool materials, the conditions of machining processes have changed, and thus the relationships should be proven.

Figure 12 shows tool life results of turning tests with a high-speed steel tool on steel DIN 100 Cr 6 in the spheroidized condition (Curve II). With the cutting speed V 60 (cutting speed with a catastrophic wear after a tool life of 60 min), a raise of sulfur contents from 0.005 to 0.020% increases machinability by about 15%. Compared with the effect of the microstructure (coarse- or fine-grained carbides), the influence of sulfur is relatively small, as can be seen from the results of a Japanese investigation (Curves I and III) [24].

As the results of turning tests with a carbide tool and cutting speeds between 200 and 315 m/min show (Fig. 13), crater wear and flank wear are practically not influenced by sulfur contents up to 0.025%. Today many kinds of inserts are available which enable easy removal of chips during machining of steels with low sulfur contents.

The investigations on machinability suggest that, normally, a sufficient machinability of the bearing steel DIN 100 Cr 6 can be achieved without higher sulfur contents.

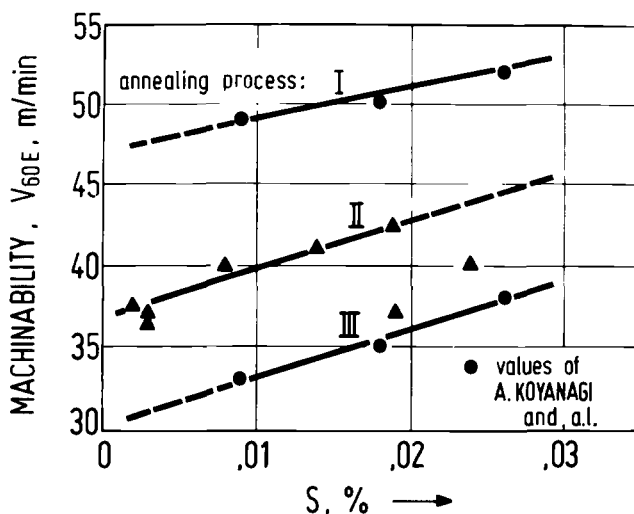


FIG. 12—Effect of sulfur on machinability of bearing steel DIN 100 Cr 6 for various spheroidized conditions (turning with high-speed tools).

Requirements on Microstructure

Structure formation with the two-stage production from ingots and concast blooms is well-established. Special attention is to be drawn to the advantage of the concast blooms with respect to less segregation. Only limited information is available about the structure of bars and wire-rods, which are rolled in a one-stage production from continuous-cast small sections

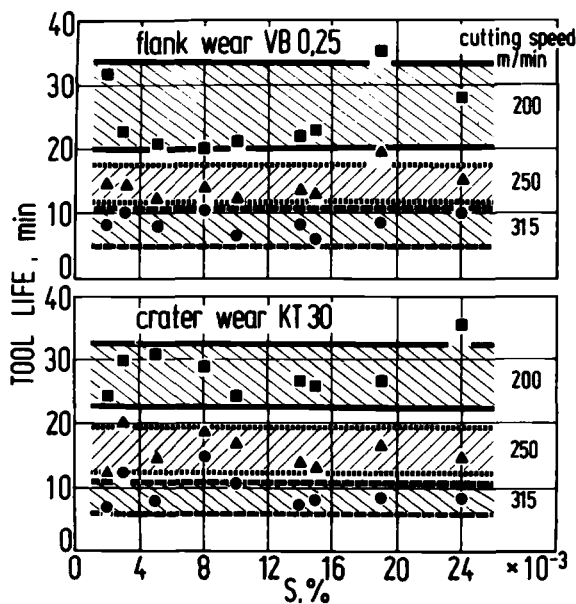


FIG. 13—Effect of sulfur on machinability (flank wear and crater wear) of bearing steel DIN 100 Cr 6 (turning with carbide tools).

without hot-charging. In this case, considerable effort is necessary to ensure formation of a uniform solidification structure without remarkable center segregation. This can be achieved by particular measures during the casting process, for example, by strict control of casting parameters and by application of electromagnetic stirring.

An example of state-of-the-art casting, Figure 14 shows results of a metallographic evaluation of carbide streaking of bars 45 to 65 mm in diameter after spheroidizing and hardening. The average of the worst fields of every specimen of a lot was determined [25]. The bars were rolled from 3 *t*-ingots (two-stage production with hot-charging) and from continuously cast billets of 150 mm² (one-stage production without hot-charging). The structure of concast material is characterized by a slightly lower carbide streaking. In a small zone in the center part, however, more pronounced carbide streaking and carbide network cannot be avoided completely due to center segregation. Up until now, the application of billets, bars, and wire-rods produced with this one-stage route has been restricted to parts such as rings and cones. To what extent other parts such as balls can be manufactured by this low-cost route in the future will depend on further developments of the process technology.

The question arises whether or not the chemical composition of DIN 100 Cr 6 steel should be altered. A steel with about 0.80% carbon, 0.50% chromium, 0.85% silicon, 0.85% manganese, and 0.1% vanadium shows a comparable hardenability and contact rolling fatigue life as well as lower tendency for center segregation [26]. In addition, the annealing time to achieve a spheroidized structure, and thus the production costs, can be reduced dramatically.

The salt-bath quenching of wire-rods directly from the rolling heat is another new method to improve product quality as well as to lower production costs (Fig. 15). By the controlled transformation of austenite, a structure can be achieved which enables a cold drawing of wire-rods without a previous spheroidizing. Also, the final annealing time can be reduced.

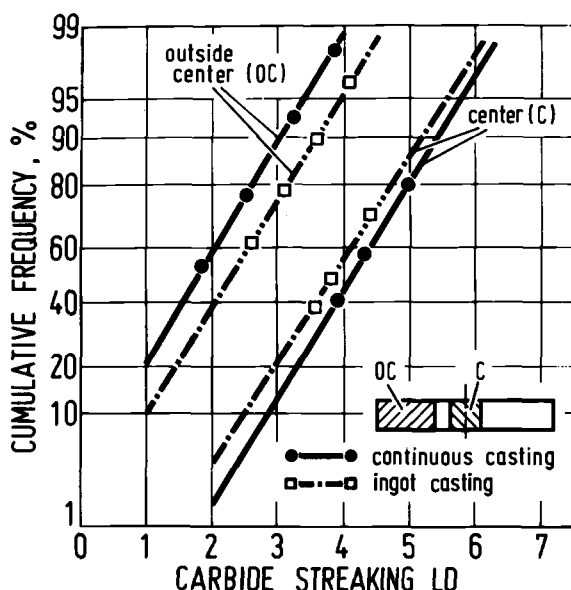


FIG. 14—Carbide streaking according to ISO 5949-1983 (average of the worst fields of every specimen of a lot) in bars of 45 to 65 mm diameter of bearing steel DIN 100 Cr 6 (comparison of ingot and continuous-cast material).

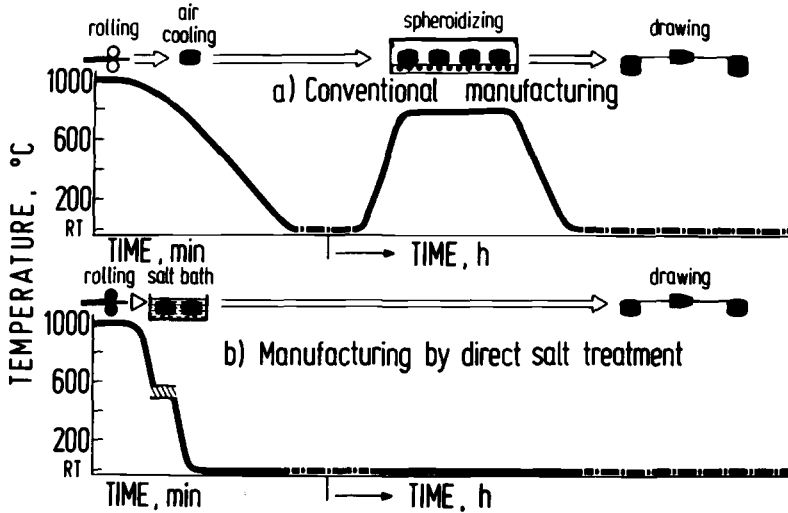


FIG. 15—Production routes for wire-rods of bearing steel DIN 100 Cr 6.

Conclusion

Developments which have enhanced product quality and reduced production costs in the manufacturing of bearing steels have been considered. The facilities, available in the works of Krupp Stahl, enable a cost-competitive production of bearing-steel products on a high-quality level. Bearing manufacturers' latest requirements for chemical composition, nonmetallic inclusions, and micro-structure can be achieved by adequate melting and casting practices as well as by rolling, heat-treatment and other processing, which are aligned to the application of the products. Further developments in melting and processing of bearing steels should be supported by reliable quantitative knowledge about the correlation between structural features and the fatigue properties of bearings.

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Oxygen Content, Oxidic Microinclusions, and Fatigue Properties of Rolling Bearing Steels

REFERENCE: Lund, T. and Åkesson, J., "Oxygen Content, Oxidic Microinclusions, and Fatigue Properties of Rolling Bearing Steels," *Effect of Steel Manufacturing Processes on the Quality of Bearing Steels*, ASTM STP 987, J. J. C. Hoo, Ed., American Society for Testing and Materials, Philadelphia, 1988, pp. 308–330.

ABSTRACT: Steels used in rolling bearing applications require high reliability and consistency. The SKF MR process developments carried out at Ovako Steel have led to the design of cleanliness specifications which correspond to such requirements. The reduction of oxygen content and harmful nonmetallic inclusions have led to significantly improved fatigue properties of rolling bearing steels.

KEY WORDS: bearing steel, cleanliness, fatigue, fracture toughness, inclusions, inclusion rating, steelmaking

The fatigue properties of steels heat treated to high hardness are to a great extent controlled by the contents of hard nonmetallic inclusions. Control of fatigue properties thus means control of the nonmetallic inclusions. High-quality steel production must be designed in such a way that the chemical compositions, the contents, and the size distributions of the nonmetallic inclusions result in a steel with good fatigue resistance.

Ovako MR Process

The Ovako MR process comprises two production steps, one melting stage, M, under oxidizing conditions and one refining stage, R, under reducing conditions. The melting operation produces a liquid raw steel which then is refined in an ASEA-SKF ladle furnace. In the ASEA-SKF ladle furnace a number of operations can be performed as illustrated in Fig. 1. After finishing, rolling bearing steels are uphill teemed to ingots at Ovako Steel.

Deoxidation

Deoxidation is carried out with aluminium. The final total oxygen content is mainly determined by the aluminium content and the effective stirring time in the ladle furnace. As aluminium is added, alumina inclusions are formed. The oxides are effectively removed by inductive stirring. Figure 2 shows the total oxygen content of two groups of heats of Society of Automotive Engineers (SAE) 52100 steel with different aluminum-contents versus time in the ladle furnace. Each group comprises seven heats.

Inductive stirring is more effective than argon gas stirring as inclusion remover in the ASEA-SKF ladle furnace. Figure 3 is a direct comparison of the oxygen content distributions of MR produced SAE 52100 steel stirred inductively and with argon, respectively.

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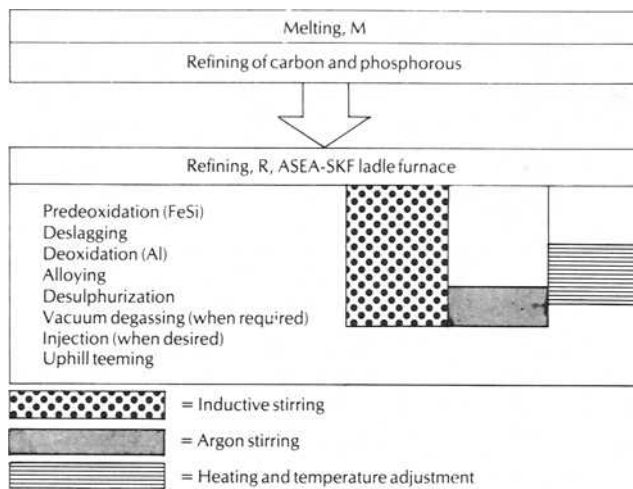


FIG. 1—MR process.

Conversely, argon stirring is more effective for desulfurization and for hydrogen removal under vacuum. In the ASEA-SKF ladle furnace, both inductive stirring and gas stirring can be used and combined.

Teeming

The bottom pouring (uphill teeming) practice used for rolling bearing steels at Ovako Steel results in a very limited oxygen pickup in the liquid steel. This is due to the short and compact teeming stream between ladle and trumpet. Additional shrouding, which has been proven effective in improving the quality of top poured steel, is not normally necessary when using Ovako Steel's practice.

To evaluate the effect of active gas shrouding, a direct comparison was made on four heats of SAE 52100, steel. These were teemed with an argon shrouding on Teeming Plate Number 1, while Plates Number 2 and Number 3 were teemed without additional shrouding. Table 1 compares the oxygen contents, nitrogen contents, and microinclusion contents of samples from Plates Number 1 and Number 2. An additional shrouding will only marginally affect the oxygen content when a well-designed bottom pouring practice is used.

TABLE 1—Comparison of contents from sample Plates Numbers 1 and 2.

Heat Number	Teeming			
	With Ar-Shrouding		Without Ar-Shrouding	
	O, ppm	N, ppm	O, ppm	N, ppm
1	11.0	79	11.0	92
2	14.0	100	13.5	107
3	12.0	100	13.0	108
4	8.5	72	9.0	79
Average	11.4	88	11.6	97

Oxygen, (ppm)

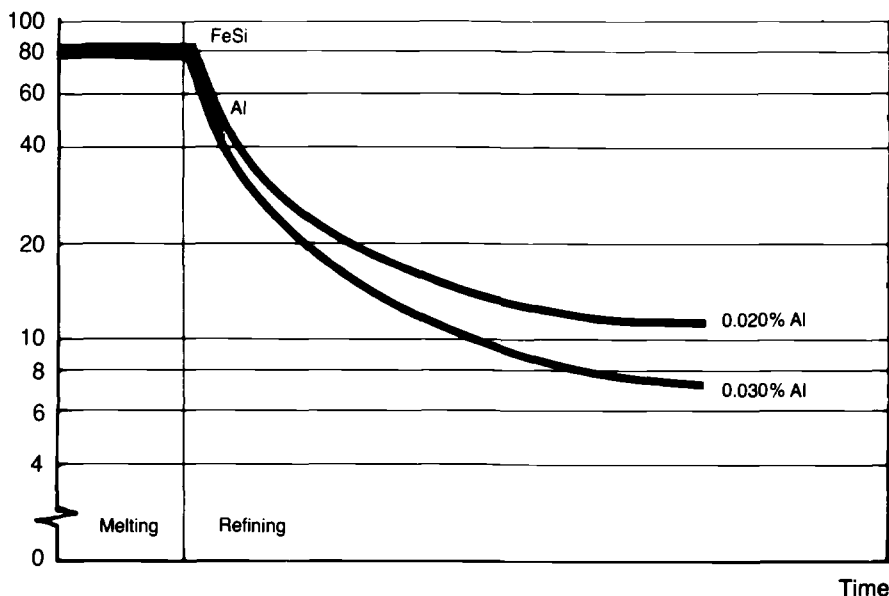


FIG. 2—Total oxygen content in liquid steel during ASEA-SKF ladle furnace treatment.

Fatigue Properties

Fatigue failures most frequently originate from hard nonmetallic inclusions as oxides and titanium-carbonitrides. Recent years have given a significant improvement in steel cleanliness.

The large influence of oxygen content on the fatigue properties has been discussed in several publications, and a summary can be found in Ref. 1. In bearings, oxygen and oxides have been proven to be harmful and directly relatable to bearing life. Sulfur and sulfides have been shown [2], not to reduce fatigue properties of bearings up to a level of at least 0.04%. Titanium-carbonitrides have been shown to initiate bearing failures.

The bearing life evolution since the 1930s is shown in Fig. 4, which gives the lives attained in testing of specially made SKF 6309 bearings under the same stress conditions over a long time period. The reduction in life during the early 1940s depends on the substitute steels used during World War II. The improvement in the life of rolling bearings attained in the last decade to a large extent has been attributed to the development in steel cleanliness.

To verify this, a direct comparison between a "modern" heat and an experimentally made old, "bad" heat was made. The steel quality of the "bad" heat is representative for the early 1960s.

The "modern" heat used in the comparison was a standard MR produced SAE 52100 steel heat. The "bad" heat was produced in an electric arc furnace using conventional two-slag practice (no ladle furnace treatment). Table 2 shows the chemical compositions and the inclusion ratings for the two heats. Steel from the two heats was processed into SKF 6309 bearings in the same way. The bearings were endurance tested under the same loading conditions as those used for the tests shown in Fig. 4. In Fig. 5, the result of the "modern" to "bad" comparison is shown. A life advantage of about ten times is noted for the MR steel. Improvements in steel quality very significantly improves bearing performance.

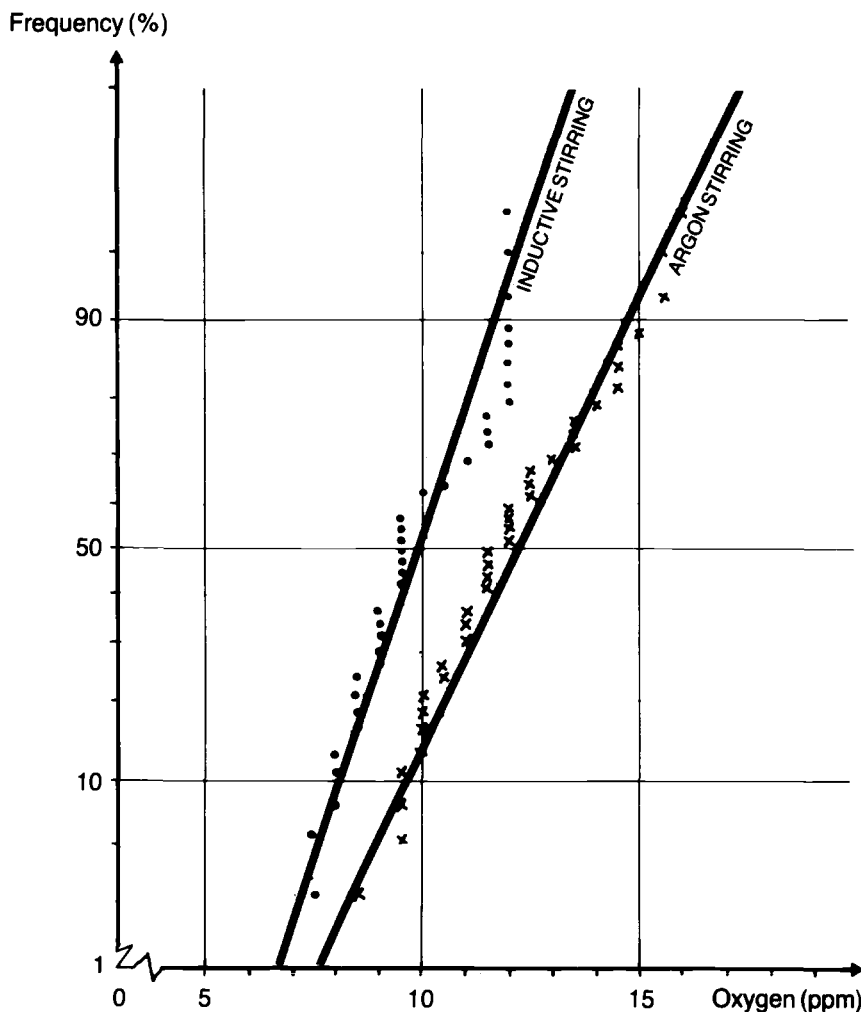


FIG. 3—Final oxygen content of 38 heats of SAE 52100 steel which were inductively stirred and 38 heats which were argon-stirred in the ASEA-SKF ladle furnace.

It is worth noting that the “bad” heat only is just above the current ASTM Specification for High-Carbon Ball and Roller Bearing Steel (A 295) (B thin is 2.5 instead of 2.0), within the ASTM A 485 and well within ASTM Specification for Carburizing Steels for Anti-Friction Bearings (A 534).

Fatigue Testing

There are a number of test methods in use today to estimate the fatigue properties of bearing steels. Bearing testing is very costly and time-consuming. Over the years, two main types of simplified rolling contact fatigue tests have become widely used. *Flat-washer* testing uses a ball set which is rotated against a flat washer-type specimen. The washers are produced from sections which are taken perpendicularly to the rolling direction. *Rod-testers* use a bar specimen, normally taken in the rolling direction. The bar is stressed either by balls or rollers.

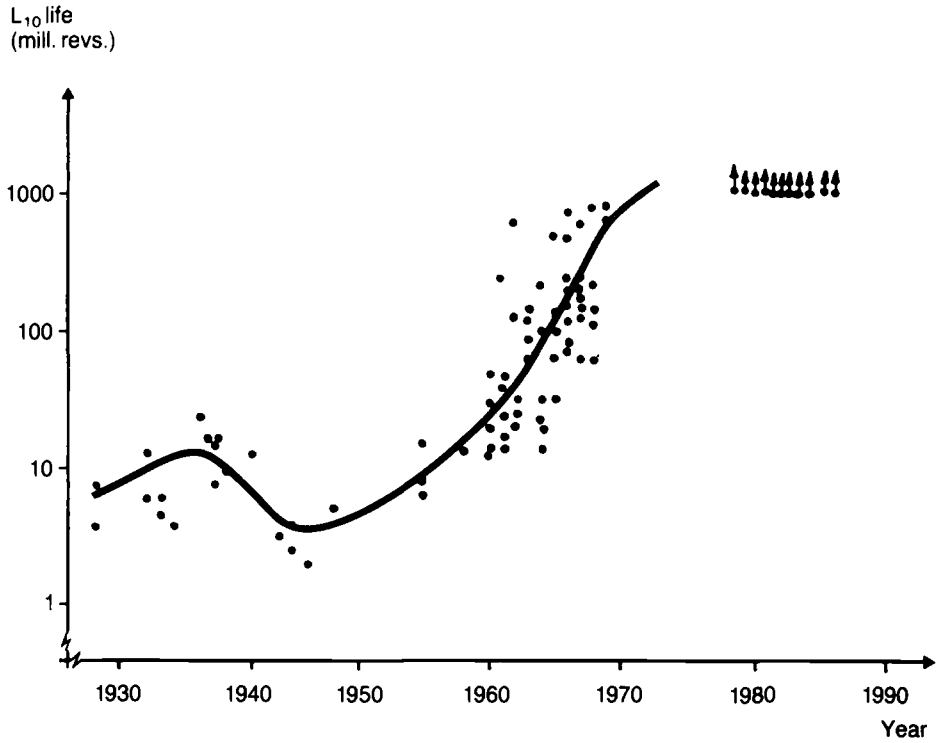


FIG. 4—Endurance tests on specially made SKF 6309 bearings. SAE 52100 steel.

TABLE 2—Chemical compositions and inclusion ratings for the two heats.

	MR	"Bad" EAF
CHEMICAL COMPOSITIONS		
Carbon, %	1.04	1.00
Silicon, %	0.23	0.21
Manganese, %	0.35	0.34
Phosphorus, %	0.010	0.010
Sulfur, %	0.013	0.013
Chromium, %	1.46	1.35
Nickel, %	0.07	0.06
Molybdenum, %	0.02	0.01
Copper, %	0.152	0.132
Aluminum, %	0.030	0.014
Oxygen, ppm	11	39
Titanium, ppm	11	10
ASTM A 295 INCLUSION RATINGS		
A: Th	1.5	1.5
He	0.8	1.0
B: Th	0.7	2.5
He	0	0.5
C: Th	0	0
He	0	0
D: Th	0	0.5
He	0	0

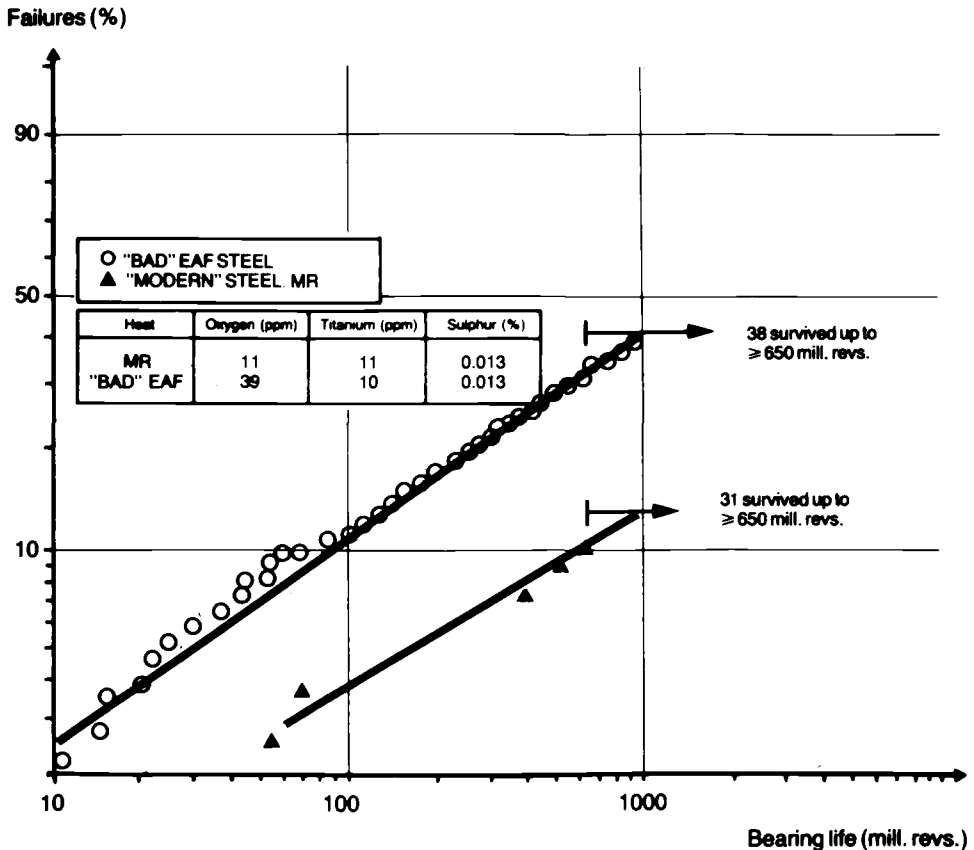


FIG. 5—Endurance tests on MR and "bad" EAF SAE 52100 steel.

There are a large number of test results published from such tests. The tests use a very high contact stress, which leads to plastic deformation and groove formation. These simplified test methods often are used to perform screening, or acceptance, tests on bearing steels. It is thus of vital importance to have some certainty that the results obtained at very high stress level do relate to bearing performance.

Flat Washer Tests

Flat washer testing was used within SKF and SKF Steel for a long time, but was abandoned about ten years ago because correlation could not be found between full bearing and washer tests [3]. In contradiction to this, several published results are available [1] which do show correlation between flat washer test results and metallurgical quality indicators as the oxygen content.

In recent times, a number of flat washer tests have been run on MR steel intentionally produced to different quality levels. The results, which are shown in Fig. 6 and Table 3, show results which contradict bearing tests. Factors other than those determining bearing life obviously can play a major role in high-stress washer testing, such as microstructural strength and composition variations.

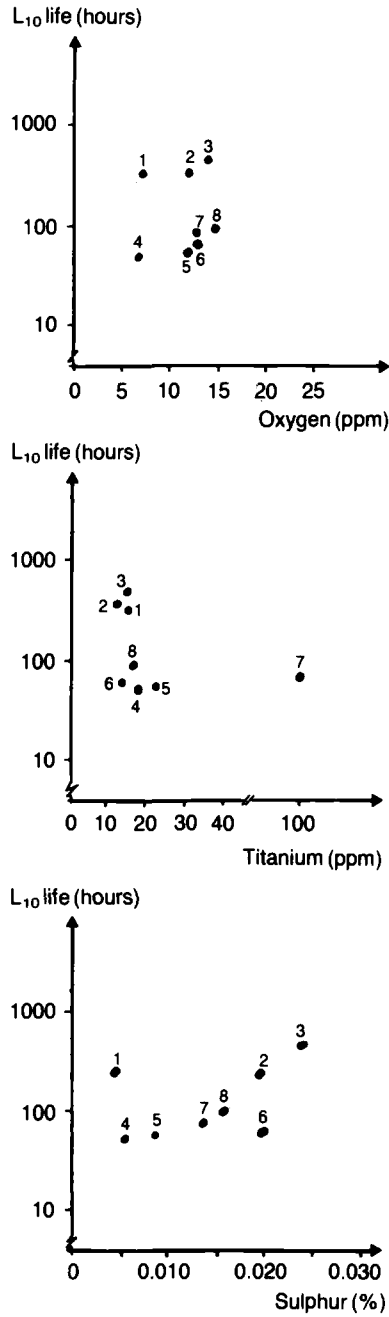


FIG. 6—Flat washer tests on SAE 52100 steel.

TABLE 3—Results of flat washer tests.

Test Number	Oxygen, ppm	Titanium, ppm	Sulfur, %	L_{10}
1	8	15	0.005	321
2	12	12	0.020	378
3	14	14	0.024	447
4	7	18	0.006	49
5	12	21	0.008	54
6	13	14	0.020	62
7	13	100	0.014	67
8	15	16	0.016	93

Rod Tests

Two types of rod testers are in use today, Polymet testers, which use a roller to bar contact, and RCF (rolling contact fatigue) testers, which use a ball to bar contact. Polymet tests were run on steel from the modern and "bad" heats earlier discussed (Fig. 5). The Polymet results, obtained under standard Polymet testing conditions, are shown in Fig. 7. Evidently, the large difference in fatigue properties between a "good" and a "bad" steel in bearing tests is not reproduced in the Polymet test. RCF tests were run under standard testing conditions on a number of SAE 52100 variants. The results are shown in Fig. 8 and Table 4.

As for flat washer tests, it seems that factors other than those determining bearing life are influencing the test results. The data published elsewhere on high-stress, simple geometry tests,

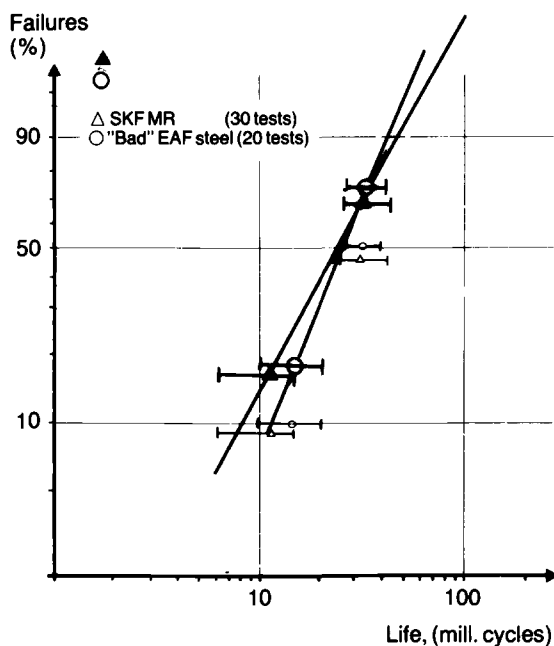


FIG. 7—Polymet tests on MR and "bad" EAF SAE 52100 steel. Same steels as in Fig. 5.

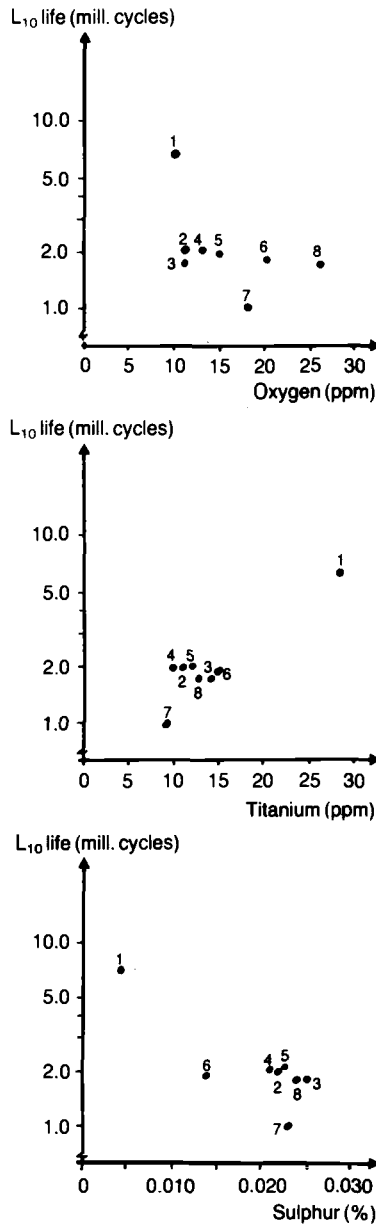


FIG. 8—RCF tests on SAE 52100 steel.

however, indicate that it can be possible to produce relevant data by such test methods. Certainly, data produced in high-contact stress testing should be viewed with considerable care when used to predict bearing performance, and more work must be devoted to establishing the correlation between bearing endurance and simplified rolling-contact test data.

TABLE 4—Results of Polymet rod test.

Test Number	Oxygen, ppm	Titanium, ppm	Sulfur, %	L_{10}
1	10	29	0.004	6.4
2	11	11	0.022	2.0
3	11	14	0.025	1.7
4	13	10	0.021	2.0
5	15	12	0.022	2.0
6	20	15	0.014	1.8
7	18	9	0.023	1.0
8	26	13	0.024	1.7

Rotating Beam Tests

Rotating beam and similar tests such as push-pull are sometimes used to evaluate rolling bearing steel performance. One advantage with this type of test is that the failure initiation cause often can be assessed. When it is a nonmetallic inclusion, its chemical composition can be determined.

The same comparison between the MR heat and the "bad" EAF heat as previously discussed was made in rotating beam testing. Figure 9 shows the results obtained. All samples

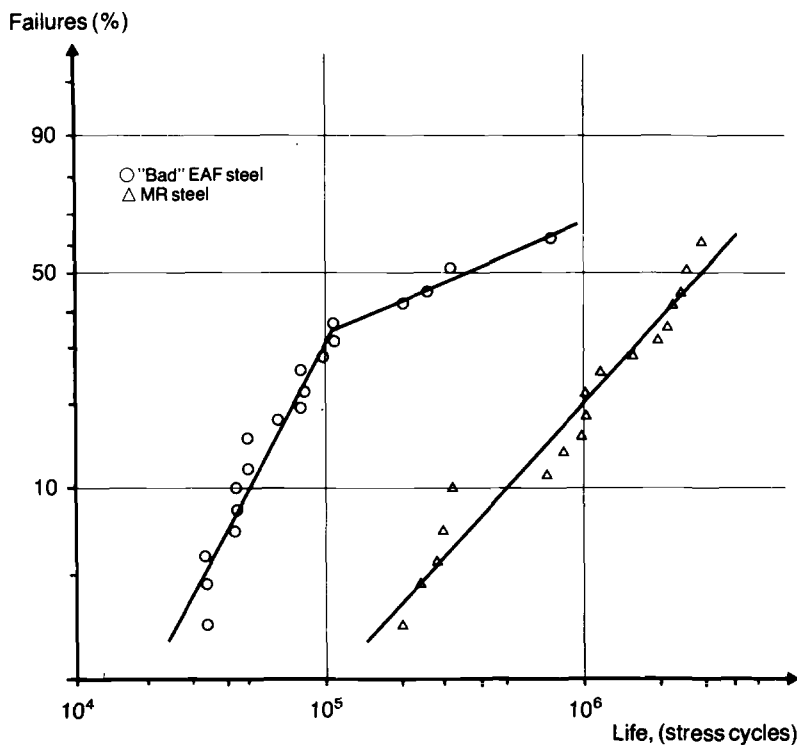
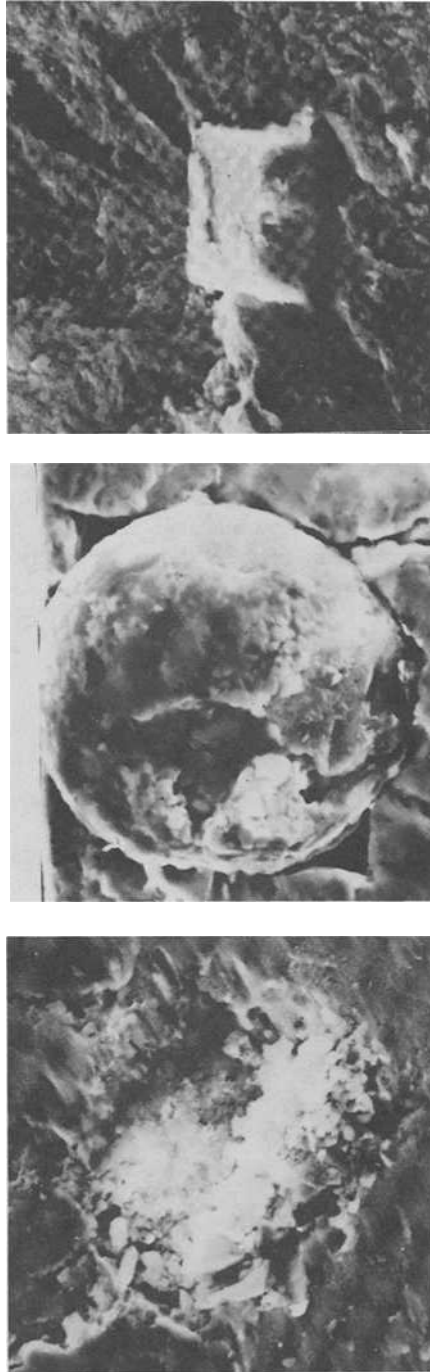


FIG. 9—Rotating beam tests on MR and "bad" EAF SAE 52100 steel. Stress level 1050 N/mm^2 .



B-type inclusion

D-type inclusion

T-type inclusion

FIG. 10—*B-type, D-type, and T-type inclusions at initiation sites in SAE 52100. $\times 710$.*

TABLE 5—Initiation frequency caused by hard metallic inclusions.^a

Sample	MR	"Bad" EAF Heat
Alumina (B-type)	35%	3%
Calciumaluminates (D-type)	20%	97%
Titaniumcarbonitrides (T-type)	45%	0%

^aC-type inclusions were not present in the steels.

were examined to determine the failure causes. For the samples where the initiation was caused by hard nonmetallic inclusions, the initiation frequency was as shown in Table 5.

Figure 10 shows examples of B-, D- and T-type initiating inclusions. By the failure analysis, it was possible to generate the joint individual failure distributions for B-, D-, and T-type inclusions.

Figure 11 shows the result. D-type inclusions are far more harmful than B- and T-type inclusions.

Fracture Toughness

Fracture mechanics considerations are sometimes required in bearing applications.² Normally, only a very small portion of a components fatigue life is spent during the crack propagation stage. In recent times, two aspects of fatigue crack propagation rates have been studied: oxidic inclusion contents and copper contents. For the MR and "bad" EAF steel variants

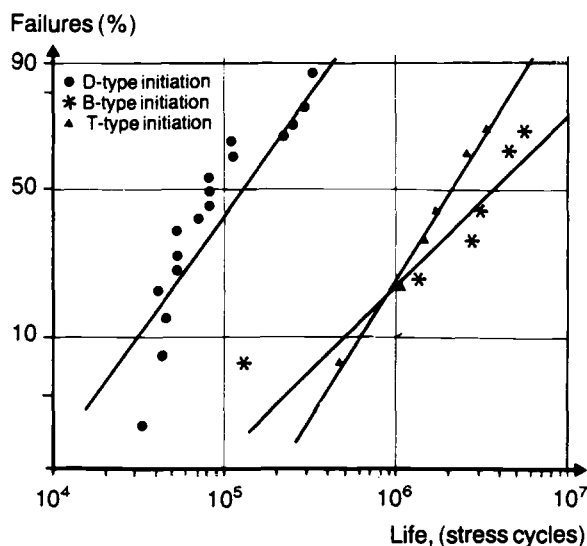


FIG. 11—Rotating beam fatigue of samples initiated by B-type, D-type, and T-type inclusions, respectively. SAE 52100 steel.

²Beswick, J., "Fracture Mechanics Concepts and Applications for Bearing Steels," *SKF Steel Technical Report*, Aug. 1984. SKF Steel Technical Reports can be obtained from Ovako Steel AB, S-813 00 Hofors, Sweden, or through the symposium chairman, J. J. C. Hoo, c/o American Society for Testing and Materials, 1916 Race Street, Philadelphia, PA, 19103.

earlier discussed, a direct comparison of crack propagation rates was made. As shown in Figs. 12 and 13, no difference exists between these variants, and neither does the difference in oxidic inclusion content influence the K_{Ic} value. In a second test, a MR heat produced with 0.37% copper was compared to the above heats. As evidenced, no difference in crack propagation rate can be noticed with the increased copper content.

Oxygen Determination

All oxygen in solidified steel is present in the form of oxides. The total oxygen content thus can be related to the total content of oxidic nonmetallic inclusions. When determining oxygen content, two factors are of prime importance: the oxygen standards used and the sample preparation technique.

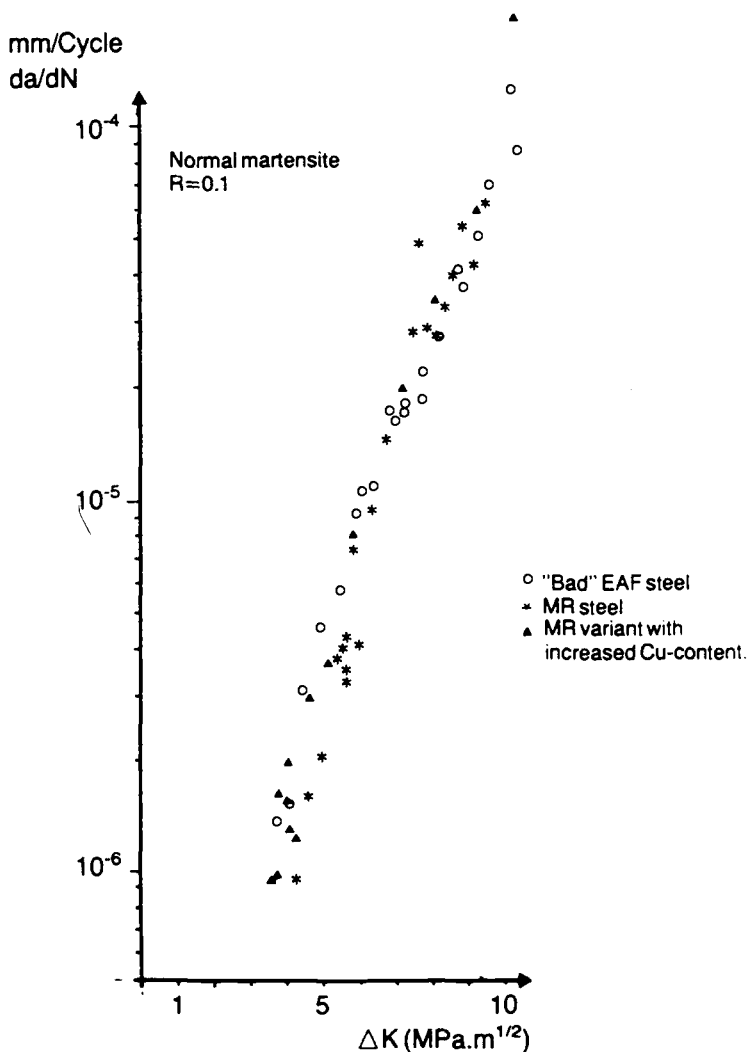
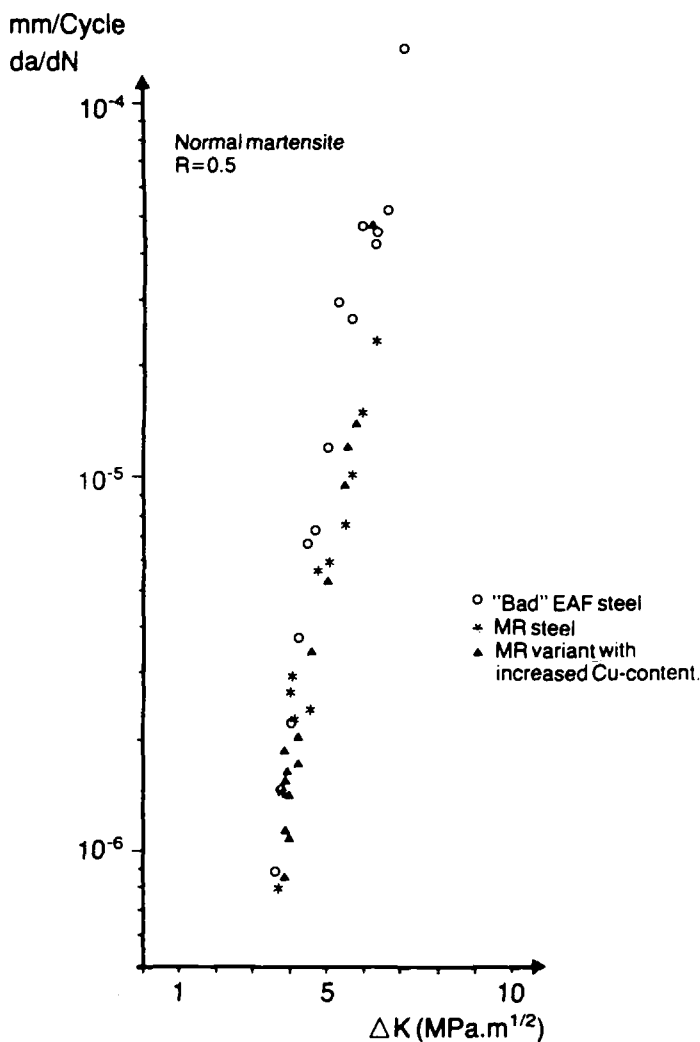


FIG. 12—Crack growth rates in SAE 52100 steel. Martensitic condition. $R = 0.1$.

FIG. 13—Same as Fig. 12 but $R = 0.5$.

Standards for Oxygen Analysis

The reference standards used must be consistent and stable. An oxygen standard is defined by a certified average value and a standard deviation. These values are based on tests carried out at several laboratories.

Sample Preparation

The sample preparation in connection to oxygen analysis is extremely critical. The sample preparation technique must in all instances prevent oxidation of the samples. The sample preparation technique used today at Ovako Steel involves a number of steps to attain consistent and reliable oxygen readings. The procedure is as follows (see Fig. 14):

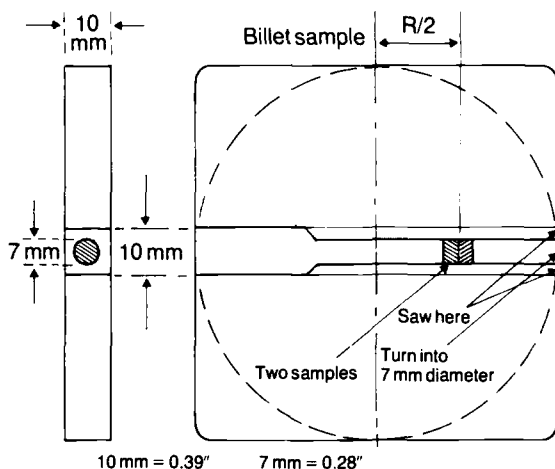


FIG. 14—Oxygen sampling.

1. Sample the billet as normal for inclusion ratings.
2. Cut off a disk of 10 mm thickness from the billet cut at the desired position. For SAE 52100 the cutting makes soft annealing necessary.
3. Cut a central portion, 10 mm thick, from the billet disk.
4. Turn the 10 by 10 mm section into a 7-mm-diameter bar.
5. After turning, hand file the bar, and cut off two samples, 3 to 5 mm long, at the half radius position. Hand file the surfaces on the cut sides. Always handle the samples with tweezers. Never touch the samples. Put the samples into protective glass or aluminium boxes.
6. Brush the samples free of debris and wash them in *n*-hexane or acetone. Dry the samples.

The sample preparation shall be made in direct connection to the analysis. This procedure has been successfully used within Ovako Steel for a long time. The correlation between the oxygen analysis performed at Ovako Steel's laboratories in Hofors and Hellefors is shown in Fig. 15.

Oxygen Distribution

To assess the variation in oxygen content within one heat, a study was made on a standard produced MR heat of SAE 52100 steel. For this heat, oxygen samples were taken from top billet positions from the 23 production ingots after rolling to 140-mm-diameter billets. In addition, one ingot from the first teeming plate was sampled at five equally spaced billet positions. Figure 16 shows the oxygen contents (average of two tests) measured within this ingot. The total range is 2 ppm, with an average of 9.3 ppm and a standard deviation of 0.8 ppm. Figure 17 shows the oxygen contents measured on the 23 ingot top positions. The total range is 5 ppm, with an average of 10.0 ppm and a standard deviation of 1.5 ppm. As indicated by Fig. 18, this implies that the oxygen content variation within one heat of MR produced SAE 52100 steel is the same as the oxygen content variation in the National Bureau of Standards (NBS) oxygen standard 1094.

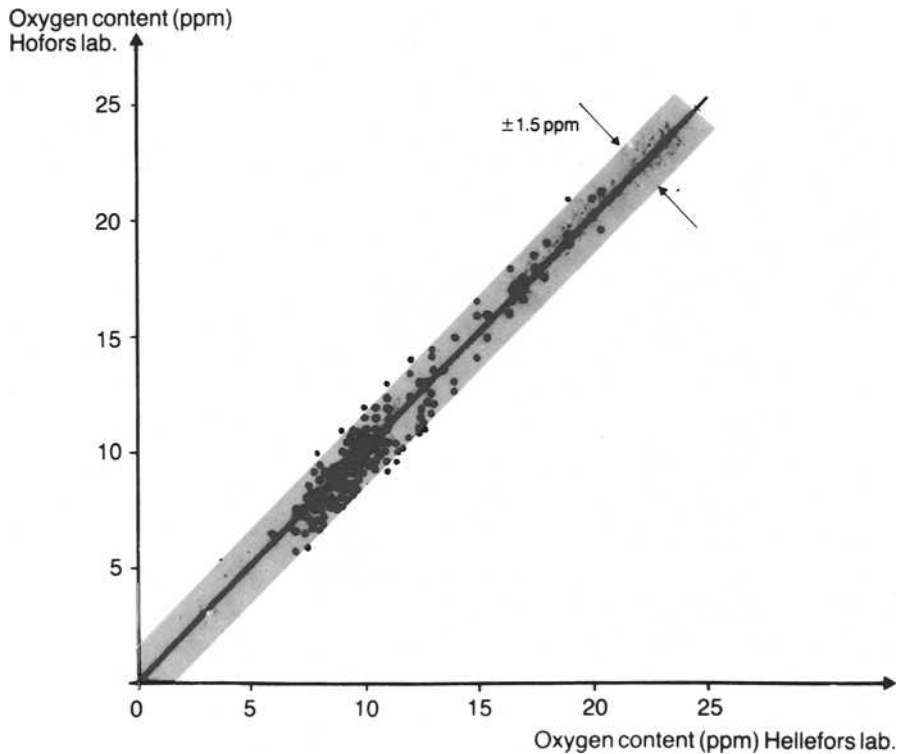


FIG. 15—Comparative oxygen determinations at two Ovako Steel laboratories. Random samples taken 1984 to 1986 on low-alloy steels.

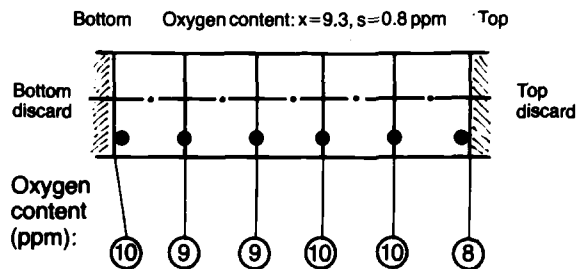


FIG. 16—Oxygen content variation within one ingot of MR produced SAE 52100 steel.

Inclusion Rating Procedures

The Ovako Steel cleanliness specifications limit analyzed contents of oxygen *and* oxidic microinclusion ratings. The microinclusion ratings are given as JK/ASTM ratings because this is the by far most common and least time-consuming inclusion rating system in use today. Inclusion ratings provide two things which oxygen analysis does not: inclusion chemistry (the

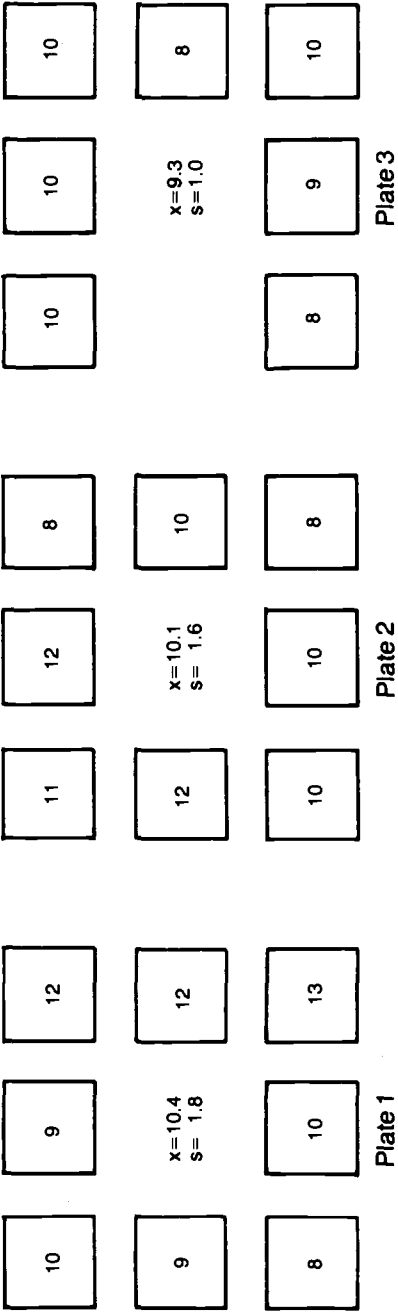


FIG. 17—Oxygen contents of billet top positions in 23 ingots of one MR produced heat of SAE 52100 steel.

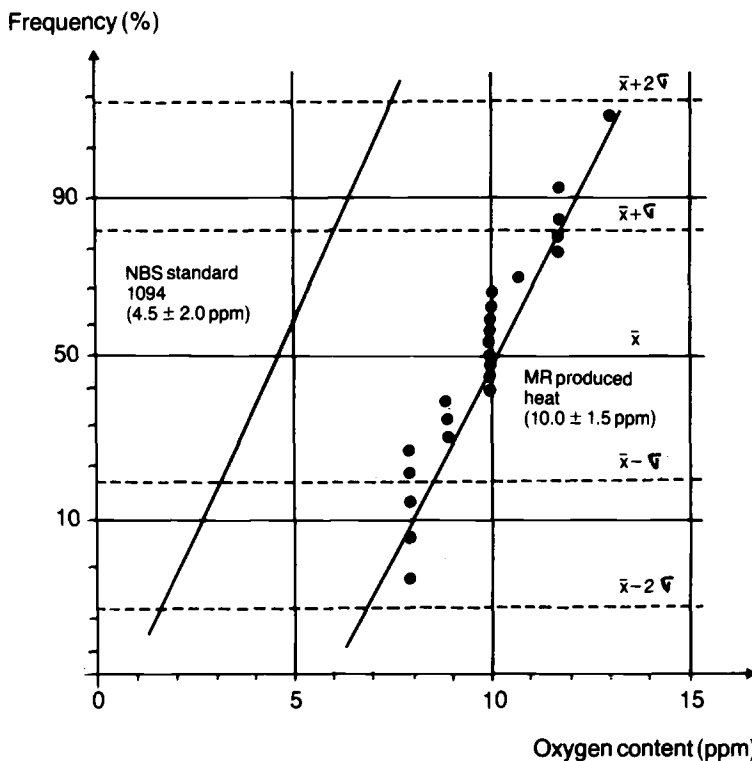


FIG. 18—Oxygen content distribution in one MR produced heat of SAE 52100 steels as measured in the top position of 23 ingots.

inclusion type classification) and an estimate of the inclusion size distribution. The combination of oxygen determination, which gives a very good measure of the total oxidic microinclusion content, and the type classification is vital to a good specification. Several inclusion rating systems are in use today, and what follows is a brief review.

JK/ASTM Ratings

The JK scale was developed at SKF Steel in the 1940s, and the JK/ASTM rating system is in worldwide use since then. In the ASTM A 295 procedure, which is an industrial standard today, a six-sample average is derived from worst field ratings on billets from the top and bottom of three different ingots per heat. Figures 19 to 21 relate JK/ASTM ratings for SAE 52100 steel produced at SKF Steel over a 15-year time period to analyzed oxygen and sulfur contents. In Fig. 19, the B-type thin rating is plotted against the analyzed oxygen content. To change the average JK/ASTM result 0.5 units, it takes a change in average oxygen content of about 5 ppm. A reduction in oxygen content gives a significant reduction in scatter in JK/ASTM rating. If *only* a JK/ASTM rating is specified, this means that a large scatter in total inclusion content (oxygen content) is tolerated. If *only* the oxygen content is specified, this effectively maximizes the JK/ASTM rating *as well as* the total oxide inclusion content.

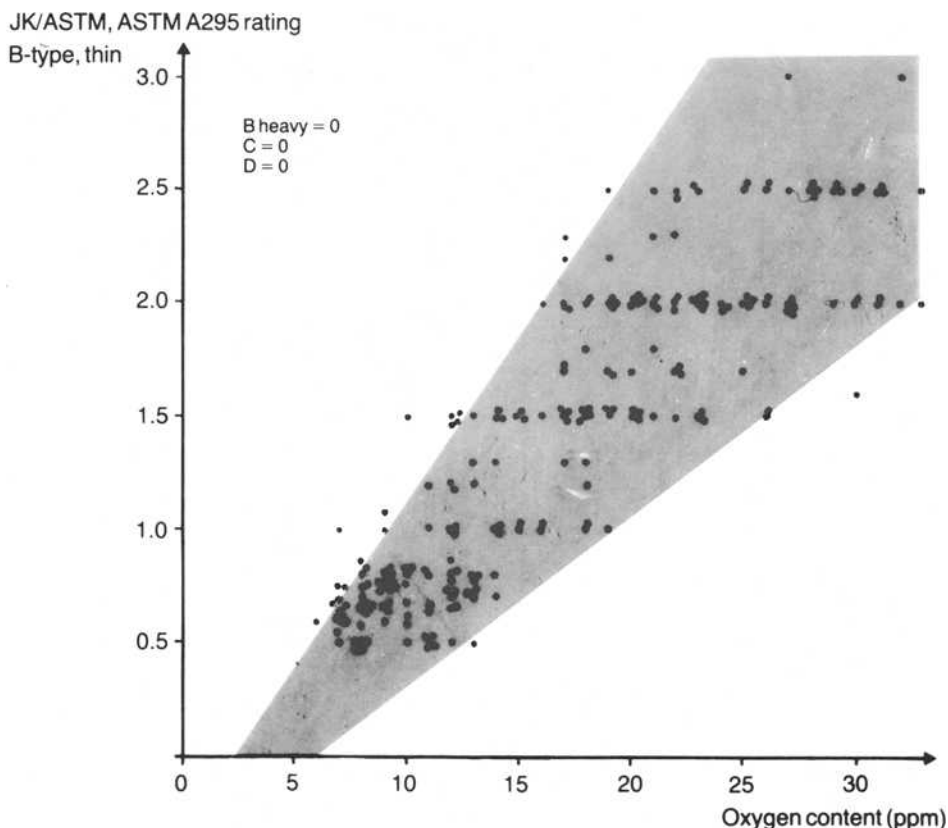


FIG. 19—ASTM A 295 B-type thin rating versus oxygen content for 200 random heats of SAE 52100 steel manufactured at SKF Steel 1971 to 1986. Three t-ingots rolled to 140-mm-diameter billets.

By the use of Figs. 19 to 21, meaningful specifications for JK/ASTM ratings and oxygen/sulfur contents can be developed. However, the inclusion rating results are affected by factors as steel composition, solidification rate (ingot size), and degree of reduction.

SAM Ratings

As rolling bearing steel has become cleaner, the JK/ASTM ratings have become less meaningful. The SAM rating developed by Sherry A. Majoy, New Departure Hyatt, adds an estimate of the inclusion content to the JK/ASTM rating. Figure 22 shows a summary of SAM ratings performed on MR produced SAE 52100. Because these heats were free from D-type and B-type heavy, only B-type thin fields larger than or equal to 1.5 are SAM counted. As earlier shown, few heats have 1.5 fields at the 10 ppm oxygen level.

SEP Ratings

The German SEP (DIN 50602) rating system attempts a further quantification by field counting and a weighting system is used in the evaluation. However, practical use has proven

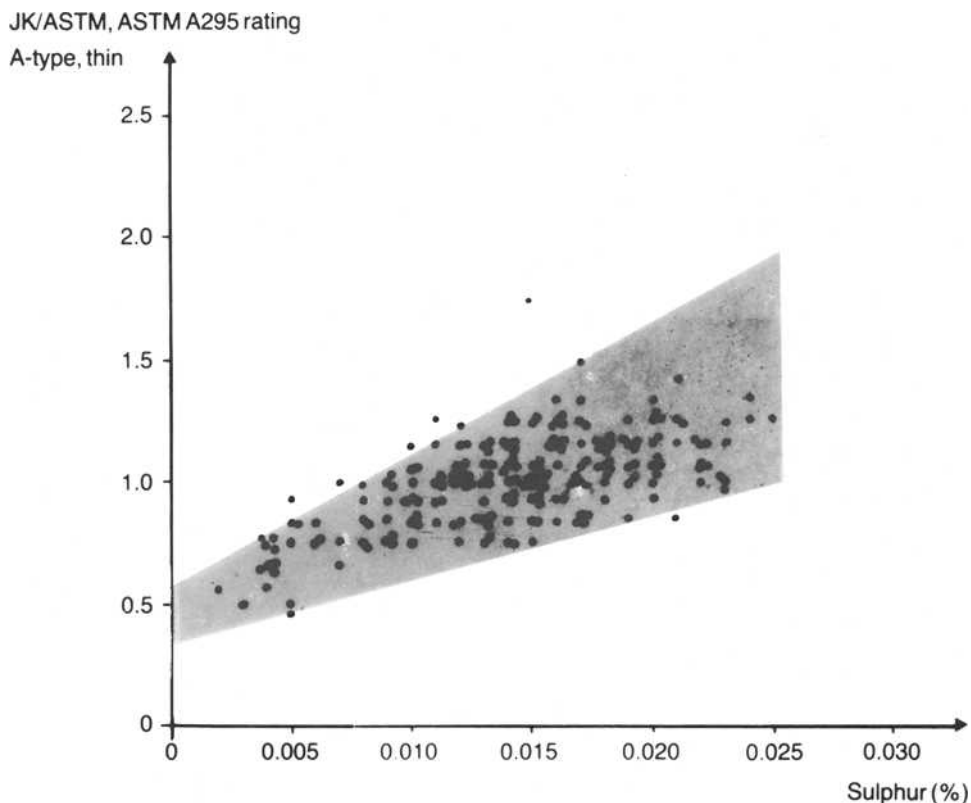


FIG. 20—ASTM A 295, A-type thin rating versus sulfur content for 200 MR produced heats of SAE 52100 steel. Three t-ingots rolled to 140-mm-diameter billets.

difficult. The large scatter experienced in manual assessments is mainly due to problems in chart interpretation.

Revised JK System

This rating system is presented in detail in Swedish Standard SS 111116. The initial tests done with the revised JK rating system at Ovako Steel indicate that the system is most versatile and certainly can be put to very good use to better quantify the inclusion content of modern steel.

Conclusions

The development of steel cleanliness in recent years has led to a very significant increase in rolling bearing steel fatigue resistance. Specifications for high-quality steels must include limits for macroinclusions, oxygen, and titanium to ensure high reliability and long service life of cyclically stressed components. The oxygen content of steel is a very good indicator of metallurgical quality, but it should be complemented with inclusion ratings in order to ascertain that C-type and especially D-type inclusions are limited to very low levels. Rolling contact tests

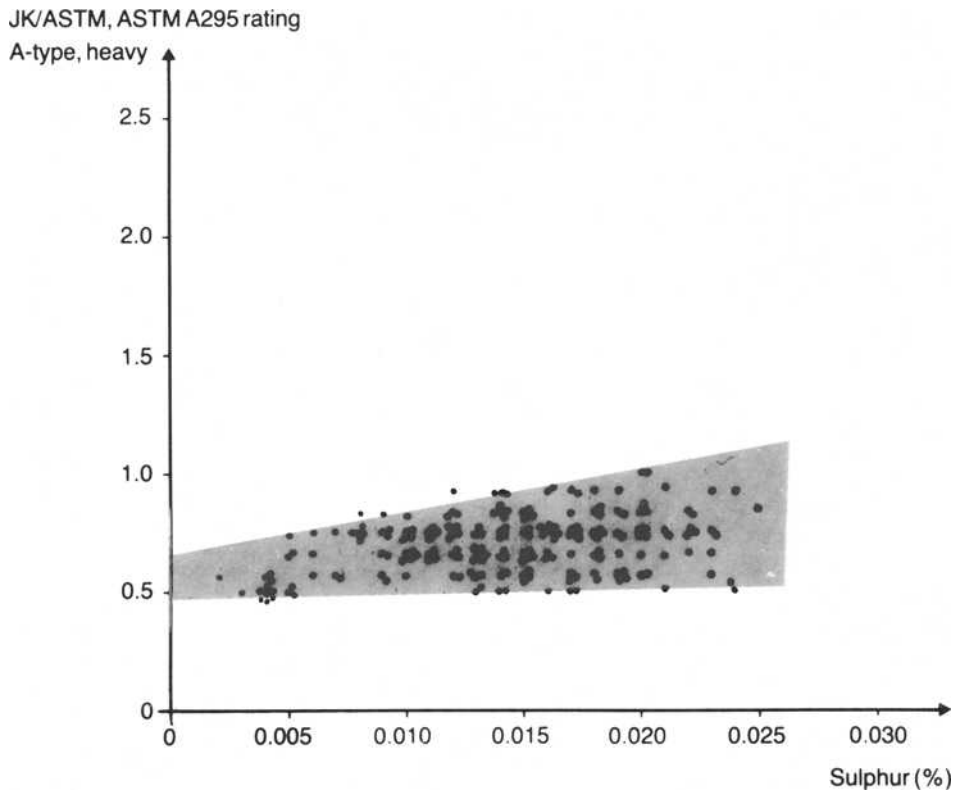


FIG. 21—Same as Fig. 22, but ASTM A 295, A-type heavy ratings.

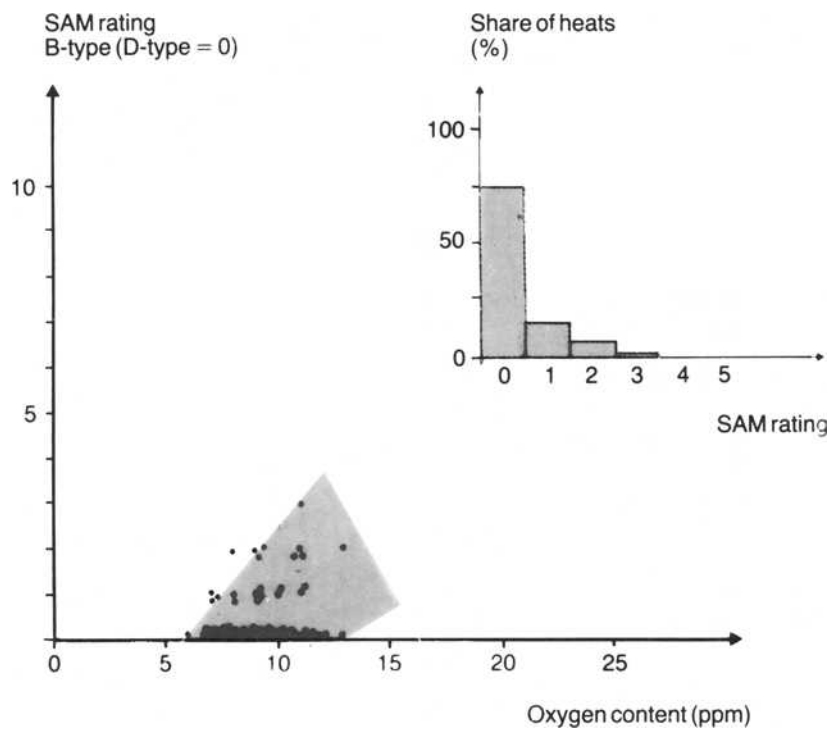


FIG. 22—SAM ratings on 100 heats of MR produced SAE 52100 produced 1985 to 1986.

performed at high stress levels should be used with great caution for prediction of bearing performance.

References

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- [2] Bäckström, S., "Effect of Sulphide Inclusions on the Endurance of Ball Bearings," *Clean Steel*, Vol. 1, Report of the Royal Swedish Academy of Engineering Sciences, Vol. 169, No. 1, Stockholm, 1971, pp. 170–177.
- [3] Johnston, G. et al., "Experience of Element and Full-Bearing Testing of Material over Several Years," *Rolling Contact Fatigue Testing of Bearing Steels, ASTM STP 771*, American Society for Testing and Materials, Philadelphia, pp. 190–205.

The Effects of Ladle Refining, With and Without Vacuum, on Bearing Steel Quality

REFERENCE: Whittaker, D. A., "The Effects of Ladle Refining, With and Without Vacuum, on Bearing Steel Quality," *Effect of Steel Manufacturing Processes on the Quality of Bearing Steels*, ASTM STP 987, J. J. C. Hoo, Ed., American Society for Testing and Materials, Philadelphia, 1988, pp. 331-347.

ABSTRACT: Three manufacturing procedures have been developed to produce bearing steels of varied cost and quality requirements: argon ladle refining, providing the lowest cost route for general bearing applications in C-Mn steels; argon ladle refining under full vacuum, for more specialized bearing applications in C-Mn and Ni-Cr-Mo alloy steels; and argon ladle refining under full vacuum plus vacuum arc remelting, for the highest cost and most critical bearing applications. Product qualities are discussed in terms of mechanical properties (hardenabilities, ideal diameters), chemistry control (C, Mn, S, O, Ti) and steel cleanliness (ASTM E 45, SAM, and AMS stepdown) ratings.

KEY WORDS: bearing steels, chemistry control, cleanliness standards, hardenability control, inclusion ratings, ladle refining, vacuum treatment

Atlas Speciality Steels produces various bearing steels, such as Society of Automotive Engineers (SAE) 1070, 52100, and 4320, for specific customers with differing end-use requirements (automotive hubs and spindles, roller bearings, and bearing races, respectively) and with differing quality testing requirements involving chemistry ranges, ideal diameter and hardenability ranges, the ASTM Practice for Determining the Inclusion Content of Steels, (E 45-85), SAM, and stepdown cleanliness maxima.

The common customer requirement is for control of the major elements, carbon, manganese, and chromium, to ensure product mechanical properties. For those elements, performance is measured by chemical limits and can be complicated by requests for ideal diameter and hardenability ranges. More recently there have been requests for control of minor elements: oxygen and titanium for fatigue life and sulfur for machinability, as measured by chemical limits alone.

Requirements discussed less but still common to all steel users are control over product size tolerances and surface conditions (freedom from imperfections, often including decarburization) and occasionally some structural limitations. Two final tasks are to produce these quality requirements with competitive prices and delivery schedules.

To be successful, the steelmaker then must:

1. Obtain the customer's desired chemistry and ideal diameter and hardenability limits at the melting/refining stage.
2. Maintain control over chemistry at the teeming stage, avoid reoxidation during pouring, entrapment of exogenous inclusions and segregation during solidification.

¹Manager, Process and Product Development, Atlas Specialty Steels Division, Rio Algom Limited, P.O. Box 1000, Welland, ON L3B 5R7 Canada.

3. Subsequently provide the proper billet/bar geometry and surface quality.
4. Minimize costs.
5. Deliver ordered quantities on time.

Equipment for Bearing Steel Manufacture

The processing routes available at Atlas Specialty Steels are shown schematically in Fig. 1. For bearing steels, three melting routes of increasing complexity and resultant quality are available beyond the initial electric arc furnace stage: open ladle refining under a slag cover (A/R or argon refining); roof covered ladle refining with arc reheating and vacuum degassing

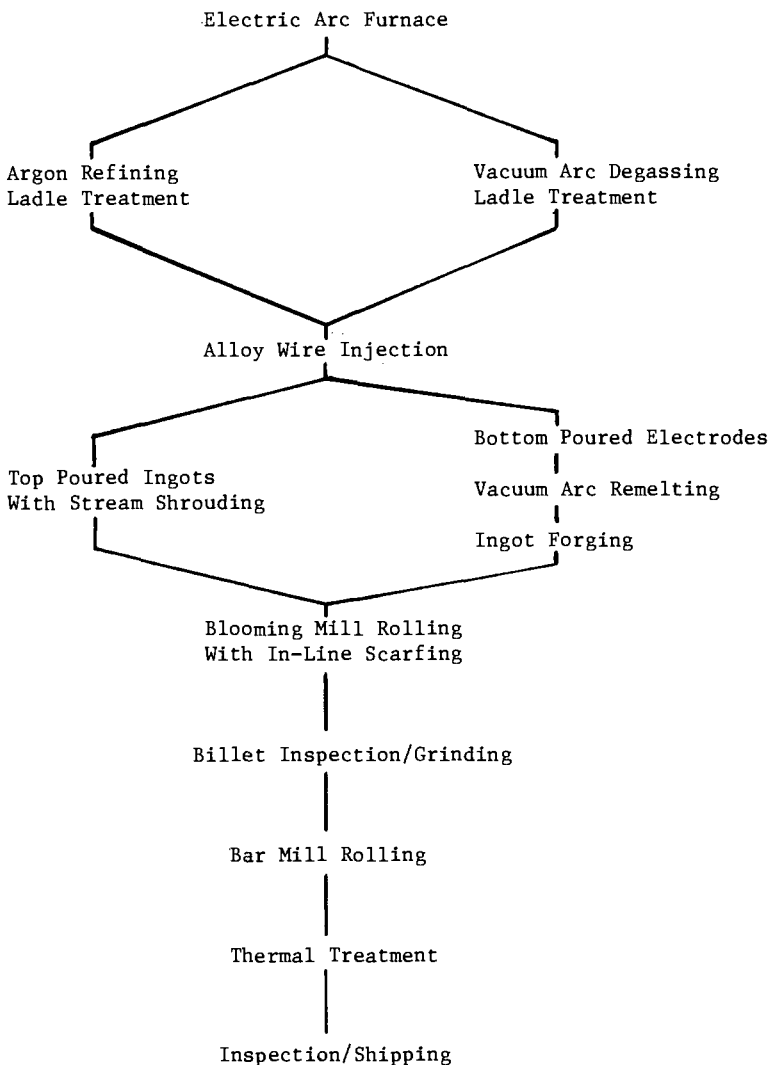


FIG. 1—Processing routes for bearing steels.

(VAD or vacuum arc degassing); and the latter procedure followed by vacuum arc remelting (VAR).

There are two Whiting Hydro Arc electric furnaces, details of which are given in Table 1. These furnaces are cold charged with scrap, alloy, and flux materials, following a DEC2060 computer calculation of least-cost charges from inventories on hand. A VAX750 computer is then used to calculate the electric energy inputs, volume of oxygen injected during meltdown and alloy additions to the furnaces as well as to the subsequent ladle refining stations. Samples are sent by pneumatic tube to a laboratory for analyses utilizing the following equipment: a Leco CS 244 for carbon and sulfur, a Leco RH101 for hydrogen, a Leco TC36 for oxygen, an ARL Quantovac emission spectrometer for titanium, and most other elements via a Phillips PW1600 X-ray spectrometer.

The standard ladle used for argon refining is shown schematically in Fig. 2. The ladle is 4 m (13 ft) high and 2.7 m (9 ft) in diameter at the upper rim and is fitted with an argon porous plug located in the ladle bottom 30 cm (12 in.) from the centerline. That site was found most effective for steel mixing, based on half-scale water model studies. The ladle is also fitted with 3.8- to 5.0-cm (1½ to 2-in.) diameter slide gate nozzles, depending on the ingot sizes to be teemed. The ladle refractories are carbon-enriched dolomite at the slag line and resin bonded dolomite for the lower walls and bottom. The argon plug is a magnesite refractory; the plug assemblies, nozzle blocks, and slide gate valves are 75 to 85% alumina refractories.

In the argon refining process, the steel is stirred with a gentle stream of argon bubbles injected through the porous plug, under a basic lime/fluorspar slag with no ladle cover and no vacuum applied. Alloys and deoxidants are added during the refining procedure, which takes approximately 35 min. The stirring action and pronounced slag/metal contact results in an effective removal of oxide inclusions, a uniform and closely controllable metal temperature before teeming, and good metal compositional control.

The vacuum arc degassing ladle is of the same basic design as that used for argon refining but is somewhat larger to handle the degassing reactions. The ladle is 4.6 m (15 ft) high, providing a 1.2-m (4-ft) free board above the slag line, and is 3 m (10 ft) in diameter at the upper rim. The same argon porous plug and slide gate nozzle assemblies are used. The ladle linings are of magnesium-oxide enriched dolomite at the slag line and burned dolomite in the lower wall and bottom.

In the VAD process two ladle covers are sequentially used, as shown schematically in Fig. 3. For reheating, the ladle is placed under a water-cooled steel roof, fitted with three electrodes, each 30.5 cm (12 in.) in diameter and together drawing 6500 kVA power. For degassing, the

TABLE 1—*Electric arc furnace details.*

Specification	Value
Nominal capacity	75 short tons
Shell diameter	5.2 m (17 ft)
Electrodes	superior grade, 51 cm (20 in.) diameter
Transformer capacity	30 MVA
Power input	27 MVA
Charging access	top charged
Tap changing	on load
Water cooled panels	cast design
Refractories:	
Roof	high alumina
Slagline	carbon magnesite
Bottom	high MgO

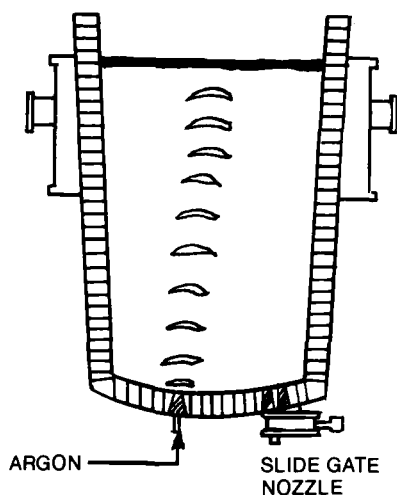


FIG. 2—Schematic view of Argon refining ladle.

ladle is placed under a second roof fitted with a sight port, vacuum offtake, oxygen lance (to allow vacuum oxygen decarburization of stainless steels), and a peripheral O-ring attached to a water-cooled annulus. For bearing steel manufacture, the oxygen lance equipment is not required. The complete roof descends such that the O-ring makes a vacuum tight seal on a 18-cm (7-in.) flange, located 18 cm (7 in.) below the ladle rim. The vacuum system consists of a water ring pump and five steam ejectors, to achieve pressures under 1 torr.

As shown in Fig. 4, a carousel system with two ladle carriages is used to provide freedom of ladle movement between the two roof stations. The two independently controlled carriages move around a central pivot such that when a ladle is standing at any one location a second ladle has ready access to all of the remaining locations.

Following the ladle refining treatments, the molten steel may be injected with sulfur cored wire using a Vallourec wire feeder. The wire is 16 by 7 mm (0.63 to 0.28 in.) in outside dimension and is fed from a reel at 100 m/min (350 fpm) to hit the bottom of the ladle before melting and releasing its pure sulfur core. Argon stirring is continuous, throughout the ladle refining and wire injection treatments.

Six 0.7 m by 10 000-kg (28 in. by 22 000 lb) electrodes may be bottom-poured on a single pouring plate for subsequent vacuum arc remelting to 0.76-m-diameter (30 in.) ingots. The VAR furnace is a Leybold Heraeus, two-station unit with power of 30 kA at a nominal 25 volts. This furnace has been equipped with the latest (1986) system for computer control of melt rates and arc gaps, the latter by sensing either voltage changes, drip shorts, or a combination of those.

Various ingot sizes may be teemed for direct rolling, a common size being a round, fluted 0.55-m diameter by 2165-kg (22 in. diameter by 4770 lb) ingot. These are top-poured from teeming cars, with the metal streams protected from reoxidation by argon shrouding. This procedure requires that the mold tops be covered and a steel pipe fitted with a diffusion bulb inserted through the cover to purge the individual molds with argon gas just before teeming. A circular "spray ring" is also fixed to the base of the slide gate assembly, and argon is blown through this ring to envelope the falling ladle stream. A third argon line floods the gate itself, to prevent oxygen aspiration between the sliding plates.

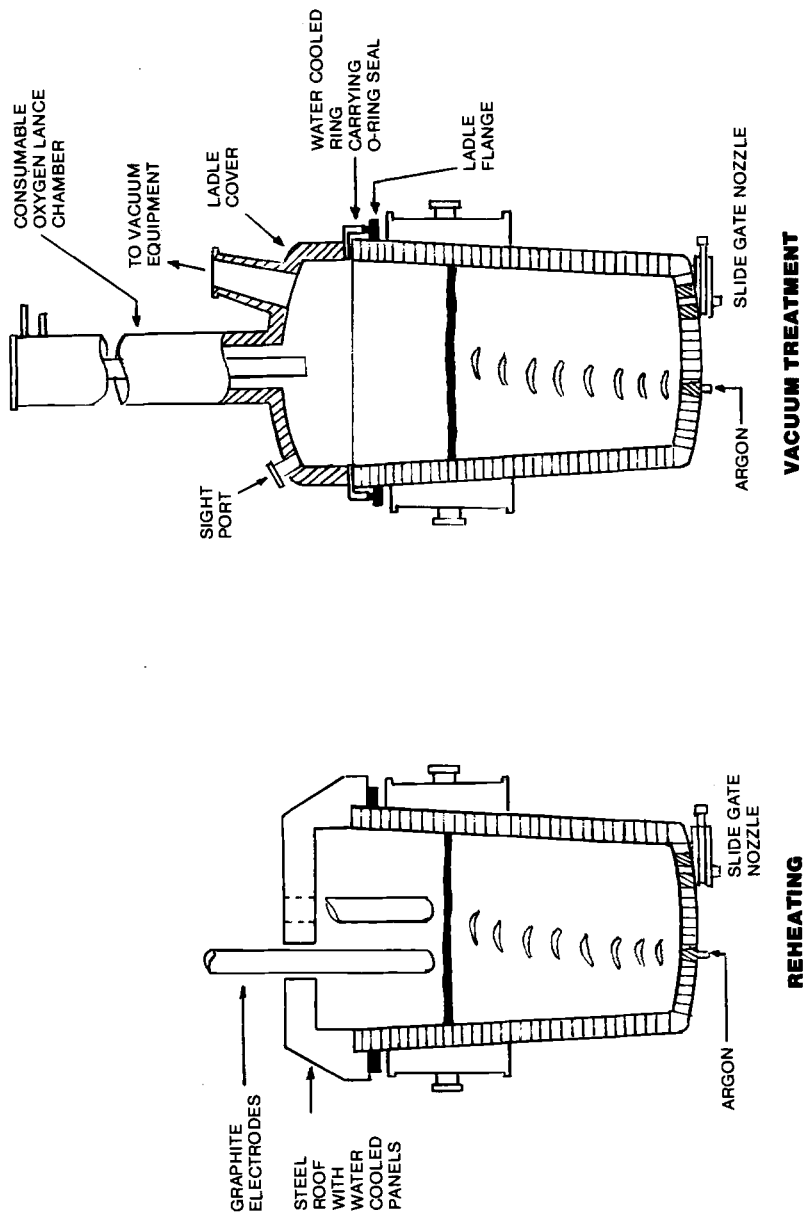


FIG. 3—Schematic view of ladle and covers for vacuum arc degassing.

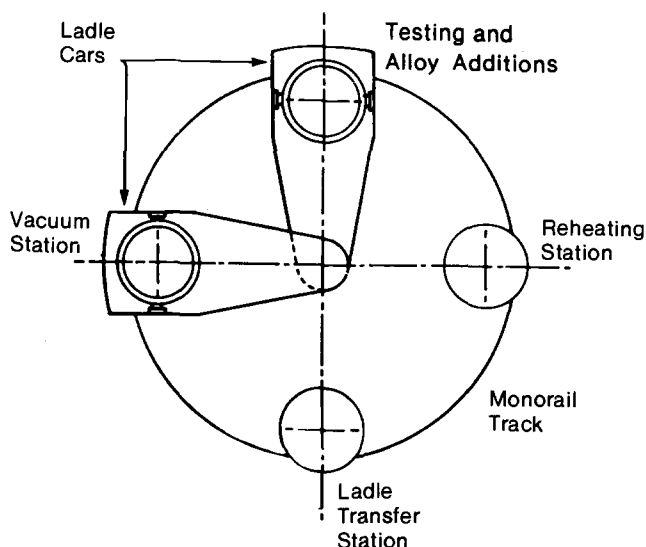


FIG. 4—Carousel system for reheating, sampling, and vacuum treatment.

Melt Shop Procedures for Bearing Steel Manufacture

Vacuum Arc Degassing

The most used procedure for manufacture of bearing steels is the VAD route shown in Table 2. This procedure is employed for both VAD-treated products and for steels subsequently vacuum arc remelted.

There are several points of interest in the VAD procedure. After melting down with both electric power and oxygen injection, carbon levels are in the order of 0.10%, and the oxygen content is high. The steel is then carbon deoxidized by injection of powdered carbon. Manganese, silicon, and chromium alloys, calculated to meet the bottom of the customer-specified range, are added to a slag-free bath just before the furnace is tapped. Since these are mild deoxidizers, some precipitation deoxidation occurs at this stage. Following this deoxidation, the balance of the carbon is added during tapping to attain the customer range.

Argon stirring is initiated in the ladle during tapping of the electric furnace and continues throughout the ladle refining process. At the ladle refining station a new lime slag is added and the steel is reheated, after which trim alloy additions are made and the steel is degassed. At this point the basic chemistry has been fixed. The thorough mixing by continued argon stirring results in excellent temperature uniformity within the ladle. The high degree of slag/metal contact very effectively removes sulfur and precipitated oxides, producing a clean steel. With degassing pressures under 2 torr for several minutes, hydrogen levels of 1.5 ppm are attained. (Final hydrogen levels of course may be higher, with extended post-vacuum exposures to the atmosphere.)

The procedure following Step 14 is optional, depending on the degree of chemistry control and hydrogen removal desired. This includes Step 20, at which point cored wire may be fed to the ladle for precise sulfur control.

TABLE 2—VAD procedure for bearing steels.

Station	Process
Electric arc furnace	1. Meltdown selected steel scrap, fluxes. 2. Test for chemistry, temperature. 3. Remove meltdown slag, inject carbon. 4. Reapply power to attain tapping temperature. 5. Test for carbon, temperature.
Ladle at electric arc furnace	6. Add Mn, Si, Cr to bottom of customer range. 7. Tap into ladle with argon flowing.
Alloy addition station	8. Add C to bottom of customer range.
Under reheating cover	9. Test for chemistry, temperature. 10. Add new flux cover. 11. Reheat with continued argon stirring.
Alloy addition station	12. Add C, Mn, Si, Cr to customer specification limits.
Under vacuum cover	13. Degas for 20 min with 14 min below 2 torr.
Alloy addition station	14. Test for chemistry, temperature.
Under reheating cover	15. Reheat as necessary.
Alloy addition station	16. Make trim alloy additions as necessary, for example, for ideal diameter requirements.
Under reheating cover	17. Reapply power to ensure alloy mixing, precise temperature control.
Under vacuum cover	18. Optional degas for 10 min to ensure low hydrogen, thorough mixing.
Alloy addition station	19. Add aluminum to ensure 0.016% in product for grain refinement.
Ladle at wire injection station	20. Add S to specification requirements. 21. Send to teeming car.

Argon Refining

The procedure for argon refining, as shown in Table 3, follows the VAD procedure only to Step 6 in the electric arc furnace. After removal of the meltdown slag and addition of alloys, a new lime/fluorspar slag is fused in the furnace, and this slag plus the metal is tapped to the argon refining ladle. During tapping, carbon is again added to the specification range. Aluminum is added during tapping for precipitation deoxidation purposes.

During the subsequent refining period, it is important to minimize metal exposure to the atmosphere by maintaining a good slag cover. Most of the features associated with the VAD process are achieved at this time: good temperature uniformity within the ladle; thorough alloy mixing for chemistry control; and rapid removal of sulfur and deoxidation products. Tapping temperatures must be higher than for the VAD process, to allow for final trim alloy additions before teeming and to ensure sufficient temperature to avoid teeming problems. Despite the obvious inability to reheat and hold the steel for very precise alloying, this process has shown a capability to meet most customer specifications. Its major shortcoming is control over hydrogen levels, such that downstream flake prevention cycles may be necessary for large billets.

Vacuum Arc Remelting

The long electrodes used for vacuum arc remelting are bottom-poured and hot-topped for surface quality and internal soundness. The hot top region is torch-cut off and the electrodes

TABLE 3—*Argon refining procedure.*

Station	Process
Electric arc furnace	1. Meltdown selected steel scrap, fluxes. 2. Test for chemistry, temperature. 3. Remove meltdown slag, inject carbon. 4. Reapply power to attain tapping temperature. 5. Test for carbon, temperature. 6. Add Mn, Si, Cr to bottom of customer range, plus lime/fluorspar fluxes.
Ladle at electric arc furnace	7. Tap into ladle with argon flowing.
Ladle held adjacent EAF for argon refining	8. Add C to bottom of customer range, Al for deoxidation. 9. Test for chemistry, temperature. 10. Maintain a gentle argon flow for 20 min or more. 11. Add trim alloy additions as necessary. 12. Test for chemistry. 13. Add aluminum to ensure 0.016% in product for grain refinement, when temperature 11°C (20°F) above teeming temperature.
Ladle at wire injection station	14. Add S to specification requirements. 15. Test for temperature. 16. Send to teeming car.

100% ground to remove surface oxides before remelting. The electrodes are also remelted in the top-up position so that any non-metallics in the extreme top position are confined to the remelted ingot top, for later removal by shear cropping during forging/rolling. The actual remelting procedure is computer-controlled from the power initiation to final hot topping stages.

Major Element Chemistry Control

Using the VAD Practice

Carbon—For 1070 steel the American Iron and Steel Institute (AISI) range is 0.65 to 0.75% carbon. A major customer cites that range and in addition requests an “aim range” of 0.69 to 0.71% carbon. As shown in Fig. 5, the actual results are all between 0.68 and 0.73%, with the customer’s aim carbon range achieved for 85% of heats made (on 180 heats surveyed).

For 52100 Grade 1 steel, the ASTM Specification for High Hardenability Bearing Steels (A 485-79) and one major customer specification allow a 0.90 to 1.05% carbon range, which is readily achieved with all results falling between 0.96 and 1.02%; that is, the same 0.06% carbon working range is obtained for both the 1070 and 52100 steels.

Manganese—The AISI 1070 range for manganese is 0.60 to 0.90%. In this case a customer has set a higher “allowed-range” at 0.80 to 1.10%, with a tighter “aim range” of 0.90 to 1.00%, basically to achieve narrowed ideal diameter requirements. The actual results have been between 0.80 and 1.05%, with the aim range achieved in 76% of cases, as shown in Fig. 6.

For SAE 52100, the ASTM A 485-79 range is 0.95 to 1.25% manganese. One customer has requested tighter limits of 1.00 to 1.20%, and these are readily achieved.

Silicon—For 52100 steel, both ASTM A 485-79 and a customer specify a 0.45 to 0.75% silicon range. Our experience indicates one-half of that range is achievable.

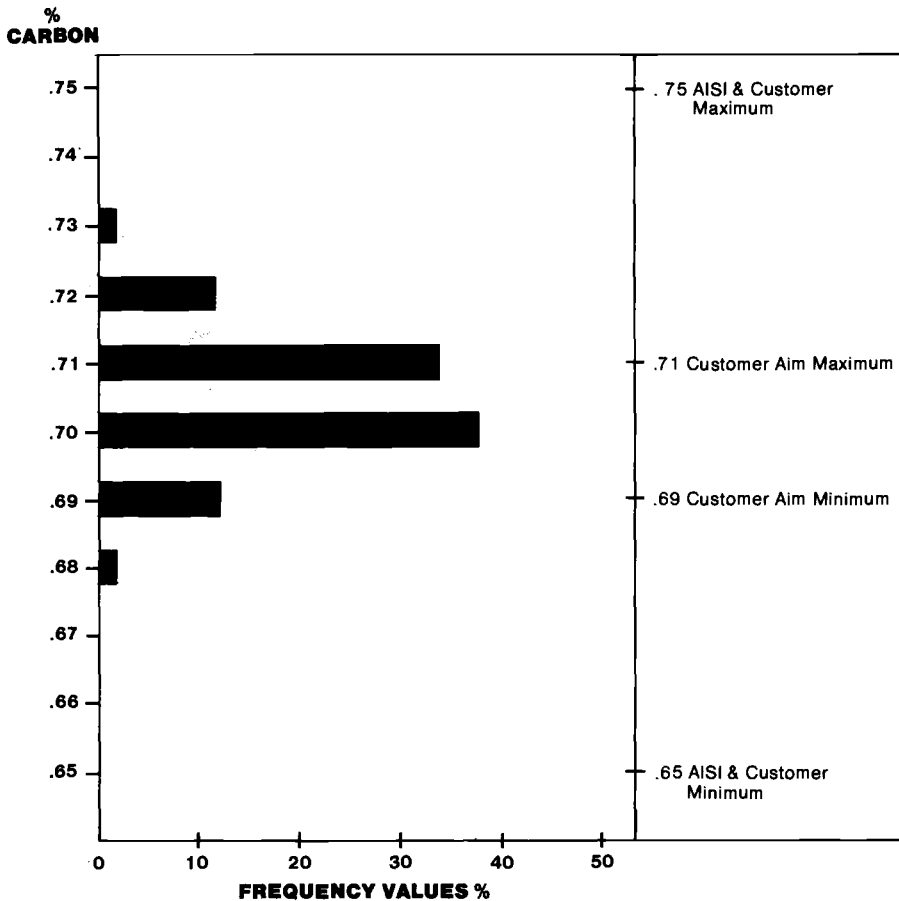


FIG. 5—Carbon contents of AISI 1070 heats.

Chromium—For 52100 steel, ASTM A 485-79 requires 0.90 to 1.20% chromium, but one customer has requested a 0.95 to 1.15% range. Chromium is relatively easy to control using ladle furnace techniques, and better than one-half of the latter range is achievable.

Using the Argon Refining Practice

As practiced by Atlas Specialty Steels, argon refining does not allow reheating and prolonged holding of the steel once it has been tapped from the electric arc furnace. Major element control therefore is not as precise, but is still well within the usual customer requirements for high carbon bearing steels. Table 4 indicates the ranges within which the above elements are controllable by both the VAD and A/R processes.

Ideal Diameter and Hardenability Results

The steelmakers' improving ability to control major elements by ladle treatments, correspondingly improves control over ideal diameter and hardenability results. Shown in Fig. 7 are

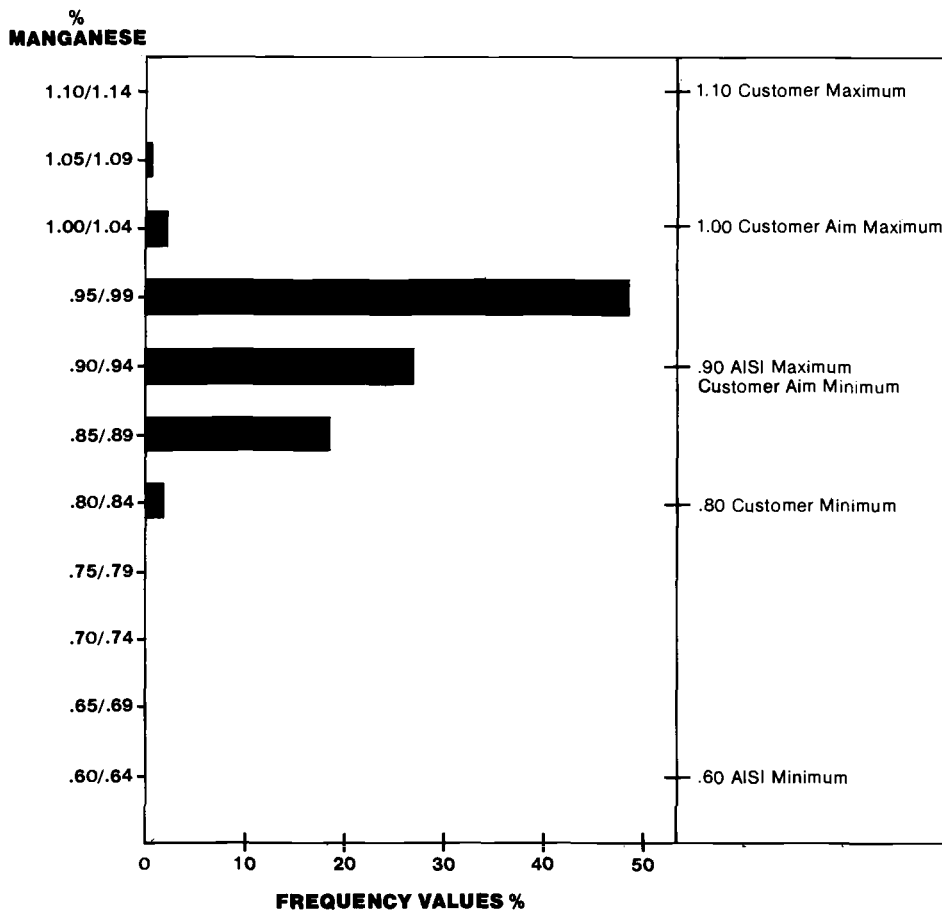


FIG. 6—Manganese contents of AISI 1070 heats.

four ranges in ideal diameter values for 1070 steel. The extreme limits possible from the overall chemical specification are 1.54 to 3.12. Based on end-use requirements, the customer specifies two ranges from 2.00 to 2.30 and 1.90 to 2.20. It is desirable, therefore, that the melt shop meet the limited range 2.00 to 2.20, and that has been achieved with 98% consistency on the latest 180 heats surveyed.

Jominy hardenability values are normally specified as a part of 52100 steel requirements. ASTM A 485-79 does not specify limits for hardenability, but the customer specifying the

TABLE 4—Control ranges for major elements in 1070 and 52100 bearing steels achieved by both the VAD and A/R process.

Element	Range, %
Carbon	±0.03
Manganese	±0.12
Silicon	±0.05
Chromium	±0.03
Sulfur	±0.007

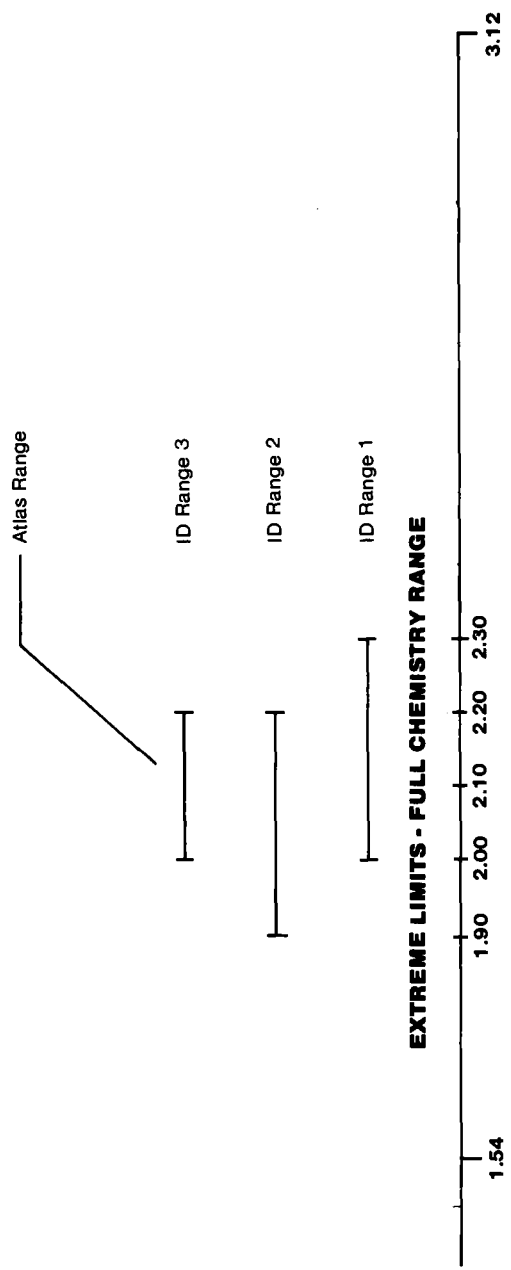


FIG. 7—Ideal diameter ranges for SAE 1070 bearing-quality steel.

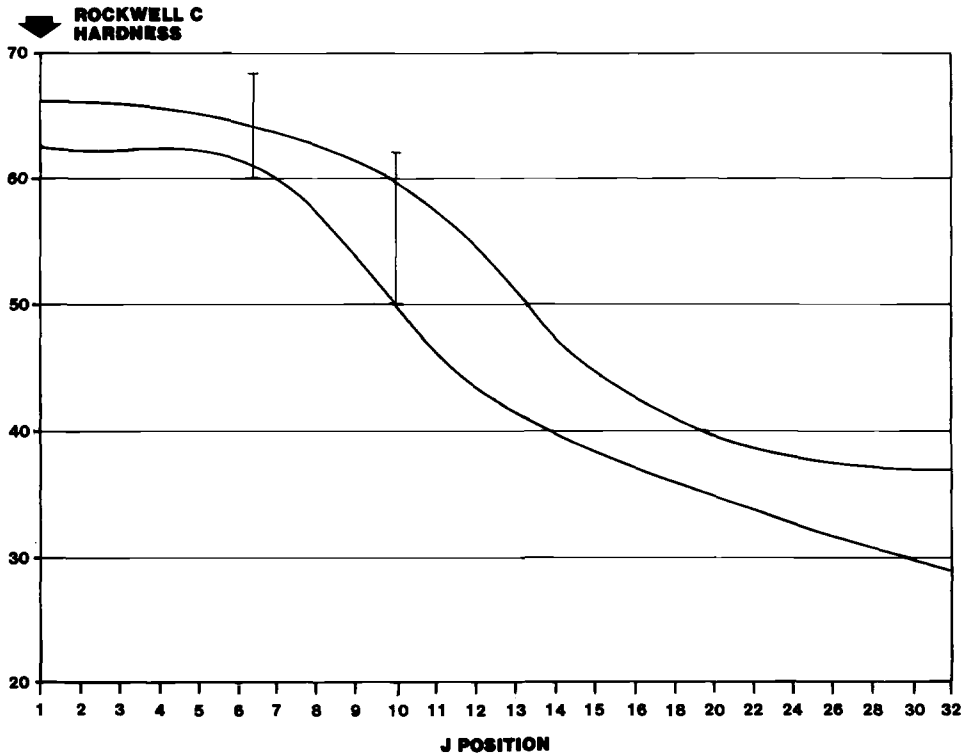


FIG. 8—Hardenability band for 52100 Grade 1 steel.

major element control range discussed above, has requested end quench hardenability limits of J6-60 minimum and J10-50 minimum. In Fig. 8 these values are shown in relation to the full ranges of Jominy results produced from recent VAD-treated heats.

Minor Element Chemistry Control

Sulfur—Although up to 0.050% sulfur is allowed in AISI 1070, most bearing steel users specify 0.025% sulfur maximum, as given in ASTM A 485-79 for 52100 steels. Some customers may actually specify a range for machinability purposes. Our major customer for 1070 bearing steel requests a range of 0.010 to 0.025% sulfur with a narrower aim range of 0.015 to 0.020%.

As shown in Fig. 9, the VAD ladle treatment plus sulfur wire-feeding technique routinely produces the customer allowed range; 92% of results also fall within the aim range. The allowed range is considered similarly achievable using the argon refining plus wire feeding process.

Oxygen—Oxygen limits are not quoted for AISI 1070 or in the ASTM specifications for 52100 steels. One of our customers specifies a 20 ppm maximum heat average for a modified AISI 4320 steel and a 13 ppm Aim average content for ASTM 52100 Grade 1 steel. Six tests are taken per heat, requiring billet samples from the top and bottom of the first, middle, and last usable ingots teemed.

Recently performed oxygen analyses for the three grades mentioned, made by the VAD practice, are shown in Fig. 10. The distribution of oxygen contents for the 52100 and 4320

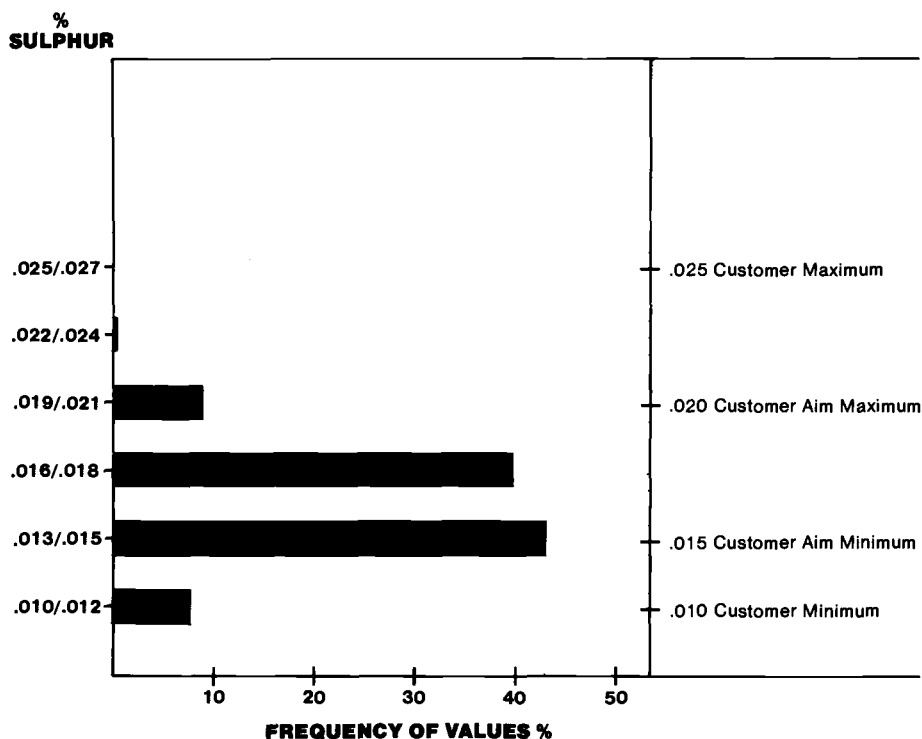


FIG. 9—Sulfur contents of AISI 1070 heats.

steels are limited, most likely because of the few heats sampled. Further data for 20 heats of a 4122 steel (0.22% chromium, 0.9% manganese, 0.2% silicon, 0.7% chromium, 0.25% molybdenum, 0.025% maximum sulfur), again made to a bearing quality VAD practice, therefore have been added for interest.

The results show individual billet samples can range from 7 to 32 ppm, with average values in the 12 to 18 ppm range, regardless of grade. The customer requirements for the 52100 and 4320 steels are also being achieved by the VAD method.

Comparisons of oxygen contents obtained by the VAD and A/R steelmaking processes, teemed with and without ladle stream shrouding and taken from the top versus bottom ingot positions, are of interest. Results showing the variation with melting method and ingot position are given for 1070 steel in Fig. 11. The differences in maximum to minimum oxygen values for the A/R heats are not wide, most likely because of the limited sample size. A rather consistent 14 to 16 ppm average oxygen level is attained, however, regardless of the melting method and ingot position tested.

In a recent, limited test, ingots of VAD treated 52100 steel were teemed from the same heat, with and without the argon shroud being active. A 7-ppm increase in residual oxygen, again regardless of position within the ingot, was obtained when the teeming stream was not argon shrouded; that is, 12.5 ppm average with the stream shrouded, 19.5 ppm average without stream shrouding.

Titanium—Residual titanium levels are not quoted in AISI or ASTM bearing steel specifications. However, a 30 ppm Aim maximum has been requested by one customer for 52100 Grade 1 steel, and 30 to 40 ppm levels are being achieved.

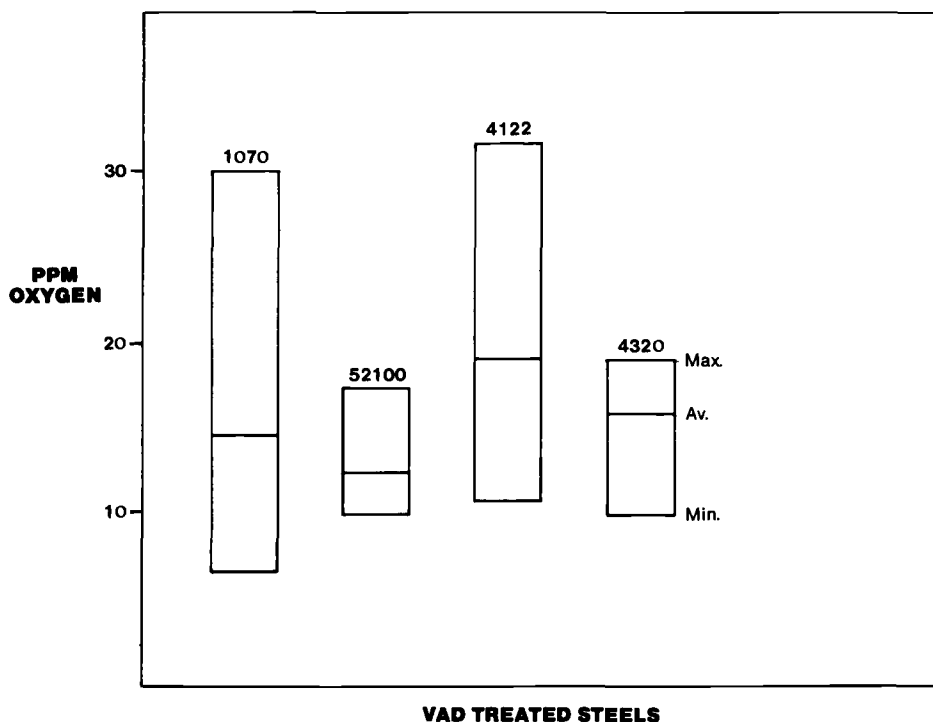


FIG. 10—Comparison of oxygen contents in four VAD-treated bearing steels.

With the Atlas Specialty Steels' electric arc furnace procedure utilizing oxygen blowing, titanium levels are near zero ppm at meltdown. However, increases in titanium result from slag and refractory/metal contact and from ferroalloy additions. Ferromanganese, ferrosilicon, and pure silicon alloys are generally low in titanium, so residual titanium is a lesser problem in 1070 bearing steel, relative to the chromium-bearing 52100, 4300, and 4100 steels. Recent analyses of certain alloys have shown values of 0.16 to 0.27% titanium in high carbon charge chromium and ferrochrome silicon, 0.06% titanium in lump silicon, and 0.01% titanium in a more expensive low carbon ferrochrome. Flux and alloy selection therefore is an important part of titanium control.

Steel Cleanliness

Jernkontoret Method

In North America, the method most used by steelmakers for measuring steel cleanliness is to rate billet samples to the Jernkontoret chart given as Plate 3 in the ASTM Specification E 45. That method is quoted in four associated specifications for bearing steels:

1. ASTM Specification for High-Carbon and Ball Roller Bearing Steels (A 295-84),
2. ASTM Specification for High Hardenability Bearing Steels (E 485-79),
3. ASTM Specification for Carburizing Steels for Anti-Friction Bearings (A 534-79), and
4. ASTM Specification for Special-Quality Ball and Roller Bearing Steel (A 535-79).

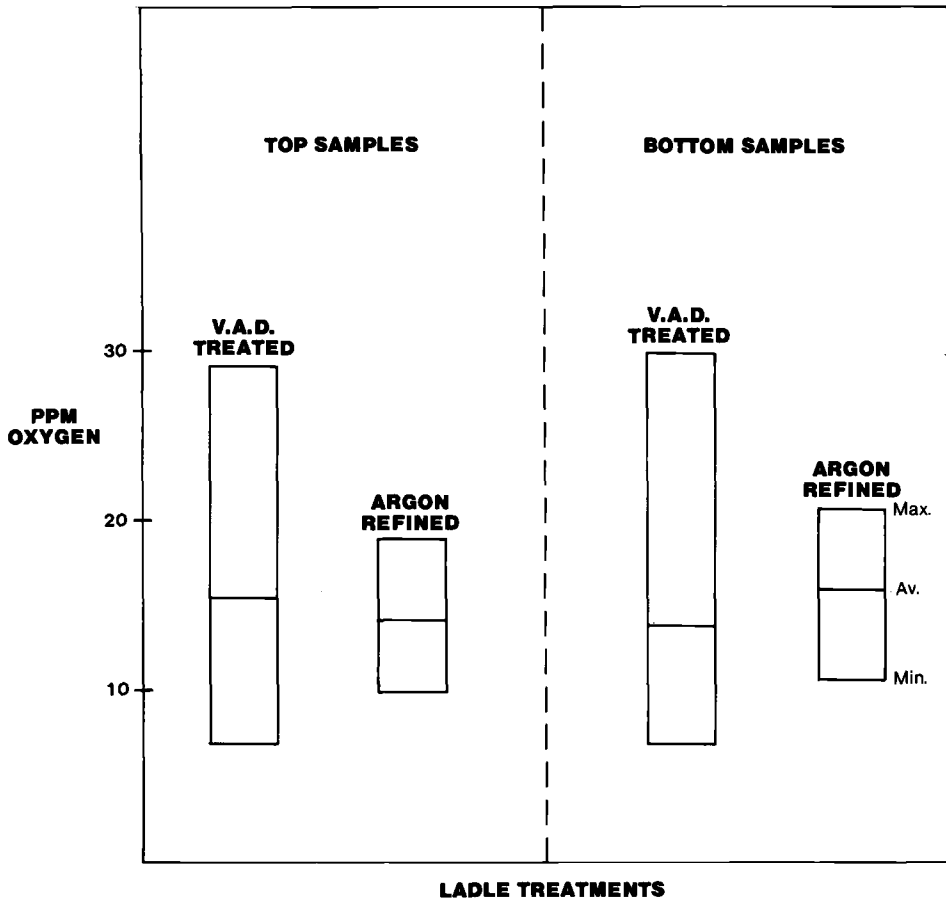


FIG. 11—Comparison of oxygen contents in SAE 1070 steel, by ladle treatment and ingot position.

The inclusion limits allowed by those four specifications are shown in Table 5, together with examples of three customer requests which combine certain elements from those specifications or, as is usually the case, are more restricting.

Also shown in Table 5 are recent inclusion counts averaged from large numbers of heats, for four bearing-quality steels made by differing methods.

Type A sulfide inclusion counts are largely dependent on residual sulfur levels; that is, the 1070, 4320, and 4120 steels all contain a nominal 0.015% sulfur, whereas the 52100 steels contain a nominal 0.002% sulfur. Because the Atlas Specialty Steels' practice relies on highly basic refractories and slags and on final aluminum additions, Type C silicate inclusions are virtually nonexistent. Occasionally Type D-Heavy globular oxides are encountered; Type D-Thin ratings of 1.0 are standard.

The Type B-Thin aluminate category is generally given the greatest emphasis. Here it is interesting to note there is rather little difference in B-Thin counts for all four steels made by the VAD process, or between the 1070 steels made by the VAD versus A/R process. However, a pronounced difference in steel cleanliness is shown for 52100 steels produced by vacuum arc remelting.

TABLE 5—*Cleanliness data.*

Specification	A		B		C		D	
	Thin	Heavy	Thin	Heavy	Thin	Heavy	Thin	Heavy
ASTM A 295-84 for 52100 and others	2 $\frac{1}{2}$	1 $\frac{1}{2}$	2	1	$\frac{1}{2}$	$\frac{1}{2}$	1	1
ASTM A 485-79 for 52100 Grade 1	2 $\frac{1}{2}$	1 $\frac{1}{2}$	2 $\frac{1}{2}$	1 $\frac{1}{2}$	2 $\frac{1}{2}$	2	2	1 $\frac{1}{2}$
ASTM A 534-79 for 4320 and others	3	2	3	2 $\frac{1}{2}$	2 $\frac{1}{2}$	1 $\frac{1}{2}$	2	1 $\frac{1}{2}$
ASTM A 535-79 for 52100 ball and roller bearings	1 $\frac{1}{2}$	1	1 $\frac{1}{2}$	1	1 $\frac{1}{2}$	1	1 $\frac{1}{2}$	1
Customer Specification for 52100 steel	2 $\frac{1}{2}$	1 $\frac{1}{2}$	2	1	1	1	1 $\frac{1}{2}$	1
Customer Specification for 1070 steel	2 $\frac{1}{2}$	1 $\frac{1}{2}$	2	1	$\frac{1}{2}$	$\frac{1}{2}$	1 $\frac{1}{2}$	$\frac{1}{2}$
Customer Specification for 4320 steel	3	2	1 $\frac{1}{2}$	1	1 $\frac{1}{2}$	1	1 $\frac{1}{2}$	1
VAD 52100	1.4	0.4	1.2	0	0	0	0.8	0.1
VAR 52100	0.4	0	0.3	0	0	0	1.0	0.1
VAD 1070	1.9	1.3	1.0	0	0	0	1.0	0
A/R 1070	2.1	1.2	1.2	0	0	0	1.0	0
VAD 4320	1.9	1.0	1.0	0	0	0	1.0	0.1
VAD 4122	2.0	1.0	1.1	0	0	0	1.0	0

SAM Method

The SAM method for rating steel cleanliness is described in ASTM A 295-84. Shown in Table 6 is a comparison of SAM ratings for AISI 1070 steel made by the vacuum arc degassing and argon refining processes. The data indicate the A/R procedure results in marginally higher SAM counts, but both steelmaking techniques produce very clean steels, well below the specification maxima of B Type-15 and D Type-10 requested by one customer.

Magnetic Particle Inspection Method

The AMS 2301G method for magnetic particle inspection is generally requested for aircraft quality steels. Its application is unusual for bearing steels. The specification allows frequency/severity counts of 0.34/0.25 maximum average and 0.67/0.55 maximum individual tests, for steels of greater than 0.25% carbon content.

For AISI 1070 steel, both the average and individual frequency/severity values are normally 0.00/0.00, with the occasional maximum individual test with values less than 0.06/0.13.

It has been our experience that values better than 20% of the AMS 2301G limits are generally achieved on a very broad range of steels, using both the VAD and A/R practices. Seldom are indications found when magnetic particle testing vacuum arc remelted products to the AMS 2301G specification.

TABLE 6—*SAM ratings for AISI 1070 steel.*

Specification	B Type		D Type	
	Average	Maximum	Average	Maximum
Customer limits	...	15	...	10
VAD method	0.3	4.3	0.2	1.5
A/R method	1.3	3.4	0.4	1.7

Conclusions

The argon refining and vacuum arc degassing processes both readily meet the requirements of existing bearing steel specifications. The VAD process offers an ability to consistently meet stricter limits for steel chemistry and cleanliness, due to lesser exposure of steels to the atmosphere and to the longer holding times available for alloy adjustments and metal contact with basic slags and refractories. The major role of the vacuum system is to reduce product hydrogen contents.

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Fatigue Life of High-Carbon Chromium Ball Bearing Steel Produced by Electric Furnace—Vacuum Slag Cleaner—Ladle Furnace—RH Degassing—Curved Continuous Caster

REFERENCE: Kumagai, K., Takata, Y., Yamada, T., and Mori, K., "Fatigue Life of High-Carbon Chromium Ball Bearing Steel Produced by Electric Furnace—Vacuum Slag Cleaner—Ladle Furnace—RH Degassing—Curved Continuous Caster," *Effect of Steel Manufacturing Processes on the Quality of Bearing Steels*, ASTM STP 987, J. C. C. Hoo, Ed., American Society for Testing and Materials, Philadelphia, 1988, pp. 348–359.

ABSTRACT: Rolling contact fatigue life of continuous-cast ball-bearing steel was 1.5 to 2 times as long as that of conventional ingot-cast one. Continuous-cast ball-bearing steel, when made by a double ladle furnace process with lower contents of oxygen, lower residuals (sulfur, phosphorus, and titanium), and lower number of finely scattered inclusions, showed fatigue life over six times as long as that of conventional ingot-cast ball bearing steel. Studies have been conducted to determine the effect of residuals and inclusions on fatigue life. The reason of longer fatigue life is thought to be due mainly to a lower content of finely scattered oxide inclusions.

KEY WORDS: rolling contact fatigue, bearing steels, continuous casting, inclusions influence, residuals influence

The rolling contact fatigue life of high-carbon ball-bearing steel is affected by many metallurgical factors, such as nonmetallic inclusions, segregation of chemical compositions, and residuals. For these reasons, high-grade ball-bearing steels which need a longer fatigue life are remelted by electro slag remelted (ESR) or vacuum-arc melting (VAR). Both methods produce fewer finely scattered oxide inclusions and have a homogeneous structure with less segregation.

On the other hand, in mass production, the metallurgical process development has resulted in a gradual decrease in oxygen content and an improvement of fatigue life of ball-bearing steel year by year, as shown in Fig. 1.

In January 1982, we installed a new steelmaking process, electric furnace (80TEF)—vacuum slag cleaner (VSC)—ladle furnace (LF)—Ruhrstahl-Heraeus degassing unit (RH)—curved type continuous caster (CC). By this process, we tried to produce clean speciality steels, and now, high quality demands of alloy steels can be satisfied. In addition, we noticed that fatigue life of continuous-cast ball-bearing steel is longer than that of conventional ingot-cast one. Further studies have been undertaken to develop ball-bearing steel having still longer fatigue life. The outline of the 80TEF-VSC-LF-RH-BL/CC process is in Table 1.

¹Manager, assistant manager, manager, and managing director, respectively, Aichi Steel Works, Ltd., Arao-machi, Tokai-shi, Aichi-ken, Japan.

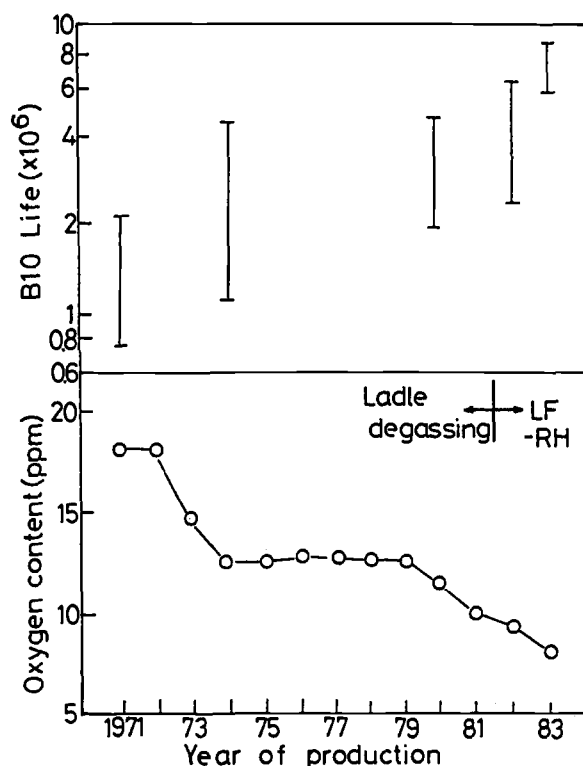


FIG. 1—Oxygen content and rolling contact fatigue life of ball bearing steels (JIS SUJ2) from 1971 to 1983 (Aichi Steel Works).

Fatigue Life of Continuous-Cast Ball-Bearing Steel

Materials

The steels used in this study consist of two series. Chemical compositions of steels are shown in Table 2. Both series were melted by the new process, but Steel A was ingot-cast (2.6T), and B-series steels were continuous-cast (370 by 480 mm bloom). To protect the reoxidation of molten steel during continuous casting, we used a long nozzle and an immersion nozzle, and the tundish was purged enough by Argon gas. Ingots and blooms were rolled to bar in 65-mm diameter and annealed after homogenized soaking.

Procedure

Fatigue life tests were run using a thrust-type rolling contact fatigue tester [1]. Test specimens were machined from bars to 60 by 5.5-mm disks in transverse and longitudinal sections. The track of the running ball on the transverse section specimens does not cross the center of bar, but the track on the longitudinal section specimens crosses the center of the bar at two points. Test specimens were hardened and tempered

(850°C × 30-min oil quench, 170°C × 90-min air cool)

TABLE 1—Outline of new steelmaking process.

Item	Specification	
Electric furnace	melting capacity	nominal 80T
	transformer capacity	75 MVA, maximum 90 MVA
Vacuum slag cleaner	sucking capacity	maximum 600 Kg/min
	cooling water	226 T/h
Ladle furnace	type	LF80-16D
	ladle inner diameter	3560 mm, 0.85 H/D
	transformer capacity	16 MVA
	temperature rising speed	average 4°C/min
RH	type	ladle lifting type
	vessel inner diameter	2300 mm
	heating device	graphite electrode, 800 kW
	lift gas	maximum 800 L/min
CC	degree of vacuum	0.1 torr
	type	radius type with two points unbending
	radius	16.0, 30.4 m
	length	67.2 m
	number of strands	2
	casting size	370 × 480 mm ²
	casting speed	maximum 0.86 m/min
	tundish	20-T capacity, 800 mm depth
	mold	built-up curved mold
	secondary cooling zone	5 zones

The hardness of specimens was HRC62-63.

The test conditions were

Maximum nominal herzian stress	$P_{\max} = 5207 \text{ N/mm}^2$
Lubricant	No. 60 spindle oil
Stress cycle	900 cpm
Diameter of ball	9.5 mm (0.375 in.)

Test Results

Figures 2 and 3 show the Weibull plots of lives of Steel A and B-series steels in transverse and longitudinal sections. The fatigue life of continuous-cast ball bearing steel was 1.5 to 2 times as long as that of ingot-cast steel, both in transverse and longitudinal sections specimens. When continuous casting was first applied to ball-bearing steel, there was concern that center segregation of continuous-cast steel might be detrimental to fatigue life. Flaking sites on longitudinal section specimens tended to come together around the center of the bar in both ingot-cast and continuous-cast specimens, but there was no correlation between fatigue life and flaking sites [2].

TABLE 2—Chemical compositions of A- and B-series steels, %.

Steel	C	Si	Mn	P	S	Cu	Ni	Cr	Mo	Al	O	Ti	Cast
A	0.97	0.27	0.41	0.017	0.009	0.14	0.06	1.47	0.02	0.022	0.0009	0.0023	ingot
B1	0.98	0.28	0.46	0.013	0.005	0.08	0.04	1.47	0.01	0.023	0.0008	0.0021	continuous
B2	0.97	0.27	0.37	0.019	0.005	0.12	0.06	1.47	0.01	0.031	0.0009	0.0019	continuous
B3	0.96	0.26	0.38	0.013	0.007	0.10	0.06	1.43	0.01	0.021	0.0006	0.0019	continuous

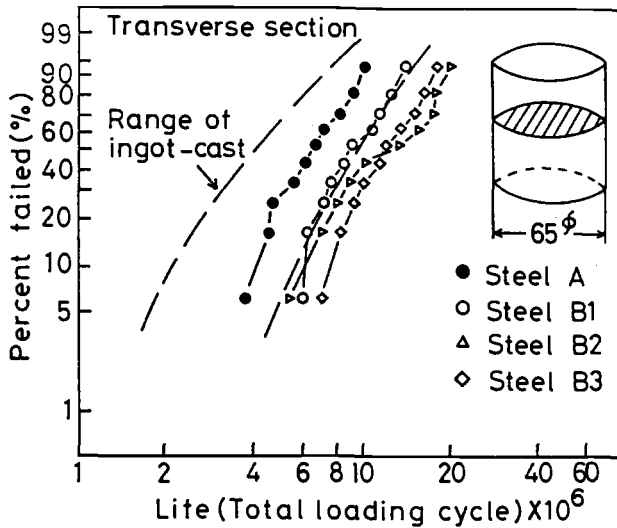


FIG. 2—Results of rolling fatigue tests of A- and B-series steels (transverse section specimens).

Improvement of Fatigue Life

It was found that the fatigue life of continuous-cast ball-bearing steel was longer than that of ingot-cast steel. As mentioned previously, steels remelted by ESR have fewer finely scattered inclusions and less segregation. Therefore, in mass production of ball-bearing steel, in order to get the same long life as produced by ESR, residuals (sulfur, phosphorus, and titanium) and inclusions may need to be decreased much more.

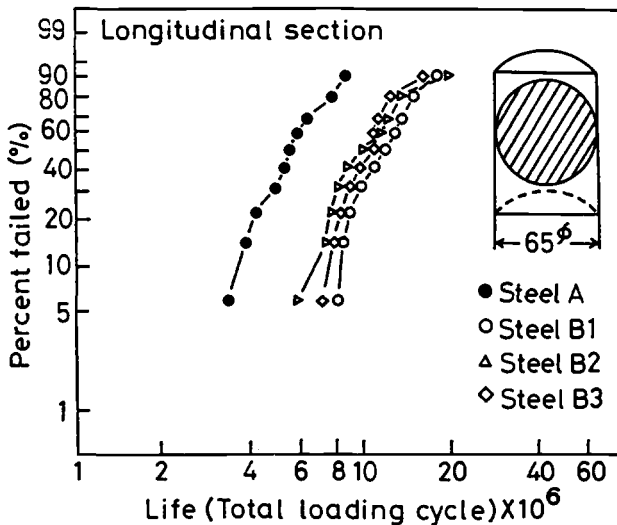


FIG. 3—Results of rolling fatigue tests of A- and B-series steels (longitudinal section specimens).

Materials

Steels used in this study consist of two series. Chemical compositions of steels are shown in Table 3. C-series steels were tried in order to refine cleaner steels with lower amounts of residuals (sulfur, phosphorus, and titanium) and fewer finely scattered inclusions. The investigation to decrease oxygen, sulfur, and titanium contents by a new process has been reported [3], but is not described in this paper in detail. The outline of this process is shown in Table 4. This process is characterized by double LF process to eliminate larger inclusions. D-series steels were remelted by ESR.

Test Results

The test procedure was the same as stated above, but test specimens were machined from only the transverse section. Figure 4 shows Weibull plots of lives of C- and D-series steels. Fatigue lives of C-series steels were as long as those of D-series steels remelted by ESR.

Figure 5 reviews the results of fatigue lives in Figs. 1, 2, and 4 and summarizes the relationship between B10 fatigue life and oxygen content in ball-bearing steels. For the conventional ingot-cast and continuous-cast steels, the relationship between B10 life and oxygen content was given as a straight band with some scatter, and the lower the oxygen content, the longer is the B10 life. B10 lives of continuous-cast B-series steels were located at the longer life side than those of ingot-cast steels in the band when compared at the same oxygen content level.

For C- and D-series steels, the relationship between B10 life and oxygen content was given as another band-situated longer-life side, as compared with conventional ingot-cast and continuous-cast steels. When compared at the same oxygen content level, B10 lives of C-series steels were over six times longer than those of conventional ingot-cast steels.

Fatigue lives evaluated at a ball-bearing maker were five to seven times longer for C-series steels and four times longer for ball bearing produced by C1 steel than those of conventional ingot-cast ball-bearing steels.²

Discussion

Nonmetallic inclusions are well-known to be the significant factor to fatigue life of ball-bearing steel. The solidification rate of continuous casting is faster than that of ingot casting, so inclusions of continuous-cast steel are expected to become finer than those of ingot-cast one. Such a case was reported in case-hardening steel [4].

As shown in Fig. 2, the fatigue life of continuous-cast ball-bearing steel was longer than that of ingot-cast one. When compared at the same oxygen content level in Fig. 5, fatigue lives of continuous-cast ball-bearing steel were found to be situated on the longer-life side of the band, which showed the relationship between B10 life and oxygen content.

Figure 6 shows the relative inclusion content of steels tested. Continuous-cast steel B1 had fewer heavy B-type and heavy D-type inclusions than Steel A. This is thought to be a reason for the longer fatigue life of continuous-cast ball-bearing steel as compared with that of ingot-cast one.

In Figure 5, when compared at the same oxygen content level, fatigue lives of C- and D-series steels were apparently longer than those of conventional ingot-cast or continuous-cast ball-bearing steels. Thus it appears that other factors in addition to oxygen content affect fatigue life.

²Mori-hara, G., KOYO, Kokubu, Kasihara-shi, Ousaka-fu, Japan, personal communication, 1960.

TABLE 3—Chemical compositions of C- and D-series steels, %.

Steel	C	Si	Mn	P	S	Cu	Ni	Cr	Mo	Al	O	Ti	Cast
C1	1.02	0.22	0.35	0.009	0.002	0.05	0.04	1.46	0.01	0.006	0.0007	0.0012	continuous
C2	0.97	0.20	0.40	0.010	0.001	0.11	0.06	1.48	0.02	0.013	0.0007	0.0015	continuous
C3	1.01	0.25	0.33	0.010	0.002	0.10	0.06	1.45	0.01	0.019	0.0006	0.0014	continuous
D1	0.99	0.25	0.47	0.012	0.003	0.05	0.03	1.46	0.01	0.021	0.0009	0.0013	ESR
D2	1.00	0.27	0.46	0.013	0.003	0.06	0.04	1.49	0.01	0.015	0.0011	0.0017	ESR

TABLE 4—Outline of melting C-series steels by double LF process.

EF	VSC	LF	RH	LF	CC
scrap blending with low residuals	dephosphorization treatment	refined by low oxygen potentials and high basicity slag low titanium ferro-alloy	degassing with high circulation rate and long treating time	soft blowing (elimination of larger inclusions)	prevention of reoxidation from air and refractory homogeneous solidification
low titanium low phosphorus	low phosphorus	low oxygen low sulfur low titanium	low oxygen	fine inclusions	fine inclusions

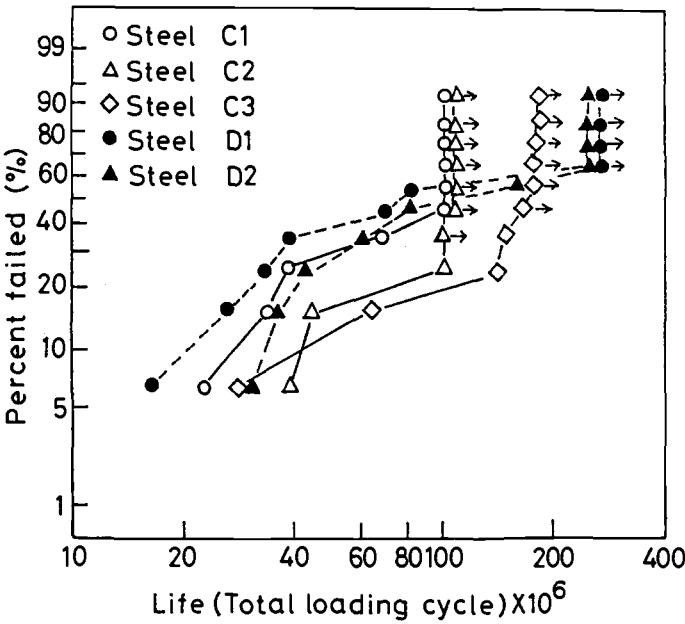


FIG. 4—Results of rolling fatigue tests of C- and D-series steels.

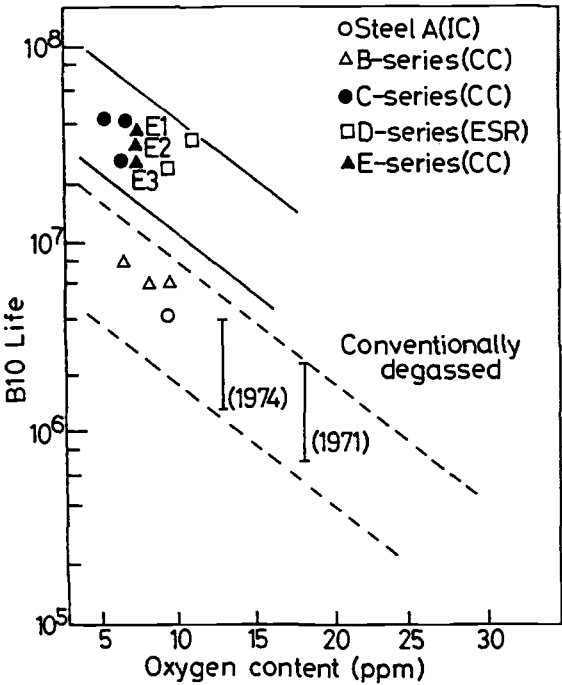


FIG. 5—Relationship between rolling contact fatigue life and oxygen content.

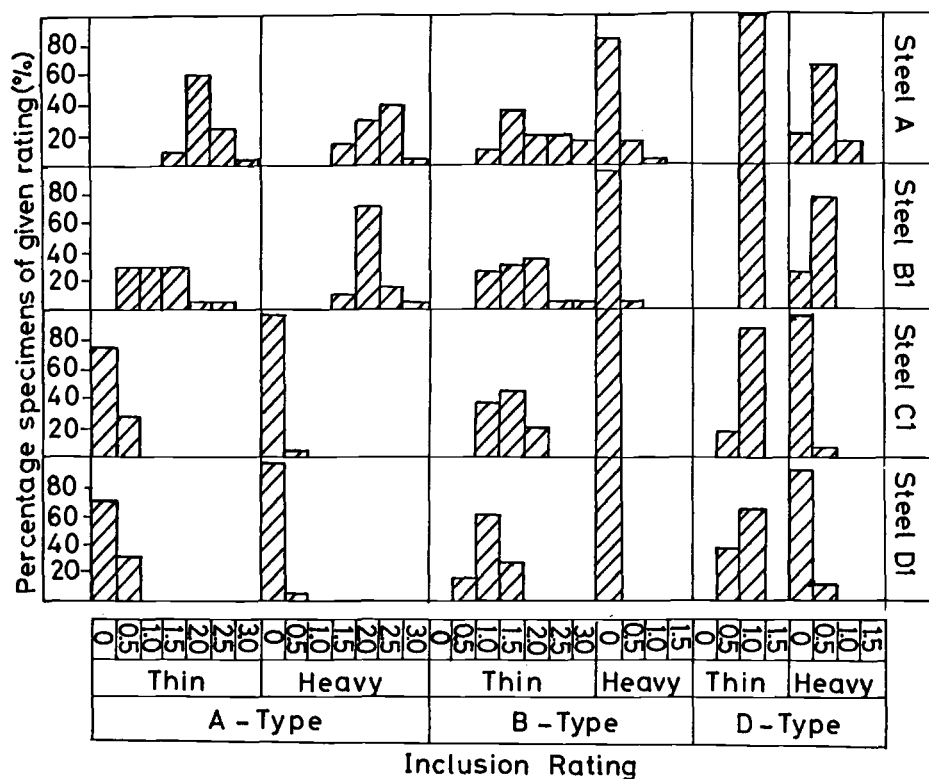


FIG. 6—Relative inclusion rating of steel tested.

The ball-bearing steel remelted by ESR is characterized by finer inclusions and less segregation. In Fig. 6, both C- and D-series steels showed much fewer heavy B-type and heavy D-type inclusions. Characteristically, C-series steels have lower contents of sulfur, phosphorus, and titanium, and D-series steels have lower sulfur content.

In order for us to investigate the effect of residuals on fatigue life, fatigue life tests were run for E-series steels with higher contents of sulfur, phosphorus, or titanium. Chemical compositions of E-series steels are shown in Table 5. The melting and casting processes of the E-series steels were the same as those of C-series steels.

B10 lives of E-series steels are plotted in Fig. 5. E-series steels have fatigue lives as long as those of C-series steels. These results show that the lower sulfur, phosphorus, or titanium content is not the essential reason for longer fatigue lives of C-series steels. Instead, the longer fatigue lives of C-series steels are thought to be due to their finer oxide inclusions.

Conclusion

1. The rolling contact fatigue life of continuous-cast ball-bearing steel is 1.5 to 2 times as long as that of ingot-cast one.
2. Continuous-cast ball-bearing steel, used to melt a lesser amount of finely scattered inclusions and lower residuals by double LF process, demonstrates a fatigue life over six times longer than that of conventional ingot-cast steel.

TABLE 5—*Chemical compositions of E-series steels, %*

Steel	C	Si	Mn	P	S	Cu	Ni	Cr	Mo	Al	O	Ti	Cast
E1	0.95	0.27	0.47	0.014	0.006	0.06	0.06	1.46	0.02	0.019	0.0007	0.0011	continuous
E2	0.95	0.30	0.45	0.016	0.007	0.05	0.04	1.47	0.03	0.027	0.0007	0.0011	continuous
E3	0.99	0.26	0.37	0.018	0.004	0.16	0.05	1.49	0.02	0.015	0.0007	0.0018	continuous

3. Residuals such as sulfur, phosphorus, or titanium are found not to be the essential reason for longer fatigue life.
4. Longer fatigue life is thought to be due to the finer oxide inclusions.

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Properties of Through-Hardening Bearing Steels Produced by BOF Blowing Metallurgy and by Electric Arc Furnace with Ladle Metallurgy

REFERENCE: Baum, R., Böhnke, K., Boeckers, T., and Klemp, H., "Properties of Through-Hardening Bearing Steels Produced by BOF Blowing Metallurgy and by Electric Arc Furnace with Ladle Metallurgy," *Effect of Steel Manufacturing Processes on the Quality of Bearing Steels, ASTM STP 987*, J. J. C. Hoo, Ed., American Society for Testing and Materials, Philadelphia, 1988, pp. 360–374.

ABSTRACT: This paper discusses the cleanliness of through-hardening bearing steels produced by different melting and deoxidation processes. Macroscopic cleanliness is tested by the step-down and blue fracture test, and microscopic cleanliness is classified according to three rating methods. Cleanliness values of the air-melted heats are compared with remelted steels. Oxide inclusions are related to the oxygen contents of the heats, titanium-carbonitrides to the titanium contents, and sulfide inclusions to the sulfur contents of the heats. Machinability is effected more by carbide grain sizes than by sulfur contents. Ingot- and continuous cast steels are compared on cleanliness and segregation.

KEY WORDS: melting procedures, aluminium deoxidation, calcium treatment, macroinclusions, microinclusions, rating methods, fatigue behavior, machinability, ingot and continuous casting, segregation, basic oxygen furnace (BOF)

Thyssen is one of the major European producers of through-hardening bearing steels, with an annual production of 160 000 tons. Figure 1 shows the product-mix.

Steel is melted in basic oxygen furnaces (BOFs) by Thyssen Stahl AG (TST) at Duisburg and in an electric arc furnace (EAF) by Thyssen Edelstahlwerke AG (TEW) at Witten and rolled in blooming mills in both places. For smaller sections and wire there are rolling mills at TEW Witten and Krefeld. Tubes are rolled by Wälzlagerrohr GmbH (WRG) at Krefeld, a subsidiary company from TEW and Mannesmannröhren-Werke AG. Remelting facilities are at TEW Krefeld and consist of electro slag remelting (ESR) and vacuum-arc remelting (VAR) furnaces.

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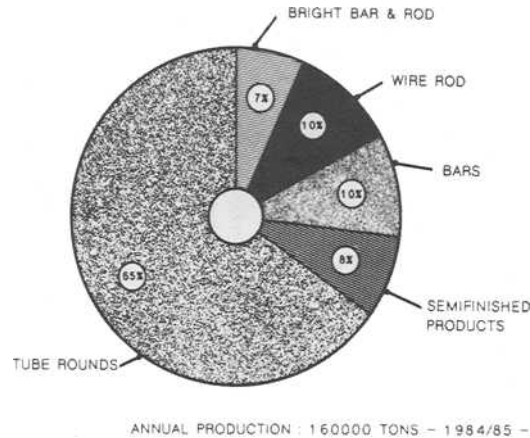


FIG. 1—Product-mix of Thyssen bearing steels.

Process Routes

Figure 2 shows the different process routes within the Thyssen group. There are two different melting routes: the EAF, ladle degassing, ingot route at Witten and the BOF-TBM, Ruhrstahl-Heraeus (RH) degassing, continuous and ingot cast route at Duisburg.

Melting and Ladle Treatment

At TEW Witten, the melting device is a 75-MVA, 110-ton furnace equipped with water-cooled side walls and roof. This furnace was the first one with a centric bottom tap hole. Recently, a porous plug for bath agitation has been installed in order to homogenize the liquid metal in the furnace and to remove the slag. During and after meltdown, the metal is dephosphorized and, under permanent oxygen injection, heated to the desired tapping temperatures under foamy slags. The steel is tapped with low carbon and low manganese contents. During tapping, small amounts of silicon and the necessary amount of ferrochromium as well as the slag-forming agents are added into the dolomite-lined ladle. During ladle treatment, vacuum carbon deoxidation (VCD) is carried out. The final deoxidation with aluminium is done under vacuum. After the vacuum treatment, smooth rinsing by argon removes deoxidation products.

In the TBM route (Thyssen-Blowing-Metallurgy), combined blowing BOF technique, a liquid metal containing 0.03 to 0.05% carbon is slag-free tapped after processing in a 140-ton TBM furnace. During tapping, the heat is alloyed and deoxidized with the full amount of aluminium into the tapping stream. Ferrochromium with a maximum content of 0.05% titanium is premelted in a net frequency furnace in order to keep the titanium content low. After tapping, the ladle is treated in two steps. The first step is vacuum treatment in an RH degasser, where hydrogen contents below 2 ppm are achieved. Simultaneously, final alloying according to the specification is carried out. During vacuum treatment, deoxidation products are washed out and dissolved in the covering slag. The second step is a so-called "soft bubbling argon treatment" in the dolomite-lined ladle under defined slags to achieve a further reduction of the deoxidation products.

WÄLZLAGER ROHR -
GESELLSCHAFT GmbH

THYSSEN EDELSTAHLWERKE AG

THYSSEN STAHL AG

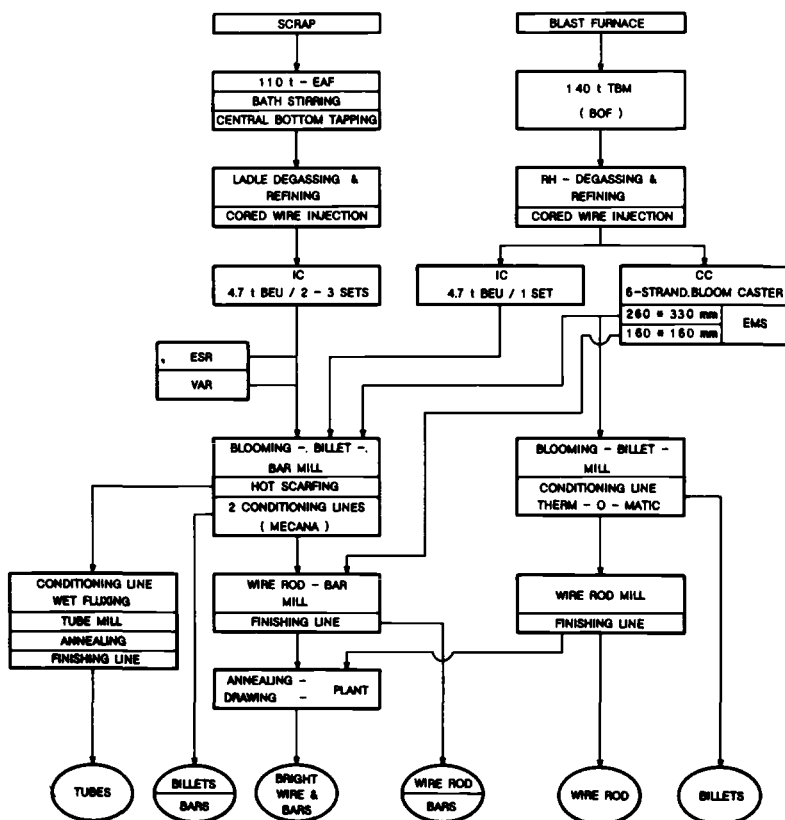


FIG. 2—Thyssen process routes for bearing steels.

For both routes, EAF and TBM, there is a cored wire injection equipment to add calcium-silicon, aluminium, and sulfur. The addition of calcium-silicon to bearing-quality heats is used only for trials.

Casting and Solidification

Ingot are up-hill teemed in large sets up to 30 molds. A heat of about 140 tons is cast in big-end-up (BEU) ingots of 4.7 tons. A six-strand bloom caster is used for the solidification of TBM steel as well. The specific data of the caster are as follows:

Type of machine: bow-type caster with 12-m radius

Number of strands: 6

Bloom size and casting speeds:

$260 \times 330 \text{ mm}^2$	} 0.6 to 0.9 m/min
$265 \times 385 \text{ mm}^2$	
$160 \times 160 \text{ mm}^2$	1.2 to 1.8 m/min

EMS: linear motor type operating with 50 Hz directly underneath the mold.

Between ladle, tundish, and mold there are submerged ceramic tubes to prevent reoxidation. The steel surface in the tundish and mold is covered with specific insulating casting powders.

Rolling

Ingots and continuous cast (CC) blooms are reduced to square sections in a two-high 1000-mm blooming mill and in a two stand open two-high mill of 800 mm to the smallest sections of 50 mm square and 53 mm round. Maximum sizes are 220 mm square or round. The prerolled blooms are hot-scarfed in order to eliminate the decarburized surface; simultaneously surface defects are removed. Billets for the wire mills and for customers follow the same production route. Before being rolled to final dimensions, all billets are inspected by two dry fluxing lines (MECANA) where ultrasonic testing is carried out simultaneously. Products for the tube mill are wet fluxed, and defects are removed by grinding. Round and square sections larger than 200 mm are forged in a long forging machine (GFM LSX 55).

Quality Tests

Macroscopic Cleanliness

Macroscopic cleanliness is determined by ultrasonic testing, blue fracture, and step-down tests.

Ultrasonic testing is a 100% method, which ensures that material exhibiting larger macroinclusions, piping, or porosity because of process failures is marked and eliminated. Macroinclusions revealed by the following blue fracture and step-down tests are thus usually below the applied ultrasonic sensitivity. The blue fracture and step-down tests are random tests. Sampling is done in accordance with the ASTM Specification for High-Carbon Ball and Roller Bearing Steel (A 295-84) and Stahl-Eisen-Prüfblätter (SEP) 1584 and 1580. Remelted ingots are each top- and bottom-checked after sufficient cropping. Specimens are taken from rolled sizes equivalent to a 30- to 35-fold reduction of the ingot or strand cast.

Blue Fracture Test—The spheroidized annealed, unnotched specimens are fractured at a temperature of about 400°C. Inclusions visually recognized by the unaided eye or by low magnification are classified into ten classes based upon procedures established in SEP 1584. These classes range from 0 to 9, where Class 0 represents defect freedom and Class 9 the presence of several inclusions > 5 mm in length and > 1 mm in width.

In Figure 3, the relative frequency of the various classes is shown for EAF-ingot cast (IC), TBM-IC, and CC steel. For IC steel of both melting processes, EAF and TBM, the same number of 95% defect-free specimens is achieved. The number for CC steel from TBM is approximately the same. It must be emphasized that the very small inclusions of Class 1, that is, < 5 mm in length and < 0.3 mm in width, are only visible by the described special blue fracture test and not by the fracture test according to the ASTM Practice for Determining the Inclusion Content of Steel (E 45-76). If we assume that these very small inclusions revealed in the semifinished size are no longer detectable after the subsequent reduction to final sizes, the shares of Class 0 and 1 can be added. By this, a number greater than 97% for all three groups is achieved.

In remelted steels, no macroinclusions could be detected. In interpreting these results it must be considered that the nonuniformly distributed macroinclusions are tested by the random test. The inspected six specimens represent only approximately 0.003% of the air-melted ingot-cast heats and 0.04% of the remelted ingots. It is obvious that the results of this random testing cannot give proper figures of the macroinclusion content, but can be considered only

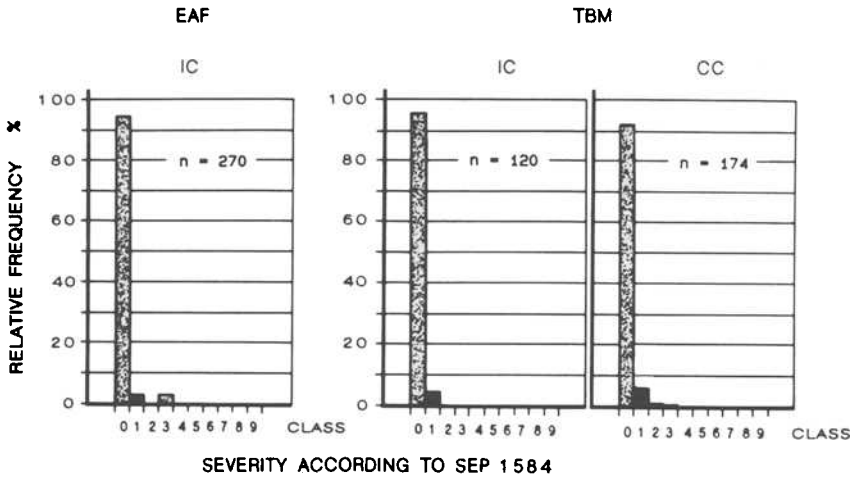


FIG. 3—Blue fracture test results from EAF, TBM IC, and CC steel.

as a statistical probability of occurrence in size and number. However, these results allow comparisons on a statistical basis of steels produced by different melting and casting methods. It has to be mentioned that the difference between the number of macroinclusions for air-melted steel and that for remelted steel is larger than the described figures may indicate. The 100% defect freedom of remelted steel is ensured by the remelting process itself, whereas non-remelted steel can achieve the described amount of defect freedom only as a statistical probability and thus cannot be guaranteed for the whole lot.

Step-Down Test—In order for us to investigate the sensitivity for determining macroinclusions for a lot of heats, step-down tests were carried out from the same samples used for the blue fracture tests. Macroinclusions detected according to SEP 1580 on the machined surfaces of the specified diameters after the magnetic particle-test are counted also, starting with > 1.5 mm in length. The total length of the inclusions is related to the examined area per square meter. In Fig. 4, results are shown for VAR, ESR, EAF, and TBM, both IC steel. For remelted steel, the defect freedom is again confirmed. EAF and TBM-IC steels exhibit approximately the same total length of 0.8 mm/m². Compared to the preceding blue fracture test,

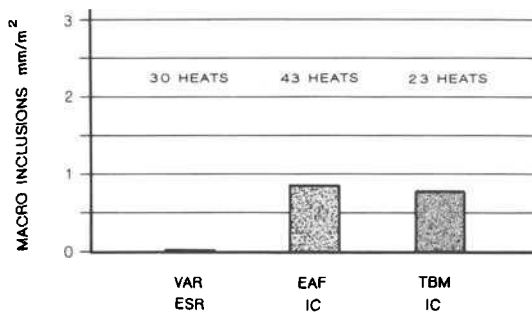


FIG. 4—Step-down test results from VAR, ESR, EAF, and TBM-IC steel.

where a total inclusion length of about 250 mm/m² was detected, the greater sensitivity of the blue fracture test becomes obvious. For that reason, the step-down test has been omitted.

Microscopic Cleanliness

Microinclusion rating is carried out in the same random test as described for blue fracture tests. The specimens are examined simultaneously according to SEP 1570-71, ASTM E 45-76, and recently to the SAM method.

Oxides: SEP 1570-71—Air-melted heats usually are rated according to Method K 4; that is, only microinclusions of size Class 4 and larger are considered. For remelted steel, evaluation starts already from size Class 1 (Method K 1). In order for us to compare these different groups, a number of air-melted heats was tested, also following the K 1 method. The *K*-values of each group are summarized as a cumulative frequency. Their 50%-(*H*₅₀) and 95%-(*H*₉₅)-frequencies are shown in Fig. 5. If we consider the K 1 and K 4 values, vacuum arc remelted (VAR) exhibits the lowest cleanliness values, followed by EAF-IC, TBM-CC, TBM-CC + Ca, and TBM-IC. As a rough comparison, both methods confirm that the differences between remelted and air-melted steels are small and that TBM-CC results in lower cleanliness values than TBM-IC.

In order for us to illustrate more clearly these differences, the occurrence of number and sizes of the different oxide inclusion types are shown in Fig. 6. Due to the aluminium deoxidation, no silicate-type inclusions occur; therefore only alumina- and globular-type inclusions are to be considered. It becomes obvious that the higher K 1 and K 4 values of TBM-IC are caused mainly by the greater number of small and larger inclusions of the alumina-type and that the lower cleanliness values of TBM-CC are achieved by the decrease of these inclusions.

As a consequence on the treatment of liquid metal in basic-lined ladles under basic slags, the mainly calcium-induced globular-type inclusions are present in all steel groups. Their total amount is lower compared to alumina-type inclusions. VAR steel exhibits the lowest amount, followed by the air-melted steels with approximately the same amounts; that is, there is no significant difference between IC and CC steel. Calcium treatment causes the highest amount of globular inclusions, but mainly in the smaller sizes, below Size 4.

It must be emphasized that the distribution curves can define statistical probabilities of occurrence of the various microinclusions only because they are inhomogeneously distributed

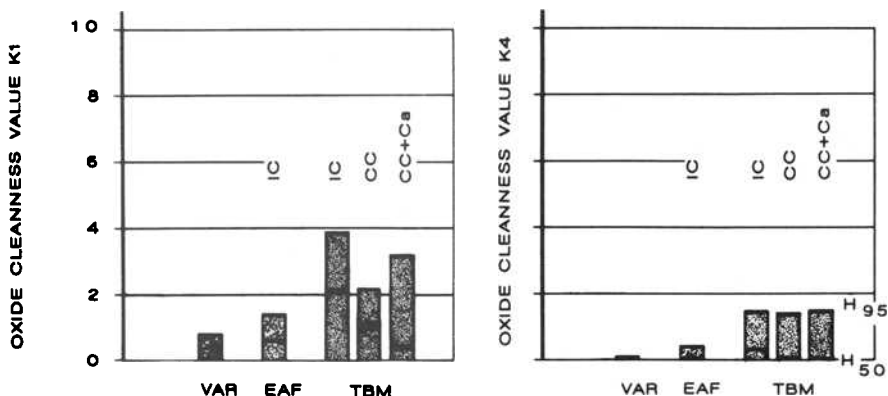


FIG. 5—Oxide cleanliness values K 1 and K 4 (SEP 1570-71),

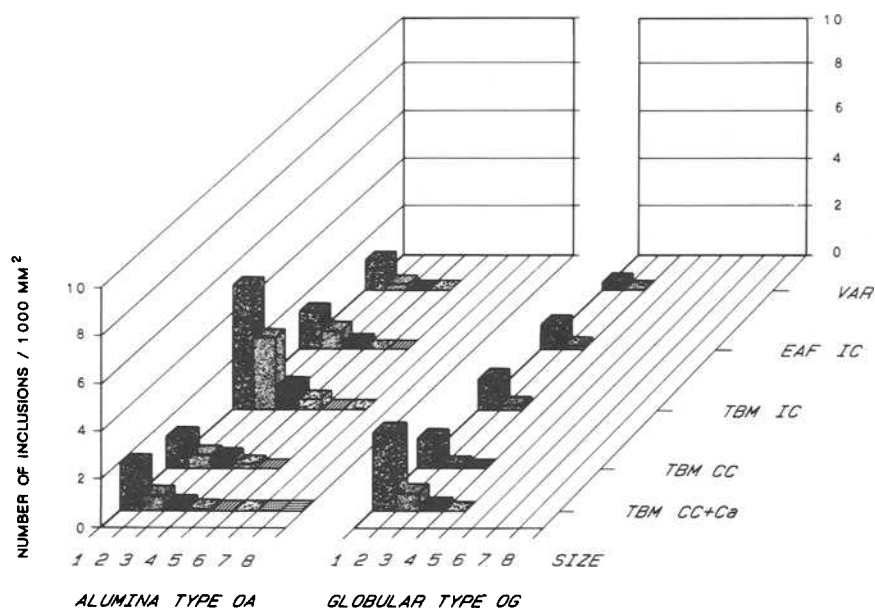


FIG. 6—Number and size of oxide inclusions (SEP 1570-71).

and rated in random tests, as are the macroinclusions. This may be exemplified as follows: the occurrence of one alumina-type inclusion of size Class 4 is probable in a tested area of 25 cm² in TBM-IC steel, in EAF-IC steel it is probable just in 330 cm², and in VAR steels it is probable only in 1000 cm² of the tested area. Thus, the probability of occurrence can be calculated for all types and sizes of microinclusions.

Oxides: ASTM E 45-76, Method A—In Fig. 7, the H_{50} - and H_{95} -cumulative frequencies of the mean values according to ASTM E 45-76 Method A are shown. They also confirm the superiority of TBM continuous casting versus TBM ingot casting. For D-type oxides, the calcium-treated CC steels exhibit no higher values than untreated CC steels, contrary to the preceding SEP K 1 method.

Oxides: SAM Inclusion Rating System—Recently, the specimens have been also evaluated according to the SAM method, as described in ASTM A 295-84. Up to now, only results of VAR, EAF, and TBM-IC steels have been available. Their H_{50} - and H_{95} -frequencies are shown in Fig. 8. For both B- and D-type values, the order of the steel groups is in accordance with the SEP method. This is confirmed by comparing the SAM and SEP (K 4) rating values of the single specimens, as shown in Fig. 9. There is a correlation between the B-type values of both rating methods. However, no correlation could be found between the D-type values, because according to SEP (K 4), there occur nearly no inclusions of size classes ≥ 4 , whereas the SAM method already considers inclusions below that size class.

Oxides: Oxygen Content and Microscopic Cleanliness—Recently, customers have used information on the oxygen contents as a criterion for cleanliness. These are determined as an average of the analyses on the six specimens, used already for the cleanliness rating. The significant cumulative frequencies are listed in Table 1. For oxygen contents, VAR, EAF-IC, TBM-IC, and TBM-CC steels follow the same evaluation sequence as metallographic cleanliness values. As shown in Fig. 10, oxygen contents correlate with oxide cleanliness values K 4 (SEP 1570-71), only valid for the aforementioned degree of reduction after rolling or forging.

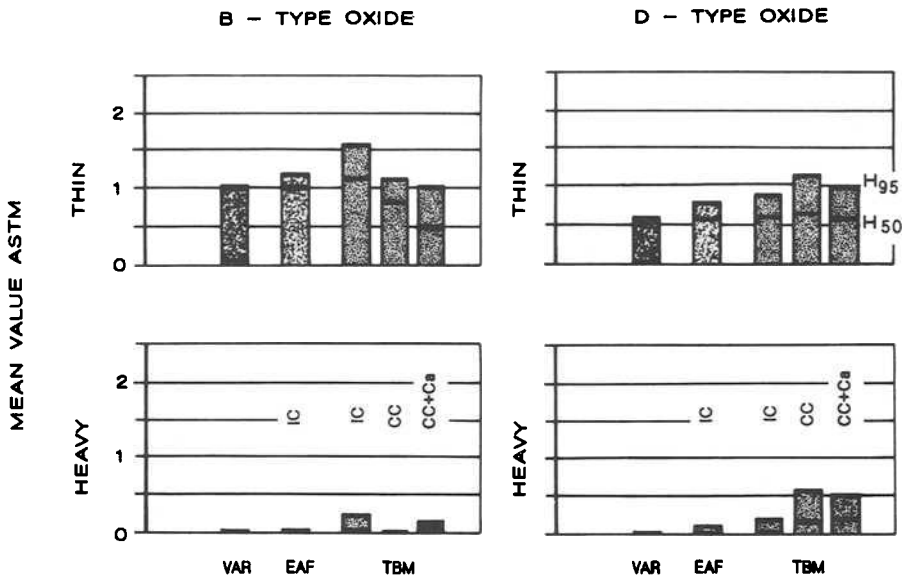


FIG. 7—Oxide inclusions according to ASTM E 45-76 Method A.

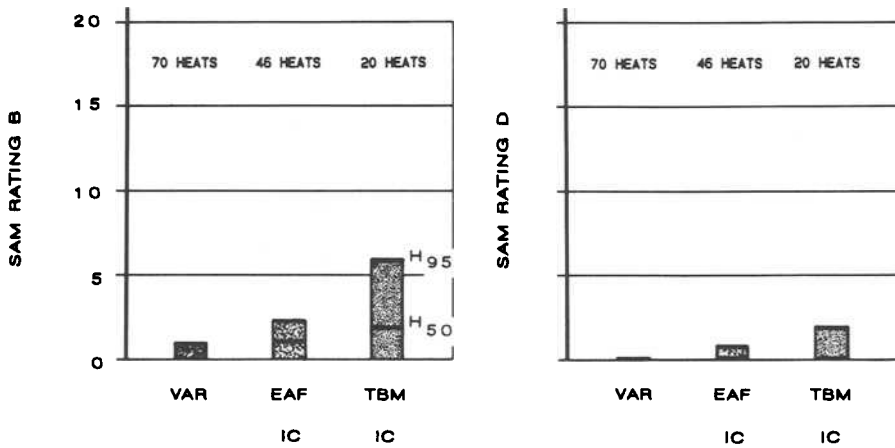


FIG. 8—Oxide cleanliness according to SAM rating method.

TABLE 1—Oxygen contents, in ppm, of various bearing steels (SAE 52100).

	VAR	EAF-IC	TBM-IC	TBM-CC
H ₅	3	6	8	7
H ₅₀	6	9	10	10
H ₉₅	9	12	12	12

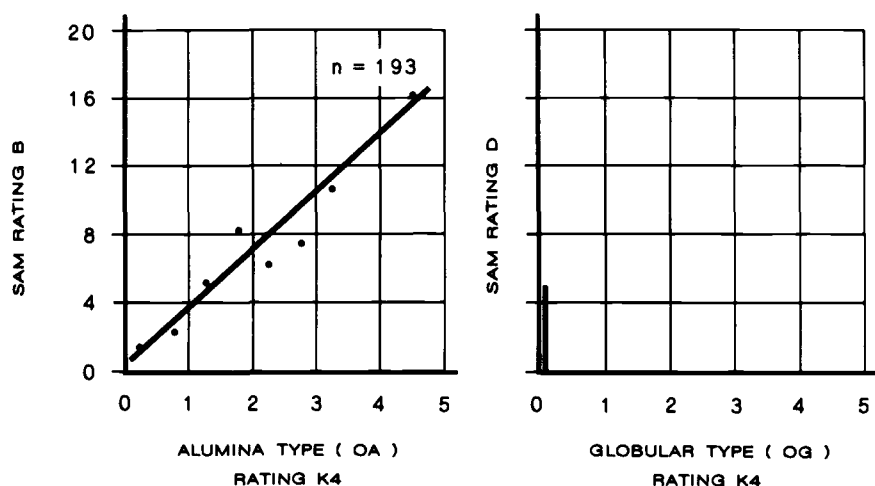


FIG. 9—SAM rating method in comparison with SEP 1570-71.

Based on this correlation it might be possible to substitute the metallographic determined values of the oxygen contents to define the quality in terms of cleanliness. Random test sampling, however, remains unchanged. It should be also taken into account that hereby no information is given about inclusion distributions of sizes and types.

In order for us to investigate the effect of calcium treatment, the composition of inclusions was analyzed microchemically after electrochemical isolation. Results are listed in Table 2. Metallographic methods used in the past had not been able to demonstrate to such an extent

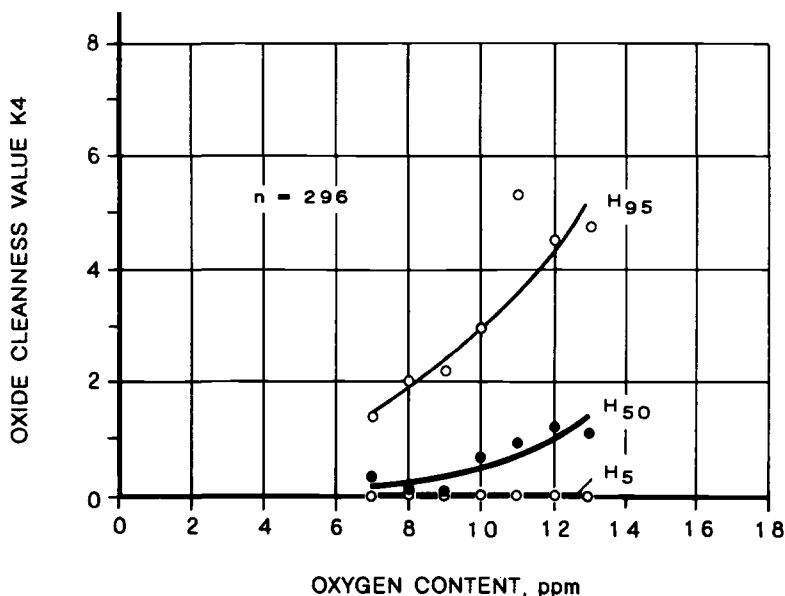


FIG. 10—Effect of oxygen on oxide cleanliness K 4 (SEP 1570-71).

TABLE 2—Chemical composition, in weight %, of oxide inclusions in various bearing steels (SAE 52100).

	EAF-IC	IC	TBM-CC	CC, calcium-treated
SiO ₂	31	35	45	26
Al ₂ O ₃	52	58	45	31
MnO	<1	<1	1	<1
FeO	5	2	2	1
CaO	2	2	4	37
MgO	2	1	2	2

the considerable difference between the CaO-contents of calcium-treated and that of nontreated steels as demonstrated by these new methods.

Titanium-Carbonitrides—Since the oxide inclusion content has been permanently lowered by ladle metallurgy, today customers demand lower titanium contents in order to minimize hard titanium-carbonitrides, which are said to reduce the fatigue life of the bearings. Up to now, titanium-carbonitrides have not been determined by the microscopical inclusion rating methods. Figure 11 represents the correlation between titanium contents and the number of the titanium-carbonitrides per square millimetre. Low titanium content of 10 ppm shows an average number of eight inclusions; from 40 ppm onwards the number increases progressively.

Sulfides—The sulfur content in bearing steels demanded by the customers is about ten times higher than the low oxygen contents aimed for, thus the amount of sulfides is higher than of oxides. As can be seen in Fig. 12, the sulfide inclusion values K 4 (SEP 1570-71) correlate closely with the sulfur contents. A similar correlation is found according to the Practice for

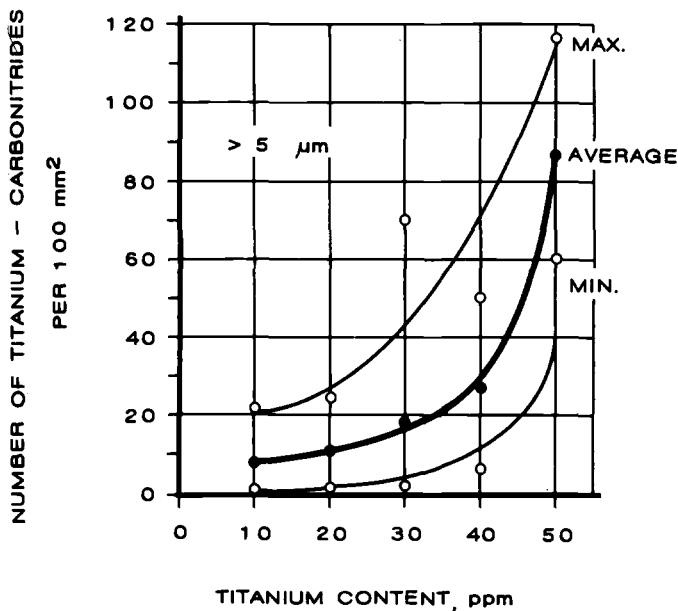


FIG. 11—Effect of titanium on number of titanium-carbonitrides.

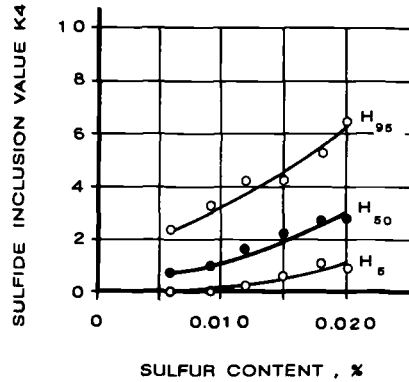


FIG. 12—Effect of sulfur on sulfide cleanliness K 4 (SEP 1570-71); IC steel.

Determining the Inclusion Content of Steel ASTM (E 45-76) Method A rating, as shown in Fig. 13.

For microscopic cleanliness in general, it has to be emphasized that the presented results of frequency distribution evaluations cannot be expected to be accepted as maximum values for each individual heat. Due to statistical distribution, a decisively larger amount of samples for each heat would have to be tested than usually done in everyday practice.

Effect of Sulfur and Carbide Grain Size on Machinability

In order to investigate the influence of sulfur on the machinability of bearing steels, we tested two heats with sulfur contents of 0.003 and 0.015%. Both heats were annealed to a coarse

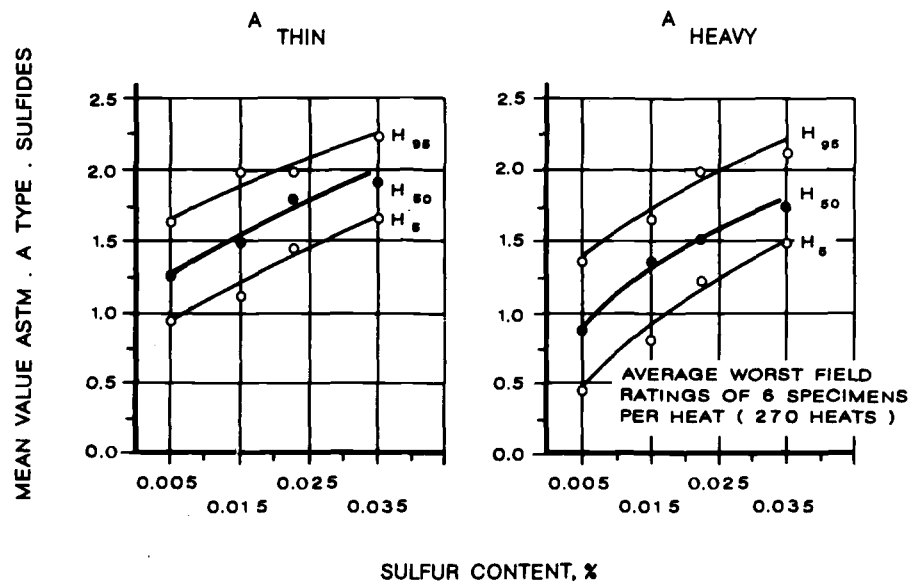
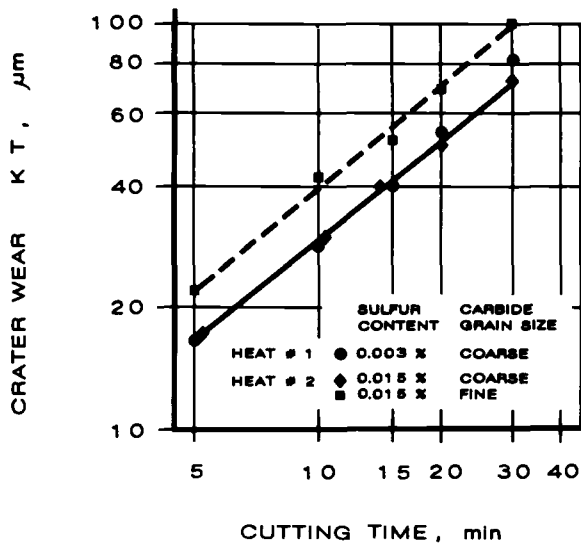


FIG. 13—Effect of sulfur on sulfide ratings according to ASTM Method E 45-76 Method A; IC steel.



SAE 52100 (100 Cr 6) . 65 mm ^ø

SPHEROIDIZED ANNEALED

CUTTING SPEED $V = 150$ m/min

FEED $S = 0.45$ mm

DEPTH OF CUTTING $A = 2.0$ mm

CARBIDE TOOL (P 10)

FIG. 14—Effect of sulfur and carbide grain size on tool wear.

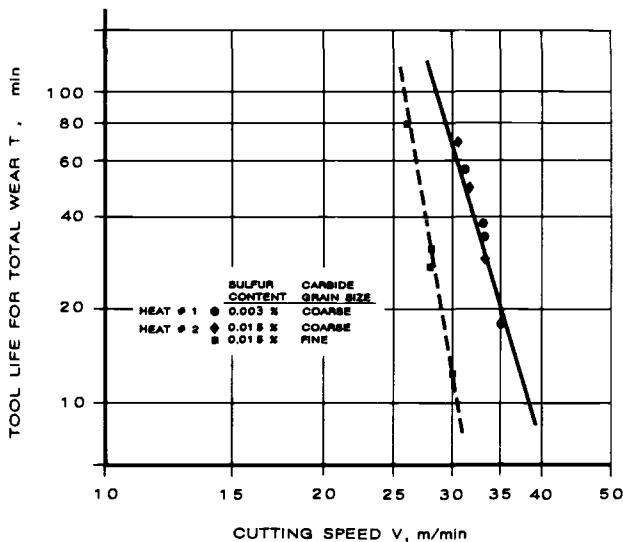
carbide grain size, and in addition the heat with the higher sulfur content was tested in a fine carbide grain size condition. Figure 14 represents the crater wear of tool in the machinability test at different cutting times. Both heats in the coarse grain condition exhibit equal crater wear in spite of their different sulfur contents. The fine grain condition results in more crater wear even though the sulfur content is higher.

The prevalence of the carbide grain size on machinability is also confirmed by the temperature tool life turning test, as is evident in Fig. 15. The coarse grain size effects significantly higher tool life times, independent of the sulfur contents within the investigated range.

Cleanliness and Fatigue Life Behavior

As described before, applied macroscopical and microscopical random test methods for inclusion rating are still useful as instruments for quality inspection, both for producer and customer. However, none of these methods can determine the total amount of inclusions, their sizes, and variety in chemistry. It is the authors' opinion that the metallographic rated data could not have been convincingly correlated to the existing fatigue life data.

In consideration of this deficiency, the majority of the customers specify maximum limits of cleanliness values according to the reported methods. They do random fatigue life tests only



SAE 52100 (100 Cr 6) . SPHEROIDIZED ANNEALED
 FEED $S = 0.25$ mm
 DEPTH OF CUTTING $A = 2.0$ mm
 HIGH SPEED STEEL TOOL (3207)

FIG. 15—Effect of sulfur and carbide grain size on tool life for total wear.

on the finished bearings to prove the sum of all parameters such as cleanliness, structure, geometry, and lubrication. Other customers do special random fatigue life tests on the semifinished products. This method gives equivalent results for integral comparison of steel quality.

In order to improve the fatigue properties of bearings, more investigations are necessary into the influence of critical sizes and the chemistry of inclusions.

Segregation in Ingot and Continuous Cast Steel

Continuous cast bearing steel has always been limited in application by customers because of its generally expected higher severity of segregation. Figure 16 shows the carbon segregation in the centerline of an as-cast 4.7-ton BEU-ingot, revealing amplitudes of positive and negative segregation in the upper part of the ingot, effected by V-segregation. They decline in the lower part changing into a steady negative segregation.

However, the above-mentioned heterogeneous coarse carbide distribution is homogenized by an appropriate reheating of the ingots, so that a sufficiently fine carbide size distribution according to SEP 1520-78 is achieved.

Figure 17 exhibits carbon segregation in the center lines of a 260 by 330 mm² CC-bloom and a 160 by 160 mm² CC-billet. The amplitudes of positive and negative segregation are present throughout the length of the strand. The average altitude, that is, the segregation degree, is significantly lower than in the upper part of the ingot. Therefore, continuous-cast steel should be allowed also the same application as ingot cast steel in this concern.

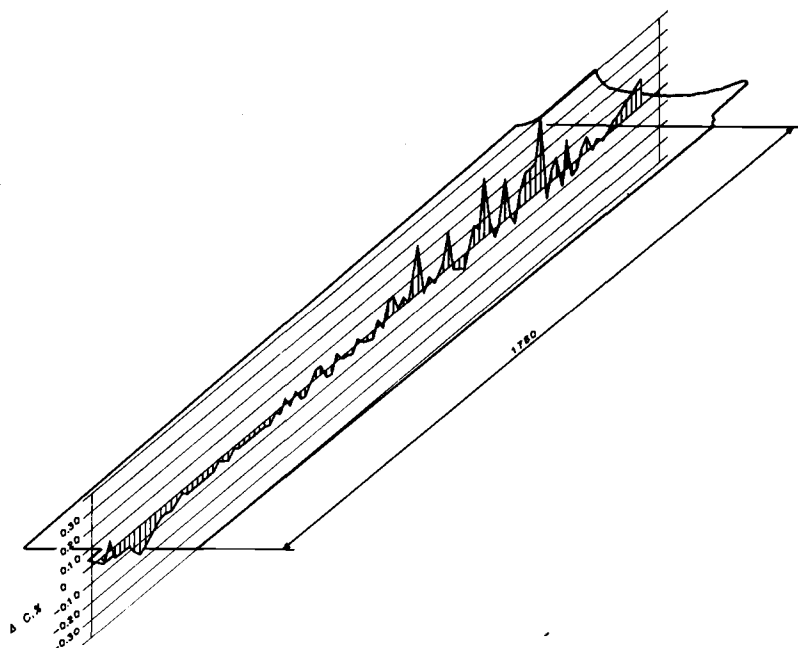


FIG. 16—Center carbon segregation in a 4.7-ton BEU-ingot (SAE 52100).

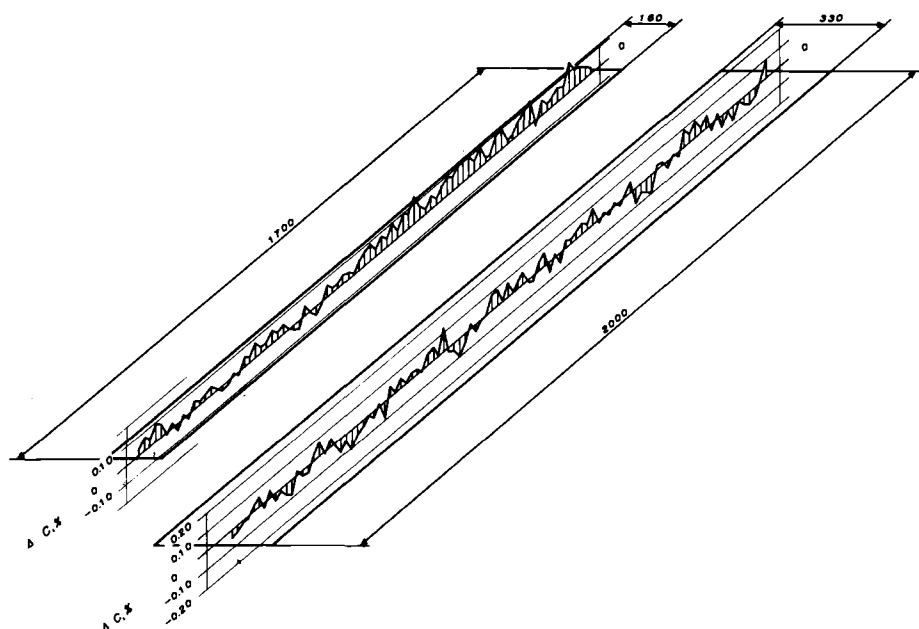


FIG. 17—Center carbon segregation in a CC-bloom (260 by 330 mm²) and a CC-billet (160 by 160 mm²).

Conclusions

Melting in EAF and BOF-TBM and different methods of aluminium deoxidation in the subsequent ladle treatment result in a similar degree of cleanliness. Air-melted ingot and continuous cast steels achieve nearly the same level of microinclusions as remelted steels. There still remains a decisive difference for macroinclusions; however, remelted steel is ensured to be free from macroinclusions, whereas some of them occasionally occur in air-melted steel.

The random test and metallographic evaluation systems do not allow uniform classifications of the various steel groups. None of the applied methods can determine definite inclusion contents. The number and sizes of the evaluated types of inclusions only can define statistical probability of their occurrence in a larger lot of heats, and the authors are not convinced that these can sufficiently be related to the fatigue behavior of the steels. Therefore, further investigation is necessary to reveal the effect of critical sizes and chemistry of the various inclusions on the fatigue behavior. First tests reveal that machinability seems to be governed by carbide grain sizes more than by sulfur contents in the investigated range of 0.003 to 0.015%. Comparison of the test results in terms of cleanliness and segregation shows that continuous casting should allow the same application as ingot casting.

Ladle Refining: An Integral Part of Bearing Steel Manufacture

REFERENCE: Davies, I. G., Clarke, M. A., and Dulieu, D., "Ladle Refining: An Integral Part of Bearing Steel Manufacture," *Effect of Steel Manufacturing Processes on the Quality of Bearing Steels*, ASTM STP 987, J. J. C. Hoo, Ed., American Society for Testing and Materials, Philadelphia, 1988, pp. 375-389.

ABSTRACT: The entire bearing steel output of Stocksbridge Engineering Steels is produced through a ladle furnace route which encompasses the major process elements necessary for the achievement of high product-cleanliness levels. The equipment and process routes involved are described, together with the vacuum arc remelting capability and its associated, high-integrity ultrasonic inspection unit. Improvements in steel cleanliness achievable with these manufacturing facilities have been assessed in terms of lower oxygen levels, better microcleanliness, and higher fatigue lives.

KEY WORDS: bearing steels, ladle furnace, vacuum arc remelted (VAR), steel cleanliness, SAE 52100, SAE 1070, SAE 1552, carburizing steels, ultrasonic inspection, rolling contact fatigue, oxygen content, inclusion assessment

Stocksbridge Engineering Steels, a Division of United Engineering Steels, is the major producer of bearing steels in the United Kingdom. As such, it has long recognized the need to satisfy ever more demanding product cleanliness levels in a consistent manner.

Most of the plant developments undertaken recently in its No. 2 Melting Shop therefore have had as twin goals the improvement of plant productivity and the attainment of ever cleaner steels. The latter improvements have been planned and conducted in close cooperation with bearing manufacturers, who have sought to exploit the benefits of material improvements in the design and use of higher efficiency components.

This paper briefly reviews recent plant enhancements in the No. 2 Melting Shop of Stocksbridge Engineering Steels and examines the improvements in product cleanliness and component fatigue life which resulted from them. Reference is also made to the manufacture and properties of vacuum arc remelted (VAR) bearing steels destined for use in the more critical applications. These VAR steels are almost exclusively produced from feedstock melted and cast in the No. 2 Melting Shop.

Melting Shop Developments

Processing Requirements

The effective production of the cleanest steels, sometimes referred to as ultra-clean steels, demands that several key factors of the melting and refining processes be meticulously followed

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and controlled. Among these factors are:

1. The complete removal from the steel of the oxidizing slag created during melting in the furnace.
2. Steelmaking in a secondary vessel equipped for
 - (a) refining and de-oxidation,
 - (b) desulfurisation,
 - (c) alloying to narrow compositional limits,
 - (d) effective removal of de-oxidation products,
 - (e) degassing to remove hydrogen, and
 - (f) temperature control.

Having produced a very clean steel, it is equally vital to avoid contamination by its environment. The lining of the secondary steelmaking vessel must be inert and erosion resistant, while reaction between the bulk melt and the atmosphere must be prevented by means of a full slag cover or by the use of an inert gas atmosphere above the melt in the secondary vessel.

High levels of steel cleanliness can only be retained during ingot casting or continuous casting if the ladle stream is fully shrouded by an inert gas or by close-fitting refractory tubes, as appropriate, and if erosion-resistant refractories are employed in the ingot holloware and for the tundish linings.

Plant Developments

Details of recent developments in the No. 2 Melting Shop have been reported elsewhere [1]. Briefly, however, the Shop is now equipped with two electric arc furnaces (100 liquid tonnes, 500 kVA/tonne, and 150 liquid tonnes, 330 to 440 kVA/tonne), each of which incorporates a submerged taphole and operates a hot heel practice to prevent carryover of oxidizing slag on tapping.

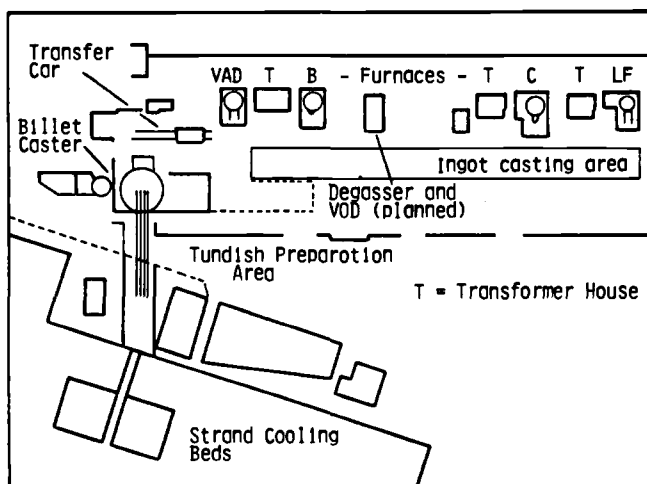


FIG. 1—Stocksbridge Engineering Steels, No. 2 Melting Shop, plant layout.

Steel from the 100-tonne furnace is processed in a vacuum arc degassing (VAD) unit, equipped for argon stirring, arc heating, vacuum degassing, and alloying, while melts from the 150-tonne furnace are processed in a ladle furnace (LF) equipped for argon stirring, arc heating, and alloying. A tank degassing unit is also available with additional alloy trimming and argon stirring facilities. This unit is currently being modified for vacuum oxygen decarburization (VOD) operation. A schematic illustration of the plant configuration is shown in Fig. 1, while plant details are summarized in Table 1. The principal features of the two furnace routes for the production of clean steels are summarized in Table 2.

The treatment ladles are lined with high alumina refractories and employ graphitized magnesite (MgO) slag lines. Sidewall-mounted canned plugs and directionally porous bottom plugs are used for argon stirring of the ladle-treated melts. The entire output from the LF is currently ingot cast, whereas up to 70% of the VAD-treated melts are continuously cast on a four-strand vertical machine capable of casting round or square sections up to a section size of about 200 mm square [2].

The VAD was commissioned in 1981 and the LF in 1985. All the output from No. 2 Melting Shop is now refined in these secondary steelmaking vessels.

TABLE 1—*Steelmaking facilities, No. 2 Melting Shop: Stocksbridge Engineering Steels.*

	B Furnace	C Furnace
Shell diameter, m	6.15	6.75
Water-cooled panels	yes	yes
Water-cooled roof	yes	yes
Nominal tap weight, tonnes	80 to 100	150
Transformer rating, MVA	50	50 (60)
Supplementary burners	no	3
Secondary voltage, V	235 to 295	235 to 593
Secondary current, KA	50	48
Electrode diameter, mm	500	600
Fume extraction	semi-direct to precipitators	direct to bag house
Slag separation	submerged tap-hole (or reladling)	submerged taphole
SECONDARY STEELMAKING FACILITIES		
	B Furnace Casts	C Furnace Casts
Unit	VAD	LF
Manufacturer	Vacmetal (UK) Ltd.	Stocksbridge Engineering Steels
Commissioning date	Sept. 1981	Jan. 1985
Ladle diameter	3.1 m	oval
Capacity	100 tonnes	150 tonnes
Transformer	15 MVA	35 MVA
Secondary voltage	231V (plus 4 other taps)	240V (plus 8 other taps)
Secondary current	30 KA per phase	32 KA per phase
Electrode size	350 mm	350 mm
PCD	1100 mm	900 mm
Alloying facility	13 Hopper System	13 Hopper System
Degassing facility	integral, by means of 4-stage steam ejector system	separate, tank degasser
Stirring method	argon, at up to 15 Nm ³ /h through sidewall-mounted, canned plug	argon, at up to 15 Nm ³ /h through sidewall-mounted, canned plug and bottom, directional porosity plug

TABLE 2—Process elements for clean steels, Stocksbridge Engineering Steels' capability.

Process Route Element	B Furnace Route	C Furnace Route
Slag-free tapping	submerged taphole	submerged taphole
Ladle stirring	argon through sidewall plug	argon through sidewall and bottom plugs
Slag control	specific synthetic slags	specific synthetic slags
Re-heating	arc heating in VAD	arc heating in LF
Degassing	integral part of VAD	separate tank degasser
Alloying	via 13 alloy hoppers	via 13 alloy hoppers
Ladle refractories	75 to 85% Al_2O_3 walls and bottom 92% MgO slag line	82 to 85% Al_2O_3 walls and bottom 96% MgO slag line
Ingot teeming shroud	inert gas	inert gas
Ingot holloware	60% Al_2O_3	60% Al_2O_3
CC ^a ladle to tundish	graphitized Al_2O_3 shroud	not applicable
CC tundish	700-mm deep, MgO-lined	not applicable
CC tundish to mould	graphitized Al_2O_3 submerged entry shroud and flux, or inert gas	not applicable

^aCC = continuous caster.

Product Cleanness

Progressive improvements in the microcleanness of alloy carburizing bearing steels and reductions in the oxygen content of through-hardening bearing steels, associated with the introduction and development of the ladle furnaces, are shown in Figs. 2 and 3, respectively. Figure 2 shows how the JK B thin inclusion rating of carburizing alloy steels has been reduced following the introduction of the ladle furnaces. The mean oxygen levels and ranges, as depicted in Fig. 3, have been obtained by assessing six representative product samples from each cast and carrying out the determinations in the Stocksbridge Engineering Steels Analytical Laboratory, which has also participated in the round-robin oxygen determination exercise reported by Green et al. [3].

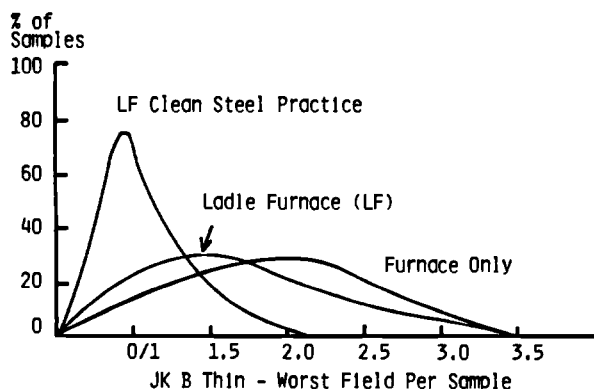


FIG. 2—Effect of steelmaking practice on cleanliness.

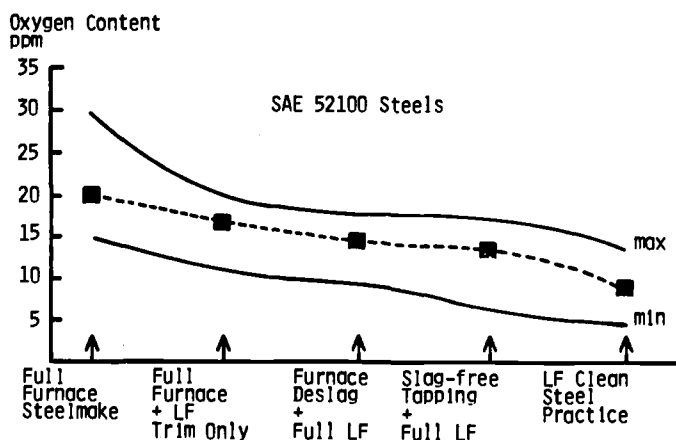


FIG. 3—Effect of steelmaking practice on oxygen content of SAE 52100 steels.

Cleanliness and Fatigue Properties

Microcleanliness

Rating of through-hardening bearing steels by the SAM method has shown that adoption of the LF practice brought about a reduction in the incidence of B-type inclusion stringers and a drastic improvement in the level of D-type random oxide particles.

A more detailed evaluation of the effects of introducing the newer practices has been made, using a manual, quantitative optical assessment method, aimed at characterizing the oxide inclusion population. The principles of this method are given in Appendix I. Its main features are that examinations are made at $\times 500$ and $\times 100$ on transverse and longitudinal sections. The combination of the $\times 500$ magnification and manual assessment permits discrimination between oxide and sulfide phases.

The distribution of oxide particle sizes in transverse sections of SAE 52100 steel tube-billets, made before and after the adoption of the full LF practice, is shown in Fig. 4. The principal observation is that the maximum size of oxide particle observed was reduced from 32 to 13 μm , with the mean oxide particle size, in the range measured, falling from 2.3 to 1.6 μm (Table 3). Measurement of the percentage of oxide particles encapsulated by sulfide showed that this increased in the steels of lower oxygen content (Table 4).

The longitudinal sections revealed that the incidence of aligned inclusions, or stringers, was similar in both steels, in respect of both the projected length values and the mean lengths of individual stringers (Table 5).

All samples contained a low level of titanium carbonitride [Ti(C,N)] particles up to a maximum size of 8 μm . The source of Ti(C,N) is likely to have come both from scrap and from ferro alloy additions: controls have been maintained on both these sources throughout the introduction and development of the ladle furnaces, and therefore it is not surprising the level and nature of Ti(C,N) has remained constant. The proportion of Ti(C,N) does increase as the oxide level drops, but the authors have no information to show that, at the levels reported, the influence on fatigue performance is significant.

The principal findings for the samples of alloy carburizing steels were similar to those for the Society of Automotive Engineers (SAE) 52100 material. Again, the use of the full LF practice reduced the numbers of larger oxides present, the maximum size observed being reduced from

TABLE 3—Oxide particle populations in bearing steels as measured on transverse sections.

Steel Type and Practice	Particle Size, μm		Oxygen Content, ppm	Area of Count, mm^2
	Oxide Mean	Largest Observed		
SAE 52100				
pre-LF	2.3	32	22	2700
post-LF	1.6	13	13	
post-LF	1.6	4	9	
VAR	2.3	4	5	240
Ni-Cr-Mo carburizing grade				
pre-LF	2.6	25	...	240
post-LF	2.1	16	...	240
SAE 1552 continuous cast and forged VAD treated	1.6	26	...	240

TABLE 4—Oxide inclusion association with and encapsulation by sulfide (% by number).

Grade	Practice and Sulfur Content, weight %	Percent Single	Percent Associated	Percent Encapsulated
SAE 52100	pre-LF 0.020	12	81	7
SAE 52100	post-LF 0.017	30	51	20
SAE 52100	post-LF 0.015	5	65	30
Alloy carburizing steel	pre-LF 0.016	34	59	8
Alloy carburizing steel	post-LF 0.024	74	26	0
AISI 1552	VAD continuously cast 0.024	13	72	16

TABLE 5—Longitudinal section measurements of oxide inclusions present as stringers.

Steel Sample	Projected Length Stringers, $\mu\text{m}/\text{mm}^2$	Mean Stringer Length, μm	Average No. Particles/Stringer
SAE 52100, pre-LF	10	80	5
SAE 52100, post-LF	12	90	4
SAE 1552 billet continuously cast and forged and VAD treated	2	50	6
Ni-Cr-Mo carburizing grade pre-LF	6	70	4
Ni-Cr-Mo carburizing grade, post-LF	5	90	9

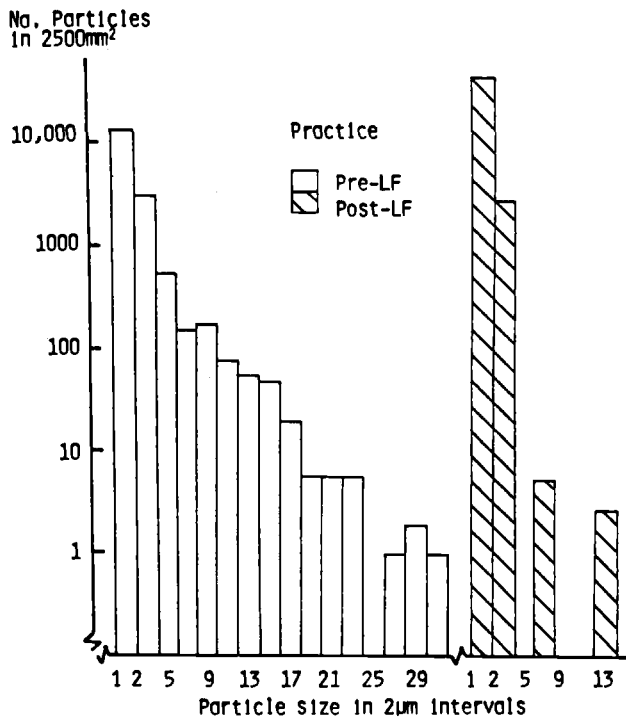


FIG. 4—Distribution by size of all oxide particles (transverse section) above $1\text{ }\mu\text{m}$ in SAE 52100 bearing steels.

25 to $16\text{ }\mu\text{m}$ and the average particle size from 2.6 to $2.1\text{ }\mu\text{m}$ (Fig. 5). For these steels, any difference in the level of sulfide encapsulation was masked by a difference in sulfur content (Table 4). The projected length of stringers and the mean stringer length were again similar for both practices (Table 5).

For comparison, similar inclusion counts were made on a sample of SAE 1552 steel, continuously cast as billet and forged to 80 mm round. The particle size distribution, Fig. 6, and Tables 3 and 5 show that an average oxide particle size similar to that of the LF SAE 52100 material was obtained, but with the maximum particle size observed larger at $26\text{ }\mu\text{m}$, for a smaller area sample.

The extent of sulfide encapsulation of the oxides was intermediate between that of the pre- and post-LF practices for the SAE 52100 steels.

There was a reduction in both the projected and mean stringer lengths in this material when compared with the SAE 52100 steels. This reflects the significant difference in the solidification conditions and reduction ratios between the two routes.

For all steels, the sulfide encapsulation of oxides was generally limited to particles of $6\text{ }\mu\text{m}$ and smaller.

Steel Cleanliness and Rolling Contact Fatigue Performance

The correlation of inclusion content with fatigue performance is discussed in detail in a paper by Hampshire and King [4].

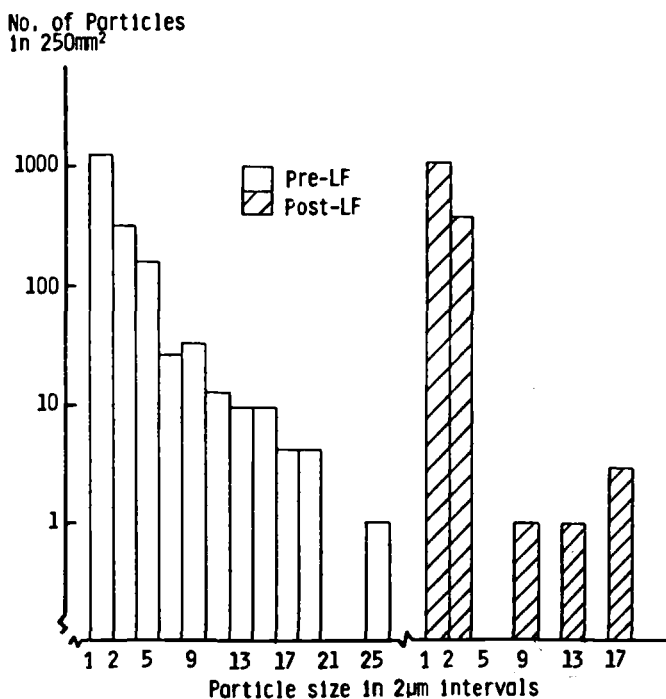


FIG. 5—Distribution by size of all oxide particles (transverse section) above 1 μm in alloy carburizing steels.

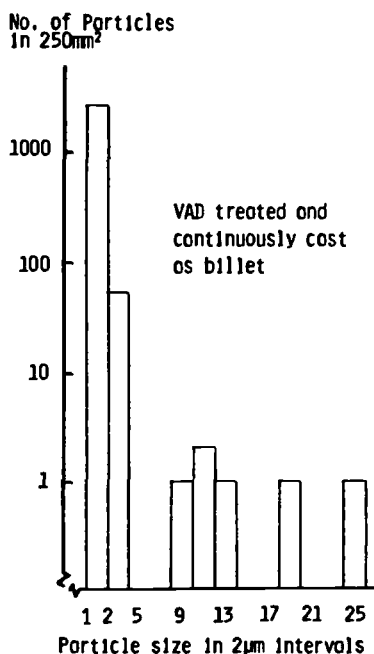


FIG. 6—Distribution by size of all oxide particles (transverse section) above 1 μm in SAE 1552 steel, continuously cast and forged.

Stocksbridge Engineering Steels makes use of the "Unisteel" washer type rolling contact fatigue test, the basic features of which are summarized in Appendix II.

The correlation of the B_{10} life observed in this test with the Ransome, Hoffman, and Pollard (RHP) summed B rating [4] for through-hardening steels is shown in Fig. 7. The case carburizing steels follow the same trend at the higher cleanliness levels, but have a relatively higher fatigue life as the inclusion content increases. The correlation with the B rating holds for steels of relatively low D ratings, although the B_{10} life is impaired by the presence of D-type particles.

The error bars in Fig. 7 are calculated for the test conditions shown. They serve to emphasize the problem of quantitative assessment of the fatigue properties of clean steels, in terms of test duration, without resorting to unrealistic test loads.

Nature of the Oxide Inclusion Phases

On the optical scale, the only duplex oxides seen in the SAE 52100 steels were isolated, large ($> 15 \mu\text{m}$) particles found in the sample representing pre-LF practice. In this material, the oxides were, principally, Al_2O_3 .

Nondispersive X-ray energy analysis of particles in steels from the LF indicated that the (apparently) single-phase oxides, in the size range 4 to $12 \mu\text{m}$, contained magnesium. The level

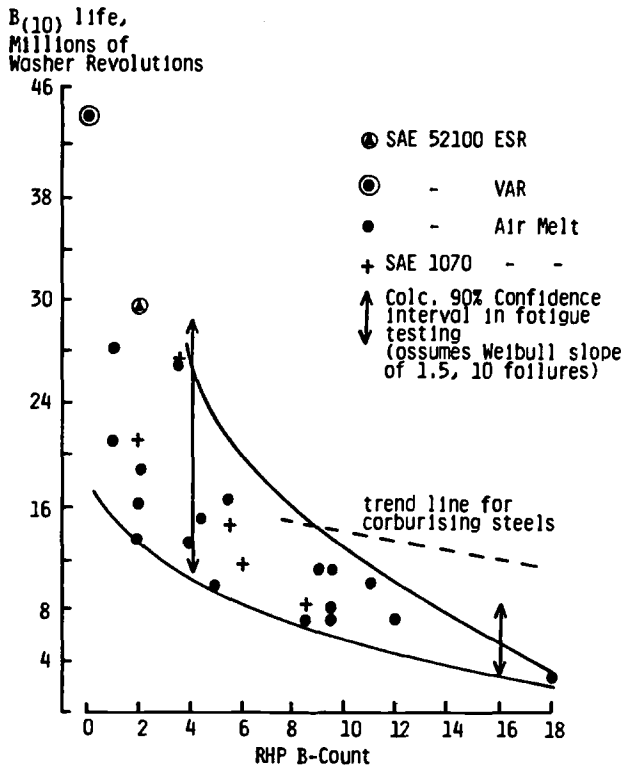


FIG. 7—Correlation between the RHP B-count for oxide inclusions and rolling contact fatigue life in the unisteel washer test.

was variable, but was too low for the oxide to be a pure spinel of the type $\text{MgO} \cdot \text{Al}_2\text{O}_3$. It is probable that the particles are in fact duplex of the $\text{Al}_2\text{O}_3 + \text{MgO} \cdot \text{Al}_2\text{O}_3$ type, the magnesium originating in the refractories used for the ladle slag line.

The free sulfide particles were found to be manganese sulfide, (MnS) , whereas those associated with oxides were found to be $(\text{Ca,Mn})\text{S}$ containing 1 to 2% calcium, arising from the synthetic slag used during ladle refining.

Production of Bearing Steels by VAR

For the most demanding product applications, clean bearing steels made via the VAD or LF route are further refined at Stocksbridge Engineering Steels by vacuum arc remelting (VAR). An outline description of the remelting facilities available is given in Table 6. The exceptionally clean steels produced via this two-step process route are currently assessed by conventional analytical and optical methods. Cleanliness levels achieved are typified by the product oxygen levels shown in Fig. 8.

While microcleanliness standards are consistently achieved under conditions of precise process control, steel users are requiring that the absence of isolated, large inclusions in the product be verified. For this reason, there has been interest in the use of electron beam, button melting methods. Such an approach is costly. Moreover, the volume examined, as with metallographic assessments, is relatively small, and the interpretation of results poses substantial problems.

To provide assurance of the overall integrity of material, Stocksbridge Engineering Steels has installed and is commissioning a full-immersion, ultrasonic testing unit for the total volume inspection of its remelted products, which is capable of resolving discontinuities on the scale of 300 to 400 μm . This facility, the specifications of which are given in Table 7, complements existing methods for product cleanliness verification.

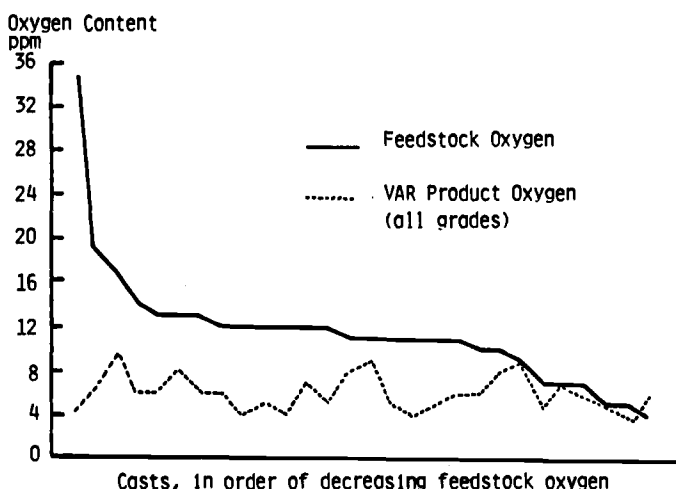


FIG. 8—Relationship between feedstock and VAR product oxygen contents for all grades produced by SES.

TABLE 6—*Remelting units, Stocksbridge Engineering Steels.*

	Electroslag Refining	Vacuum Arc Remelting ^a
Commissioning date	1976	1982
Manufacturer	Birlec Ltd.	Lectromelt Ltd.
Electrical rating	2000 KVA	up to 90 V and 17.5 KA (DC)
Operating voltage	60–80 (AC)	25–30
Operating current	12 to 17 KA	10 to 14 KA
Ingot size	560-mm square or 635-mm diameter	510-mm or 660-mm diameter
Maximum ingot weight	8 tonnes	5 tonnes

^aFurther installation planned.TABLE 7—*Immersion ultrasonic unit plant specification.*

Commissioning date	Oct. 1986
Size range	60 to 300 mm diameter 1 to 11 m length
Piece weight (maximum)	6.5 tonnes
Bar surface condition	> 100 CLA
Temperature	ambient
Inspected area	3-mm surface dead zones, full length cover
Sensitivity	U.S. MIL-STD-2154 Class AA $\frac{2}{64}$ in. (0.8 mm) flat bottom hole (FBH) Class AAA $\frac{1}{64}$ in. (0.4 mm) FBH—size limit Rolls Royce RPS707

Problem of Cleanliness Assessment in Bearing Steels

It is generally accepted that chart-based worst-field assessment methods are of limited value for quantitative specification of the microcleanliness of modern bearing steels. Quantitative estimates based on chart sizing intervals, as in the Stahl und Eisen Pruefblatt (SEP), SAM, and RHP methods, are a significant improvement. However, all optical methods suffer from the drawback that they are very time-consuming if a large area of a sample is to be examined at magnification sufficiently large to permit resolution of the different inclusion types. (Even automated methods require significant times for full-field data acquisition.) Manual methods may be subject to variations in the performances of a single operator, as well as between different operators. Automatic systems based on image analysis also require standardization of image recognition conditions, together with stringent and reproducible specimen preparation.

An indication of the problem of specimen size in the examination of bearing steels can be gained from the data in Fig. 4. If it is assumed that the frequency of occurrence by size of the oxide microinclusions is distributed according to the Poisson relationship, then the specimen area to detect a particle of a given size can be estimated. Figure 9 shows that the observed distribution for the sample of pre-ladle furnace SAE 52100 steel can be represented by a Poisson distribution of the form

$$E_{(n > x)} = N_e - \frac{x}{3.5}$$

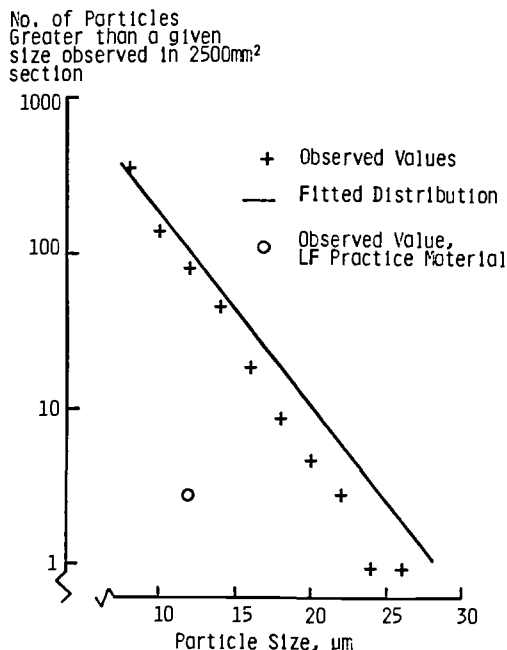


FIG. 9—Size distribution for all oxide particles greater than 8 μm in sample of SAE 52100 steel made by pre-LF practice.

where $E_{(n > x)}$ is the expected number of particles of size greater than x , and N is the total population size. This relationship may be used to estimate the area of specimen which must be examined to have a 90% probability of finding an inclusion of a given size (Fig. 10).

It must be emphasized that there is no evidence to justify the assumption that all oxide inclusions present in a steel specimen form part of the population assessed on the microscale. Extrapolation of the line in Fig. 10 to larger particle sizes is thus not justified by experimental observations. However, the data indicate that areas in excess of 100 cm^2 must be examined to detect an oxide particle of 30 μm section. For the improved, post-ladle furnace particle, the required area is not estimable.

There is thus a gap in the range of detectable oxide inclusions between those assessed by conventional metallographic sampling and those by larger volume surface indication or production ultrasonic methods. The gap represents at least an order of magnitude in particle size and probably covers particles which may originate both from deoxidation or reoxidation reactions and from exogenous sources.

As clean steel manufacturing techniques continue to evolve, it is clear that the verification of product standards will rely increasingly on precise and consistent process control. The record of the exercise of this control will itself provide the evidence for the avoidance of harmful inclusions from deoxidation and reoxidation reactions and exogenous sources. This stage is being approached, as indicated by the generally consistent microcleanliness levels achieved by major steelmakers. However, monitoring of product oxygen levels and, for some products, the application of full ultrasonic monitoring, remain necessary to provide customers with the confidence given traditionally by the optical assessment of small specimens.

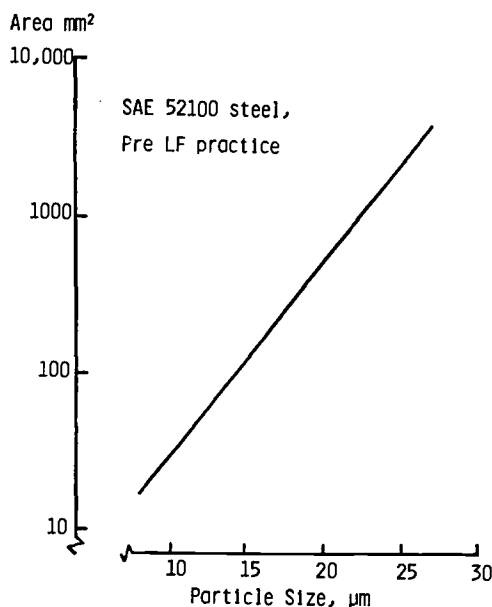


FIG. 10—Distribution of Fig. 9 used to predict area to be examined for 90% probability of locating an oxide particle equal to or greater than a specified size.

Summary and Conclusions

Stocksbridge Engineering Steels' entire liquid steel production is refined in ladle furnaces, thus permitting strict adherence to the process control standards needed to achieve high levels of steel cleanliness in a consistent manner. Further precautions are taken during ingot teeming and continuous casting to prevent contamination of the steel from refractories and from an oxidizing environment. The resulting product may be further refined by vacuum arc remelting.

In the case of bearing steels, oxygen contents have been reduced from typical values of 20 ppm to less than 10 ppm by the introduction of ladle refining, with further reduction to less than 6 ppm being achieved by vacuum arc remelting.

These improvements have been accompanied by a reduction in the number of larger microinclusions and have been reflected in a marked increase in rolling contact fatigue life.

The problems associated with the qualitative assessment of the low inclusion content found in modern bearing steels have been discussed. There remains an inspection gap between state-of-the-art production inspection methods based on ultrasonics and microcleanliness assessments using small sample metallographic procedures. It is considered that adequate microcleanliness levels are achieved by modern steelmaking methods. Given the difficulties in verifying product cleanliness at current levels, increased emphasis must be given to the assurance of product quality, on the basis of strict adherence to clearly defined manufacturing routes. Harmful inclusions in bearing steels must be engineered out by process control, and the progressive introduction and use of statistical process control in the melting shop, mill, and testing areas will bring about further improvements.

Acknowledgments

The authors wish to thank D. R. Stone, Managing Director, Stocksbridge Engineering Steels, and Dr. R. Baker, Director of Research, British Steel Corporation, for permission to publish this work. They also acknowledge the valuable assistance given by M. J. Worrall and P. F. Morris.

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APPENDIX I

Optical Inclusion Assessment Methods

A detailed investigation into the effects of changes in steelmaking practice (such as the introduction of LF operation), on the resulting inclusion population necessitates information on the nature, size distribution, and frequency of occurrence of the inclusions. In the work reported here, manual optical assessments have been used.

The assessment has been based on three counts. The basic inclusion population was determined on transverse sections at a magnification of $\times 500$, sizing particles larger than $1\text{ }\mu\text{m}$ into $1\text{-}\mu\text{m}$ class intervals based on an approximation of the particle $2\text{-}\mu\text{m}$ section to the diameter of a circle by means of a graticule. The particles were subdivided on the basis of optical appearance into one of the following categories:

- (a) single-phase oxides,
- (b) single-phase oxides, associated with sulfide or sometimes $\text{Ti}(\text{C},\text{N})$,
- (c) single-phase oxides, encapsulated by sulfide,
- (d) multi-phase oxides, distinguished as above,
- (e) single-phase sulfide,
- (f) sulfide associated with oxide, or sometimes $\text{Ti}(\text{C},\text{N})$,
- (g) $\text{Ti}(\text{C},\text{N})$ particles, and
- (h) $\text{Ti}(\text{C},\text{N})$ particles, associated with sulfide or oxide.

Particles greater than $8\text{ }\mu\text{m}$ were sized at a magnification of $\times 100$. In all cases, the sizes reported are for the oxide phase only.

To supplement these results, longitudinal sections were also prepared and examined at $\times 100$ to investigate the alignment of oxide inclusions and $\text{Ti}(\text{C},\text{N})$ particles as stringers. Inspection at $\times 500$ was used to confirm the inclusion phase identity. From the longitudinal sections, the following parameters were derived:

- (a) projected length of oxide inclusion stringers, ($\mu\text{m}/\text{mm}^2$),
- (b) mean oxide stringer length, and
- (c) average number of oxide particles in a stringer.

A stringer is defined as a row of oxide particles, greater than $38\text{ }\mu\text{m}$ (B thin $\frac{1}{2}$), which is aligned within the diameter of the field at $\times 100$. Where oxides were observed encapsulated in

or associated with sulfide particles, they were measured as oxide stringers, the presence of the sulfide being ignored.

Fragmented Ti(C,N) particles were also recorded in the same manner as oxides, where the trail length exceeded the B thin $\frac{1}{2}$ length rating.

The total area examined at $\times 500$ was 10 mm² for the transverse sections. The area examined at $\times 100$ was 240 mm² except for two samples of SAE 52100 steels, where the area was 2700 mm². The longitudinal areas examined at $\times 100$ and $\times 500$ were 240 mm² in each case. Where counts have been presented as histograms, results from individual counts have been normalized to the field area shown.

APPENDIX II

Unisteel Rolling Contact Fatigue Test Conditions

Flat Washer Test: 62-mm-diameter test track.

Test speed: 1500 revolutions/minute.

Contact area: 0.42-mm diameter.

Nine balls: each 7.9-mm diameter.

Load per ball: 371 N.

Maximum calculated Hertzian stress: 3725 N/mm².

Surface finish: aim 0.1 μ m CLA.

Lubrication: Shell 717 (total loss) operating at about 50°C.

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ISBN 0-8031-0999-7