ABSTRACT

Steel suitable for use as rolling elements, such as bearings and gears, and a method for producing the steel are disclosed. The surface layer of the steel, which is directly responsible for the rolling-fatigue strength of the steel, is composed of a structure comprising quasi-carbide dispersed in the martensite matrix (hereinafter referred to as "quasi-carbide structure").

5 Claims, 28 Drawing Figures
FIG. 1

[C\% - C after Annealing
[C\% - 2 After Rapid Heating and Quenching
[Cr\% - Cr after Annealing
[Cr\% - 2 After Rapid Heating and Quenching

Solid Dissolution Limit of C in Austenite

Matrix (Martensite)

(Fe-M)\textsubscript{4-11}C

Retained Austenite

Outer Shell

MICRONS

0\textsubscript{1}

0\textsubscript{3}
FIG. 2

FIG. 3
FIG. 4

x 5000
5% picral etched

FIG. 5

\[ \text{Quenching Temperature: } 950^\circ C \]
\[ \text{Average Heating Rate: } 5^\circ C/\text{sec.} \] x 5000
x 10,000 5% picral etched

FIG. 6

x 10,000 5% picral etched
(C% = 1.05%)

FIG. 7
FIG. 8

x 5000 5% picral etched

FIG. 9

x 5000 5% picral etched
FIG. 12

Average Heating Rate 50-120 °C/sec above 850°C
Surface of Carburized Layer: C%: 1.36%
FIG. 13

1. Cementite

2. Cementite + Diffused Layer

3. Higher Temperature
   - Quasi-Carbide (earlier stage)

4. Convexed
   - Quasi-Carbide (middle stage)

5. Quasi-Carbide (final stage)
   - Two quasi-carbides combined together

6. Quasi-Carbide (final stage)
   - Martensite in the matrix to grow coarse

7. Retained Austenite
FIG. 14

(a) $t = 0$  
$C_0 - C_C$  
$C_C$ in Cementite

(b) $t = t_1$
$C_0 - C_C$  
$C_C$ in Cementite

(c) $t = t_2$ ($t_2 > t_1$)
$C_0 - C_C$  
Microns
FIG. 17

Average Life Ratio as Compared With Material Quenched in an Electric Furnace

Max. Heating Temperature (°C)

FIG. 18

Average Life Ratio as Compared With Material Quenched in an Electric Furnace

Life Ratio

Residual Stress

Max. Heating Temperature (°C)
FIG. 19

Average Life Ratio as Compared With Material Quenched in an Electric Furnace

Max. Heating Temperature (°C)

Quasi-Carbide Structure

FIG. 20

C% vs. Depth from Surface (mm)
FIG. 23

Average Life Ratio as Compared With Material Quenched in an Electric Furnace

- Preheating Temperature (°C)
- Quasi-Carbide Vol %
- Accicular Martensite
- Retained Austenite
- Small Amount of Quasi-Carbide and Cementite
FIG. 24

Quasi-Carbide Structure Average Life Ratio as Compared With Material Quenched in an Electric Furnace

Max. Heating Temperature (°C)

FIG. 25

Depth from Surface (mm)

C, N%, N%
<table>
<thead>
<tr>
<th>Designation</th>
<th>Heating Method</th>
<th>Heating Temperature</th>
<th>Hardness after Tempering</th>
<th>Retained Austenite</th>
<th>Slide Wear Condition</th>
</tr>
</thead>
<tbody>
<tr>
<td>I</td>
<td>Electric Furnace</td>
<td>840°C</td>
<td>63 RC</td>
<td>10%</td>
<td>V=1.25pp, No Lubrication</td>
</tr>
<tr>
<td>II</td>
<td>&quot;</td>
<td>&quot;</td>
<td>840</td>
<td>60</td>
<td>&quot;</td>
</tr>
<tr>
<td>III</td>
<td>&quot;</td>
<td>Induction</td>
<td>1000</td>
<td>63</td>
<td>&quot;</td>
</tr>
<tr>
<td>IV</td>
<td>&quot;</td>
<td>&quot;</td>
<td>1100</td>
<td>61</td>
<td>&quot;</td>
</tr>
<tr>
<td>V</td>
<td>&quot;</td>
<td>&quot;</td>
<td>1170</td>
<td>60</td>
<td>&quot;</td>
</tr>
</tbody>
</table>

**FIG. 26**

- Sliding Distance (m)
- Wear Amount (mm³)
STEEL SUITABLE FOR USE AS ROLLING ELEMENTS

CROSS-REFERENCE TO PRIOR APPLICATION:

This is a continuation-in-part of Ser. No. 297,644 filed Oct. 16, 1972 and now abandoned.

FIELD OF INVENTION

The present invention relates to steel for rolling elements, such as bearings and gears, exhibiting improved rolling fatigue strength characteristics.

BACKGROUND INFORMATION

For improving the rolling fatigue strength of steel, various techniques have been developed. In the field of steelmaking processes proper, vacuum degassing and vacuum remelting have been proposed for this purpose. In the field of heat treatments, mar-stressing is often times performed, while in the field of steel-working specific fiber orientation by hot-working and aus-forming has been attempted.

However, the improvement in the rolling-fatigue life obtained by these conventional procedures is very limited and these procedures cause various problems, such as increased production cost, low productivity and limitation of steel grades.

SUMMARY OF INVENTION

The present invention remarkably improves the rolling-fatigue life by combining a particular melting method with a particular working method, while at the same time enhancing the productivity. The invention is applicable both to through hardened steel and case hardened steel.

The present invention provides a novel steel for rolling elements and a method for producing the same. Briefly, the steel is characterized in that the surface layer which is directly responsible for the rolling-fatigue strength has a structure which is dispersed with quasi-carbide (a granulated substance as hereinafter defined) in the martensite matrix (hereinafter called a “quasi-carbide structure”).

The present invention will be explained by referring to the attached drawings.

FIG. 1 is a graph showing schematically the structure and chemical composition of the quasi-carbide,
FIG. 2 shows results of line-scanning profile by X-ray microanalysis of C and Cr contents in the cementite and their back scattered electron images,
FIG. 3 shows results of line scanning profile by X-ray microanalysis of C and Cr contents in the quasi-carbide and their back scattered electron images,
FIG. 4 is an electron-microscope photograph showing a bearing steel as annealed,
FIG. 5 is an electron-microscope photograph showing a quenched structure of a bearing steel (JIS G 4805 SUJ2),
FIG. 6 is an electron-microscope photograph showing the quasi-carbide structure of the bearing steel (JIS G 4805 SUJ2) as rapidly heated and quenched,
FIG. 7 is an electron-microscope photograph showing the quasi-carbide structure of Cr—Mo steel as rapidly heated and quenched,
FIGS. 8 and 9 are electron-microscope photographs of the quasi-carbide structures of the bearing steels SUJ3 and SUJ2 (JIS G 4805), respectively;
FIG. 10 shows the characteristic X-ray image of the cementite by an X-ray microanalyser,
FIG. 11 shows the characteristic X-ray image structure of the quasi-carbide by an X-ray microanalyser,
FIG. 12 shows the relation between the maximum heating temperature and the C and Cr contents in the quasi-carbide;
FIG. 13 explains the destruction process of the cementite in case of rapid heating;
FIG. 14 explains the dissolution process of the cementite in the austenite in case of an ordinary heating;
FIG. 15 explains the dissolution amount of the cementite into the austenite in case of an ordinary heating and rapid heating;
FIG. 16 explains the temperature-time curve for a rapid heating;
FIG. 17 shows the relation between the maximum heating temperature and the rolling-fatigue life of the bearing steel (JIS G 4805 SUJ2) rapidly heated and quenched;
FIG. 18 shows the relation between the maximum heating temperature and the rolling-fatigue life of the bearing steel (JIS G 4805 SUJ2) as quenched by high frequency rapid induction heating;
FIG. 19 shows the relation between the maximum heating temperature and the rolling-fatigue life of a carburized Cr—Mo steel as quenched by high frequency rapid induction heating;
FIG. 20 is a distribution curve of carbon contents of the carburized layer of Cr—Mo steel;
FIG. 21 shows the relation between the maximum heating temperature and the rolling-fatigue life of a carburized Ni—Cr—Mo steel as quenched by high frequency rapid induction heating;
FIG. 22 shows the relation between the average heating rate from room temperature and the rolling fatigue life;
FIG. 23 shows the relation between the pre-heating temperature in a quenching by rapid heating and the rolling-fatigue life as well as the amount of the quasi-carbide;
FIG. 24 shows the relation between the maximum heating temperature and the rolling-fatigue life of a carbonitrided Cr—Mo steel as quenched by high frequency rapid induction heating;
FIG. 25 shows curves of carbon and nitrogen concentrations in the carbonitrided layer of the Cr—Mo steel;
FIG. 26 shows the relation between the sliding distance and the wear amount in the wear tests made on the bearing steel (JIS G 4805 SUJ2) as ordinarily quenched and the same as quenched by high-frequency induction heating.

The term quasi-carbide structure used in the present invention has the following meanings:
The quasi-carbide structure is a structure in which quasi-carbide as defined hereinafter is chiefly dispersed in the martensite matrix and satisfies the following conditions:
1. Volume % of the quasi-carbide is 15 – 80%.
2. Size of the quasi-carbide (regarded as spherical) is about 0.2 – 10 μ in diameter.
The quasi-carbide is produced not by slowly dissolving the granular cementite above the A1 temperature into austenite, but rather by converting the granular cementite by one step into a new compound which is similar to but, in fact, different from the cementite.
The characteristics of the quasi-carbide are:
It is almost spherical conceptionally and is composed of a core portion and an outer shell portion. As shown by the curve 2 in FIG. 1 and the lineal analysis of C and Cr % by X-ray microanalyser in FIG. 3, the quasi-carbide has lower C and Cr contents than the C and Cr contents of the cementite indicated by the X-ray microanalysis in FIG. 1 and FIG. 2, and yet has an increasing C content toward the core portion and a decreasing C content toward the outer shell portion. This is in contrast to the uniform composition of the cementite proper. According to the measurements by the X-ray microanalyser, the C content in the core portion is about 2 - 5% which is not characteristic of a cementite nor a retained austenite produced by a conventional quenching method. Thus, this is considered to be a completely new structure.

The outer shell portion of the quasi-carbide is composed of retained austenite, but the carbon dissolved therein decreases continuously toward the matrix.

Alloying elements, such as Cr, Mn and Mo, which have strong affinity to carbon are also diffused, but not as much as carbon, and take a similar distribution pattern to that of carbon.

The carbon content in the boundary zone between the core portion and the outer shell portion is estimated to be around 1.7% in view of the fact that the outer shell portion is composed of retained austenite and the dissolution limit of carbon in austenite is about 1.7%.

The differences in the composition between the cementite proper and the quasi-carbide are as mentioned above. Thus, when observed by an electron microscope, the differences are clearly seen by comparing FIGS. 4 and 5 for the cementite and FIGS. 6, 7, 8 and 9 for the quasi-carbide.

As seen in FIGS. 4 and 5, the cementite is thus observed as a granular matter having a clear boundary with the martensite matrix. By contrast, as seen in FIGS. 6, 7, 8 and 9, the quasi-carbide has an obscure boundary with the martensite matrix, although the degree varies slightly depending on the heating conditions and the steel grades.

In the quasi-carbides shown in FIG. 6 and FIG. 8, the boundary with the matrix is obscure and a linear etching pattern produced in the grain is observed in the quasi-carbide formed from a large cementite. It is one of the characteristics of the quasi-carbide that a large number of the linear etching patterns are observed when a quasi-carbide is formed from a large cementite. FIG. 7 shows a microstructure of case-hardening steel which has been carburized and annealed and then converted into the quasi-carbide structure. The center portion having a relatively clear boundary like a cementite, corresponds to the core portion as mentioned before. In this case, the center portion is particularly hard to corrode because the concentration of Cr and Mo in the center portion is high and also because the discrepancy between the diffusion zone of C and the diffusion zone of Cr and Mo is remarkable.

FIG. 9 shows a typical quasi-carbide which was produced by heating the steel composition SUJ2 (JIS G 4805, similar to AISI 52100) to 1080°C at an average heating rate of about 400°C/sec.

Observations showing the above-mentioned characteristics of the quasi-carbide are shown in FIGS. 10a, b and FIGS. 11a, b. FIG. 10 is a characteristic X-ray image (photographed by an X-ray microanalyser) of the primary cementite in the steel C_2 of Table I, which was carburized and quenched, and FIG. 10a shows the image of CrKα, and FIG. 10c shows the image of CrKα. When FIGS. 10a, b and c are superimposed on each other, the Cr-rich region and the C-rich region coincide. This indicates that the cementite contains a large amount of Cr and C.

FIG. 11 shows the characteristic X-ray image of a quasi-carbide structure converted from the cementite shown in FIG. 10 (C_3 - Q_3, Table 2). When the image of CrKα of FIG. 11a and the image of CrKα of FIG. 11b are superimposed, it becomes clear that Cr is contained in a large amount in the quasi-carbide, but the C content has been remarkably reduced.

2. The quasi-carbide takes various modes depending upon heating conditions.

Referring to the carbon content (maximum) in the core portion of the quasi-carbide structure, only, the carbon content decreases as the heating temperature increases, as shown in FIG. 12, but to a much lesser degree as compared with the chromium content, for example. This indicates that the diffusion of chromium is remarkably slow as compared with that of C.

Together with the above structural changes, the microstructures viewed by an electron microscope also change from 3 to 7 in FIG. 13 as the heating temperature increases. In this case the effects of the average heating rate are that the change of the quasi-carbide shifts to the higher temperature side as the heating rate is increased; 3 to 7 of FIG. 13 show the quasi-carbide. FIG. 7 corresponds to 3 of FIG. 13, FIG. 6 corresponds to 3 of FIG. 13, FIG. 6 corresponds to 4 of FIG. 13, FIG. 8 corresponds to 4 and 5 of FIG. 13, and FIG. 9 corresponds to 5 of FIG. 13.

Actually, individual cementites in the same steel material vary from each other and it is not possible that all of the cementites take the same mode; thus, quasi-carbides coexist in a stepwise manner. For example, a structure composed mainly of the quasi-carbide of 3 and 4 together with a small amount of the carbides of 2 and 6 may be obtained. The forming mechanism of the quasi-carbide is explained in the following:

In respect of a steel containing more carbon than corresponds to the eutectoid, and when the heating temperature exceeds the Ac_3 transformation point and the matrix has been transformed into austenite, the cementite is dissolved into the austenite and the temperature further increases. In this case, if the heating rate is slow, the cementite becomes smaller in an equilibrium state maintaining the constant composition of (a) → (b) → (c) in FIG. 14. On the other hand, if the temperature is raised suddenly, the free energy of the cementite becomes so large that it is no longer possible to reduce the free energy suddenly by the equilibrium diffusion as shown in FIG. 14, and thus the cementite phase disappears at once as if concentration of carbon did not exist as the cementite phase, but only an irregular distribution of concentration existed in one phase as shown in FIG. 1.

Regarding the change in the cementite amount, two different forms are taken depending on the heating rate bordering on a certain heating rate — (dT/dt)c — as shown in FIG. 15.

The curve 1 in FIG. 15 is commonly seen in the conventional austenitizing treatment, and the curve 2 is seen in case of rapid high temperature heating.

T' represents a temperature at which the quasi-carbide starts to form, which temperature is higher than the Acem point in a hyper-eutectoid steel and higher than the As point in a hypo-eutectoid steel.
Figuratively speaking, the cementite explodes at a certain temperature in case of a rapid high temperature heating, and the wreckage of the explosion is regarded as the quasi-carbide. This explosive phenomenon is considered to cause the carbon distribution as shown in FIG. 1, which is different from that shown in FIG. 14.

The requirements of the quasi-carbide structure for improving the rolling fatigue-life will now be explained hereinafter.

No substantial improvement in the fatigue-life can be obtained if the amount of the quasi-carbide produced by the rapid heating is less than 15 volume % as is clear from the Examples set forth hereinafter, while if the quasi-carbide amount is above 80 volume %, the retained austenite becomes excessive so that the hardness required by rolling elements cannot be obtained. The range of the quasi-carbide is thus limited to 15 - 80 volume %. The most preferable range of the quasi-carbide is 20 to 50 volume % depending on the chemical composition, heating rate and heating temperature.

Regarding the size of the quasi-carbide, although a very fine size of the quasi-carbide can be effective for improving the fatigue-life, the size of the carbide of the material obtained by pretreatments, such as annealing or tempering, is the finest carbide precipitate, which has a diameter of about 0.1 μ. The quasi-carbide produced from this carbide precipitate will have a diameter of about 0.2 μ. Thus, the lower limit of the size of the quasi-carbide is defined to be about 0.2 μ in diameter.

As will be explained hereinafter, if the quasi-carbide grows too large, the improvement of the fatigue-life will be lost, but when a quasi-carbide of about 10 μ is formed by rapidly heating a carburized steel containing primary carbide of about 5 μ by carburization, improvement of the life is achieved, as shown in Examples set forth hereinafter. Thus, the upper limit of the size of the quasi-carbide is limited to about 10 μ. The most preferable range of the size of the quasi-carbide is about 0.4 to about 0.4 μ depending on the chemical composition, heating rate, heating temperature when the pretreatment or carburization is done properly.

The method for producing the steel for rolling elements according to the present invention will now be explained in detail.

The features of the method of the present invention reside in that a steel material, whose surface layer has one of the following basic compositions (1) or (2) at a temperature below 850°C and contains 5 - 40 volume % of granular carbide composed mainly of cementite, is heated to a temperature between about 1000° and 1250°C at an average heating rate above 10°C/sec. above 850°C, and then rapidly cooled to obtain the hardness required by the rolling elements.

The reasons for limitation of the above elements are as follows.

C is an important element for forming the quasi-carbide. With a carbon content below 0.65%, the amount of the quasi-carbide is too small to give satisfactory results. The lower limit of the carbon content is thus defined as 0.65%. With too high a carbon content, the carbide becomes too large, thus hindering the uniformity of the structure. The upper limit is thus set to 1.4%. The most preferred range of the carbon content is 0.8 - 1.2%.

Mn is added in order to improve the hardenability of the steel. Its lower limit is set as 0.04% which is regarded as an impurity in ordinary steel-making. If the manganese content is too high, it will negatively affect the general mechanical properties. Thus, the upper limit is set as 1.5%. The most preferred range of the manganese content is 0.25 to 1.2%.

Si is added to improve the heat resistance of the steel. The lower limit of Si is set as 0.04% which is regarded as an impurity in ordinary steel-making. The upper Si limit is set as 2.0%, which is considered as the addition limit. The most preferred range of the silicon content is 0.2 to 1.5%.

Cr is necessary for forming the carbide and at least 0.2% of Cr is necessary for forming the quasi-carbide. Too much chromium, however, will form carbides other than the cementite carbide, so that the characteristics of the carbide will not be uniform. Thus, the upper limit of chromium content is set as 2.5%. The most preferred range of the chromium content is 0.3 to 2.0%.

Mo is also effective as a carbide former and its lower limit of 0.06% and its upper limit of 0.6% are set for similar reasons as for Cr. The most preferred range of the molybdenum content is 0.08 to 0.6%.

In the present invention, chromium and molybdenum may be added singly or in combination. If added in combination, the lower limit of each of Cr and Mo is set as 0.06%, while the upper limit of Cr is set as 2.5% and the upper limit of Mo is set as 0.6%. The total of Cr and Mo should be more than 0.2% for obtaining the desired results. The most preferred range of Cr and Mo when used in combination is 0.3 to 2.0% for Cr and 0.08 to 0.6% for Mo and the total content is more than 0.4%.

Ni is effective for improving hardenability and mechanical properties of the steel. At least 0.25% of Ni is necessary for the above effects, but too high a nickel content increases the amount of retained austenite so that required surface hardness of the rolling elements cannot be obtained. The upper limit thus set as 5.0%. The most preferred range of the nickel content is 0.4 - 4.5%.

V is a carbide-forming element and is also effective for improving the impact resistance of the steel. Its lower limit of 0.03% and its upper limit of 0.2% are set for similar reasons as for Mo. The most preferred range of V is 0.10 - 0.15%.

W is effective in the same manner as Mo, and its lower limit of 0.10% and its upper limit of 0.8% are set for similar reasons as for Mo. The most preferred range of W is 0.3 to 0.6%.

| Composition (1) | C | 0.65 - 1.4% |
| Mn | 0.04 - 1.5% |
| Si | 0.04 - 2.0% |
| Cr | 0.20 - 2.5% |

| Balance Fe and unavoidable impurities, or |
| Composition (2) | C | 0.65 - 1.4% |
| Mn | 0.04 - 1.5% |
| Si | 0.04 - 2.0% |
| Mo | 0.06 - 0.6% |

| Balance Fe and unavoidable impurities |

The above basic compositions may contain for specific purposes:

| Ni | 0.25 - 5.0% |
| V | 0.03 - 0.2% |
| W | 0.10 - 0.8% |
| B | 0.001 - 0.01% |
B improves the hardenability of the steel and is added in an amount of 0.001 to 0.01%, which is commonly considered as an optimum addition range. The most preferred range of B in the present invention is 0.002 – 0.005%.

The heating conditions for obtaining the quasi-carbide structure according to the present invention will now be explained in detail.

1. As the starting material to be treated according to the method of the present invention, various (a) through-hardened steels, pretreated by treatments such as annealing, normalizing, quenching, tempering, thermal refining and isothermal transformation or (b) case-hardened steels subjected to carburization or carbonitriding down to below a depth required by rolling elements and further subjected to the above pretreatments may be used as long as the steels satisfy the condition that 5 to 40 volume % of granular carbide (mainly cementite) is present at a temperature below 850°C in the matrix composed of one or more of ferrite, austenite, martensite, etc. These starting materials are preheated to a temperature below 850°C by an ordinary heating or rapid heating, and are then heated to a prescribed temperature between about 1000°C and about 1250°C at an average heating rate of more than 10°C/sec. above 850°C, whereafter the steels are rapidly cooled to obtain the hardness required by rolling elements. This takes a temperature-time curve as shown by A in FIG. 16. In this case, the prescribed temperature approaches closer to 1000°C as the average heating rate approaches 10°C/sec., and approaches closer to 1250°C as the average heating rate increases beyond 10°C/sec.

$$\text{The average heating rate} = \frac{T_1 - 850°C}{t_2 - t_1}$$

wherein

- $T_1$: maximum heating temperature
- $t_2$: starting time of cooling
- $t_1$: time at which 850°C is reached

2. The quasi-carbide structure may also be obtained by rapidly heating the starting material as defined before from room temperature to a prescribed temperature between about 1000°C and 1250°C at an average heating rate of 25°C/sec. and then rapidly cooling the material. In this case, the temperature time curve is as shown at B in FIG. 16.

In this case,

$$\text{the average heating rate} = \frac{\text{maximum heating temperature} - \text{room temperature}}{\text{starting time of cooling} - \text{starting time of heating}}$$

In both of the methods (1) and (2), heat treatments such as tempering are affected after the quenching according to the specific use to which the steel is to be put. The upper limit of the average heating rate may be increased up to the practical upper limit of a rapid heating device.

The reasons for the upper limit of 850°C for the preheating and the average heating rate of more than 10°C/sec. above 850°C or the average heating rate of more than 25°C/sec. from the room temperature are that almost no quasi-carbide is formed outside the above range as understood from the Examples set forth hereinafter.

In the most preferred embodiment of the present invention, a steel material containing 10 to 30% of granular carbide composed mainly of cementite is used. Regarding the heating conditions, it is preferable that the average heating rate above 850°C is more than 20°C/sec. and the steel is heated to a temperature between about 1050°C and 1200°C.

Under the above-mentioned heating conditions, the surface layer of the through-hardened steel and case-hardened steel will be imparted with the quasi-carbide structure and will then exhibit excellent rolling fatigue characteristics.

By quenching from the heating temperature between 1000° and 1250°C, most of the structure is transformed into the quasi-carbide structure. When a carbide former such as Cr and Mo is contained in the steel in an amount close to its upper limit, the cementite predominates in the annealed stage, but a small amount of special carbides mingles therewith and these special carbides coexist in a very small amount with the quasi-carbide.

For further improvement of the rolling-fatigue strength the following considerations may be taken into account.

1. When a carbide former such as Cr, Mo, V and W or Si is present in a relatively large amount, the heating temperature should be somewhat higher or the heating rate should be lower within the range of the present invention.

2. The heating rate should be somewhat higher and the heating temperature should be slightly lower within the range of the present invention, when the carbon content is relatively low and if other elements are maintained on the same levels.

3. When the nickel content is relatively high, the retained austenite increases on the higher temperature side and thus the hardness is decreased. Further, the martensite then tends to grow in slightly excessive manner. Therefore, the heating temperature should be limited to the lower side or the heating rate should be higher.

However, when a carbide former such as Cr, Mo, V and W is contained in a large amount, the above adjustments are not necessary.

Explanations are now given for the reasons why the life of the rolling elements obtained by the present invention is improved.

It is assumed that the following factors either singly or in combination are responsible for the increased life of the quasi-carbide structure (martensite + quasi-carbide, including retained austenite, + small amount of retained austenite).

1. The quasi-carbide in the present invention is a granule of about 0.2 to 10 μ diameter, and its core contains a phase of the composition of (Fe, M)_{x-11}C, which phase is surrounded by the retained austenite. This produces the quasi-carbide structure in which the quasi-carbide granules are dispersed in great number in the martensite matrix. This means that the retained austenite, which forms the outer shell of the quasi-carbide, is dispersed very uniformly and finely in the hard martensite matrix, thus preventing a strain increase of the martensite due to rolling contact-stress and the progress of fatigue cracking.

2. The cementite has a clear-cut boundary with the martensite matrix. It is believed by some authorities
that the boundary between the cementite and the martensite is weak and tends to cause cracking and promotes its development.

When the cementite is, however, transformed into the quasi-carbide, the boundary between the quasi-carbide and the martensite is composed of austenite and martensite, so that the boundary is not weak as contrasted to the cementite.

3. When the quasi-carbide is present, the martensite does not grow excessively large. Therefore, it is possible to heat the steel to a high temperature without excessively enlarging the martensite and to increase the amount of elements dissolved in the martensite.

By the combined effects of the above factors, the rolling-fatigue life is remarkably improved.

Explanations will now be given on the differences between the present invention and the conventional induction hardening. In the induction hardening, the average heating rate may be, in special cases, 25°C/sec. in order to thin the hardening depth or due to the shape of the work pieces, but the heating temperature is determined mainly in view of the hardness, microstructure, tensile strength, repeated bending, fatigue strength. For example, the Japanese journal "Heat Treatment", Vol.5, No.2, pages 126 to 136, teaches a temperature range below 900°C as an optimum heating temperature (maximum heating temperature) in an ordinary induction heating. Further, "Induction Heating and Fatigue Strength" published by Nikkan Kogyo Shinbun (1963), page 170, teaches an optimum heating temperature range (or heating time) based on the relation between the maximum heating temperature and strength. Both publications indicate an optimum heating temperature range of 750°C - 1000°C for steels containing 0.65 - 1.4% of carbon. This is completely different from the present invention in which the quasi-carbide is formed efficiently by the complex combination of the chemical composition, the structure, the heating temperature and the heating rate. The steel products obtained by the present invention is also completely different from those obtained by the conventional induction heating.

Further, the steel of quasi-carbide structure obtained by the present invention shows satisfactory properties (for example, bending strength, etc.), in addition to the improved rolling-fatigue life.

Steels for rolling elements are required to have good wear resistance as well as good rolling-fatigue life. The steels obtained by the present invention show also satisfactory wear resistance, as shown in FIG. 26.

As explained above, the several properties of the steel obtained by the present invention all satisfy the requirements for rolling elements.

Examples of the present invention will be set forth hereinafter.

Table 1 shows chemical compositions of the raw steel materials to be used in the present invention.

In Table 1, J₁ - J₆ and C₂ are high-carbon steels, which are pretreated so as to form granular carbide structures and are rapidly heated. C₁ - C₂ and N₁ - N₄ are low-carbon steels which are carburized or carbonitized to increase the carbon content in the surface layer to 0.65 - 1.40%, whereafter the steels are annealed to precipitate granular carbide and are rapidly heated.

All of the steels listed in Table 1 have been pretreated so as to contain the amount of granular carbide required in the present invention.

Table 2 shows the steel products obtained by the present invention.

The cementite of the material J₁ as spheroidized is shown in FIG. 4, and the cementite of the material J₁ as ordinarily heated and quenched is shown in FIG. 5, which is the same as the structure of a conventional steel.

### Table 1

<table>
<thead>
<tr>
<th>Designation</th>
<th>C</th>
<th>Si</th>
<th>Mn</th>
<th>Ni</th>
<th>Cr</th>
<th>Mo</th>
<th>V</th>
<th>W</th>
<th>B</th>
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<td>—</td>
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<td>N₁</td>
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### Table 2

<table>
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<tr>
<th>Designation</th>
<th>Raw material</th>
<th>Average Heating rate(°C/sec.)</th>
<th>Maximum heating temperature(°C)</th>
<th>Quasi-Carbide</th>
<th>Corresponding Figs.</th>
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<tr>
<td>J₁ - O₁</td>
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<td>300 - 500</td>
<td>1000 - 1250</td>
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<td>100 - 200*</td>
<td>1000 - 1200</td>
<td>1.3</td>
<td>25 - 45</td>
</tr>
<tr>
<td>J₃ - O₃</td>
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<td>10 - 20</td>
<td>1000 - 1100</td>
<td>1.3</td>
<td>25 - 45</td>
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<tr>
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<td>1000 - 1180</td>
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<td>&quot;</td>
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<td>0.5 - 10</td>
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<tr>
<td>C₅ - O₅</td>
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<td>10 - 20</td>
<td>1000 - 1100</td>
<td>0.2 - 2</td>
<td>20 - 40</td>
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<tr>
<td>C₆ - O₆</td>
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<td>40</td>
<td>1050</td>
<td>1 - 3</td>
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In the following Examples, all of the comparative steels were heated to a quenching temperature between 800° to 850°C, at an average rate of less than 7°C/sec., from the room temperature and then quenched. The structure of the inventive steel J1 – Q1 obtained by quenching the material J1 by a rapid direct heating by electric current (direct heating) contains the quasi-carbide structure according to the present invention, as shown in FIG. 6. The structure of the inventive steel C1 – Q3 obtained by carburizing the material C1 gives carbide of 1.05% to the surface and quenching the carburized surface by a high frequency rapid induction heating results in the quasi-carbide structure of the present invention.

When the materials J2 and J3 are quenched by a high frequency rapid induction heating to obtain the steels J2 – Q and J3 – Q, the quasi-carbide structures as shown in FIG. 8 and FIG. 9 are obtained.

When the material C3 is carburized to give a surface carbon content of 1.36% and quenched by an ordinary heating, the chemical composition of the primary cementite shows almost a constant composition as shown in FIG. 2 and FIG. 10, but when the carburized surface is subjected to the high frequency rapid induction heating and quenching according to the present invention to obtain the steels C5 – Q1, the primary cementite in the surface disappears and the composition gradient peculiar to the quasi-carbide appears, as shown in FIG. 3.

The relation between the carbon and chromium contents of the quasi-carbide in the steel C5 – Q1 and the maximum heating temperature is shown in FIG. 12, from which it is clear that the quasi-carbide has a chemical composition clearly different from that of the cementite in the temperature range of about 1000° to 1250°C. The relation between the maximum heating temperature and the rolling-fatigue life was sought for in connection with the steel J4 – Q2 which was subjected to a direct rapid heating and quenching from room temperature at a heating rate above 25°C/sec. The results are shown in FIG. 17. The rolling-fatigue test was done, using a needle roller of 3 mm diameter with maximum contact pressure of 420 kg/mm², and a repeated stress rate of 8 × 10^5/min. and using 120 spindle oil lubricant.

When the relative life is expressed by the average life ratio based on the life obtained when the same materials are subjected to an ordinary heating and quenching in an electric furnace, the life will be improved several times, as shown in FIG. 17, in the range of about 1000° to 1250°C, but no quasi-carbide structure is obtained below about 1000°C and above about 1250°C and thus no improvement of the life is obtained.

FIG. 18 shows the results obtained from the rolling-fatigue test made on the steel J1 – Q2 obtained by quenching the material J1 at a heating rate of more than 25°C/sec. from room temperature.

The rolling-fatigue life test in this case was done by using a thrust-type life tester, with a maximum contact pressure of 500 kg/mm² and a stress repeating rate of 1500/min. and using 60 spindle oil lubricant. Regarding the relation between the average life ratio to the material quenched by ordinary heating and the maximum heating temperature, the life is improved under the heating condition of 1000° – 1170°C due to the quasi-carbide structure, as shown in FIG. 18. In case of induction hardening, compressive residual stress is retained in the surface layer and the rolling fatigue life is increased. But actual measurements of the residual stress after the rolling-fatigue test at the position of the maximum shear stress under the rolling track shows that the improvement of life is based on the quasi-carbide structure, since it has almost no relation with the maximum heating temperature, as shown in FIG. 18.

The material C1 was carburized, spheroidized and quenched at a heating rate of more than 25°C/sec. from room temperature by the rapid induction heating and quenching to obtain the steel C5 – Q1.

The relation between the maximum heating temperature and the rolling-fatigue life obtained by the thrust-type tester under the same testing conditions as mentioned before, were sought for and the results are shown in FIG. 19. Also in this case, the life is improved under the heating condition of 1000° – 1230°C in which the quasi-carbide structure is obtained. The carbon content in the carburized layer is about 1.2% in the surface, as shown in FIG. 20.

The rolling-fatigue test was made on the steel N2 – Q under the same test conditions as before. The steel N2 – Q was obtained by carburizing the material N2 to give a carbon content of 0.05% at the surface, spheroidizing the steel, and quenching it from the room temperature at a heating rate above 25°C/sec. by high frequency rapid induction heating. The results are shown in FIG. 21. In this case, too, the life is improved under the heating condition of 1000° – 1180°C in which the quasi-carbide structure is obtained.

Test pieces were prepared from the steels J1 – Q1, J1 – Q2, J3 – Q, C4 – Q, C5 – Q, C6 – Q, N1 – Q and N2 – Q for both needle roller type life testing and thrust-type life testing, and life tests were conducted under the same testing conditions as mentioned before. The above steels were obtained by quenching the materials J1, J2, J3, C4 and the carburized materials C5, C6, N and N2 with various different average heating rates from room temperature by rapid direct heating, high frequency rapid induction heating and salt-bath heating. The results are shown in FIG. 22. At the average heating rates above 25°C/sec. the quasi-carbide structure is obtained and an increased rolling-fatigue life is observed.

The material J1 and the material C5 having a carburized surface were preheated to various temperatures in an electric furnace, held at the temperature for 30 minutes and subjected to the rapid induction heating...
and quenching at an average heating rate of 10° - 20°C/sec. up to the maximum heating temperature of about 1100°C, to obtain the steels J₁ - Q₁ and C₂ - Q. Rolling-fatigue tests were made on these steels under the same testing conditions as mentioned before, by a thrust-type tester, and the results are shown in FIG. 23.

As clearly understood from FIG. 23, the quasi-carbide structure is obtained when the preheating temperature is below 850°C and improvement of the life is observed; but if the preheating temperature is above 850°C, an over-heated structure predominates and no substantial improvement of the life is obtained even if the quasi-carbide is retained locally in a small amount.

The material C₁ was carbonitrided, spheroidized and quenched at a heating rate of above 25°C/sec. from room temperature by a high frequency rapid induction heating to obtain the steel C₁ - Q₂. The relation between the maximum heating temperature and the rolling-fatigue life was determined by the thrust-type life tester under the same testing conditions as mentioned before. The results are shown in FIG. 24.

As shown in FIG. 25, the carbon and nitrogen contents in the carbonitrided layer of steel C₁ - Q₂ are tempered hardness range of HRC 60 - 63, and the present inventive steel shows similar or better wear resistance than the conventional steel.

Rolling-fatigue tests were conducted on the steels J₁ - Q₂, J₂ - Q₁, C₁ - Q₁ and C₂ - Q under the same test conditions as mentioned before. The steels J₁ - Q₂, J₂ - Q₁, C₁ - Q₁ and C₂ - Q were obtained, respectively, from the material J₁, J₂, C₁ and C₂. The material J₁ has the same composition as J₂, but is obtained by vacuum remelting and electro-slag remelting. The material C₁ was obtained by surface carburization for obtaining a carbon content between 1.1 and 1.2% and then spheroidization, and the material C₂ was obtained by vacuum remelting steel of the same composition as the material C₁. The results are shown in Table 3.

A similar level of life improvement is obtained by the present inventive method in spite of the differences of the steelmaking process.

Thus, the steel J₂ - Q obtained by the present invention shows a life which is 36 times longer than the ordinary quenched bearing steels (similar to AISI 52100) as air melted, and the steel C₂ - Q shows a life which is 30 times longer than the same.

<table>
<thead>
<tr>
<th>Steel Making Method</th>
<th>Through Hardened Steel (High Carbon Steel)</th>
<th>Case Hardened Steel (Low Carbon Steel)</th>
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<tr>
<td>Ordinary Heating and Quenching</td>
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<td>Vacuum Degassing + Electroslag Melting</td>
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<td>Designation</td>
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<td>J₂ - Q</td>
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<td>Average Life Ratio to Vacuum Degasting and Ordinary Heating and Quenching Material</td>
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<td>Rapid Heating and Quenching (Present Invention)</td>
<td>Average Life Ratio to Air Melt Bearing Steel and Ordinary Heating and Quenching Material</td>
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</table>

What is claimed is:

1. Steel suitable for use in rolling elements and exhibiting improved rolling-fatigue strength, the surface layer of said steel consisting essentially of the following chemical composition;
   - C, 0.65 - 1.4%; Mn, 0.04 - 1.5%; Si, 0.04 - 2.0% and one member selected from the group consisting of 0.06 - 0.6% Mo; 0.30 - 2.5% Cr; and a combination of 0.06 - 2.5% Cr and 0.06 - 0.6% Mo wherein Cr + Mo ≥ 0.2%; with the balance being iron and unavoidable impurities, said surface layer having a martensite matrix and having dispersed within said matrix 15 - 80 volume % of quasi-carbide particles composed of a core portion and an outer shell portion, said core portion having a carbon content

about 1.0% and about 0.6%, respectively. In case of carbonitridation also the quasi-carbide structure is obtained and the life is improved under the inventive heating condition of 1000°C to 1180°C.

Wear tests were made on an ordinary quenched material obtained by quenching the material J₁ under ordinary heating in an electric furnace and the inven-
tive steel J₁ - Q₂ quenched by high frequency rapid induction heating at a heating rate above 25°C/sec. from room temperature, and the relation between the wear amount and the sliding distance was determined. The results are shown in FIG. 26.

It is understood that there is no substantial difference in wear amount between the ordinary quenching and the present inventive method within the quenched layer.
of about 2 – 5%, said outer portion being composed of retained austenite having a carbon content which decreases continuously toward the outer shell portion and which further possesses lower carbon and chrome contents than cementite said quasi-carbide being in the form of particles of about 0.2 – 10μ in diameter.

2. A method for producing steel for rolling elements which comprises heating a steel material to a predetermined temperature between about 1000°C and 1250°C at an average heating rate of more than 10°C/sec. above 850°C and quenching the thus heated steel material to obtain a steel wherein the surface layer has a martensite structure with 15 to 80 volume % of quasi-carbide particles distributed through the martensite structure, said particles being in the size range from about 0.2 to 10 microns and having a core portion and an outer shell portion, said core portion having a carbon content of about 2 to 5% and said outer portion being composed of retained austenite and having a carbon content which decreases continuously toward the outer shell portion which further possesses lower carbon and chrome contents than cementite, the surface layer of said steel starting material consisting essentially of the following chemical composition:

C, 0.65 – 1.4%; Mn, 0.04 – 1.5%; Si, 0.04 – 2.0%; one member selected from the group consisting of 0.06 – 0.6% Mo; 0.30 – 2.5% Cr, and a combination of 0.06 – 2.5% Cr and 0.06 – 0.6% Mo wherein Cr + Mo ≥ 0.2%, the balance being iron and unavoidable impurities, said surface layer containing 5 – 40 volume % of granular carbide composed mainly of cementite at a temperature below 850°C.

3. A method for producing steel for rolling elements which comprises heating a steel material to a predetermined temperature between about 1000°C and 1250°C at an average heating rate of more than 10°C/sec. above 850°C and quenching the thus heated steel material to obtain a steel wherein the surface layer has a martensite structure with 15 to 80 volume % of quasi-carbide particles distributed through the martensite structure, said particles being in the size range from about 0.2 to 10 microns and having a core portion and an outer shell portion, said core portion having a carbon content of about 2 to 5% and said outer portion being composed of retained austenite and having a carbon content which decreases continuously toward the outer shell portion which further possesses lower carbon and chrome contents than cementite, the surface layer of said steel starting material consisting essentially of the following chemical composition:

C, 0.65 – 1.4%; Mn, 0.04 – 1.5%; Si, 0.04 – 2.0%, one member selected from the group consisting of 0.06 – 0.6% Mo; 0.30 – 2.5% Cr, and a combination of 0.06 – 2.5% Cr and 0.06 – 0.6% Mo wherein Cr + Mo ≥ 0.2%, the balance being iron and unavoidable impurities,
said surface layer containing 5 – 40 volume % of granular carbide composed mainly of cementite at a temperature below 850°C.

4. The method of claim 3 wherein the average heating rate is more than 20°C/sec.

5. Method according to claim 2 in which the steel material is rapidly heated from room temperature to a temperature between about 100°C and 1250°C with an average heating rate of more than 25°C/sec.
UNITED STATES PATENT OFFICE
CERTIFICATE OF CORRECTION

Patent No. 3,929,523 Dated December 30, 1975
Inventor(s) Masao Kinoshii et al

It is certified that error appears in the above-identified patent and that said Letters Patent are hereby corrected as shown below:

In the heading of the Patent, [73] should read as follows:

--[73] Assignee: Nippon Seiko Kabushiki Kaisha, Tokyo, Japan--.

Signed and Sealed this Sixth Day of July 1976

RUTH C. MASON
Attest:
Attesting Officer

C. MARSHALL DANN
Commissioner of Patents and Trademarks