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Weight Reduction of Automotive Parts by Use of High-strength Steel Bars and Rods

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Abstract

Needs for the fuel economy improvement and the weight reduction of automobiles are ever increasing to enable us to cope with global environmental problems and especially to reduce CO_2 emissions that are a main factor in global warming. It is urgent to address and solve these problems. As far as the weight reduction of automobiles is concerned, aluminum and other new materials are highlighted. The responsibility of ferrous materials is higher than ever in that they can provide a variety of properties by a combination of chemistry, heat treatment and fabrication technology, and they can be stably supplied at low cost. The efforts made by Nippon Steel to increase the strength of special steel bars and rods that support functional automotive parts are described.

1. Preface

Triggered by global environmental issue, the reduction of body weight and improvement of fuel consumption of cars became important tasks to tackle, and the development of special steel bar and wire rod products for car applications which can bring about stronger and lighter car components is being required.

This paper describes Nippon Steel's special steel bar and wire rod products responding to the demands for materials for stronger car components such as high strength steels for gears of power trains, high strength shafts and CVJ's (constant velocity joints), high strength as rolled steels for engine and undercarriage parts, high strength bolts, and bearings, etc.

2. Steels for High Strength Gears

Fatigue strength and wear resistance are required of gears, which are important mechanical parts for cars and industrial machines, and in this sense carburized casehardened steels are generally used for gears [generally]. These materials are specified under JIS (Japan Industrial Standard) as Low Alloy Steels for Machine Structural Use SCr 420, SCM 420, etc. Demands for high strength gears have become stronger recently as requirements for fuel economy and weight reduction of cars grew more than ever. Fatigue strength of carburized gears is considered in two aspects: dedendum strength (bending fatigue strength of gear tooth) and tooth surface strength (pitting strength).

2.1 Philosophy of enhancing fatigue and pitting strengths of carburized steel material

Photo 1 shows a microscopic fractography of carburized conventional JIS SCM 420 material after a rotary bending test by a Ono tester. Fatigue cracks start from oxidized grain boundary layer formed during the heat treatment of the carburizing process and propagate as grain boundary fracture, leading to the final failure of the test piece. **Fig.1** shows an observation of fatigue crack development by oxidation method, interrupting the fatigue test of a carburized test piece after a predetermined lapse of time[1]. It is observed that fatigue cracks begin to occur at an early stage in relation to the fatigue life of the entire test piece and that the number of bends till the occurrence of fatigue cracks is smaller as the load stress becomes larger. From these results, it is understood to be effective for enhancing fatigue strength of carburized steel material to control formation of the oxidized grain boundary layer where fatigue cracks start and also decrease effective load stress by giving residual stress and/or increasing surface hardness.

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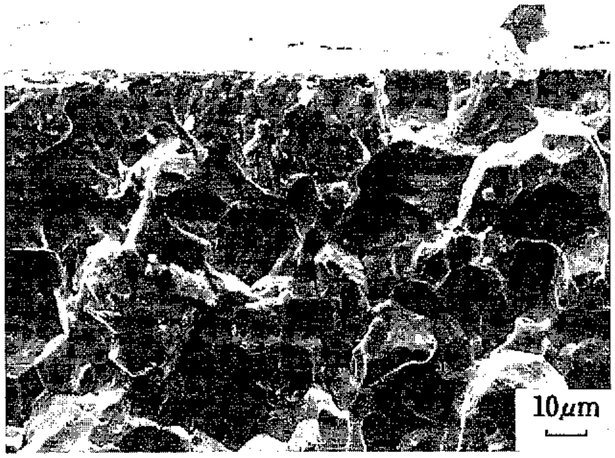


Photo 1 Fracture of a rotary bending test piece tested by an Ono tester

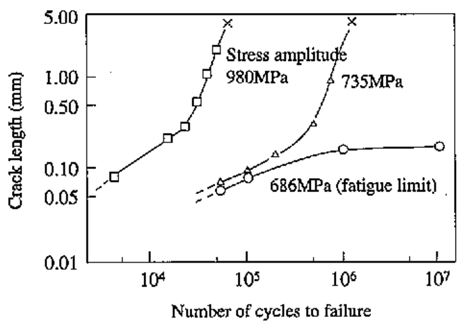


Fig.1 Initiation and propagation behavior of fatigue crack under bend fatigue progress[1] (SCM 420, contact stress: 2,942MPa)

In Fig. 1 showing the case of stress amplitude of 686 MPa, which is the fatigue limit, it is observed that 99% of the entire fatigue life is accounted for by propagation and arrest of fatigue cracks, and it is found that arrest behavior of fatigue cracks strongly affects the fatigue strength. From the viewpoint of fracture mechanics the fatigue crack arrest condition is expressed as follows[2]:

$\Delta K_{eff} \leqq \Delta K_{eff\text{-}th}$ (= const.)

where,

$\Delta K_{eff} = \sigma_a (\pi a)^{1/2} + K_{res} + K_{clos}$

where, σ_a: fatigue stress amplitude,

a: crack length,

K_{res}: stress intensity factor for residual stress

K_{clos}: stress intensity factor for closure of crack end

That is to say, because threshold stress intensity factor $\Delta K_{eff\text{-}th}$ of 0.8 % C steel (amount of C in carburized layer) is nearly constant, it is necessary for increasing fatigue strength to control effective stress intensity factor ΔK_{eff} at the crack end not to exceed $\Delta K_{eff\text{-}th}$. It is deduced that for this purpose it is effective to reduce oxidized grain boundary layer, which represents the initial crack length, and cancel effective tensile force at the crack ends by giving compressive residual stress for helping the closure of crack ends.

As for the pitting fatigue behavior, **Fig.2**[3] shows an observation of fatigue crack development in a sectional plane of a test piece of carbuzised conventional JIS SCM 420 steel after a fatigue test by a twin cylinder type roller pitting tester for a pre-determined time. It was observed that fatigue cracks started from oxidized grain boundary layer on the surface as early as within less than 1/1,000 of the total fatigue life. Their propagation speed was found to be slow in

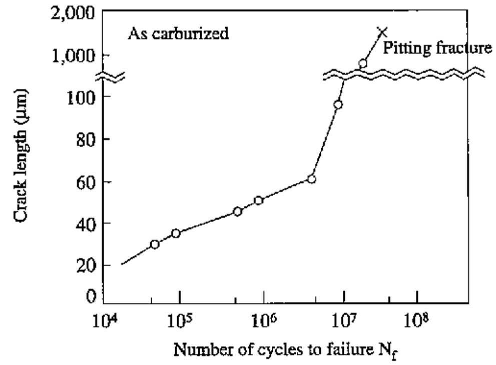


Fig.2 Initiation and propagation behavior of fatigue crack under pitting fatigue progress[3]

the initial propagation stage but it was greatly accelerated in the later stage towards the end of the pitting fracture life. From these observations it is considered effective for enhancing pitting strength to control formation of the oxidized grain boundary layer for decreasing occurrence of fatigue cracks and raise surface hardness.

2.2 Steels for high strength gears

An example of chemical composition of high strength gear steel developed on the basis of the above fundamental studies is shown in **Table 1**[4]. Si, which helps formation of the oxidized grain boundary layer, is decreased and hardenability is maintained by Mo and V. The V also refines grain size. For controlling unisotropy of fatigue behavior caused by long MnS, which often forms, the shape of MnS is controlled by addition of Ca to form calcium aluminate core and have it covered with (Ca, Mn)S restraining at the same time the formation of hard Al_2O_3.

Tooth bending fatigue strength tests and roller pitting tests were done on the above developed steel and conventional JIS SCM 420 on two test pieces each, one carburized at 930°C for 5 hours and the other carburized likewise and then shot peened. As seen in **Figs.3 and 4**, the developed steel showed higher values in tooth bending fatigue strength and pitting strength than the conventional steel in

Table 1 Example of chemical composition of high strength gear steel[4]

(wf%)

C	Si	Mn	S	Cr	Mo	V	Ca	O
0.20	Reduced	0.60	0.015	0.50	0.75	0.10	Added	Reduced

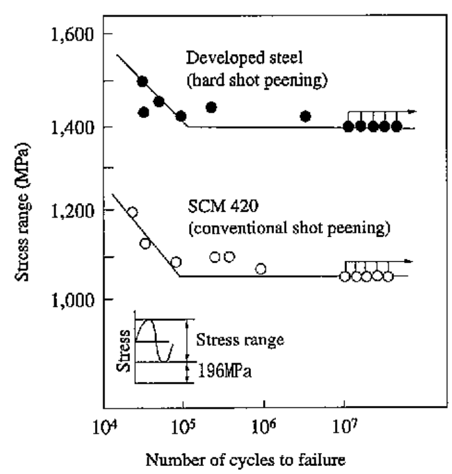


Fig.3 Gear tooth bending fatigue strength of new high-strength gear steel[4]

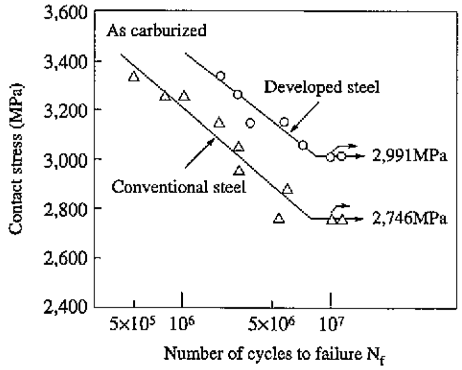


Fig.4 Pitting strength properties of high strength gear steel[4)]

either of the test pieces[4)].

As described above, it is important for enhancing the performance of carburized parts in both tooth bending fatigue strength and pitting strength to secure optimum conditions of the steel materials and gear manufacturing. That is, on the steel material side it is important to control grain boundary oxidation during the carburizing treatment to the minimum, design the alloying chemistry so that deterioration of surface hardness is minimized even with small amount of inevitable grain boundary oxidation and raise the surface hardness especially for pitting strength. On the gear manufacturing side it is extremely effective to give an appropriate amount of compressive residual stress to the surface and enhance smoothness of tooth face contact.

3. Steels for High Strength Shafts

As the shape of various shafts is suitable for induction hardening by induction heating, their surface is hardened in most cases by that method. Medium carbon steels such as Carbon Steels for Machine Structural Use S 43 C, Marganese Steels for Machine Structure Use SMn 443, etc. under JIS are used for shafts. Because the main function of shafts is transmission of torque, the principal characteristic required of the materials is torsion strength (static torsion strength and torsion fatigue strength) and demands for higher strength shafts are becoming stronger and stronger, too.

3.1 Decisive factors of torsion strength of induction-hardened mechanical parts

Three aspects are considered about hardness of induction-hardened parts, i.e. hardness of the hardened layer, depth of the hardened layer and hardness of the core. Static torsion strength depends on all of these three. As shown in **Fig.5** induction-hardened parts normally fractures in a plane perpendicular to the longitudinal axis (Mode III fracture) presenting shear fracture in most of the fracture surface. It has been known that static torsion strength in the Mode III fracture can be approximated by equivalent hardness Heq defined as "mean hardness weighted by radius squared" as described below[5)].

$$H_{eq} = \frac{3}{a^3}\int_0^a H(r)\,r^2\,dr$$

where, H(r) is hardness, a is radius, r is distance from thee center.

Fig.6 shows relation between the equivalent hardness and static torsion strength of steel materials having various hardness distributions. When Mode III fracture occurs the equivalent hardness and static torsion strength are in good correlation. It has to be noted, however, that when the equivalent hardness is increased to 700 or

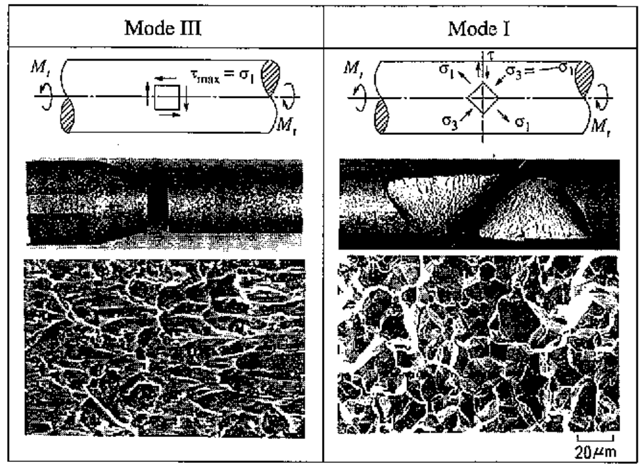


Fig.5 Static Torsion Fractures[5)]

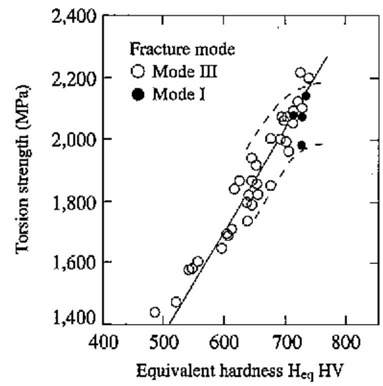


Fig.6 Relation between equivalent hardness H_{eq} and torsion strength[5)]

more Mode I fracture (fracture in a plane 45° to the longitudinal axis) begins to occur and static torsion strength decreases instead.

3.2 Philosophy of hardness enhancement of induction-hardened parts

A typical example of Mode I fracture is shown in the right hand half of Fig.5. The fracture initiates at the surface from a grain boundary cracking. A typical brittle fracture of a martensite structure tempered at low temperature is seen on the fracture surface of crack propagation area of the hardened portion. Because Mode I fracture is brittle fracture the torsion strength of materials showing this type of fracture is determined by its critical strength of brittle fracture. In other words, as the fracture is triggered by a grain boundary crack the torsion strength is determined by grain boundary strength.

Fig.7 shows a result of studies on the effects of phosphorus reduction and boron addition on Mode I fracture strength[6)]. All the Mode I fractures are brittle fractures initiated by grain boundary crack. Mode I fracture strength decreases as the hardness of the hardened layer increases, but when the hardness is the same it increases with addition of boron. The effect of boron addition is equal to or larger than that of a phosphorus reduction from 0.013% to 0.002%. **Fig.8** shows an analysis result of grain boundary characteristics by Auger Electron Spectroscopy (AES). As boron precipitates in grain boundary, grain boundary precipitation of phosphorus becomes less than that of the steel without boron. This fact indicates that grain boundary cleaning effect of boron increases grain boundary strength and

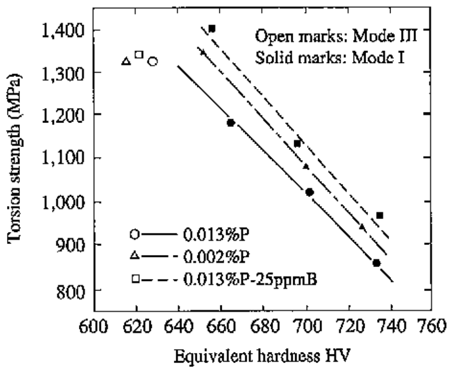


Fig.7 Effects of phosphorus reduction and boron addition on torsion strength[6)]

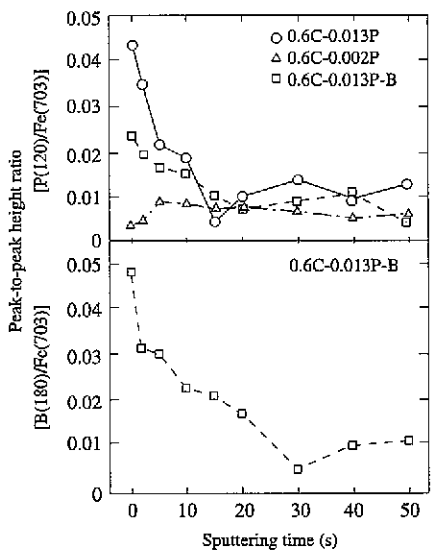


Fig.8 Grain Boundary Segregation Behavior of phosphorus and boron[6)]

enhances Mode I fracture strength. The reason why the effect of boron addition is greater than phosphorus reduction is that boron itself strengthens inter-granular bond.

Behavior of torsion fatigue fracture is also markedly influenced by hardness distribution. In the case of materials having a thick hardened layer or under high stress the torsion fatigue fracture starts from the surface, while if the hardened layer is thin or the stress is low it starts from inside. Under the surface-starting conditions the torsion fatigue life depends on the equivalent hardness and grain boundary strength[7)].

As described above strengthening of shaft materials can be achieved by raising grain boundary strength and increasing hardness factors in an appropriate manner.

4. High Strength Non-Tempered Steel

Principal properties required of automobile engine and undercarriage parts are strength and toughness and so-called high toughness steels such as Alloy Steels for Machine Structural Use SCM 440 by JIS are used for these parts. Most of them are manufactured by hot forging and, for the purpose of giving desired strength and toughness, quench-and-temper treatment is necessary after hot forging. For reduction of manufacturing costs and energy consumption, however, steels not requiring the quench-and-temper treatment are being looked for. In response to this demand various kinds of so-called hot forging non-tempered steels have been developed. Use of the non-tempered steels can eliminate cracks and strain caused by the quenching process besides eliminating the costs of the process.

4.1 Philosophy of toughness enhancement of hot forged products

It is comparatively easy to obtain strength equivalent to the quenched-and-tempered products skipping the heat treatment process. In the case of hot forged non-tempered steels the mechanism of precipitation hardening by adding vanajium is commonly used for the purpose. Products as hot forged, however, have coarse grain structure and thus it is very difficult to secure good toughness. For this reason, the main technical challenge regarding the hot forged non-tempered steels is how to refine the as-hot-forged grain structure to enhance toughness.

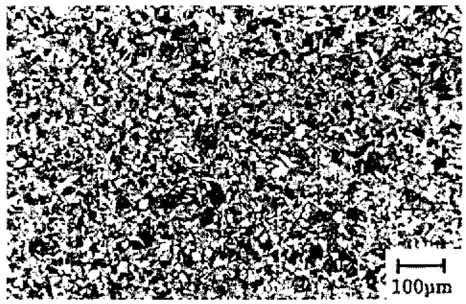
As the hot forging temperature is usually as high as 1,250°C, grain refining by thermo-mechanical treatment such as controlled rolling is impracticable. Control of carbides/nitrides by addition of micro-alloy elements has been the normal practice for refining austenite grain size. For controlling γ grains at the temperature of 1,250°C for hot forging, use of TiN is most effective and this method has been commercially used in TiN type non-tempered steels[8)]. Even in this case, however, the size of γ grains is about 40μm, not fine enough to obtain toughness comparable to the quenched-and-tempered products.

As a countermeasure, a new method is recently attracting attention as a third generation grain refining technology whereby, in case the prior austenite grains cannot be made finer than a certain limit, the effectual structure is virtually refined through formation of intra-granular ferrite in quantity. Intra-granular ferrite is a form of ferrite texture which is nucleated and grow inside the prior austenite grains. It was first discovered as acicular ferrite texture having TiO series oxides as transformation nucleus inside crystal grains in the heat affected zone of welded plates, and became established as a means for improving toughness[9)]. Amongst various developments of its use thereafter as "oxide metallurgy" in various fields, a new granular shape intra-granular ferrite having MnS-VN as transformation nucleus was discovered and was commercially applied as a measure for enhancing toughness of the hot forged non-tempered steels[10)].

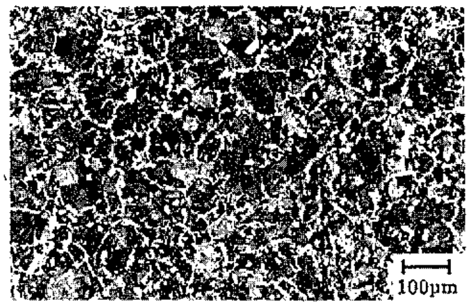
4.2 High-strength hot forged non-tempered steels having excellent toughness

As-hot-forged microstructures of intra-granular ferrite-type and conventional TiN-type non-tempered steels are shown in **Fig.9** and their impact toughness in **Fig.10**. The grain structure of the intra-granular ferrite-type steel is very well refined and its impact toughness is better than the quenched-and-tempered products.

Fig.11 shows structure of an intra-granular ferrite-type non-tempered steel after quenching at 640°C interrupting the course of cooling after hot forging. Intra-granular ferrite is observed as indicated by an arrow, and a deposit suspected to be the ferrite transformation nucleus is seen in the intra-granular ferrite. This is a compound deposit of MnS and VN. The VN and (001) plane of the intra-granular ferrite are in the Baker-Nutting relation, and thus VN can act as transformation nucleus for the intra-granular ferrite due to high lattice consistency[11)]. The essential point of structure refinement of the intra-granular ferrite-type non-tempered steel is to widely disperse a great quantity of MnS acting as precipitation nucleus of VN so that a great deal of intra-granular ferrite is precipitated.



(a) Intra-granular ferrite-type non-tempered steel (0.07S-0.13V-0.015N)



(b) TiN-type non-tempered steel (0.016S-0.1V-0.008N)

Fig.9 Micro structures of intra-granular ferrite-type and TiN-type non-tempered steels[10]

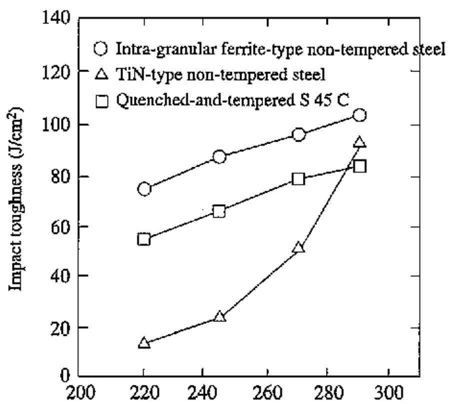


Fig.10 Comparison of impact toughness between intra-granular ferrite-type non-tempered steel and conventional steel[10]

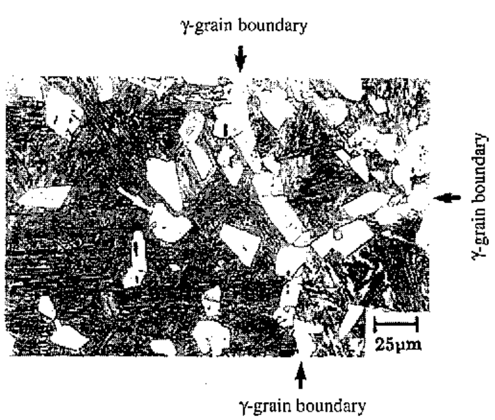


Fig.11 Structure of an intra-granular ferrite-type non-tempered steel after quenching interrupting cooling after hot forging[10] (white arrow: typical intra-granular ferrite)

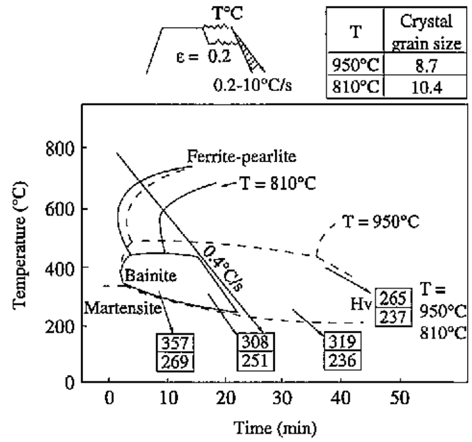


Fig.12 Effects of processing temperature on CCT curve of SCM 440 steel

Besides the ferrite-bainite-type non-tempered steels, there are other low/medium carbon non-tempered steels that use toughness enhancing mechanism of bainite or martensite structure[8]. Though these non-tempered steels have excellent strength and toughness, their low yield ratio and durability ratio have to be improved.

4.3 Soft steels for cold forging

Many small machine parts are made by cold forging work. Normally as-rolled structure of alloy steels like JIS SMC 440 is a mixed structure of bainite, which is too hard even for cutting, let alone cold forging. For this reason, simplified annealing is often done prior to working.

Fig.12 shows result of an investigation on the effects of processing temperature on bainite formation behavior. Formation of bainite is checked in the low temperature range and the material hardness is lowered as a consequence. Austenite grain size becomes small at low processing temperatures. The reason why formation of bainite is restricted under low temperature processing is that the smaller the austenite grain the more its diffusion transformation is accelerated since austenite grain boundary acts as ferrite transformation nucleus.

Various soft cold forging steels have been developed wherein bainite formation is restricted as described above by application of controlled cooling methods such as low temperature rolling, covered slow cooling, etc. in the manufacture of alloy steels, making the hitherto indispensable simplified annealing unnecessary[12].

5. Steels for High Strength Bolts

Although individual bolts are small and light, the number of them used in a vehicle is enormous. Steel materials used for bolts include Carbon Steel for Machine Structural Use S 45 C by JIS and low carbon boron steels for 700 and 800 MPa class bolts and Low Alloy Steels for Machine Structural Use SCM 435 and SCr 440 by JIS for 1,000 and 1,100 MPa class bolts. Bolts are no exception as far as the demands for higher strength as described above are concerned.

As is well known, enhancement of bolt strength is hindered by delayed fracture. Delayed fracture is caused by hydrogen generated from corrosion reaction, coming into the steel material and accumulating at grain boundaries to cause cracks[13]. For improving delayed fracture resistance, therefore, it is necessary to develop materials which (1) do not corrode, or (2) do not fracture even when hydrogen penetrates.

5.1 Development viewpoints

Corrosion inevitably occurs to low alloy steels used for ordinary bolts, sooner or later. For enhancing the bolt strength, therefore, it is effective to raise grain boundary fracture strength. The grain boundary fracture strength is determined by (1) grain boundary energy, and (2) fine deposits precipitated at grain boundary. It is known that grain boundary embrittlement is caused by segregation of phosphorus or manganese or fine deposits of carbides, etc. at the boundary [13]. The reason why fine deposits make grain boundary brittle is suspected that relaxation of plasticity at grain boundary crack ends is hindered by the fine deposits. It is, therefore, effective to make the fine cementite at grain boundary comparatively larger by tempering the bolts at high temperatures in order not to let fine deposit form at the boundary.

5.2 Methods for hydrogen measurement and evaluation of delayed fracture

As described above, the delayed fracture of bolts is caused by hydrogen which comes into steel and moves freely (diffusive hydrogen). A diffusive hydrogen measuring equipment as shown in **Fig.13** has been developed[14]. Hydrogen diffusing into steel around the room temperature can be measured quantitatively by heating the test piece and continuously measuring hydrogen discharged from the heated test piece. It is possible to quantitatively evaluate delayed fracture behavior of steel materials by measuring the amount of diffusive hydrogen using test pieces having different amounts of diffusive hydrogen prepared under different hydrogen charge conditions and, in parallel, by determining critical diffusive hydrogen amount to cause delayed fracture within a certain time limit by measuring the time to cause fracture at delayed fracture tests.

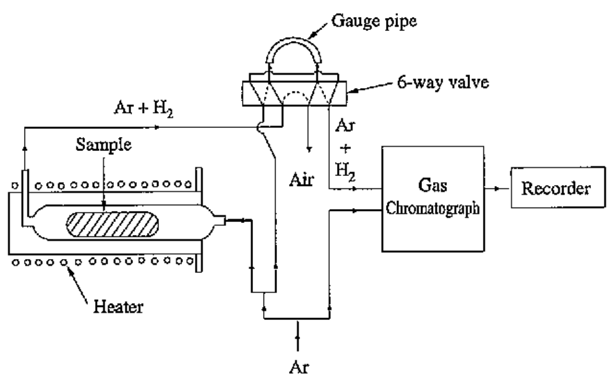


Fig.13 Steel hydrogen analyzer

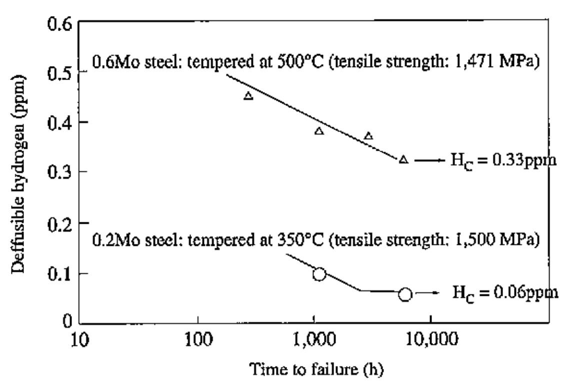


Fig.14 Delayed fracture properties of high-molybdenum, high temperature tempered steel[15]

5.3 Delayed fracture resistant steel for high strength bolts

It is important for steels for high strength bolts having good resistance against delayed fracture to reduce phosphorus, sulfur, manganese, etc. for maintaining grain boundary strength and enhance resistance against temper-softening so that the material strength is not lost under high temperature tempering. Furthermore, refinement of crystal grain size is also important. **Fig.14** shows delayed fracture behavior of the bolt materials developed from such a viewpoint[15]. High molybdenum steel tempered at a high temperature demonstrates a high critical hydrogen amount (H_c) and good delayed fracture properties.

6. Steels for High Strength Bearings

Bearings are principal mechanical components of automobiles and industrial machines and they are the most important of its required characteristics is rolling contact fatigue behavior. High Carbon Chromium Bearing Steels SUJ by JIS or casehardened steels such as SCM 420 by JIS are used for bearings. The High Carbon Chromium Bearing Steels is hyper eutectoid steel with 1% C and, in addition to rolling contact fatigue strength, an excellent wear resistance is given to it by making the final product form a martensite structure dispersed with fine residual carbides through spheroidizing annealing and quench-and-temper. Bearings made of the casehardened steels have superior rolling contact fatigue resistance against surface-initiated cracks often seen in cases like dirty lubricant conditions. This is due to large residual compressive stress at the surface (like in the case of the gear steels) and a great amount of retained austenite in the surface layer. Another advantage of the casehardened steels compared with SUJ is excellent toughness.

6.1 Philosophy of improving rolling contact fatigue strength of bearing steels

Rolling contact fatigue fracture usually initiates from non-metallic inclusions under clean lubrication conditions. For this reason extension of product life is being envisaged by decreasing non-metallic inclusions through minimizing oxygen content and making its size finer. For evaluating size and distribution of harmful non-metallic inclusions, the microscopy-extreme value statistics method has recently been introduced in place of the cleanliness evaluation method of JIS traditionally employed. This new method estimates the maximum size of non-metallic inclusion using the extreme value statistics on an assumption that its distribution follows double exponential distribution. More specifically, an extreme value statistic graph is prepared based on measurements of maximum non-metallic inclusion diameter in the reference microscopy area and the maximum non-metallic inclusion diameter in the risk area is estimated[16]. A good correlation has been reported to exist between the maximum non-metallic inclusion diameter estimated by the microscopy-extreme value statistics method and rolling contact fatigue behavior of materials made to have different non-metallic inclusion diameters under different de-oxidation conditions[17].

6.2 Philosophy of improving rolling contact fatigue strength under high bearing force

Bearings became to be used under higher bearing force conditions as the cleanliness of the bearing steels became remarkably better and the life of the products was consequently made longer. Under such a condition a new phenomenon began to be seen wherein peculiar structures as shown in **Fig.15** were formed during accumulation process of rolling contact fatigue and these structures were found to serve as a fracture propagation channel[18]. The peculiar structures are divided into a lightly colored zone called the white band and a

Fig.15 Peculiar structures and arrested crack caused by rolling contact fatigue progress[18)]

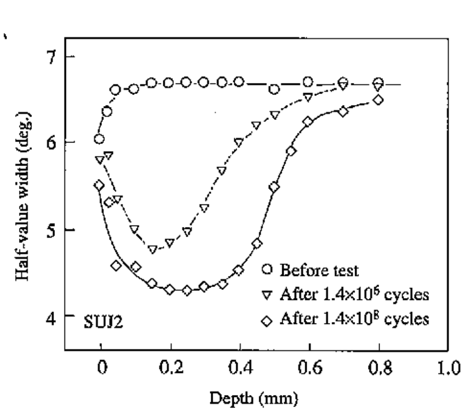


Fig.16 Change of distribution of half-value width of X-ray diffraction peaks[18)]

vertically striped zone called lamellar structure and the development of these structures is most remarkable in the depth range where shear stress is the highest.

From the above facts, it is considered that the peculiar structures are formed by strain concentration and heat generated therefrom. Fine cementite depositions are observed around both the white band and the lamellar structure, hence the lamellar structure can be defined as a pseudo-pearlite phase accompanying precipitation of cementite. The cementite precipitation seems to hint that the temperature in the layer immediately below the rolling contact surface is as high as 300°C. The white band is a kind of ferrite phase having low carbon content and low hardness.

Fig.16 shows distribution of half-value width of X-ray diffraction peaks along the depth. Change of the half-value width with the lapse of time is the most remarkable around the depth of 0.2 mm where the shear stress is the largest, and it decreases monotonously during fatigue progress. The half-value width of X-ray diffraction peaks corresponds to crystal dislocation density and can be used as a macro indicator of formation of the peculiar structures and material deterioration caused by them during development of rolling contact fatigue.

7. Conclusion

Nippon Steel's responses to the demands for reducing the weight and improving the fuel consumption of vehicles were described above focusing on the bar and wire-rod products of special steels. For achieving strengthening and weight reduction of the automobile parts, it is important to combine material developments with design technology and fabrication and working methods for fully utilizing the material characteristics in a well-balanced manner. The authors would welcome suggestions and guidance in this respect.

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