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Development of Case Hardening Steels for Automotive Transmission

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Abstract

The reduction of carbon dioxide emission has become an important issue in the prevention of global warming in the world. For automotive transmission, reduction of the size and weight of parts improves the fuel efficiency of automobiles. Therefore, it is important to strengthen gears and CVT pulleys which are the main parts of automotive transmission. Furthermore, in case hardening heat treatment, vacuum carburizing and gas nitriding in addition to conventional gas carburizing are starting to be adopted. We have developed case hardening steels for high strength transmission parts through a combination of the steel material component and the heat treatment process.

1. Introduction

In recent years, from the viewpoint of global environmental protection, there has been a strong demand for improvement of fuel efficiency of vehicles and reduction of carbon dioxide emissions. For automotive transmission, reduction of the size and weight of parts is becoming important and complex part shapes are being adopted to reduce the size of the unit. Then in addition to the conventional demands for weight reduction and further strength improvement, the gear steel is also required to have the low heat treatment distortion property in order to realize quietness. Furthermore, in order to reduce the carbon dioxide emission in the gear manufacturing process, a shift from the conventionally employed gas carburizing treatment to the gas nitriding treatment and the vacuum carburizing treatment is promoted. Thus automobile companies employ various surface hardening heat treatment processes, and provide necessary functions to their respective transmission parts. In this regard, Nippon Steel Corporation has developed case hardening steels to respond to the requirements of various surface hardening heat treatment processes and the transmission parts. The steels have been developed to provide a superior performance to those of alloy steels for machine structural use of the domestically used JIS-SCr and SCM systems. The developed steel is compositionally designed so that the hardness and the toughness of the carburized layer under the tooth surface and the hardness of the gear core portion are optimized so as to satisfy the requirements of the subject units, and designed to exert their performance after the completion of the customers' manufacturing processes.

This report introduces Nippon Steel's developed steels: high strength gear steel, high impact fatigue resistant strength gear steel, wear resistant CVT (Continuously Variable Transmission) steel, mild carburizing steel and high strength nitriding gear steel.

2. High Strength Gear Steel

Conventionally, for the manufacturing of automotive gears, JIS-SCr420 steel and SCM420 steel, to which the gas carburizing hardening treatment has been applied, are used and the incompletely hardened microstructure generated in the carburized layer remains a problem. In the gas carburizing hardening, since the steel is heated in the carburizing gas atmosphere containing oxidizing gases such as CO_2, H_2O and O_2, the oxides of Mn, Cr and Si are produced at the grain boundaries of the prior-austenite (intergranular oxidation) during heating. As a result, the alloying elements such as Mn, Cr and Si decrease in the matrix around the intergranular oxidation, and the hardenability is decreased.[1, 2] Since the diffusion velocity of oxygen in steel is lower than that of carbon, the depth of the intergranular oxidation layer is smaller than the carburized depth. Consequently, in the gas-carburized hardened steel, despite the high carbon content, the hardenability of the subsurface layer is locally decreased and the incompletely hardened microstructure with low hardness (fine pearlite microstructure) is formed in the subsurface layer. The

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incompletely hardened microstructure with low hardness becomes the origin of the fatigue fracture, and decreases the bending fatigue strength and the pitting fatigue strength, a type of rolling contact fatigue strength of the gear teeth.

To reduce the incompletely hardened microstructure, a reduction of the formation of intergranular oxidation products and an increase in the matrix hardenability are considered. To suppress the intergranular oxidation, the device on the part of the processor's side to reduce the oxidizing gas components of the carburizing gas and composition design on the steelmaker's side to reduce the elements that form intergranular oxidation are considered. For example, as **Fig. 1** shows, Si is most influential on the depth of the intergranular oxidation among the intergranular oxide producing elements. Therefore, by reducing the Si content, and by increasing the content of Mo, an element that does not produce intergranular oxide products, to compensate for the deficiency of the hardenability produced thereby, a microstructure free of incompletely hardened microstructure is developed.[3)]

Table 1 shows the chemical compositions of the developed steel CM201, and **Fig. 2** shows the microstructure of the subsurface layer. As compared with JIS-SCM420 steel, in the high strength gear steel wherein the incompletely hardened microstructure is reduced, the rotating bending fatigue strength is improved by 13%, and the bending fatigue strength of the gear is improved.

Furthermore, in order to further improve the bending fatigue strength, the compressive residual stress is imparted to the subsurface layer. The higher the matrix strength is, the larger the compressive residual stress developed by shot-peening becomes. Therefore, the shot-peening is more effectively applied to the developed CM201 steel wherein the incompletely hardened microstructure is reduced.

It is also important to suppress the pitting failure, the fatigue failure initiated at the gear tooth surface layer, as well as to suppress the gear bending fatigue failure. For instance, one method considered to improve the fatigue strength is, instead of employing the high strength gear steel, to grind off the gear tooth subsurface layer that contains the incompletely hardened microstructure. However, even in the case without the incompletely hardened microstructure, depending on the loading conditions such as contact stress, slipping speed and the like, the tooth surface temperature increases due to the friction-induced heat caused by the repetition of the meshing of the gear teeth, and the tooth surface is tempered thereby. Therefore, the hardness decreases during its operation. When the tooth surface hardness decreases, pitting failure is generated more readily, and so it is important to maintain the tooth surface hardness even during its operation.

It is reported that a good correlation is established between the pitting fatigue strength and the hardness of the carburized layer obtained with tempering at 300°C,[4)] and the pitting fatigue strength is considered to be enhanced by providing the chemical compositions that impart high tempering hardness to the matrix when tempered at 300°C. The hardness obtained with tempering at 300°C is determined by the hardened carburized layer hardness governed by the carbon content of the carburized layer and by the resistance to temper softening governed by the alloying element contents. As **Fig. 3** shows, the element Si has the highest resistance to temper softening in the temperature region around 300°C.

The process of tempering the carbon steel martensite is considered to go through the following three stages: in the first stage, the fine ε carbide ($Fe_{2-3}C$) as a transition carbide is precipitated, in the second stage, the retained austenite is decomposed to form cement-

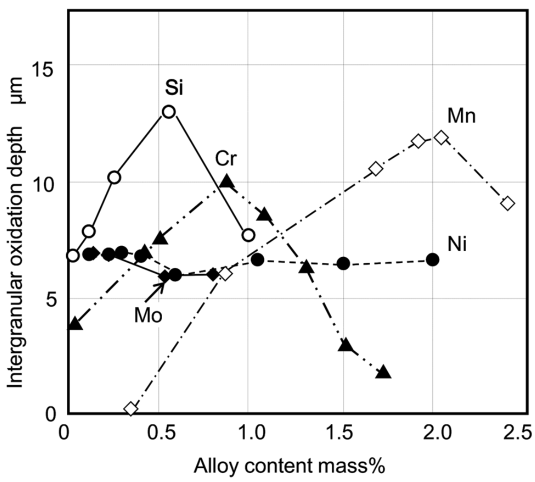


Fig. 1 Intergranular oxidation depths varying alloy contents

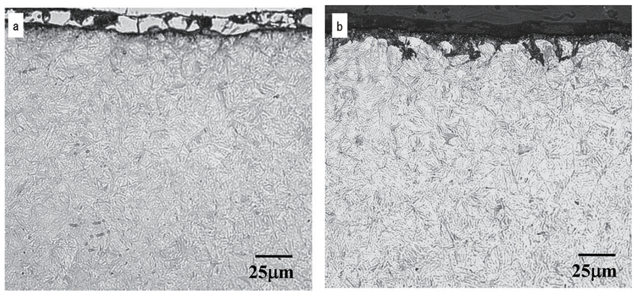


Fig. 2 Micrographs of carburizing layer (nital etching), a) CM201, b) SCM420

Table 1 Chemical compositions of developed steels

(mass%)

	Grade	C	Si	Mn	Cr	Mo	V	B
Developed steel	CM201	0.20	0.10	0.75	1.00	0.40	–	–
	XG5	0.20	0.50	0.35	1.45	0.35	–	–
	NT-B	0.20	0.25	0.40	1.65	–	–	add
	Wear resistance	0.17	0.25	0.80	1.85	–	–	–
	MSB20	0.20	0.80	0.80	0.10	–	–	0.0020
	Nitriding	0.10	0.15	0.55	1.25	–	0.17	–
Conventional steel	JIS-SCM420	0.20	0.25	0.80	1.10	0.20	–	–
	JIS-S35C	0.35	0.20	0.75	–	–	–	–

ite θ and ferrite α, and in the third stage, the ε carbide transforms to the cementite θ.[5)] The third stage of the Si-added steel is considered to move to the higher temperature side, and the transformation of the ε carbide to the cementite θ is suppressed; therefore, the decrease of the matrix hardness during the low temperature tempering is considered to be suppressed. In **Fig. 4**, the bright field image of the transmission electron microscopy (TEM) image of 0.6C-1Si steel tempered at 400°C, the selected area diffraction pattern, the key diagram and the dark field image formed at the spot indicated as DF in the diagram are shown. In the dark field image, a number of fine ε carbides several nm in size and aligned in the same orientation in what is considered to be the martensite lath are confirmed.

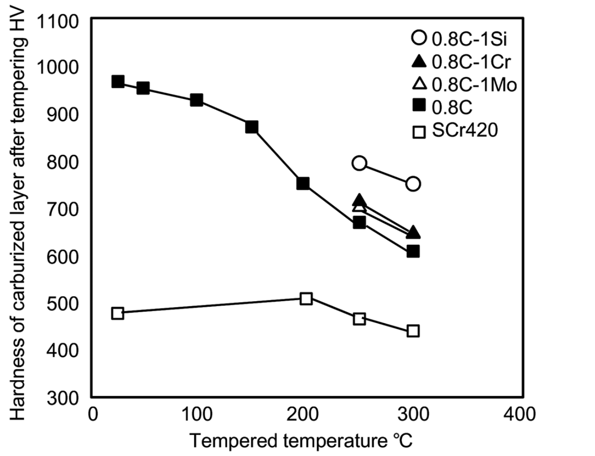


Fig. 3 Hardness of carburized layer after low temperature tempering with varying alloy element contents

The Si content is increased in the developed steel XG5 as shown in the chemical compositions in Table 1, and as **Fig. 5** shows, the pitting resistant strength obtained in a roller pitting fatigue test is improved by 21% from that of JIS-SCM420. Furthermore, by adjusting the contents of Mn and Cr appropriately, the hardness after the spheroidizing annealing treatment is reduced, and as **Fig. 6** shows, XG5 has cold forgeability equivalent to that of SCr420.

3. High Impact-fatigue-resistant Strength Gear Steel

The pinion and the side gear of the differential gear system are subjected to the repetitive impact load generated at a sudden start, and the bending fatigue fracture is caused after the very small numbers of times of impact loading from several tens to several hundreds of times, and thus to suppress the bending fatigue fracture, the low cycle impact bending fatigue strength is required. **Figure 7** shows the result of the low cycle impact bending fatigue test conducted, using a B-added steel, and SCM420 steel as the basis for comparison. The crack initiation strength after 100 repeated impacts is arranged and shown. In JIS-SCM420, the higher the core portion hardness is, the higher the crack initiation strength becomes. Furthermore, in the B-added steel, the crack initiation strength is improved significantly as compared with that of SCM420 despite the core portion hardness being the same.

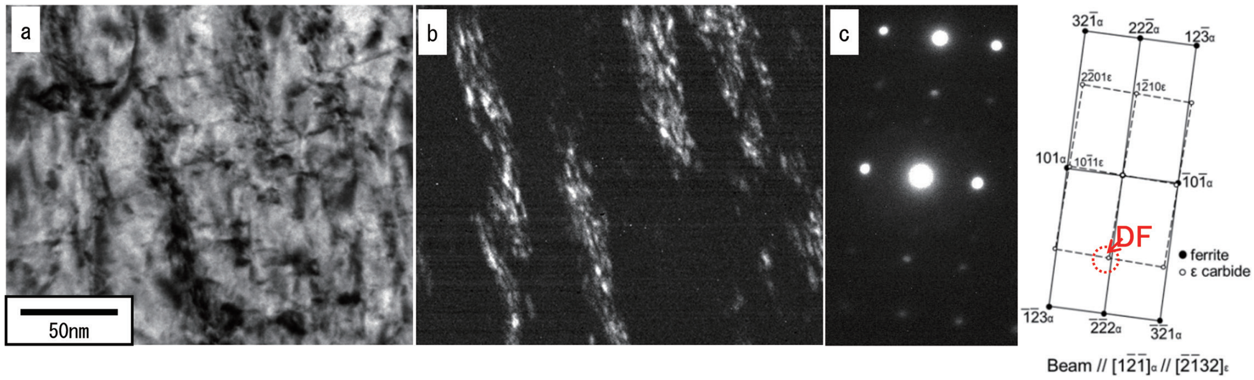


Fig. 4 TEM images of 0.6C-1Si steel tempered 400˚C, a) Bright field image, b) Dark field image, c) Selected area diffraction pattern

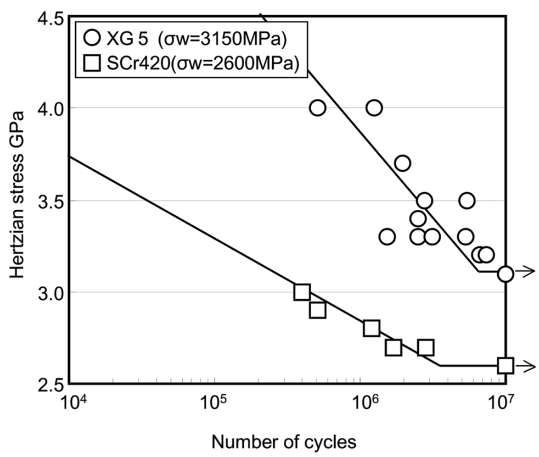


Fig. 5 Results of roller pitting fatigue test with XG5

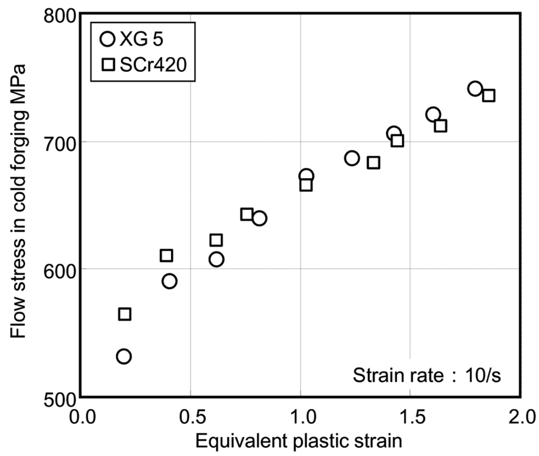


Fig. 6 Flow stress in cold forging in XG5

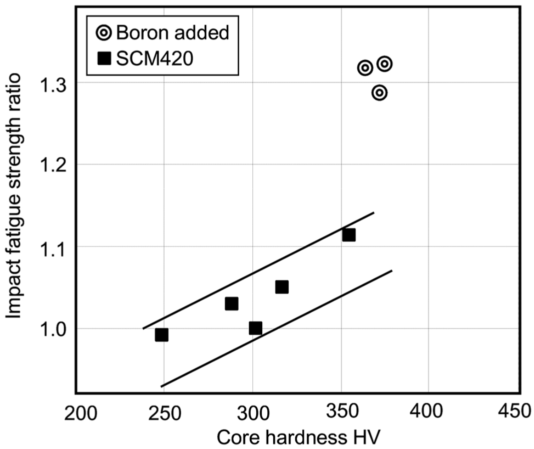


Fig. 7 Results of impact fatigue test

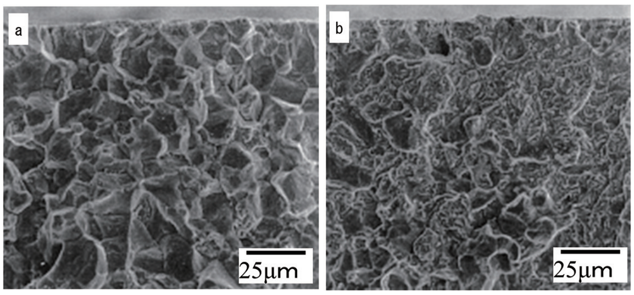


Fig. 8 SEM (scanning electron microscopy) images of impact fatigue fracture origin, a) SCM420: Intergranular fracture, b) B-added: Transgranular fracture

Conventionally, the low cycle fatigue life of steel is expressed by the Coffin-Manson Rule based on the relation between the plastic strain range and the number of times of the repetitive load, and the higher the tensile fracture ductility ε_f is, the longer the life becomes.[6] However, in the carburized gear, the hardness drops inwardly from the surface, and a yield strength distribution that corresponds to the hardness drop exists. When the gear is subjected to an impact load, since the stress generated in the core portion exceeds the yield strength, a macroscopic tooth plastic distortion occurs. When an impact bending is repeated in one direction, the tooth plastic distortion is accumulated in one direction in a ratcheting-like manner, and the mean strain ε_m developed in the subsurface layer gradually increases. According to the interpretation of Sachs-Weiss et al., the larger the difference ($\varepsilon_f - \varepsilon_m$) between the tensile fracture ductility ε_f and the mean strain ε_m is, the longer the low cycle fatigue life having the mean strain becomes.[6] When an impact load of the same magnitude is repetitively applied to gears having almost the same subsurface layer hardness but different core portion hardness, the higher the core portion hardness is, the smaller the strain developed in the subsurface layer is.

Based on the above, it is interpreted that the higher the core portion hardness is, the smaller ε_m becomes, and as a result, $\varepsilon_f - \varepsilon_m$ becomes larger, and the life becomes longer. Furthermore, the reason for B-added steel crack-initiating life becoming longer than that of the carburized SCM420 is considered to be attributed to the competition effect that reduces the intergranular concentration of P, an intergranular embrittlement element, suppresses the intergranular fracture, and improves the fracture ductility of the carburized layer thereby. The fracture surface images of the origin of the low cycle impact fatigue fracture shown in **Fig. 8** indicate the transition from the intergranular fracture to the transgranular fracture with the addition of B.

4. Wear Resistant CVT Steel

In the belt type CVT, a pair of a carburized pulleys holds the steel belt (element+ring) in between, and transmits the power via the frictional force generated between the two units. In the operation, the pulley surface is subjected to a repetitive pressing force exerted by the belt and a frictional force, and the minute wear of the depth of several μm to several tens of μm is developed on the pulley surface. The step caused by the minute wear inhibits the smooth continuous speed control that characterizes the belt type CVT and deteriorates the frictional force between the belt and the pulley surface. The minute wear of the pulley is the fatigue wear caused by the serial minute cracks. When the Hertzian contact of the minute groove on the side face of the element of the belt unit with the pulley surface is assumed, the subsurface layer of about 10 μm is considered to be subjected to the shearing stress, and by the repetition thereof, minute cracks are initiated on the pulley surface, leading to a wear after they are linked.

In the wear resistant CVT steel, the wear resistance is improved by strengthening the portion subjected to the repetition of the shearing stress by taking advantage of the shearing stress repetitively applied through the contact with the belt. Specifically, aiming at improving the subsurface layer hardness, the carbon concentration of the subsurface layer was made to be above 0.8% of the eutectoid concentration to form the retained austenite of high carbon, which was then transformed to stress-induced martensite.[7] To exploit this effect efficiently, as Table 1 shows, more Cr was added to the wear resistant CVT steel than to JIS-SCr420. Owing to the high affinity for carbon, Cr has the effects of increasing the content of carbon in gas carburizing, lowering the Ms point and increasing the amount of the retained austenite generated in the gas carburizing hardening. As compared with JIS-SCM420, in the CVT pulley applied with the developed steel, the amount of the retained austenite was increased by 1.5 times before operation. Through the operation of the unit, the pulley subsurface layer hardness was significantly enhanced, and as a result, the wear depth was reduced by 37% and excellent wear resistance was exhibited.[8]

5. Mild Carburizing Steel

Conventionally, for the gas carburized high strength gear steel, the incompletely hardened microstructure was reduced, the subsurface layer hardness was enhanced and the fatigue strength was secured from the material aspect viewpoint by adding Ni and Mo. However, Ni and Mo are rare metals, and there has been a strong need for suppressing the material cost rise due to the massive addition of these metals. The development of a new mild carburizing process is in progress that compatibly enables the saving of the alloy metal in the steel chemical compositions and the enhancement of the fatigue strength of gear units by combining the advantages of the vacuum carburizing treatment and the high-frequency quenching process.[9] In the vacuum carburizing, the carburizing treatment is conducted in a furnace wherein the furnace pressure is reduced to several kPa and C_2H_2 gas is introduced under about several tens of kPa. As oxidizing gases are not contained in the carburizing gas, the intergranular oxidation is reduced, and the incomplete hardening is also reduced. Furthermore, in the gas carburizing, when the Si content is increased, as an oxidation film is formed, the carburization is

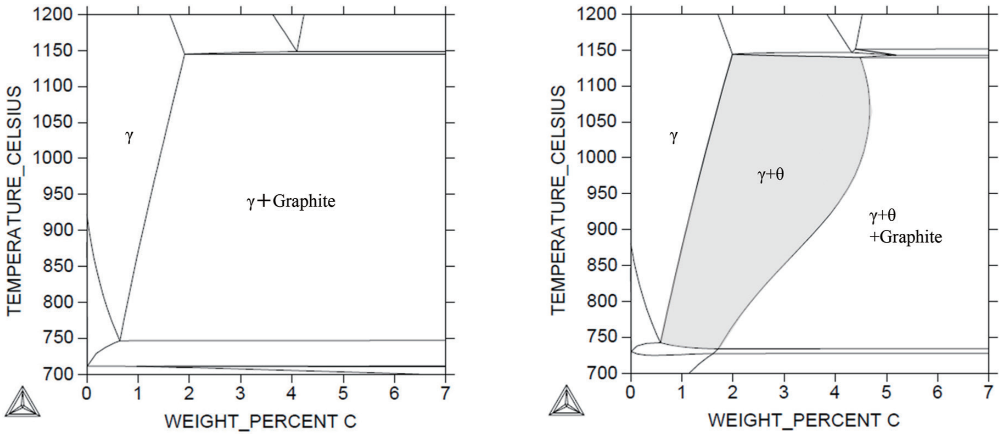


Fig. 9 Equilibrium phase diagrams of Fe-C system, a) MSB20 and b) SCM420

inhibited.[10] However, in the vacuum carburizing, as the oxidation film is not formed, the freedom of the chemical composition design of the Si content in steel is enlarged.

In the high-frequency quenching process, as heating and quenching are accomplished within a short period of time, the crystal grain is refined, and therefore, the suppression measure for crystal grain coarsening by adding the micro-alloying elements such as Nb and Ti used in the gas carburizing process is not necessary.[11] Additionally, as compared with the oil quenching that is generally used in the gas carburizing quenching, since water is used as the refrigerant, the cooling rate is high, and the minimally necessary hardenability of the steel material can be produced.

The vacuum carburizing is conducted via two stages. In the first carburizing stage, the steel units are heated under a reduced pressure atmosphere, and then carburized by C_2H_2 gas under the pressure of several tens of kPa. In the subsequent second stage of diffusion, the units are retained under a vacuum atmosphere for diffusing carbon in the said units. In the diffusion period, the carbon concentration in the neighborhood of the surface exceeds its solid-solution limit, and the cementite θ is precipitated depending on the steel chemical compositions and the temperature. Coarsened θ is precipitated at the retained austenite grain boundaries in particular, deteriorates the intergranular strength, and deteriorates the gear tooth bending fatigue strength.

Then, for the development of MSB20, the chemical compositions were studied so that the precipitation of θ does not occur at the grain boundary of the prior-austenite during the carburizing period in the vacuum carburizing. For example, Morita et al.[12] report that the subsurface layer carbon concentration of the material in the carburizing period in the vacuum carburizing is raised up to that of the carbon concentration that balances graphite, and θ is precipitated in the grain boundary when the phase is dual and comprises the γ phase and θ phase.[13] Additionally, due to the decreasing of the Cr concentration and the increasing of the Si concentration, the carbon concentration that balances graphite decreases, and the region where the θ phase exists becomes smaller in the equilibrium phase diagram of the Fe-C system. Based on these findings, in order to establish a chemical composition system to avoid thermodynamically the formation of the dual phase of γ and θ by adjusting the Si and Cr contents, the Si content was increased from the viewpoint of the aforementioned resistance to temper softening, and the Cr content was adjusted to below 0.1% from the viewpoint of suppressing θ in the grain boundary during the vacuum carburizing. Furthermore, to suppress the intergranular precipitation of the embrittling elements, the P content was decreased, and with the addition of the minute amount of B to enhance the intergranular strength, the delay of the fatigue crack propagation was sought. Table 1 shows the chemical compositions, and **Fig. 9** shows the equilibrium phase diagram of the Fe-C system.

The pitting resistant strength of the mild carburizing developed MSB20 steel obtained in a roller pitting test shows a 25% enhancement of strength as compared with that of the gas carburized SCM 420. The hardness at the depth of 25 μm of SCM420 after tempering at 300°C is HV507, while that of MSB20 is HV639. These results are considered to be attributed to the superimposed effect of the suppression of the incompletely hardened microstructure and the enhancement of the resistance to temper softening by increasing the Si content. Furthermore, the rotating bending fatigue strength is enhanced by 13%. Based on these results, with the combination of the developed steel and the mild carburizing treatment, it is clearly possible to enhance the pitting fatigue strength and the bending fatigue strength simultaneously.

6. High Strength Nitriding Gear Steel

To achieve the high strength of the gear steel, in the general carburizing treatment, transformation to martensite is used. However, the heat treatment distortion due to the transformation expansion is unavoidable. Since the heat treatment distortion causes the gear teeth meshing noise, it is a general practice to grind the gear tooth after the carburizing treatment to restore its correct profile to reduce its influence. However, in addition to being one of the cost increasing factors, the application of teeth grinding is difficult in some cases such as to internal teeth gear. Therefore, focus was placed on the lower temperature treating nitriding technique[14–16] that gives small heat treatment distortion, and a high strength nitrided gear steel was developed.

The nitriding treatment is a process in which the subject units are heated at 500–600°C, and the subsurface layer is hardened by the nitrogen atoms that are diffused to the steel through the surface. The surface layer is strengthened with the solute strengthening by nitrogen atoms dissolved in the steel and the precipitation strengthening by the formation of the alloyed nitride products. Since the subsurface layers of the tooth root and the tooth surface of a gear are subjected to high stress, it is necessary to increase the hardness of the nitrogen-dispersed layer efficiently in the nitriding process. From the viewpoint of machinability, hardness should not be very

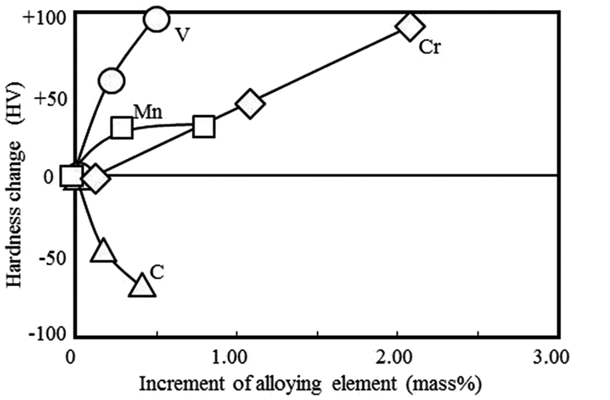


Fig. 10 Effect of alloying element amounts to the nitrided surface hardness

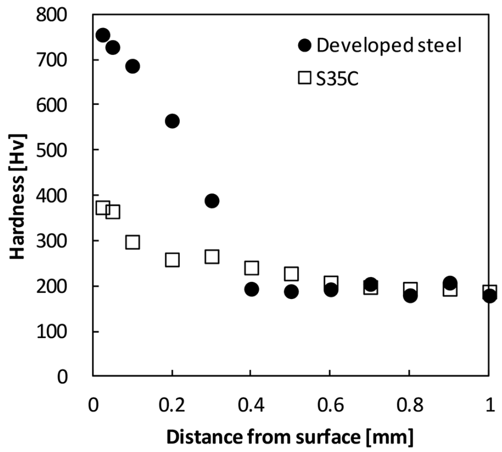
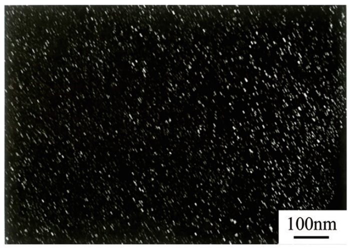


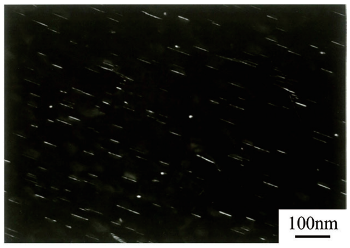
Fig. 11 Hardness profile of nitrided layers of developed steel and S35C

high before nitriding treatment. However, if the gear core portion is very soft, plastic distortion is developed in the core portion during operation, therefore, the gear material should be hardened but not be softened at the nitriding treatment temperature.

In the development of the high strength nitriding gear steel, as **Fig. 10** shows, the change of the subsurface layer hardness in relation to the changes of the alloy element contents of steel was studied. From the result, the chemical compositions indicated as Nitriding in Table 1 were adopted, wherein the contents of Cr and V, both of which are considered to be effective in improving the subsurface layer hardness, are increased, and the content of C that is considered to worsen the subsurface layer hardness is reduced. **Figure 11** shows the hardness distributions of the nitriding-treated developed steel and the nitriding-treated S35C. In the developed steel with high Cr and V content, the subsurface layer hardness reaches HV700–800. Through the TEM observation at the point of a depth of 50 μm, fine precipitates are observed in either nitrided layer. When the distribution states of the precipitates are compared, as opposed to the particles in S35C having the long side of about several tens of nm in length and dispersed coarsely, the refined particles of several nm in size are precipitated very densely in the developed steel (**Fig. 12**). It was considered that the difference in the state of the distribution of the fine precipitates developed the difference in the hardness of the nitrided layer of both steels. Furthermore, from the selected area diffraction pattern of the precipitates shown in **Fig. 13**, it was interpreted that the particle is the carbide-nitride-based composite (Cr, V)(C, N)



(a)Developed steel



(b)S35C

Fig. 12 TEM dark field images of precipitates taken at a depth of 50 μm below the surface

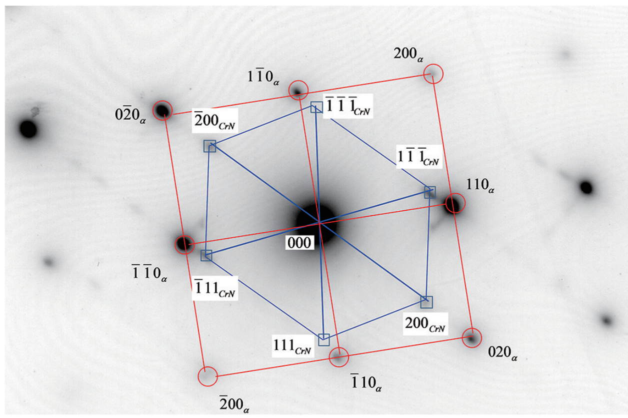


Fig. 13 Selected area diffraction pattern taken in nitrided developed steel and corresponding key diagram

of the FCC type, and contributed to hardening by being precipitated based on the orientation relationship between the base metal ferrite and the Baker-Nutting.

As compared with S35C, the developed steel exhibits 3.2 times higher rotating bending fatigue strength and 1.3 times higher fatigue strength in the roller pitting test. Such strengths are equivalent to those of the gas-carburized JIS-SCr420 units, and the developed steel is used as the gear steel replacement for the carburized gear steel[17].

7. Conclusion

This paper introduced the transmission gear steels developed by Nippon Steel. The developed steels introduced herein have already been put into actual mass production, contributing to the reduction of the carbon dioxide emission and the global environment protection through the downsizing and weight reduction of automotive parts and the optimization of the manufacturing processes.

Acknowledgements

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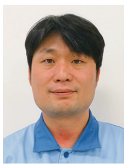

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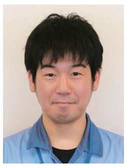

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