MATERIALS OF AN ALL-UNION SCIENTIFIC AND TECHNICAL CONFEREN

EFFECT OF THE QUALITY OF MATERIAL ON THE RELIABILITY AND LIFE OF GEAR WHEELS

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UDC 621.833.7.05:62-

The reliability and life of heavy-duty gear wheels are not determined by optimum des alone but also depend to a considerable extent on the properties of the material and thei production technology; the respective requirements were explained fairly exhaustively in [1-4]. The present article contains some information that is usually not taken into acco by designers and technologists.

The requirements that steels for gear wheels have to fulfill do not necessarily have to be confined to criteria of content of carbon and alloying elements [1-4]. The quality of the case and the strength properties of gear wheels are greatly affected by the grain size. With grain size No. 5 or more ($d \geq 60$ μm) the resistance of case-hardened steel to brittle failure is substantially impaired. This is due to the fact that the boundaries of large grains are more intensely carburized and saturated with carbides than the boundarie of small grains. During the subsequent quenching, the carbides are not completely dissolv and block the grain boundaries, thus hindering recrystallization and refinement of the grain [5

Static bending tests of specimens imitating a case-hardened tooth with module 4 showe that when the grain size is changed from No. 8 to No. 5, it leads to a considerable loweri of bending strength (σ_b) and of the rupture work (A) of steels 20Kh2N4A and 18Kh2N4MA (Tab 1). Reduced σ_b as a result of coarse grain was also found in the case-hardened steel 25Kh2GNTA [4].

Coarse grain strengthens the effect of embrittling factors such as increased thicknes of the case and excessive carburization, which cause spalling on the teeth [3]. It should be pointed out, however, that similar spalling was also found in gear wheels made of steel 20Kh2N4A with grain size No. 5 and with specified thickness of the case and carbon content in it [5]. In these cases the cause of spalling is the large grain, with which we found in the most highly saturated part of the tooth (intersection of the crest, end face, and working face) large martensite needles commensurable with the grain size. The fracture of the case had a pseudostonelike structure while the core had a standard fracture, which distinguishes it in principle from the fracture of overheated steel.

Spalling of gear-wheel teeth made of coarse-grained steel occurs after hardening and shot peening (in practically finished parts), and this causes considerable economic losses To prevent such defects, steel for gear wheels has to be stably fine-grained (Nos. 6-9 according to GOST 5639-65); this is attained by optimum melting technology and acceptance inspection of gear wheels.

Acceptance inspection of steel has to be particularly reliable because according to the conditions of melting, the steel is prone to grain growth. When such steel is used fo making gear wheels, its proneness to overheating upon pressing must be taken into account; this as well as large initial grain size impairs the mechanical properties of case-hardene steel (Tables 1, 2).

The question of optimum saturation of the case of gear wheels with carbon to ensure high static and fatigue strength, resistance to pitting and wear has been studied in sufficient detail [1-4]. Attention should only be drawn to the necessity of a differentiated approach to this by taking the nature of the alloying into account. For Cr–Ni steel, a content of 0.9-1.2% C in the case is optimal. In consequence of the proneness to oversaturation, the formation of a carbide network, and the stability of austenite, a higher carbon content in the case of this steel is unacceptable. If the steel is economically alloyed by a complex of carbide-forming elements, is stably fine-grained, has a structure of finely

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*Load Capacity and Quality of Gear Transmissions and Reduction Gears of Machines (Sept. 25-27, 1985, Alma-Ata).

Translated from Metallovedenie i Termicheskaya Obrabotka Metallov, No. 7, pp. 50-53, July, 1985.

0026-0673/85/0708-0540$09.50

TABLE 1

Steel	Size of specimens, mm	Relative thickness of case	Grain No.	σ_b, MPa	A, J
18Kh2N4MA	6 × 10 × 60	0,25	5	2000	—
			8	2830	—
20Kh2N4A			5	—	400
			8	—	600
18Kh2N4MA	3 × 10 × 60	1 (bulk carburization)	5	1870	220
			8	2600	570
20Kh2N4A			5	2330	460
			8	3000	580

TABLE 2

Overheating of pressings	Kind of fracture	σ_b, MPa	A, J
Without overheating	Ductile	1800...1950	280...335
With overheating (heating to 1250 deg C instead of 1180 deg C)	Coarsely stonelike	1460	195
	Stonelike	1450	179

TABLE 3

Hardening equipment	Tempering temperature, deg C	σ_b, MPa	f, mm	A, J
Installation STTsA	120	1880	0,09	880
	130	1880	0,10	880
	140	1950	0,10	960
	150	2120	0,15	1200
	160	2180	0,21	1230
Furnace Ts-105	150	2420	0,24	1520
Furnace SNZ-8.5	130	2380	0,23	1470
	160	2450	0,27	1570

acicular martensite and disperse separated carbides, the optimum carbon content in the case may be higher (1.5-1.8%). This ensures high heat resistance of the case connected with the alloying of the solid solution and higher stability under contact loads.

For instance, nitrogen case-hardened steel 20Kh3MVFA is not inferior in fatigue limit to the high-strength steel 20Kh2N4MA but is better in contact-fatigue strength [6]. In consequence of the high heat resistance of the case, steel 20Kh3MVFA is less sensitive to structural changes in grinding (burning) which impair fatigue strength and resistance to pitting [2, 7].

The mechanical properties of case-hardened steel are also affected by the cooling rate in quenching and by the hydrogenation in heating in protective atmospheres. The cooling rate does not depend on the composition of the coolant alone, but also on the rate at which it is mixed. In the quenching tanks of continuous case-hardening installations (STTsA) intense mixing of the oil and a high cooling rate of the parts are ensured. Brisk cooling in quenching reduces the resistance to brittle failure of case-hardened steel 20Kh2N4A (Table 3) because a structure of martensite with a high level of microdistortions forms (Fig. 1). On the teeth of case-hardened gear wheels with particularly small module ($m \leqslant 5$) spalling may then occur.

Hydrogenation in heating prior to quenching in an atmosphere of endogas impairs the strength properties of case-hardened steel in consequence of reduced ductility of the core. It should be noted that as a result of saturation with hydrogen the embrittlement of the case is less noticeable than the embrittlement of the core [8]. Holding for 1 month of parts hydrogenated in hardening leads to the desorption of the hydrogen and the restoration of the mechanical properties; this can be accelerated by raising the temperature or by lengthy tempering after hardening [8].

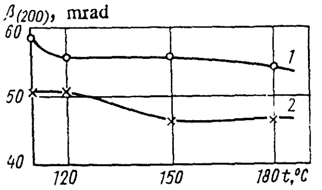

Fig. 1

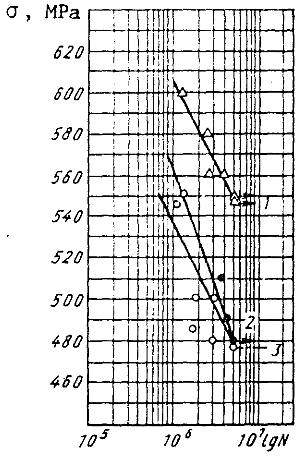

Fig. 2

Fig. 1. Effect of low tempering temperature on the width of the line of martensite $\beta_{(220)}$ of steel hardened at different cooling rates: 1) hardening with cooling in a tank with intense mixing of oil (furnace STZ of the installation STTsA); 2) hardening with cooling in a tank with calm oil circulation.

Fig. 2. Torsional fatigue strength of case-hardened steel 18Kh2N4MA. The variants of treatment are indicated next to the curves.

TABLE 4

Quenching	σ_b*, MPa, after		
	case-hardening	single-stage nitrogen case-hardening	two-stage nitrogen case-hardening
Direct	1490	1630...1790	1480...1640
After second heating	1890...2080	1990...2540	1800...2400

*Specimens of steel 20Kh3MVFA.

Economical and fairly widely used is the direct quenching of gear wheels after heating for case-hardening or nitrogen case-hardening. However, this process cannot be used with Cr—Ni steel which contains very stable austenite because high tempering before quenching is required to ensure an optimal microstructure of the case. Quenching after a second heating improves the stability of the microstructure of the case and the mechanical properties of nickel-less steels (Table 4).

It is known how subzero treatment increases the hardness of the case. However, the embrittling effect of this operation is not always taken into account; this effect is the more intense, the larger the amount of residual austenite that is subjected to γ → α-transformation at cryogenic temperatures [2, 9]. Subzero treatment impairs the characteristics of resistance to brittle failure of case-hardened steel 18Kh2N4MA in static bending (Table 5) and also of fatigue strength (Fig. 2), while contact-fatigue strength does not increase, regardless of the increased hardness [2].

The concluding operation of the cycle of thermochemical treatment, tempering after quenching (or after subzero treatment), has a considerable effect on the properties of case-hardened steel, substantially reducing the microstresses of the high carbon martensite. Thi entails a reduction of the useful compressive stresses in the case, and therefore a trend appeared considering low tempering as a useless or even deleterious operation. At the same time it was not taken into account that omitting tempering or lowering its temperature below the optimal one leads to an abrupt decrease of resistance to brittle failure of case-hardene steel. The effect of low tempering is particularly obvious when it is summed with embrittle

TABLE 5

Variant of treatment	Heat treatment after case-hardening	*HRC*	σ_b, MPa	*f*, mm	*A*, J	Microstructure of the case
1	Double high tempering + quenching at 800 deg C + low tempering	60	1870	0,10	750	Finely acicular martensite, no austenite, globular carbides (Fig. 3a)
2	Quenching at 850 deg C + low tempering	54	1680	0,10	680	Coarse and medium acicular martensite, large amount of austenite, globular carbides (Fig. 3b)
3	Quenching at 850 deg C + subzero treatment + low tempering	57	1570	0,03	545	Large martensite needles on background of finely acicular martensite, no austenite, globular carbides (Fig. 3c)
4	Quenching at 850 deg C + subzero treatment	63	810	0	360	The same

Note. Double high tempering was carried out at 640°C, low tempering at 150°C, subzero treatment at −196°C.

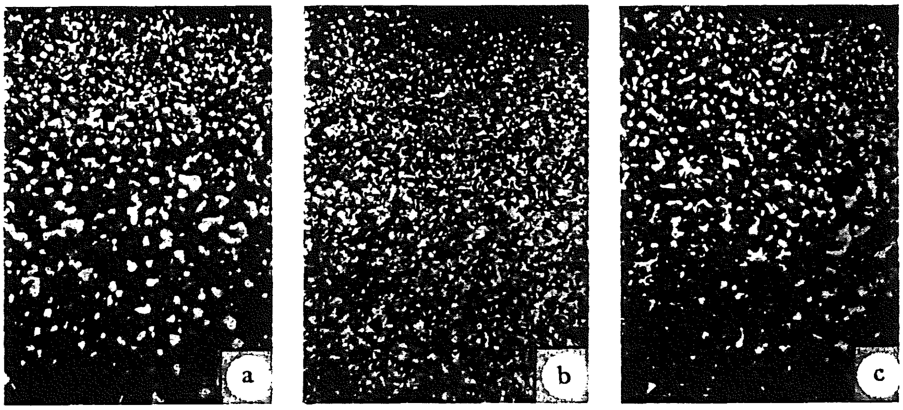

Fig. 3. Microstructure of the case of steel 18Kh2N4MA (×200): a, b, c) variants of treatment 1, 2, 3, respectively.

ment due to other factors such as increased cooling rate in quenching (Table 3, Fig. 1), thickness of the case and carbon contained in it, subzero treatment (Table 5), etc. More than once we found brittle through failure of thin-walled case-hardened parts in consequence of nonadherence to the regime of low tempering. Reduced resistance to brittle failure has an adverse effect on fatigue strength [9] and resistance to pitting of case-hardened steel.

Nitriding, used alongside case-hardening and nitrogen case-hardening for improving the strength of gear wheels, is inferior to these two methods of thermochemical treatment as regards fatigue strength and resistance to pitting [2], but under certain loading conditions the mechanical properties of nitrided steels are sufficiently good, ensuring satisfactory operational stability of gear wheels. An undoubted advantage of nitriding is the minimal deformation of parts on account of the low-temperature regime of the process (500-600°C), and also the possibility of finish-machining completed parts.

For instance, nitriding is used successfully for improving the strength of epicyclic gear wheels meshing with case-hardened parts and breaking down (before the introduction of nitriding) on account of high wear.

Nitrided gear wheels may be made of medium alloy steels (18Kh2N4MA), but also of low alloy steels, in particular steel 38KhS.

The application of plastic surface deformation (roller peening, shot peening) for improving the strength of case-hardened steel is a reserve for increasing the life of gear wheels [2]. The optimum variant of ensuring high operational stability of gear wheels includes a rational geometry of the tooth gap (with undercut), strain-hardening of shot peening, which provides a high level and stability of the residual compressive stresses, and in consequence makes for high fatigue strength. The design and production technology of gear wheels envisage grinding solely on the surfaces of the teeth, ensuring high resistance to pitting without impairing bending fatigue strength.

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EFFECT OF THE SURFACE LAYER ON THE GALLING OF THERMOCHEMICALLY TREATED GEARS

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UDC 620.18:669.14.018.298:620.186.4

There are many studies of galling on the working surfaces of heavy duty gears [1-3, etc.]. The falling reveals itself through formation of deep, wide grooves, chips and lumps on the contacting surfaces.

The galling of thermochemically treated gears can be controlled by providing the surface layer with the correct microstructure, physicochemical, physical, and mechanical properties. There are contradictory data as to the effect of the above characteristics on the galling resistance of gears.

For instance, according to the experimental results [2], the higher the frictional work hardening of the materials in contact, the lower their antigalling properties. The authors believe that hardness has only a secondary effect.

The deformation energy relaxation of steel also has a significant effect on galling control. It was shown [3] that the galling resistance of steel increases with increasing energy relaxation. High deformation energy relaxation is characteristic of Cr—Ni steels with a high content of fine residual austenite (up to 45%) in the carburized layer; while low energy relaxation is characteristic of steels with a high content (up to 60%) of coarse residual austenite, as well as steels having a predominantly martensite structure with a low content (up to 10-12%) of residual austenite.

It was shown [4] that with increasing residual austenite content in the structure, the resistance of steel to galling declines.

Translated from Metallovedenie i Termicheskaya Obrabotka Metallov, No. 7, pp. 54-56, July, 1985.

0026-0673/85/0708-0544$09.50