

The Effect of Chromium in High Carbon Bearing Steels

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Chromium is present in the 52100 bearing steel composition in the range 1.30 to 1.60 wt pct. Chromium has a significant influence on the spheroidization of cementite, finer carbides being formed due to chromium additions. The ferrite to austenite transformation temperatures are increased due to chromium. The microchemistry of the cementite in 52100 changes during heat treatment; the chromium content of secondary carbides is generally lower than that in the spheroidal (FeCr)₃C produced by soft annealing. The rate of carbide dissolution is controlled by the rate of chromium diffusion from the carbide-matrix interface. Also, the chromium content of the residual, spheroidal (FeCr)₃C increases during austenitization. The effect of substitution of chromium in bearing steel compositions is discussed. In view of the beneficial effect of chromium, only substitution by similar strong carbide forming elements should be considered for bearing steels.

I. INTRODUCTION

THE standard hypereutectoid steels used for through-hardened rolling bearings contain chromium in the range of 0.9 to 1.8 wt pct. Typical designations for these steels include SAE 52100, ASTM A485 Grades 1 to 4, AISI Modifications 1 to 4, JIS SUJ2, 100Cr6, RUL 1, CSN 414109, ShKh15, GCr15, *etc.* A review in the United States has classified chromium as the second most strategically vulnerable metal after tantalum.¹ As a result of this vulnerability, new substitute compositions for bearing steels have been reported (see Table I), to join others proposed over the years.²⁻¹⁰ Before considering an alternative composition to 52100 steel the influence of chromium needs to be defined clearly. Chromium has a critical influence on heat treatment in terms of hardenability, control of grain size, consistency of spheroidization, and resistance to decarburization.

In this paper, the particular effect of chromium on the behavior of carbides during spheroidization, and austenitization is discussed in the context of its possible substitution in the 52100 standard bearing steel composition.

II. EXPERIMENTAL PROCEDURE

Steel within the 52100 specification range in the form of 15 mm diameter hot rolled and spheroidize annealed bar was used for the investigation together with two other 1 pct carbon experimental steels. The experimental steels contained zero and 0.64 wt pct Cr, the remaining elements representing the nominal composition of 52100; see Table II. The experimental steels were hot rolled at 1100 °C down to 12.5 mm diameter with a finish temperature of approxi-

Table I. 52100 Substitute Steels

Code (If Any)	Proposal Year	Composition, Wt Pct							Reference
		C	Mn	Si	Mo	Cr	Ni	V	
—	1951	0.94	0.35	0.22	0.48	0.08*	0.05*	—	2
EX2	1965	0.70	0.41	0.28	0.11	0.28	0.91	—	3
TBS 9	1971	0.9	0.65	0.3	0.075	0.225	—	—	4
NL	1969	0.75	2.0	2.0	0.01*	0.95	0.09*	0.1*	5-6
—	1972	0.75	2.0	2.0	0.01*	1.35	—	—	7
NL-M	1975	0.71	1.69	1.44	0.81	0.6	—	0.22	8
Si-Mo	1984	1.04	0.61	0.6	0.28	—	—	—	9
Ni-Mo	1984	0.8	0.35	0.27	0.69	—	1.03	—	
—	1984	0.81	0.82	0.85	—	0.51	—	0.04	10

*Residual contents

Table II. Compositions of Steels Evaluated in This Study

Code	Composition, Wt Pct								
	C	Cr	Mn	Si	S	P	Mo	Ni	
1.42 pct Cr (52100)	1.01	1.42	0.3	0.24	0.02	0.02	0.015	0.055	
Zero Cr	1.01	0.03	0.38	0.43	0.026	0.013	0.003	0.03	
0.64 pct Cr	1.00	0.64	0.37	0.43	0.026	0.014	0.003	0.03	

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mately 900 °C. Slow cooling to room temperature resulted in normalized microstructures.

The phase transformation behavior of the three steels was characterized using a quenching dilatometer with pro-

grammed continuous heating rates. In addition to recording the ferrite-to-austenite transformation temperatures A_{c1} (A_{c1b}) and A_{c3} (A_{c1e}), the martensite start temperature was also measured after quenching from various temperatures above the A_{c1} temperature. Specimens of size 5 mm OD \times 3 mm ID \times 10 mm length were used for the dilatometry studies. Subsequently, quantitative metallographic measurements were made of the area fraction of undissolved spheroidal M_3C . The area fraction determination of residual carbide was accomplished using grids placed on photomicrographs taken at 1000 times magnification of metallographic specimens etched in picral. The mean matrix carbon content (C_γ) was calculated from the area fraction of the undissolved spheroidal M_3C .

The applied expression was as follows:

$$C_\gamma = C_t - (V_v/0.1495)$$

where C_t = total carbon content of steel

V_v = volume fraction (equivalent to area fraction) of undissolved spheroidal carbides.

From this, the volume fraction of undissolved carbides in a fully spheroidize annealed 52100 steel containing 1 wt pct carbon will be 14.95 pct, assuming zero carbon in the ferritic matrix. Inhomogeneity on a macrostructural scale is a major problem in obtaining reproducible quantitative microstructural measurements. The reproducibility was improved by performing the measurements exclusively on the transverse section of the cylindrical dilatometer specimens.

For the calculation of the matrix carbon content, a density of 7.671 g/cc for the M_3C and 7.374 g/cc for the ferrite matrix was used. Strictly speaking, a correction should also be made for the substitution of Cr for Fe in the $(FeCr)_3C$ carbides in the chromium-containing steels. However, the difference is marginal as the relative difference in atomic weight between chromium and iron is less than 7 pct. For consistency, the chromium content in the M_3C was assumed to be 9 wt pct.

As reported in a previous publication,¹¹ the prior austenite grain size can be determined in 52100 steel by metallographic etching in a saturated picric acid solution containing minor additions of HCl and "Teepol" wetting agent. Further characterization of the martensitic matrix microstructures was performed on thin foils in a transmission electron microscope.

Experimental heat treatments to produce the various carbide phases were applied to the 52100 steel using a high temperature salt bath. The chromium content of the carbides was determined in a scanning electron microprobe analyzer using carbon-extraction replicas. The technique was similar to that adopted by Estay, Chengji, and Purdy¹² to determine the carbide microchemistry in Fe-C-Mn, dual-phase steel. The extracted carbides could be observed and analyzed directly at an accelerating voltage of 17 kV, and a typical micrograph of extracted carbides is shown in Figure 1. The chromium $K\alpha$ full width half maximum (FWHM) intensity was measured relative to the iron $K\alpha$ FWHM intensity using a Si(Li) solid state energy dispersive X-ray spectrometer. By removal of the fine carbides from the ferritic matrix, a reliable measurement could be obtained of the Cr/Fe ratio after application of the atomic number, fluorescence and

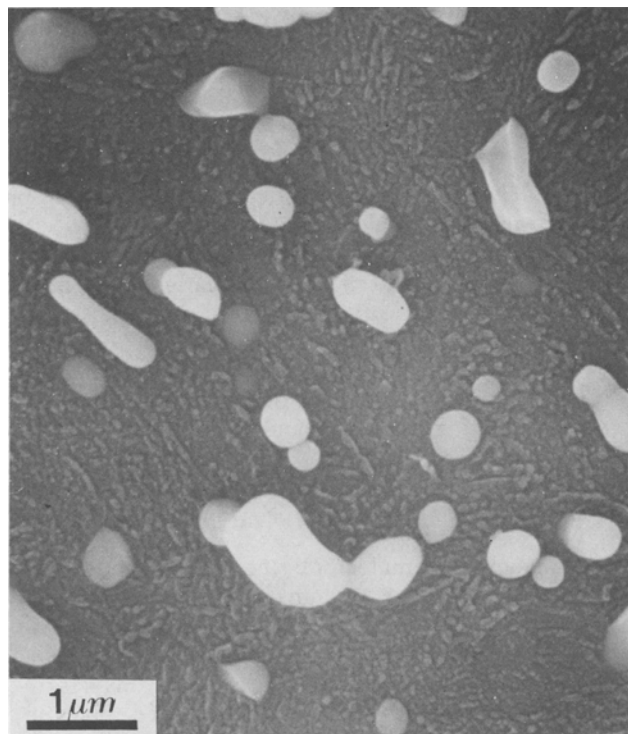


Fig. 1—Scanning electron micrograph of spheroidal cementite extracted from 52100 steel.

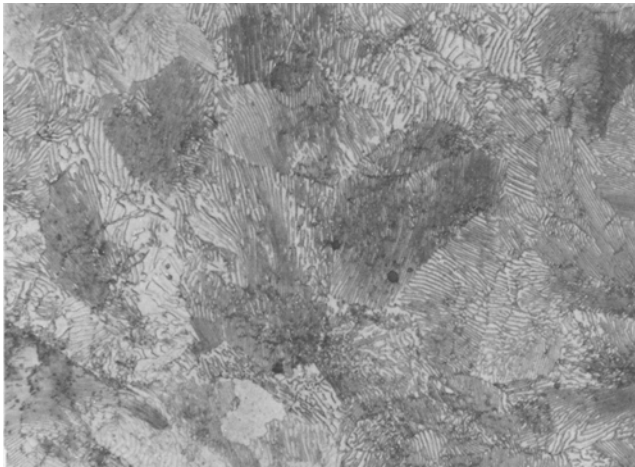
absorption corrections usually employed in bulk microprobe analysis. As cementite is stoichiometric with respect to carbon (6.67 wt pct carbon), the true weight percentage of chromium could be obtained by normalization to the 93.33 pct metallic constituent.

The influence of particle size in the analysis method was investigated. No trend toward lower or higher Cr contents was observed relative to carbide size. This observation is consistent with the results reported by Estay, Chengji, and Purdy for measurements on submicron sized carbides.¹²

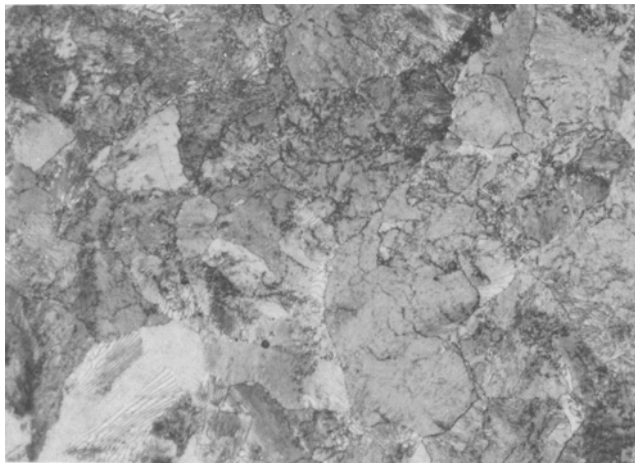
III. EXPERIMENTAL RESULTS

A. Annealing Response

The normalized microstructures of the three steels are shown in Figure 2. The applied normalization treatment was 1100 °C in a salt bath for 30 minutes, followed by cooling in still air to room temperature. A fine spheroidize-annealed structure was produced in the 1.42 pct Cr steel by austenitization at 780 °C for 30 minutes, furnace cooling to 735 °C, and cooling at a rate of 5 °C per hour to 685 °C. The zero chromium steel was spheroidized by austenitizing at 760 °C for 1 hour with subsequent cooling between 730 and 705 °C at 5 °C per hour. The 0.64Cr steel was given an identical cooling cycle but the initial austenitization treatment was 780 °C for 4 hours, the aim being to produce similar soft-annealed microstructures prior to subsequent study. This was not fully successful, because larger carbides were produced in the two low chromium experimental steels. The resultant microstructures are shown in Figure 3.



(a)

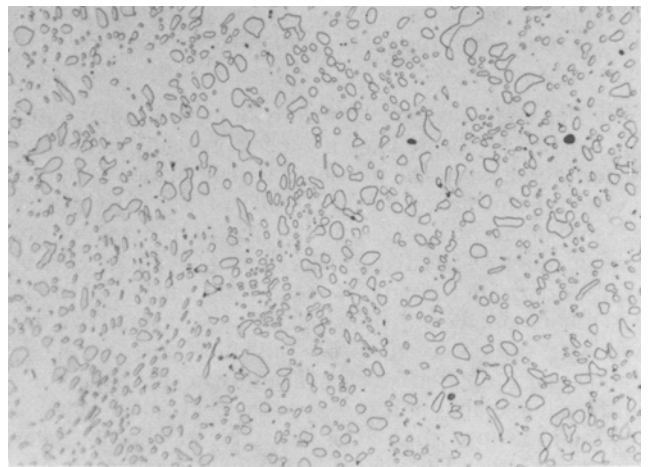


(b)

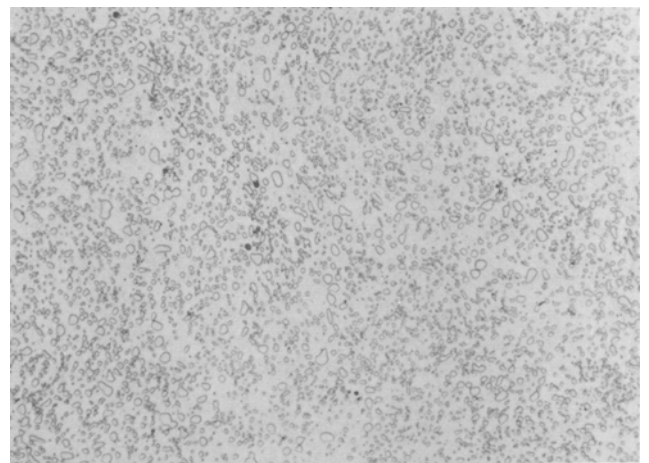


(c)

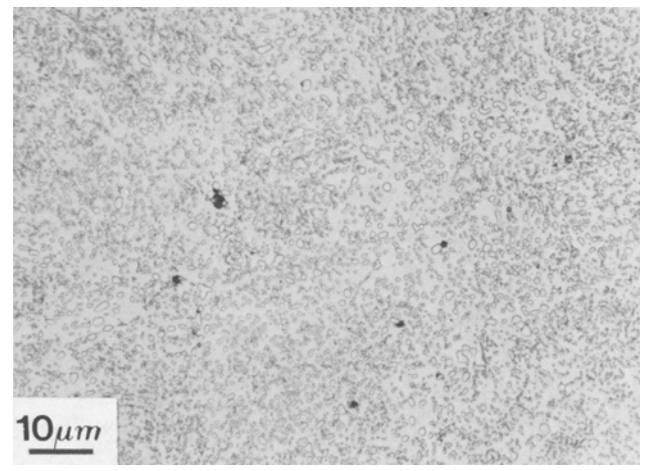
Fig. 2—Optical micrographs of normalized structures for (a) zero Cr, (b) 0.64 pct Cr, and (c) 1.42 pct Cr hypereutectoid steels (nital etch). Magnification 870 times.



(a)



(b)



(c)

Fig. 3—Spheroidize annealed microstructures for (a) zero Cr, (b) 0.64 pct Cr, and (c) 1.42 pct Cr steels containing 1 pct carbon. Magnification 850 times.

B. Influence of Chromium Content on the Ferrite to Austenite Phase Transformation under Continuous Heating Conditions

The ferrite-to-austenite transformation temperatures A_{c1} (A_{c1b}) and A_{c3} (A_{c1e}) are known to be appreciably influenced by heating rate.^{13,14} The measured transformation temperatures related to heating for the 52100 steel (1.42 wt pct Cr) are shown as a continuous heating time-temperature austenitization (TTA) diagram in Figure 4. The solid lines in the TTA diagram of Figure 4 indicate the rates used in the presently reported continuous heating experiments. In common with the observations by Schlicht,¹³ for continuous heating transformation experiments, no evidence supporting a diffusionless, allotropic ferrite-to-austenite transformation was found in this study even at the highest heating rates employed. At heating rates of 100 to 200 °C per second the start of the transformation was ill-defined due to poor demarcation in the dilatation/temperature curves. The solid curve with the faster rate (up to 950 °C) in Figure 4 is designated "fast heating rate" and the lower solid curve (up to 850 °C), the "slow heating rate". As shown in Figures 5(a) and (b), chromium significantly influences the start and finish temperatures of the ferrite-to-austenite phase transformation and widens the ferrite + austenite + M_3C phase field in these relatively low alloy steels.

C. Carbide Dissolution and Martensite Start Temperature Related to Chromium Content under Continuous Heating Conditions

The "slow" and "fast" heating rates, previously described, were used to determine the carbon matrix en-

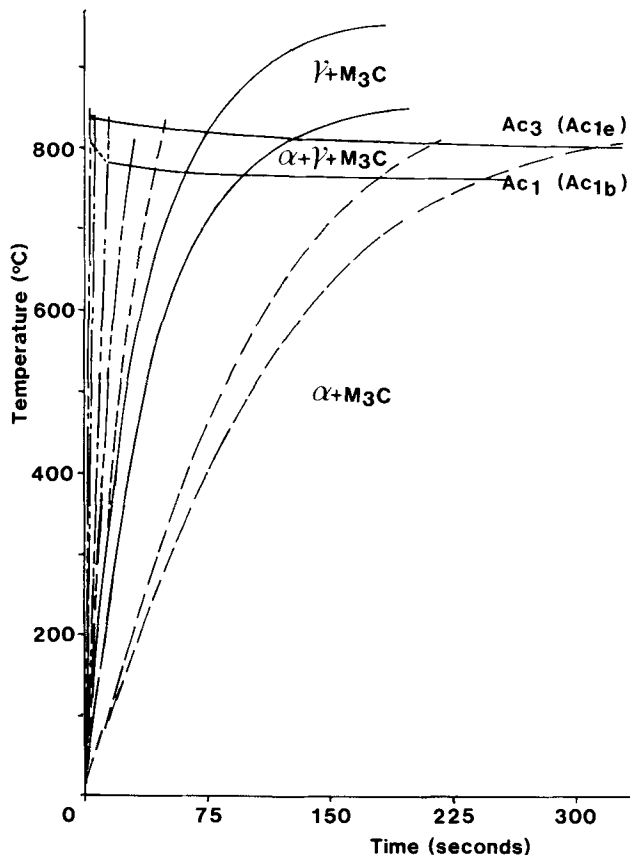


Fig. 4—Time-temperature-austenitization (TTA) diagram for 52100 steel with a spheroidize-annealed prior structure.

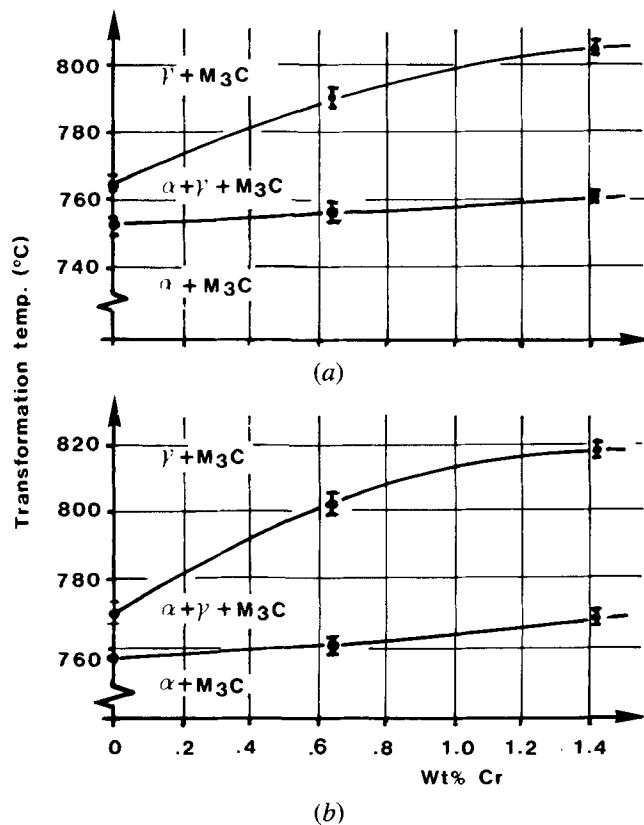


Fig. 5—Ferrite-to-austenite transformation temperature related to total chromium content (a) "slow heating rate", (b) "fast heating rate" (error bars show 68 pct confidence limits).

richment and related M_s temperatures after quenching from various temperatures. The matrix enrichment in carbon was determined from measurements of the area fraction of undissolved spheroidal carbide persisting after martensitic quenching from various temperatures above the A_{c1} temperature. For the two heating rates studied, mean matrix carbon contents and M_s temperatures vs time are shown in Figures 6 and 7. The error bars show the 68 pct confidence limits for the mean of 6 determinations of the matrix carbon content. In the case of the M_s determinations, only the mean values are plotted since the scatter in the measurement was small. The carbide dissolution with related depression in the M_s temperatures is significantly influenced by the chromium content in these steels under the continuous heating conditions studied. The lower Cr steel showed significantly faster rates of carbide dissolution with resultant lowering of the M_s temperatures upon quenching.

D. Carbide Dissolution, Austenite Grain Growth, and Carbide Chromium Contents for Isothermal Austenitization of 52100 Steel

Measurements of the rate of matrix carbon enrichment under conditions of isothermal austenitization at 850 and 950 °C are shown in Figures 8(a) and (b). The values from the Orlich¹⁴ study are included in these plots, after calculation of the matrix carbon contents from the published carbide volume fractions. The 850 °C: 30 minutes data point from Nunomura *et al.*¹⁵ is also shown in Figure 8(a), and the data points from Yajima *et al.*¹⁶ are shown in Figures 8(a) and (b).

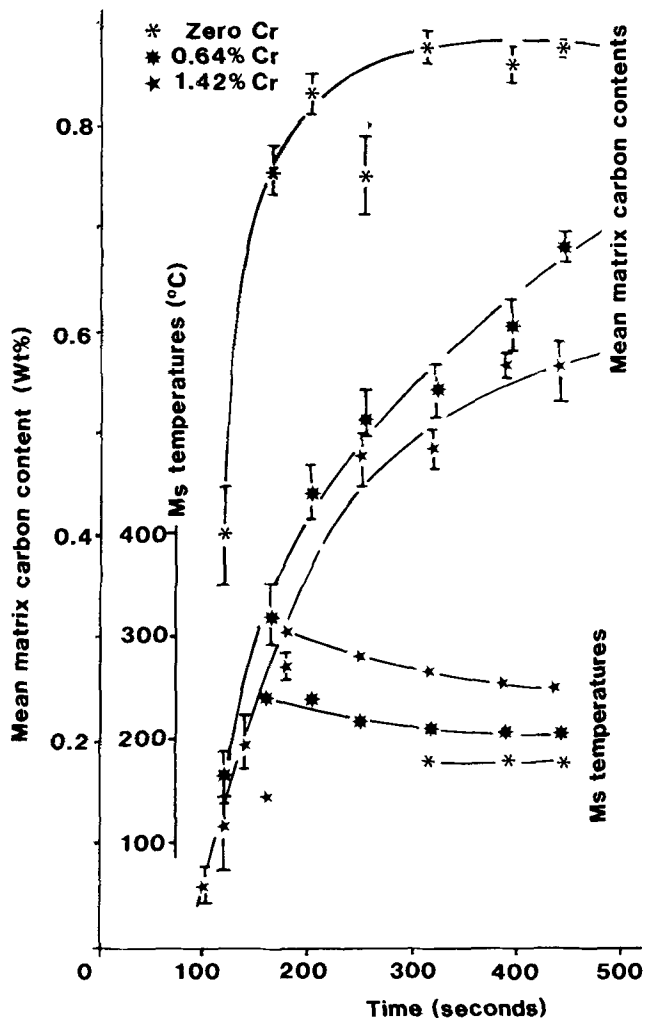


Fig. 6—Mean matrix carbon contents during “slow heating” to 850 °C and M_s temperatures upon incremental quenching for zero, 0.64, and 1.42 pct Cr steels containing 1 pct C (error bars show 68 pct confidence limits)

Plots of prior austenite grain size as a function of isothermal austenitization of 820, 850, 875, 900, and 950 °C are shown in Figure 9. With reference to the previously shown (Figure 8) carbide dissolution behavior related to temperature and time, it is apparent that significant austenite grain growth occurs only after appreciable carbide dissolution. A semi-quantitative relationship of total carbon and chromium content related to prior austenite grain size for martensitic-hardened 52100 (850 °C: 15 minutes from a soft annealed prior structure) was reported in a previous publication.¹¹ The total chromium content was observed to influence the austenite grain size stability to the square power compared with a linear effect of carbon content on grain size.

The chromium content of the residual carbides during 820, 850, 900, and 950 °C isothermal austenitization was determined using an extraction replica and electron microprobe analysis technique; see Figure 10. The distribution in the measured values for 10 minutes austenitization at 850 °C and quenching, against the distribution in the measured values in the soft annealed condition are shown in Figure 11. The median value for the ‘as spheroidize annealed’ condition was approximately 7 wt pct Cr and in the

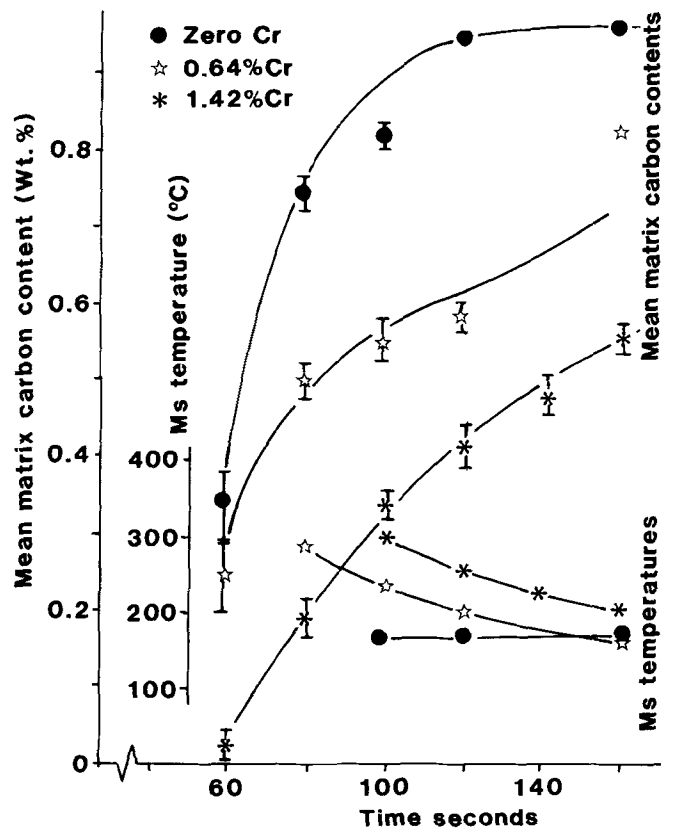


Fig. 7—Mean matrix carbon contents during “fast heating” to 950 °C and M_s temperatures upon incremental quenching for zero, 0.64, 1.42 pct Cr steels containing 1 pct C (error bars show 68 pct confidence limits).

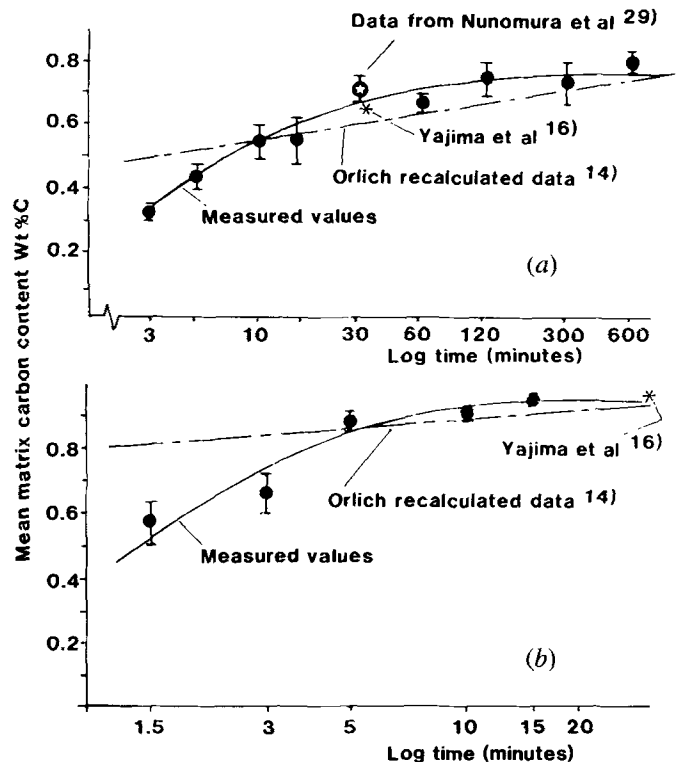


Fig. 8—Matrix carbon enrichment in 52100 steel after isothermal austenitization temperatures at (a) 850 °C and (b) 950 °C (error bars show 95 pct confidence limits unless stated otherwise).

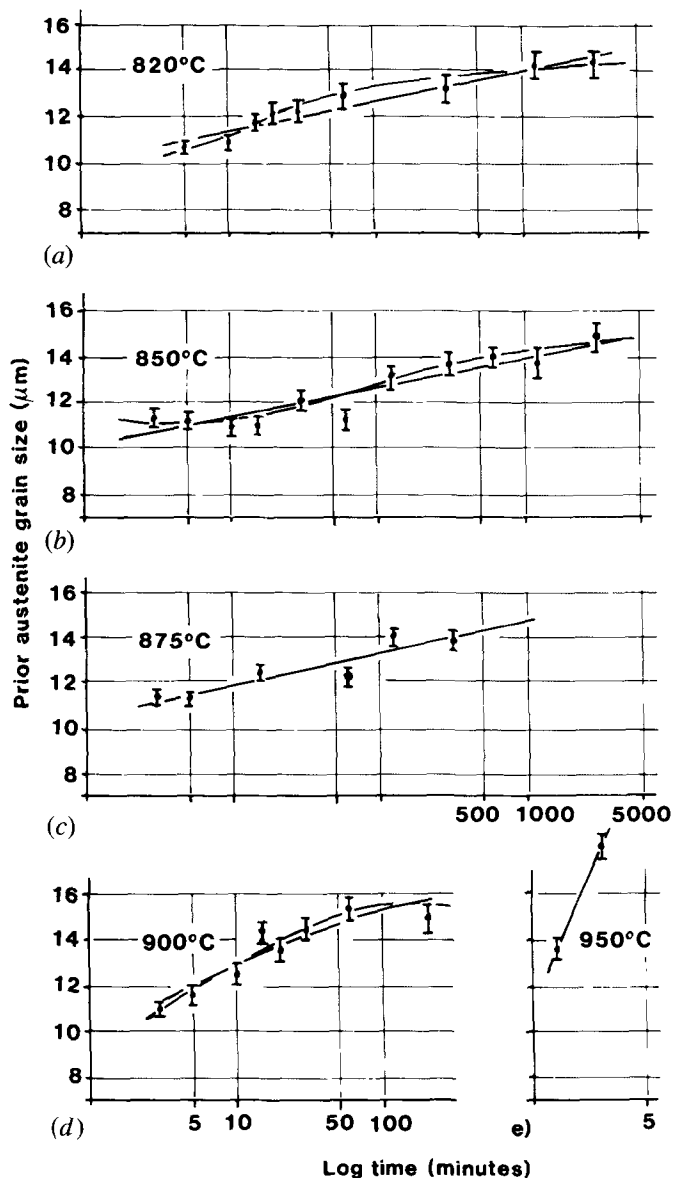


Fig. 9—Prior austenite grain size for isothermal austenitization at (a) 820, (b) 850, (c) 875, (d) 900, and (e) 950 °C of 52100 steel with a soft annealed prior microstructure.

hardened condition, 9 wt pct Cr. From the 396 individual chromium content determinations of the carbide microchemistry in the soft annealed 52100 steel, no “low” values in the range of 2 wt pct chromium were observed. “Low” values would have been measured if the low carbon secondary carbides reported by Glowacki and Barbacki,¹⁷ and Stickels¹⁸ had persisted after the applied spheroidization treatment.

E. Phase Field and Chromium Content of Secondary Carbides in 52100

The reaction time for the formation of cementite allotriomorphs in 52100 steel, austenitized at 1050 °C for 15 minutes and isothermally transformed at various temperatures is shown in Figure 12. The measured reaction time from Ando and Krauss¹⁹ is also shown in Figure 12. The chromium content of the secondary carbides was determined

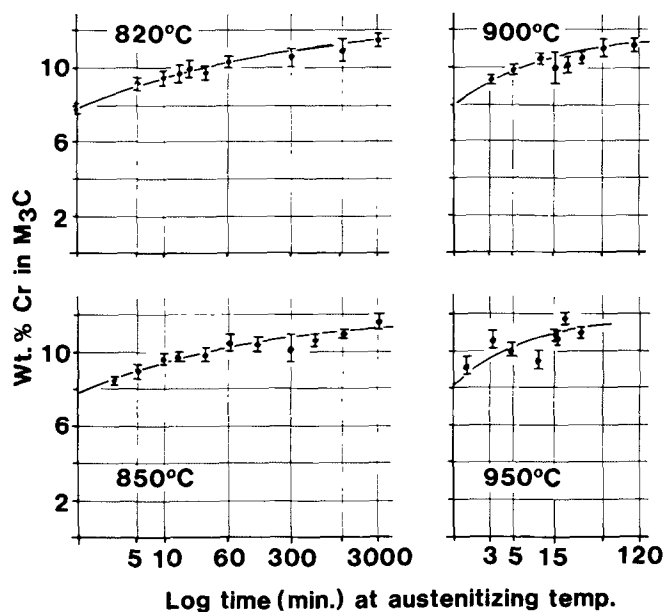


Fig. 10—Chromium content of residual carbide for isothermal austenitization of 52100 steel.

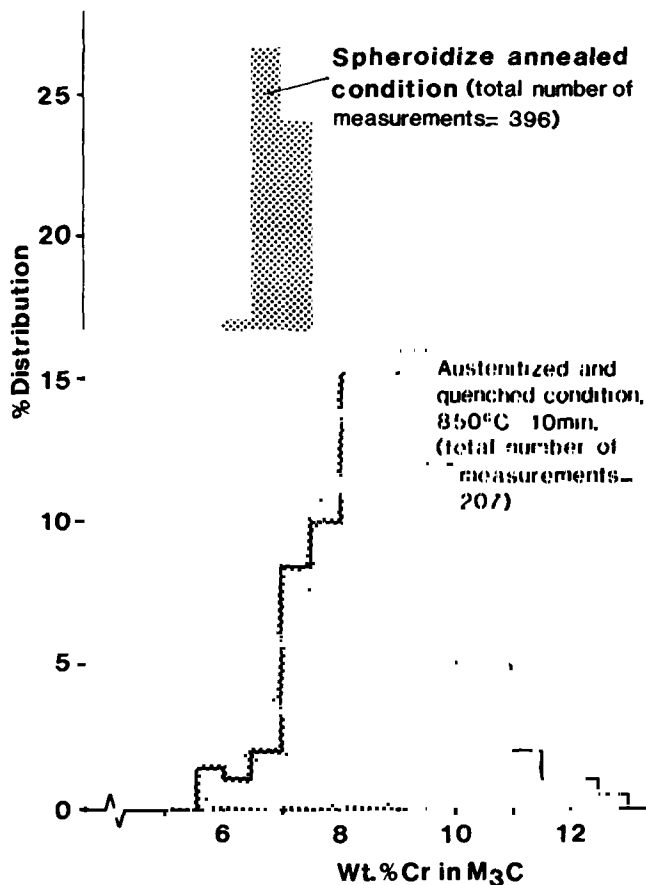


Fig. 11—Distributions in measured values for chromium content of spheroidal carbides in soft annealed and hardened (850 °C, 10 min austenitization) conditions in 52100 steel.

by direct measurement in the scanning electron microanalyzer from thick regions on the nital etched metallographic sections; see Figure 13. The results of these measurements are plotted in Figure 14 as a function of chro-

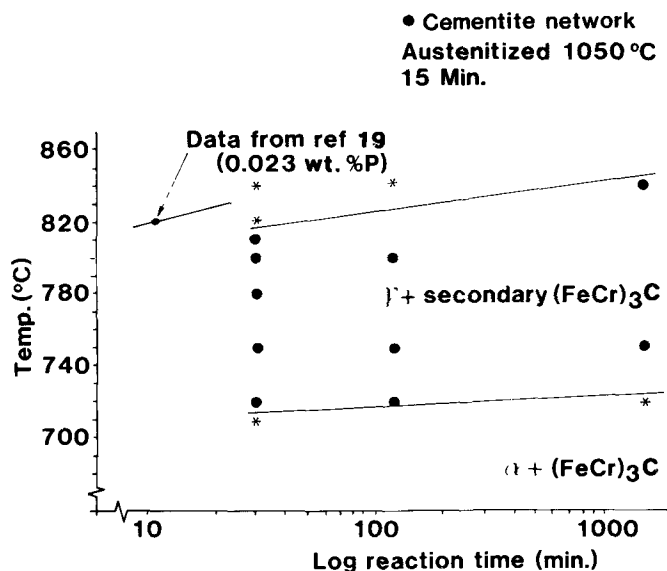


Fig. 12— Isothermal reaction time for cementite allotriomorphs in 52100 steel after austenitization at 1050 °C.

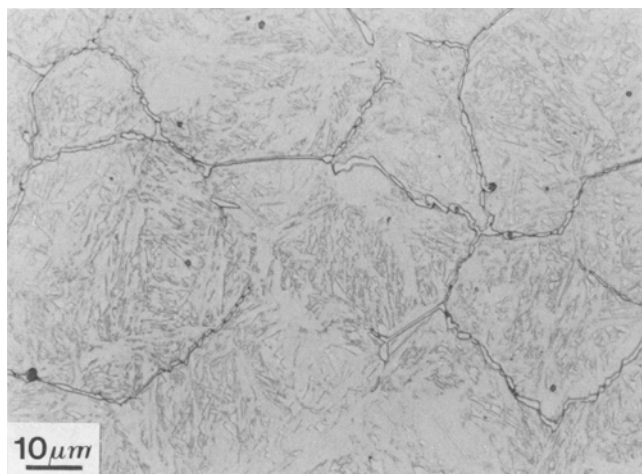


Fig. 13— Optical micrograph of secondary carbide formation in 52100 steel, nital etch.

mium content, transformation temperature, and reaction time. Chromium compositions approaching those of spheroidized annealed carbides were observed for the secondary carbides exposed to extended periods in the high temperature range of the secondary carbide phase field. For short reaction periods, chromium contents of 2 to 3 wt pct were observed which is consistent with the values reported by Glowacki and Barbacki.¹⁷ The chromium enrichment of secondary carbides, in the temperature range 720 to 800 °C, is very sluggish compared to that observed for the spheroidal carbides during austenitization in the temperature range of 820 to 950 °C. This temperature dependence is reconcilable with diffusion-controlled behavior. Upon spheroidization, the chromium content of the secondary carbides attains the same composition as the newly formed M_3C carbides during spheroidization. As stated earlier, no (secondary) carbides with low chromium contents were observed during extensive microchemistry determinations of the carbides in the soft annealed condition of the 52100 steel.

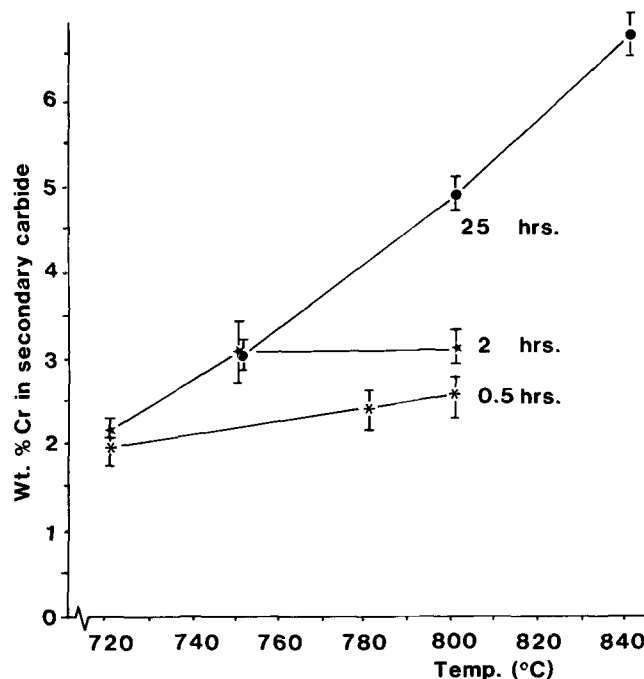


Fig. 14— Secondary carbide chromium content as a function of time and temperature of isothermal transformation.

F. Austenite Homogeneity

Conventionally austenitized and martensitically quenched 52100 steel has a microstructure consisting predominately of fine, twinned, plate martensite with some regions of lath martensite; see Figure 15. Specimens rapidly austenitized to produce an average matrix carbon content of 0.55 wt pct and quenched exhibited “ghost-type” structures; see Figure 16. These structures are common in induction hardened 52100 steel. From transmission electron microscopy the “white ghost” regions were observed to be free of spheroidal carbides. The more highly-alloyed regions account for the lighter-etching characteristics when using nital etching for light optical microscopy examination. No evidence was found by either light or electron optical techniques for a diffuse chromium-rich zone, due to rapid heating, sur-

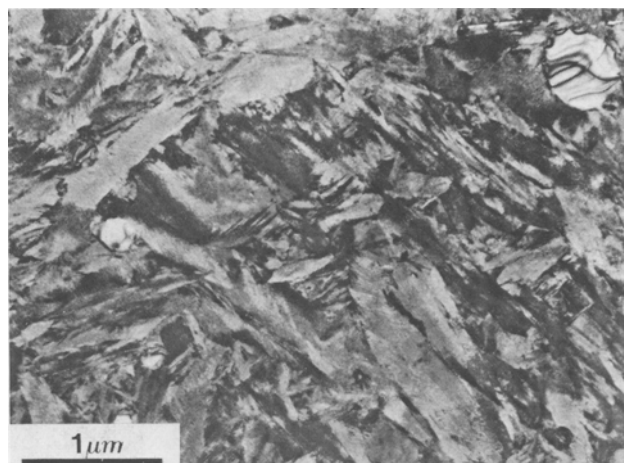


Fig. 15— Transmission electron micrograph of conventionally martensitic hardened 52100 ball bearing steel (850 °C; 15 min soak).

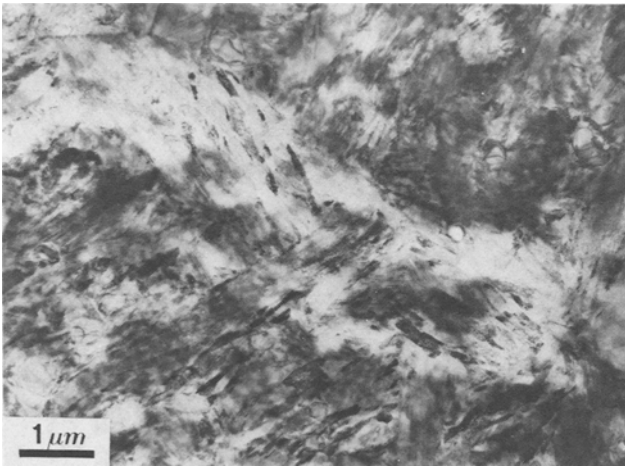


Fig. 16—Transmission electron micrograph of 'ghost structure' typically observed in rapidly austenitized and martensitically quenched 52100 steel.

rounding the residual carbides in 52100, as reported by Kinoshi and Furumura.²⁰ Soaking is common practice in heat treatment although, in many instances, unnecessary allowances are made for section thickness in steels.²¹ Soaking time is very often confused with heat-up time. Soaking the 52100 steel used in this investigation for 15 minutes at 850 °C produced no slack-quench constituents, such as upper bainite and/or fine pearlite, on quenching in the dilatometer experiments. The same steel, rapidly austenitized, using continuously increasing heating rates up to 1050 °C to produce the same average matrix carbon content of 0.55 wt pct, resulted in as much as 2.22 area percentage of slack-quench constituent using the same quenching conditions in the dilatometer. It can therefore be confirmed that in order to attain the same fully martensitic structures, "overheating" is necessary for rapid austenitization heating conditions for hardening 52100 steels as compared to 'soaking' at the austenitizing temperature.

IV. DISCUSSION

It has been demonstrated that increased chromium contents in hypereutectoid steels have a significant influence on a number of parameters, one of these being the carbide size resulting from spheroidization. The fine spheroidal structures in steels with higher chromium results from their finer pearlite lamellae.²² A study of Oyama *et al.*²³ on ultra-high carbon steels showed that the addition of 1.5 pct chromium produced a divorced eutectoid transformation upon isothermal soaking in the ferrite-austenite-cementite, three phase region. Fine spheroidal cementite was formed in 1.5 pct chromium steel, whereas an equivalent chromium-free steel produced cementite lamellae. A finer carbide distribution in the prior soft annealed structure results in not only finer residual carbides after hardening but also a finer martensitic structure. Finer carbides are reported to be favorable as regards rolling contact fatigue life in 52100; see Monma *et al.*,²⁴ Faunce and Justusson,²⁵ Luty,²⁶ and Meredith.²⁷ Fine carbide structures are beneficial from the point of view of accelerating the ferrite-to-austenite phase transformation, and improving the austenite homogeneity¹³ by shortening

the carbon diffusion path. Finer spheroidized annealed structures are produced by the use of lower temperature soft annealing cycles;²⁸ however, as reported by Hewitt,²⁹ the mechanism of carbide spheroidization is complex and cannot be rationalized in terms of temperature alone.

The second and more important influence which Cr has in hypereutectoid steels is its influence on carbide stability during austenitization. The observed systematic increase in the ferrite-to-austenite transformation temperatures with increased Cr contents was also reported by Kelley.³⁰ Kelley, however, did not observe a widening in the ferrite-austenite- M_3C phase field with Cr contents between zero and 1.3 wt pct, an observation noted in this study. This discrepancy could, however, be due to the use of dissimilar base composition in the study by Kelley, *i.e.*, 0.62 pct C, 1.7 pct Ni, and 0.75 pct Mo.

The observed chromium enrichment of the undissolved carbides during austenitization is significant and follows the behavior expected from the carbide-forming nature of chromium. Glowacki and Weirszylowski³¹ also showed Cr enrichment of M_3C for 52100 steel austenitized in the range 820 to 850 °C, but Cr depletion of the M_3C after austenitization at 950 °C. The diffusion of carbon from the carbide/matrix interface will be reduced significantly due to the relatively slow diffusion of chromium at the interface. Some chromium obviously diffuses toward the center of the $(FeCr)_3C$ particles as the average matrix carbon and chromium content enriches due to carbide dissolution. The smaller and/or boundary carbides probably dissolve in preference to the larger carbides within grains. During austenitization, the stability of the undissolved spheroidal carbides increases due to the chromium enrichment, and an equilibrium develops between carbide dissolution and matrix chromium dilution.

The carbide dissolution behavior of 52100 steel during isothermal austenitization has also been studied by Orlich,¹⁴ Peilloud and Dudragne,³² and Nunomura *et al.*¹⁵ Good agreement was observed between their data and the results of this study within the 95 pct confidence limits of the mean values for the longer transformation times. For the shorter transformation periods, no agreement was observed for the linear carbon and logarithmic time relationship shown in the Orlich study,¹⁴ and poor agreement over the whole transformation range was observed with the Peilloud and Dudragne study.³² As reported previously,¹¹ the heat treatment response of 52100 is influenced significantly by the prior microstructure, and this could account for the dissimilarities in the carbide dissolution rates among the four studies, especially for short transformation times.

The observed enrichment in $(FeCr)_3C$ with isothermal austenitization implies that the matrix is diluted in chromium with extended soaking time which would result in decreased hardenability. However, by relating the isothermal carbide dissolution behavior to the chromium content of the carbides and matrix, an equilibrium balance between matrix chromium and matrix carbon content was observed, independent of the austenitization conditions, as shown in Figure 17. The presence of stable carbides makes the heat treatment of 52100 steel relatively insensitive to austenitization time at austenitization temperatures in the range of 780 to 860 °C. The degree of stability is dependent upon the microchemistry of the cementite. According to

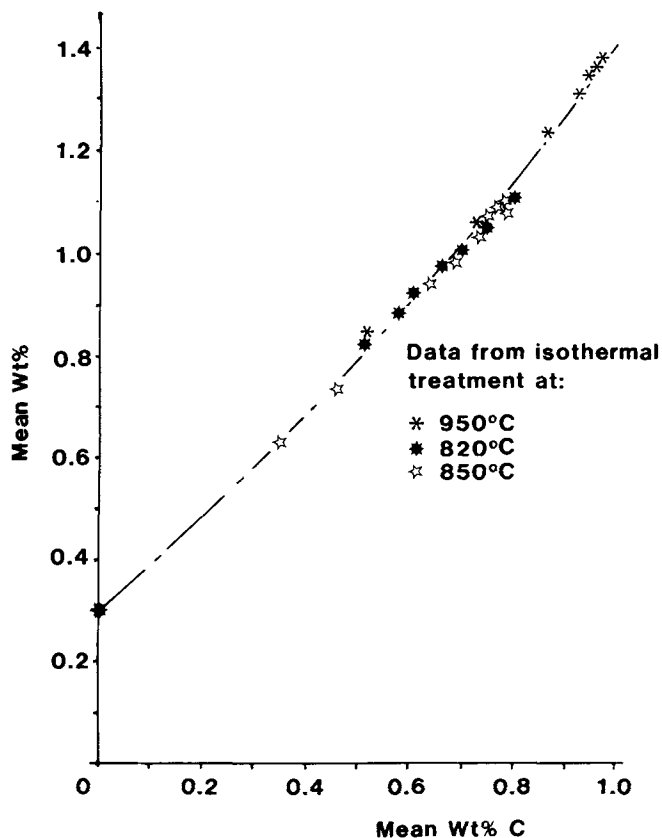


Fig. 17—Ratio of matrix carbon and chromium contents in 52100 steel during isothermal austenitization.

Woodyatt and Krauss,³³ cementite can accommodate up to 18.1 wt pct of chromium. However, the actual chromium content of the M_3C in 52100 steel is reported by Glowacki and Barbacki¹⁷ to be dependent on the matrix in which it was formed, *i.e.*, ferrite or austenite. The secondary carbide, formed from austenite, was reported to contain 2 wt pct Cr. Stickels¹⁸ also reported some difference in the cementite chromium content for carbides in equilibrium in ferrite or austenite.

The reported lower chromium content for carbides in equilibrium with austenite is surprising in view of diffusion behavior for the metallic constituent of cementite, reported by Hillert *et al.*³⁴ For alloyed cementite the rate of carbon diffusion is controlled by the rate of carbon diffusion through the austenite-carbide interface. The diffusion path of chromium is either into the matrix or toward the center of the cementite particle, chromium having a preference for carbide formation under equilibrium conditions. However, under practical treatment conditions, *i.e.*, continuous or rapid austenitization, this may not occur. It has been demonstrated that at low reaction temperatures secondary carbides can have lower Cr contents in 52100. However, after spheroidization, which is normally applied for machinability reasons, the secondary carbides attain the composition of the newly-formed carbides and no austenite inhomogeneity results during subsequent austenitization for hardening.

The amount of equilibrium chromium partitioning is known to be affected by the phosphorus content, the influence of phosphorus on the secondary carbide thickening

kinetics being reported by Ando and Krauss.³⁵ The reaction time observed in this study for secondary carbide formation did show some dissimilarity as compared to values shown in the literature. The different phosphorus contents employed in the various investigations could account for this. Secondary carbide phase formation has been shown by Ando and Krauss³⁵ to be controlled by lateral ledge migration on the partially coherent interface of the austenite boundary cementite. The partitioning of chromium makes the carbide thickening reaction sluggish, particularly as chromium enriches at the interface region in the later stages of isothermal transformation.

The dual diffusion path for Cr during carbide dissolution indicated by the results of this study followed the behavior predicted by Hillert³⁴ *et al.* for alloy carbides.

When considering the substitution of chromium in ball bearing steel, due regard should be paid to the full spectrum of requirements. The substitution should not only be based on requirements such as the attainment of equivalent through-hardenability or equivalent high stress level rolling contact fatigue performance. These factors are significant, but both spheroidization and hardening heat treatment response and consistency in machinability are just as paramount in ball bearing applications.

Substitution of Cr with other carbide forming elements should be made with care. Technically, Mo is a very good substitute for Cr but a very strong carbide former such as Ti is not. Even minute addition of titanium lowers the fatigue strength of 52100 steel due to the formation of brittle carbonitrides.³⁶ Vanadium additions have been suggested as also reducing fatigue strength in bearing steels,³⁷ but experience with valve spring steels³⁸ has not shown any effects to suggest a negative influence on fatigue properties. Manganese additions do not produce carbide stability and also tends to produce coarse, difficult-to-spheroidize, pearlite lamellae in hypereutectoid through hardening bearing steel composition.³⁹

A further requirement of a general through-hardening steel is its response to bainitic (isothermal) hardening. Bainitic hardening is applied for 52100 steel in preference to martensitic hardening for some applications.^{40,41} Manganese and particularly silicon are known to extend significantly the time required for transformation to lower bainite. Molybdenum and chromium have much less effect on the time required for the lower bainitic transformation, but are very effective in extending the pearlite and upper bainitic reaction times.

All of the alternative steels proposed therefore suffer from one or more undesirable features. It is clear that, based on the present understanding, the 52100 composition for through-hardened bearings is the best compromise in properties for a given through-hardening requirement. This is not to say, however, that the situation will always be the same. If chromium, for any reason, has to be substituted in ball bearing steel compositions in the future, the past research in this area has provided usable compositions for through-hardening bearing manufacture.

V. CONCLUSIONS

1. Chromium in the range zero to 1.42 wt pct in 1 pct carbon hypereutectoid steel increases the ferrite-to-austenite

- transformation temperature and widens the three phases ferrite + austenite + (FeCr)₃C region.
- The pearlite lamellae are finer in 1 pct C steel with increased Cr contents which results in finer carbides and improved spheroidization characteristics when compared to steels with increased Mn contents.
 - The chromium content in secondary carbides of 52100 is of the order of 2 to 3 wt pct for short reaction times. For long reaction times in the upper temperature range to secondary carbide, chromium content approaches that of the (FeCr)₃C produced by soft annealing, *i.e.*, 7 wt pct Cr.
 - The chromium content in the undissolved carbide in 52100 increases as carbide dissolution occurs during austenitization for hardening. An equilibrium is formed between the carbon and chromium matrix composition irrespective of the austenitization conditions.
 - Carbide stability is essential to produce the fine grained medium carbon martensites generally required for through-hardening ball bearing steels. Chromium is a relatively cheap effective element and should only be substituted by suitable carbide forming elements for through-hardening bearing steel applications.
 - The manufacturing aspects of a steel composition are vital for a successful through-hardening ball bearing steel composition. The 52100 steel composition is a good compromise between machinability, product reliability, cost, rolling contact fatigue, and structural performance for standard through-hardening bearing applications.
 - For heavy section through-hardened bearings, Mo and Mn additions should be made with a corresponding Cr balance to maintain the necessary carbide stability.

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