

Some Problems on Quenched Steels. III : On the Mechanism of Temper Hardening in Some Quenched Special Steels

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Some Problems on Quenched Steels. III On the Mechanism of Temper Hardening in Some Quenched Special Steels*

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Synopsis

The mechanism of temper hardening in Mn-steels and high C-Cr-steel was investigated by means of hardness, dilatometric and microscopic observations. When the retained austenite was tempered, carbides were precipitated from it, and the precipitation could be observed as the contraction in dilatometric curves and be conformed directly by microscopic analysis. The precipitation of carbide played an important rôle in the hardening, independent of A_r'' transformation. The γ - α transformation of retained austenite took place after some amount of carbide had been precipitated.

I. Introduction

The mechanism of temper hardening in a quenched high speed steel and that of the decomposition of residual austenite were reported in the previous paper.¹⁾ The present research was carried out to ascertain whether the author's theory on the decomposition of residual austenite in high speed steel is, in general, applicable to other steels or not. At first sight, some age-hardenable ferrous alloys of the type of α -phase, for example, Fe-W alloy, or those of the type of γ -phase, for example, age-hardenable 18-8 stainless steel may be considered to be relevant to the present work, but both are concerned merely with age-hardening phenomena. The temper hardening or the decomposition of retained austenite is a very complicated phenomenon, for the γ - α transformation is superposed on the precipitation of carbide. A large amount of austenite is retainable by quenching in steels containing a certain amount of Mn or Cr. So, with these steels, the author examined the mechanism of the decomposition of retained austenite by the same way as that reported in the previous paper¹⁾.

* The 646th report of the Research Institute for Iron, Steel and Other Metals.

This paper was read at the 15th meeting (1935) of Iron and Steel Institute of Japan.

(1) K. Monma, Sci. Rep. RITU, **3** (1951).

II. Hardness change by tempering

The compositions of the samples used in this research are shown in Table I. All the samples were melted in vacuum with high frequency furnace and cast in rectangular metal mould, 200 mm in length, and then forged and machined to the required dimensions. They were quenched in oil at 1000° and tempered at the required temperature for 30 min, and then Rockwell "C" hardness were measured.

Table I.

No.	Mn %	C %	Fe %
12	6.08	1.23	Bal
10	5.85	1.08	"
8	5.89	0.81	"
6	4.37	0.67	"
4	6.44	0.45	"
C-Cr	11.59%	2.17	"

Fig. 1 (A) show the results of the sample No. 12. The curve (a) is the case of successive tempering, and the curve (b) is that of the virgin sample at respective temperings. They are similar to each other, but the hardness in (b) is higher than that of (a) up to 400°, owing to the mass effect of the sample. As shown in the figure, no

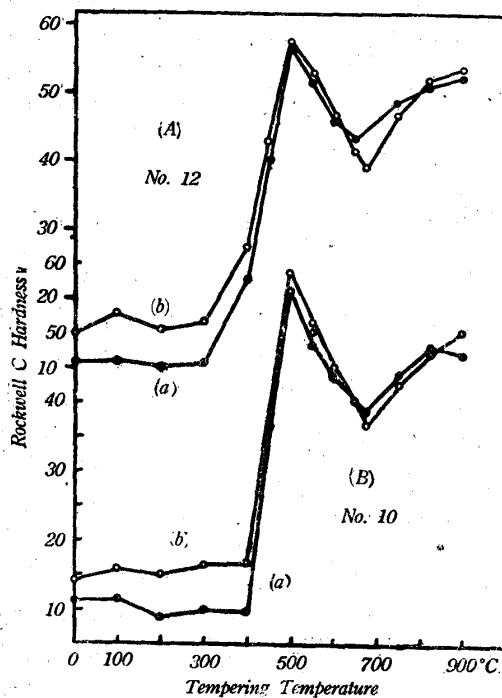


Fig. 1. Effect of Tempering on the Hardness of Quenched Mn-Steels

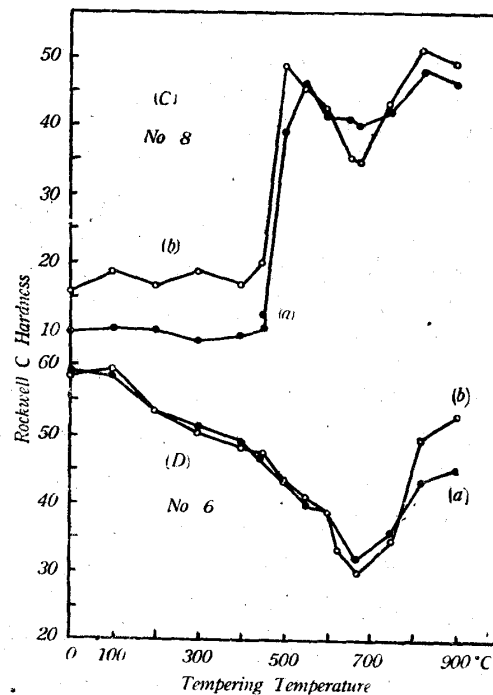


Fig. 2. Effect of Tempering on the Hardness of Quenched Mn-Steels.

change appeared till 300°, but in the temperature range of 400~500°, the hardness increased steeply. The curves (B) in Fig. 1 and (C) in Fig. 2 show the results of the samples No. 10 and No. 8 respectively, in which the curves (a) are the cases of

successive tempering, while the curves (b) are those of virgin samples at respective temperings. The sample No. 10 showed no change up to 400°, and in the tempering at temperatures higher than 450° the hardness increased steeply. In the case of No. 8, the hardness slightly increased by the tempering at 450°, and in the tempering at 500° its hardness increased steeply. Both results showed the maximum hardness at 500°. The curve (D) in Fig. 2 shows the results of the sample No. 6, in which the hardness was highest in the quenched state and decreased by tempering, except at 100°. No remarkable difference was observable between curves (a) and (b). The results of the sample No. 4 were similar to those of No. 6.

From these results it will be seen that the samples containing more than 0.8% carbon are very soft when quenched at 1000°. In fact their microscopic structures were wholly austenitic as shown in Fig. 7. The samples containing less than 0.6% carbon hardened to the maximum by quenching, and gradually softened by tempering as shown in curve (D) in Fig. 2. It is, therefore, certain that the temper hardening in curves (A), (B) in Fig. 1 and (C) in Fig. 2 had some relations to the decomposition of retained austenite and that the higher the carbon in the sample is, the lower the tempering temperature to increase the hardness becomes.

Fig. 3 shows the change in the hardness of high C-Cr-steel quenched at 1150° in oil and tempered. The details of experiments were quite similar to those in the former case of Mn-steel. As shown in the figure, in the tempering at the temperature lower than 450°, no change in hardness occurred, and in the tempering at 500°, the hardness was slightly increased, the maximum occurring at 550°, and in the tempering at the temperature higher than 550°, the samples softened steeply.

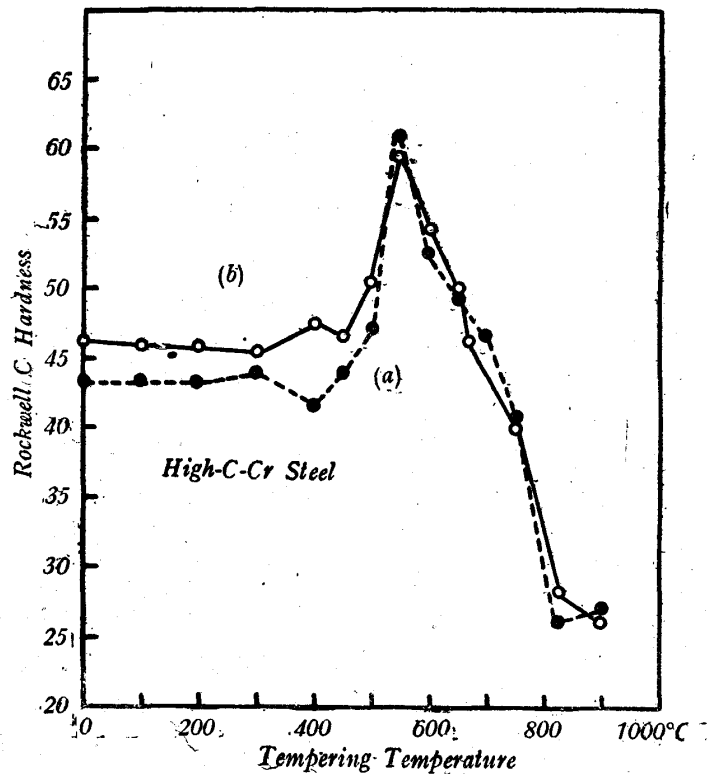


Fig. 3. Temper Hardness Change of High C-Cr-steel. (1150°O.Q.)

III. The dilatometric measurements

Figs. 4 and 5 show the results of dilatometric measurements of the steel No. 12 quenched in oil. The curve *a b c* in Fig. 4 is the case of the tempering at 395°, and is smooth both in heating and cooling, but, during 30 minutes that it was kept at 395°, it contracted somewhat, and so the cooling curve is slightly below the heating curve. The dilatometric curve for the temperature lower than 395° was

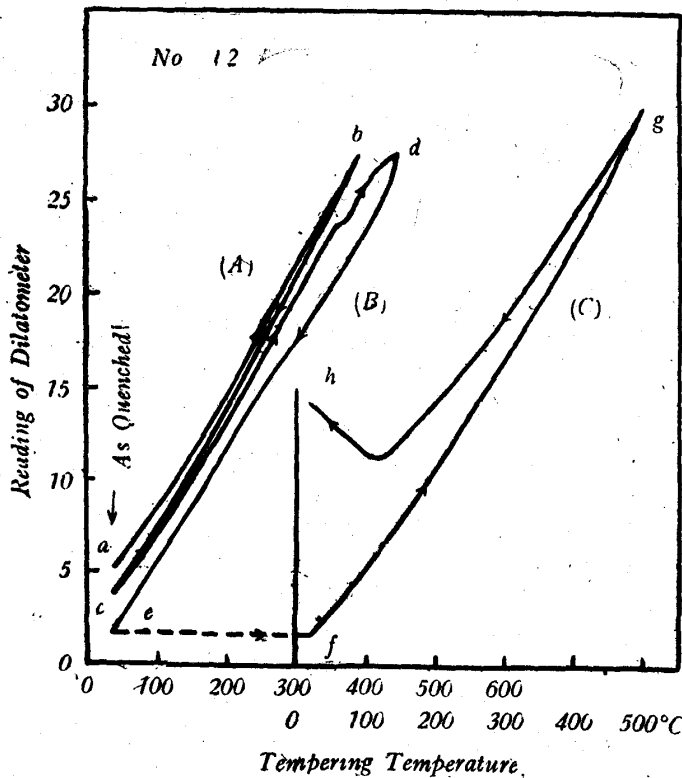


Fig. 4. Tempering Dilatometric Change in Quenched Sample No. 12. (1000° O.Q.)

the same as the heating curve for the tempering at 395°. The curve *c d e* is the case of the successive heating up to 450°, and is smooth till about 350°, and then is contract obviously at about 370°, and, during 30 minutes that it was kept at 450°, a slight contraction again took place. The cooling curve is smooth and below the heating curve. Corresponding to this curve the hardness was surely increased as shown by the curve (A) in Fig. 1. The curve *f g h* is the case of the successive tempering at 500°. As

in the former example, the heating curve is smooth and the cooling curve is above the heating curve in the range 500°~200°, and an expansion begins to appear at about 200°, and it takes place abruptly at 125°, that is, by heating at 500°, the retained austenite came to change into martensite in cooling.

The curves in Fig. 5 represent the dilatometric experiments on the sample No. 12 tempered at temperatures higher than 525°. All the curves are smooth till 400° in heating, but an obvious contraction occurs in the range 400~500°, and the heating at the temperatures above 500° caused distinct expansion. In the case of the tempering at 525° (curve (A)), the expansion at temperatures above 500° was small, and so the cooling curve came slightly below the heating curve, and an abrupt expansion due to the martensitization of the retained austenite appeared at 120°. In the case of the tempering at 565°, as shown in curve (B),

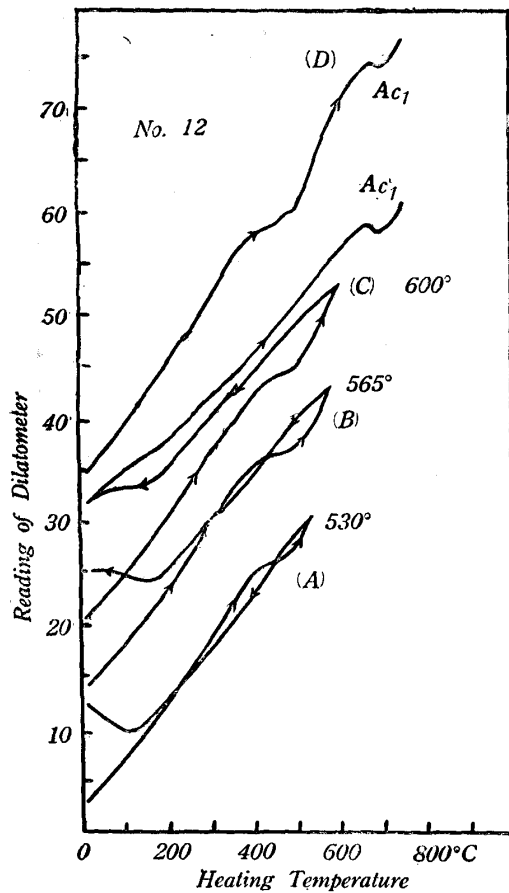


Fig. 5. Tempering Dilatometric Curves in Quenched No. 12 Samples. (1000° O.Q.)

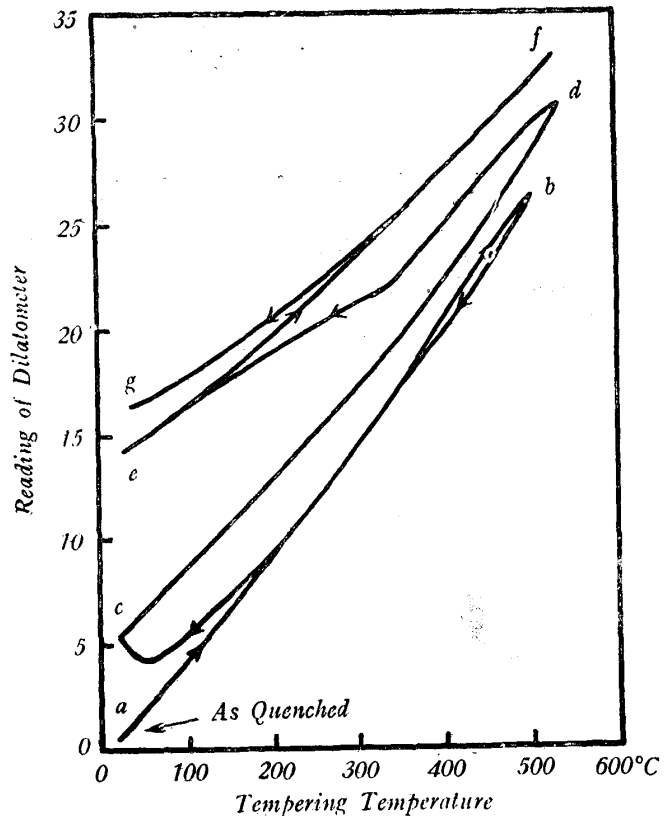


Fig. 6. Tempering Dilatometric Curves in High-C-Cr-Steel. (1150° O.Q.)

the cooling curve was above the heating curve, because the tempering temperature was higher than that in the former case, and the Ar'' expansion became small. The curve (C) represents the result of the virgin sample heated up to 600°. In this case, a large expansion appeared in the range 500~600°, and consequently, the cooling curve was far above the heating curve, and the Ar'' expansion was very small. Successive heating of the same sample up to 750° showed no change except A_3 transformation. The curve (D) represents the results of the virgin sample heated up to 750°, for which there will be no explanation necessary. The results for the samples No. 10 and No. 8 were almost similar to that of the sample No. 12.

The dilatometric measurements on a high C-Cr steel are shown in Fig. 6. The heating curve of quenched sample was smooth till 400°, and, in the range 400~500°, it slightly contracted with the rise of temperature as shown in the curve and so the cooling curve was slightly below the heating curve. The cooling curve was smooth till about 60°, at which the Ar'' expansion occurred. On successive heating up to 540°, the heating curve remained smooth, but the cooling curve was above the heating curve, and this fact would show that the Ar' transformation or the decomposition of the retained austenite occurred in heating. In cooling the Ar''

expansion occurred at about 330° . In the successive heating up to 540° , no remarkable change was observable. Comparing these results with those of the manganese steel shown in Fig. 4 it will be seen that the features of curves are quite similar to one another.

IV. Microscopic examination and consideration of results

Photo. (A) in Fig. 7 represents the microstructure of the sample No. 12 quenched at 1000° in oil. It was wholly austenitic structure. Photo. (B) is the structure of the same sample when tempered at 450° for 30 mins. This structure corresponds

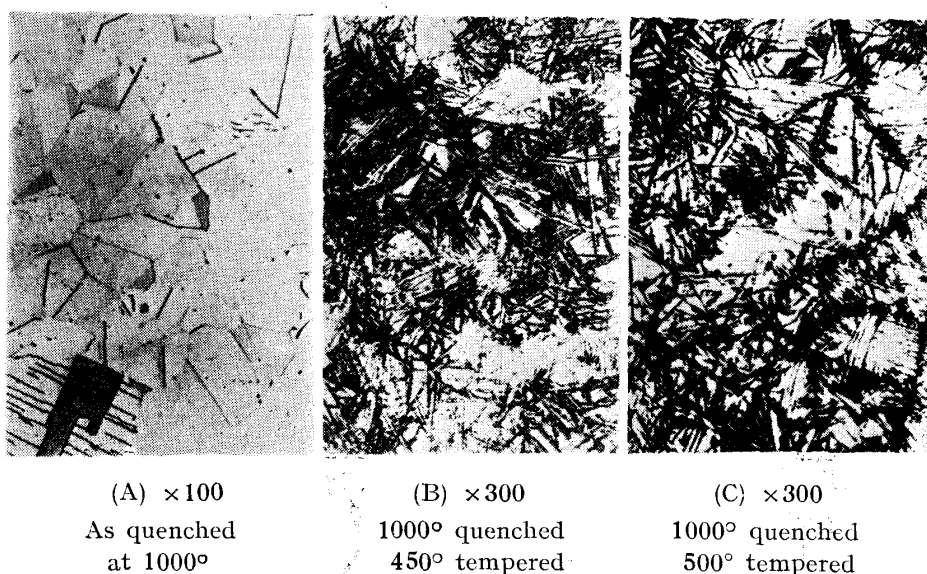


Fig. 7. Microphotographs of Mn-steel No. 12

to the case of the abnormal contraction in the heating curve shown in Fig. 4, and to the case in which the hardness was remarkably increased in spite of Ar'' transformation. Numerous needle-like crystals of carbide could be seen in austenite matrix. Photo. (C) is the microstructure of the same sample when tempered at 500° , which corresponds to the dilatometric curve (C) in Fig. 4. In this case, the precipitated carbides could also be seen as in the former case, but the ground mass had transformed into martensite, and, moreover, some network troostite was visible. This troostite was produced by the Ar' transformation at about 500° , which corresponded to the higher position of the cooling curve than the heating curve in Fig. 4.

From the above results, it may be concluded that the temper hardening of manganese steel is caused not only by the martensitization of retained austenite and that the precipitation of carbide plays also an important rôle in it. Next, the mechanism of the decomposition of retained austenite can be explained as follows: the retained austenite precipitates carbide on tempering and becomes

unstable after the precipitation of a certain amount, and finally the γ - α transformation takes place. This process of the decomposition is quite similar to that in a high speed steel previously reported.¹⁾ In the present case of a Mn-steel, the carbide precipitation was ascertained from the dilatometric curve and microphotographs, whereas, in the case of a high speed steel its direct confirmation failed.

In the case of a high C-Cr-steel, the mechanism of temper hardening can be explained as follows: comparing two curves in Figs. 3 and 6 with each other, it will be seen that the tempering of 500°, the hardness was somewhat increased, and the dilatometric curve showed a small Ar'' expansion in cooling; hence, the cause of the temper hardening may be considered as being due to the martensitization of the retained austenite. But, considering the features of the dilatometric curves which were indicative of the fact that the martensitization occurred only in the cooling course of tempering, it will rather be natural to conclude that a certain amount of carbide first precipitate from austenite by the tempering at 500° allowing the austenite to become unstable, and then the γ - α transformation comes to occur in the course of cooling. Accordingly, the mechanism of the temper hardening will be the same as that in Mn-steel.

If the above-stated mechanism of the decomposition of retained austenite is valid, the same should be applicable to the decomposition of austenite in the course of cooling from temperature above A_3 point. From such a view point, the author, using some Mn-steels, obtained cooling curves by dilatometer. Fig. 8 represents the results of various Mn-steels cooled in furnace at 900°.

The curves of the low carbon samples, Nos. 8, 6 and 4, were all smooth and Ar'' transformation occurred at about 100°, which was very small in the case of No. 8. In the case of the samples 12 and 10, a large contraction due to carbide precipitation appeared in the temperature range of 550~600°, and Ar'' expansion occurred at about 170° and 120°, respectively. These samples contained much carbide dissolved in austenite, and so the carbide was precipitated to some extent in the course of cooling. This was the reason why the contraction appeared in the curves of high carbon samples. Austenites from which carbide has been sufficiently precipitated become unstable and will change into α -phase or into troostite at that temperature, and the expansion due to troostite formation will be cancelled by the contraction due to carbide precipitation; other austenites from which carbide has been precipitated to a certain extent but insufficient for the occurrence of Ar'' transformation, will change into martensite at relatively low temperatures. Austenite without carbide precipitation will remain as it is down to room temperature. In the case of the sample No. 8 the carbon content was moderate and so the tendency of a carbide precipitation was not sufficient to decrease the carbon content in austenite, and, hence, the structure could not transform into martensite and remained as austenite.

From the above consideration the cooling curves in Fig. 8 can be explained

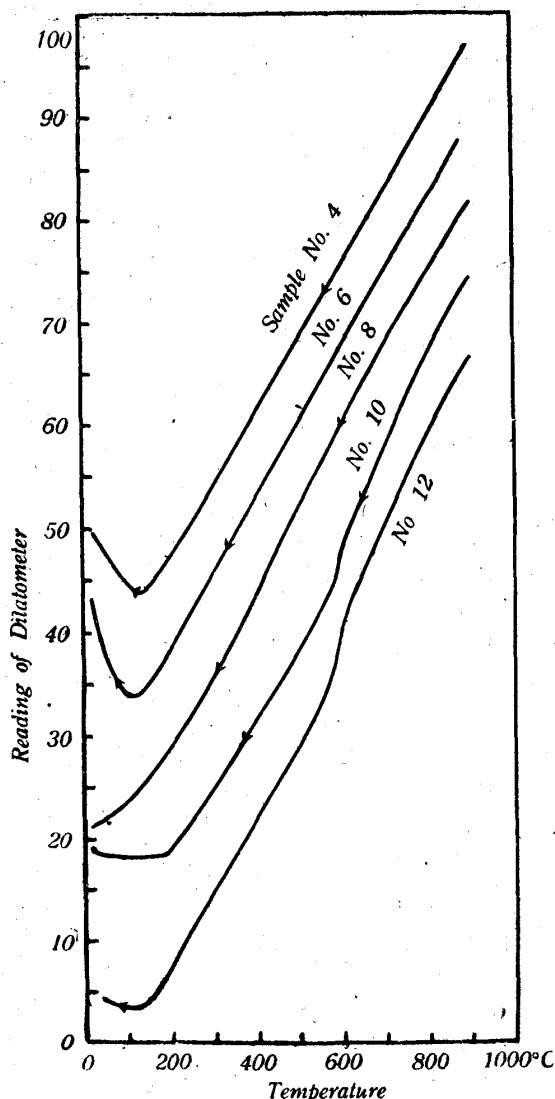


Fig. 8. Dilatometric Cooling Curves in Manganese Steels.

fall of temperature and it scarcely takes place in the vicinity of room temperature. Hence, it follows that the transformation of austenite will be suppressed when the cooling is too quick to perform the diffusion, provided that the sample contains a sufficient amount of carbide; in other words, what is suppressed by a quick cooling is not the lattice transformation of the γ -phase but the precipitation of carbide dissolved in austenite. Thus, the reason why the high carbon Mn-steel retains austenite at relatively slow rate of cooling as shown in Fig. 8 will be understood.

Summary

The results of the present investigation may be summarized as follows:

- (1) The quenched Mn-steel of high carbon content was hardened by tempering,

smoothly. Thus, it can be ascertained that the author's theory on the decomposition of retained austenite is, in general, applicable to the mechanism of the decomposition of austenite.

In addition to these, the reason why austenite is retained will briefly be mentioned. From the results of the above experiments, it may be said that the retainableness of austenite is not due merely to the quick rate of cooling compared with of transformation. As repeatedly stated, the transformation of austenite is directly related to the precipitation of carbide dissolved in it and takes place only when the amount of dissolved carbide has become proper. In case the carbide content is small from the outset, the austenite will easily be transformed into martensite or into troostite by quenching. The precipitation itself, however, is, phenomenon accompanied by diffusion of atoms and the time necessary for its completion becomes long with the

the higher the carbon content was, the lower the temperature at which the hardening occurred was. The maximum hardness could be obtained by tempering at 500°.

(2) The hardness of high C-Cr-steel was relatively low when quenched from high temperatures, and was increased by tempering.

(3) The structure of quenched Mn-steel was wholly austenitic, from which carbide was precipitated by tempering. The precipitation played an important rôle in the temper hardening.

(4) The precipitation of carbide caused an obvious contraction in dilatometric curve, and the precipitated carbide could be ascertained by microscopic observation.

(5) The austenite, which had precipitated carbide in a certain extent, changed into troostite in heating. When the temperature was not so high as to produce troostite, but was sufficiently high to precipitate carbide to a certain extent, the austenite changed into martensite in cooling indicating Ar" expansion.

(6) In the case of high C-Cr-steel, the carbide precipitation from austenite was not directly observable. But, from the similar forms of dilatometric curves, it may be natural to conclude that the mechanism of temper hardening is the same as that in Mn-steel.

(7) The reason for obtaining retained austenite in Mn-steel is as follows: the cooling rate is too quick to precipitate a sufficient amount of carbide from austenite, and the austenite which dissolved more carbide than a certain amount cannot be transformed into martensite; that the rate of cooling is quicker than that of the lattice transformation of austenite is not the cause.

In conclusion the author expresses his cordial thanks to Professor T. Ishiwara, the ex-Director of the Research Institute for Iron, Steel and Other Metals, Tohoku University, for his kind guidance.