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Residual Stress in Iron Alloys Variously Heat-Treated

By

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I. Introduction

Concerning the stress restored in alloys variously heat-treated, a number of interesting papers have been hitherto published[1]. The present writers have carried out a series of similar experiments on some iron alloys and steels, and the results which are somewhat different from those previously published, are briefly described in the present paper.

II. Experimental Procedure

In the present experiments, a small prismatic specimen ($4 \times 9 \times 100 mm^3$) prepared from 3 kg of forged ingot, was chiefly used to eliminate the mass effect as much as possible. Measurement of the stress restored in the specimen after heat-treatment was performed by Prof. T. Nishihara's etching method[2]; as shown in Fig. 1, two needles were attached to both ends of the specimen and the apartures of these needles at their upper and lower points were measured with a comparator with 1/100mm precision after each removal of 0.1mm of the specimen in its thickness by etching off one side of the surface of $9 \times 100 mm^2$ with an adequate reagent. The thickness was measured by means of a micrometer with 1/100mm precision. The stress which had been restored in each removed part of the specimen is calculated as follows.

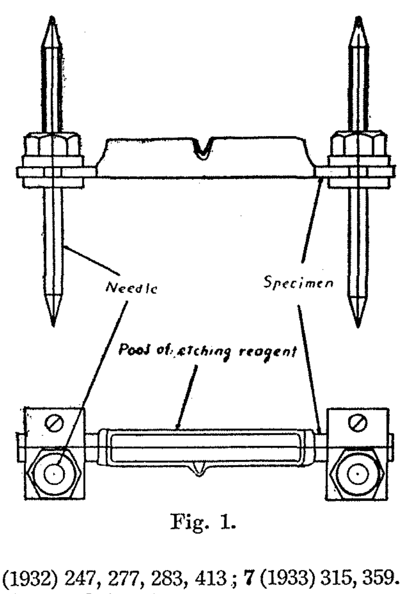

Fig. 1.

1. H. Bühler: Archiv. f. Eisenhüttenwes., **6** (1932) 247, 277, 283, 413; **7** (1933) 315, 359.
   T. Kawaguchi: Jap. Soc. f. Met., **8** (1944) 578; **9** (1945) 20.
   T. Sugeno: J. Iron & Steel Inst. Jap., **30** (1944) 19, 52.
2. T. Nishihara & S. Tahira: Trans. Jap. Soc. Mech. Eng., **14**, No. 49, (1948) 2.

The radius of curvature ρ can be obtained from the readings of the apartures of the needles by the formula:

$$\rho = \frac{(l_1 + l_2)}{2\Delta} g\,, \text{ with } \Delta = (P - Q) - (P_0 - Q_0)\,,$$

where l_1, l_2: lengths of the two needles,
g: length of the part to be etched (50 mm in this case),
P_0, P: apartures between the upper points of the needles before and after etching respectively,
Q_0, Q: apartures between the lower points of the needles before and after etching respectively.

Then the bending moment M_n is calculated from ρ above obtained as follows:

$$M_n = \frac{E}{P} \cdot \frac{a}{12}(2b - n\Delta h)\,,$$

where a: width of the specimen (9 mm in this case),
$2b$: thickness of the specimen (4 mm in this case),
Δh: depth removed at one time (0.1 mm in this case),
M_n: bending moment after n times removal of Δh,
E: Young's modulus.

The stress σ_n restored in n-th removed part is connected with the bending moment as follows:

$$M_n = a\left\{\sigma_n + \frac{\sum_{n=1}^{n}\sigma_{n-1}\Delta h}{2b - (n-1)\Delta h}\right\}\int_{2b-n\Delta h}^{2b}\left(\frac{2b - n\Delta h}{2} - t\right)dt$$

Therefore

$$\begin{aligned} M_n \doteq -\,a\Bigg[&\sigma_1 \frac{2b + (n-1)\Delta h}{2}\Delta h \\ &+ \left(\sigma_2 + \frac{\sigma_1}{2b - \Delta h}\Delta h\right)\frac{2b + (n-3)}{2}\Delta h + \cdots\cdots\cdots\cdots \\ &+ \left\{\sigma_m + \frac{\sigma_1 + \sigma_2 + \cdots + \sigma_{m-1}}{2b - (m-1)\Delta h}\Delta h\right\}\frac{2b + \{n - (2m-1)\}\Delta h}{2}\Delta h \\ &+ \left\{\sigma_n + \frac{\sigma_1 + \sigma_2 + \cdots + \sigma_{n-1}}{2b - (n-1)\Delta h}\Delta h\right\} \times \frac{2b - (n-1)\Delta h}{2}\Delta h\Bigg]\,, \end{aligned}$$

where t: thickness of the remaining part of the specimen after n-th removal of its surface.

Thus the mean value of the longitudinal stress which had been restored in each removed part of the specimen is obtained and may be plotted against the thickness of the specimen and thus the stress distribution curve is obtained.

Some iron alloys were used as specimens, which undergo no change of phase during the heat-treatment, in order to know the nature of the merely cooling stress. The effects of the cooling rate and of the plasticity of the alloy have been studied in this series of experiments. In the next place, some iron-nickel alloys as well as steels were used as specimens in order to know the effects of the phase change upon the residual stress. There are found many interesting papers about the residual stress in the case of iron nickel alloys[3], but to the present writers it seems very difficult to analyse the results without solving the nature of the irreversibility of the $\alpha-\gamma$ transformation in this alloy, about which a thorough study has been recently made in the senior writer's laboratory in Sendai[4].

Accordingly, the explanation of the results obtained are made very clear as will be seen later.

When steels are heat-treated, various structures are obtained according to the conditions, and hence in the present study the stress restored in steels variously heat-treated was measured and finally in those of the specimens carbulized or decarbulized.

III. Results

(1) α or γ phase iron alloys undergoing no phase change during cooling: Some Fe–V and Fe–Cr alloys were used as α alloys and 36 % Ni–Fe alloys as γ. The specimens were cooled from 300~1200° C by different cooling rates, and the elastic stress restored in the specimen was measured. Figs. 2 and 3 show some of the results obtained for example. In the case of furnace cooling (5° C per min.), no stress remained in all cases, but when the cooling is accelerated, the stress appeared. The stress distribution curves (half of them are drawn) shown in these figures are simple in their type, no minimum point being found on them. The stress is highest on the surface and is compressive and gradually decreases towards the interior of the specimen and at last it changes into tension. The highest tensional stress in the interior is always less than the compressive one on the surface. As the cooling velocity or the temperature from which the specimen is cooled increases

3. H. Bühler: Archiv. f. Eisenhüttenwes., **6** (1932), 283.
4. K. Iwase & S. Takeuchi: Jap. Soc. f. Met., **6** (1942), 361.

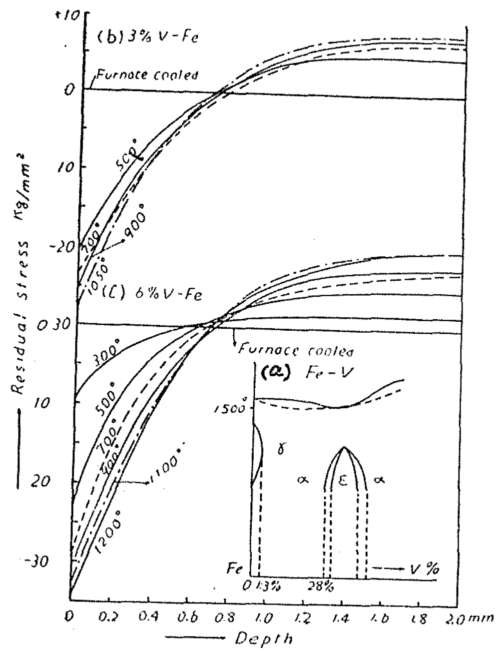

Fig. 2. Stress-depth curves of Fe-V alloys quenched from various temperatures.

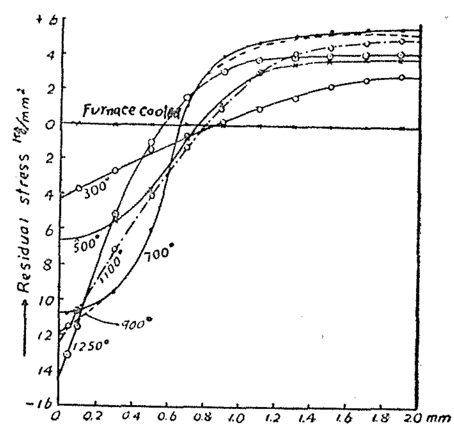

Fig. 3. 36 % Ni-Fe.

the stress becomes greater and when these cooling conditions are the same, the stress becomes greater as the strength of the alloy increases (The alpha alloy is stronger than the gamma alloy and 6 % V-alloy than the 3 % one). As the size of the specimen in the present study is very small, the residual stress is not so great.

(2) γ **phase iron alloys undergoing the lattice transformation during cooling :** In the foregoing section, the stress restored in the specimen was merely the cooling stress. In the present section, the quenching or the transformation stress will be explained. For this purpose 3~ca. 30 % Ni alloys have been used as specimens. According to the result of the investigation recently made by one of the writers[4], in these alloys, as the rate of diffusion is very feeble, the ordinary A_3 transformation hardly takes place and in place of it the lattice transformation commences on passing the curve GA on cooling and GB on heating without any change of the composition (in Fig. 4). This lattice change continues for a certain range of temperature and is completed only by passing through this range. This temperature interval of the lattice transformation increases as the Ni content in the alloy increases. It has been also ascertained that the furnace cooling is sufficient to prevent the diffusion in these alloys whose Ni content

are not less than ca. 5 %, and that it results only in the lattice transformation.

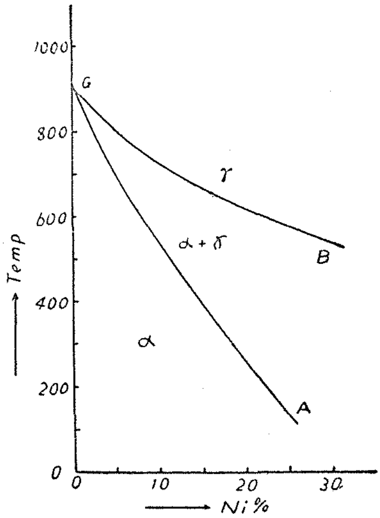

Fig. 4. γ–α of Fe-Ni alloys equilibrium.

According to the results of investigations found in literature, the residual stress of the quenched Fe–Ni alloy is compressive on the surface and the expansion caused by the transformation acts as additive to the cooling stress which had been restored in the specimen before the transformation, and the stress distribution curve have a minimum or maximum point in general. These results seem to the present writers to be attributed to the mass effect of the specimen so large as $\phi(50\sim70) \times 350\,\mathrm{mm}^3$. As the lattice change above said is accompanied by a remarkable expansion, the transformation stress may be thought as tensional on the surface of the specimen, because the merely cooling stress was compressive as shown in the preceding section. The results obtained by the present writers have justified this view as the following sections show.

(a) 29.8—31.5 % Ni–Fe alloys.

In the first place, it has been intended to make clear the effect of the lattice transformation upon the residual thermal stress of the alloy. For this purpose 29.8 % Ni alloy was used as a specimen, because it remains in the gamma state at the room temperature when it is cooled from a higher temperature as shown in Fig. 4. By cooling the alloy in the liquid air it transforms into alpha completely. Therefore, the alloy was cooled in the furnace on the one hand and quenched in ice water on the other, and the cooling stress has been measured respectively

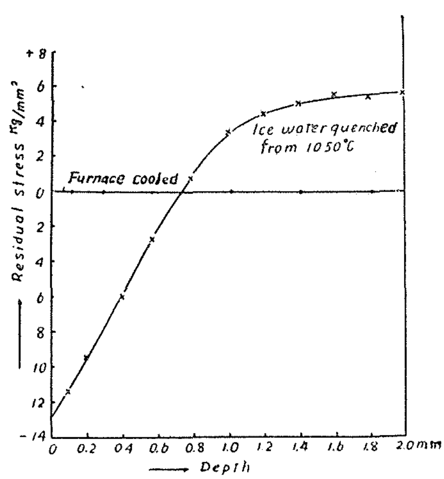

Fig. 5. 29.8 % Ni–Fe.

with the result that scarcely no stress remained in the furnace cooled specimen while not small amount of the stress (Fig. 5) remained in the quenched specimen. By cooling these two sorts of specimens, similarly heat-treated as before, to the liquid air temperature with various speeds, the lattice change from gamma into alpha was completed. The stress restored in these specimens has been measured at the room temperature. As shown in Fig. 6, the distribution curves in these cases are also simple and are near to each other regardless of their heat-treatments. The stress on the surface are tensional in all cases, the quicker cooling by the liquid air temperature resulting in slightly greater stress though the cooling rate prior to the lattice change had little effect upon the residual stress. Thus the large cooling stress of the quenched specimen, which existed before the lattice change, seems to have been released by the lattice change. As this fact is very important it was confirmed with the specimen containing 31.5 % Ni as shown in Fig. 7. This fact as well as the fact that the stress on the surface is tensional do not coincide with the results obtained by previous investigators probably because the mass effect predominates in the cases on very large specimens.

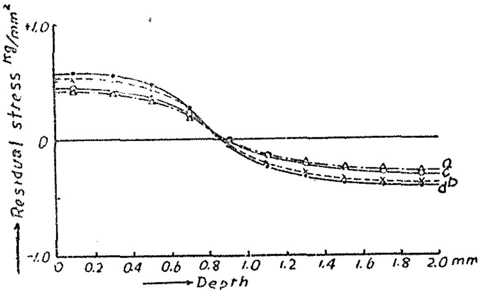

Fig. 6. a. △ Slow cooled till −185° from 23° C after slow cooled in furnace.
b. × Quenched into −185° from 23° C after slow cooled in furnace.
c. ○ Slow cooled till −185° from 23° C after ice water quenching.
d. ● Quenched into −185° from 23° C after ice water quenching.

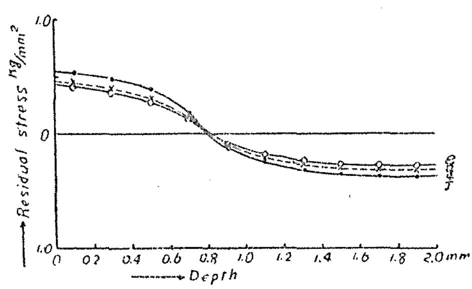

Fig. 7. e. ○ Slow cooled till −185° from 23° C after ice water quenching.
f. ● Quenched into −185° from 23° C after ice water quenching.
g. × Quenched into −185° from 23° C after slow cooled in furnace.

As the lattice transformation is a kind of recrystallization, if it happens on the surface layer of the specimen, the cooling stress existing within that layer may be released by the transformation and the stress existing within the remaining part of the specimen will be influenced thereby and the distribution curve of that part will be somewhat altered while the free expansion followed by the transformation may be restricted by the rigidity of the remaining part and thus also the distribution curv of that part may be again altered. In this meaning the transformation stress may be additive to the cooling stress which existed beforehand. If the size of the specimen is small, the time lag of the occurrence of the transformation in the centre as compared to that on the surface is very small and as the cooling stress is small, and so all the cooling stress which existed beforehand in the whole specimen may be released by the lattice change as the results shown in Figs. 6 and 7 represent.

(b) The effect of the transformation velocity.

As shown in the preceding section, the cooling stress becomes greater as the cooling velocity increases. The transformation stress also becomes greater as the velocity with which the transformation proceeds in the specimen from the surface into the interior increases as the obtained results in Fig. 8 show. The specimen in this case contains 23 % Ni and the lattice change commences at about 160~170° C and is completed before the temperature reaches 0° C. In this figure the curves a and c show that the cooling stress existed before the transformation is released, and so the restored stress shown in these four curves may be taken as they represent the transformation stress not influenced by the cooling stress which existed beforehand.

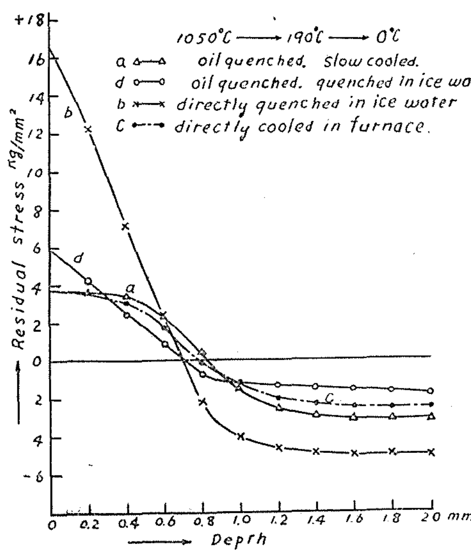

Fig. 8. 23% Ni-Fe.

(c) The restored stress in the α phase specimen containing Ni in various percentage.

As the Ni content of the alloy increased, the transformation point gradually decreased as shown in Fig. 4, and so when these alloys are quenched from 1050° C to room temperature or so, the plasticity of the alloy at the temperature of the transformation as well as the interval of the temperature after the transformation, through which the alloy is quickly cooled, decrease. These facts may naturally influence the stress restored in the quenched specimen. The specimens containing 3~25 % Ni have been quenched in ice water or furnace cooled (5° C per min.) from 1050° C and the stress restored in them as well as their hardness have been measured and the results obtained are shown in Figs. 9 and 10, the tensional stress and the Vickers hardness on the surface respectively are represented in the figures. On the curves drawn in these

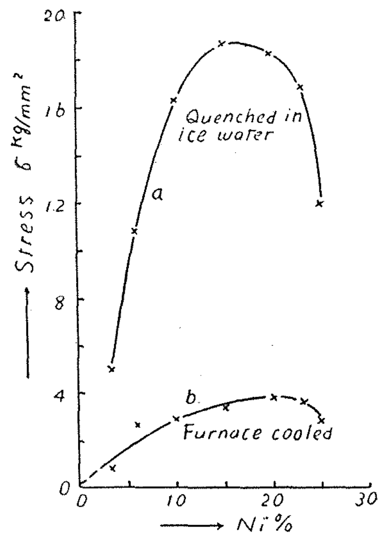

Fig. 9.

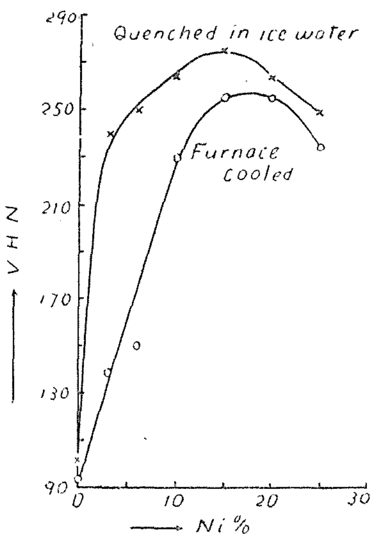

Fig. 10.

figures, the maximum points are found and those on the stress curves move to the left or right as the quenching medium is changed as shown in Fig. 11. The decrease of the plasticity of the alloy at the temperature

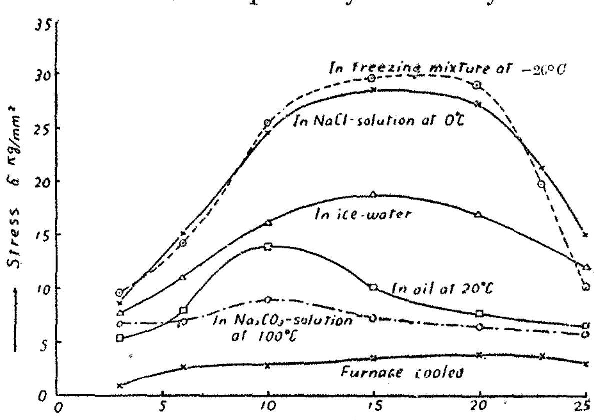

Fig. 11.

of the transformation, as the Ni content of the alloy increased, may cause the increase of the residual stress while the increase of the temperature range of the transformation, as the Ni content of the alloy increased, may cause the decrease of the stress, and the quicker cooling may weaken the effect of the increase of the range. The existence of the range of the temperature of the lattice transformation may result in the occurrence of the transformation at the same time in the different parts of the specimen whose temperatures are different and this may diminish the stress. It is to be noticed that the hardness curve of the furnace cooled alloys is closely situated to that of the quenched ones as shown in Fig. 10 while the two corresponding stress curves are remarkably different as shown in Fig. 9.

In all of these specimens, no trace of the gamma phase remained at the room temperature, being completely transformed into alpha phase. The influence of the existence of the gamma phase in the partly transformed specimen has been studied in the next experiments.

(d) The restored stress in the specimen partly transformed.

The lattice transformation point of the 29.8 % Ni alloy is below the room temperature and the alloy is completely transformed into alpha by immersing it in liquid air, and so when this alloy is cooled to some points between these temperatures, the alloy may consist of alpha and gamma phases, varying the amount of each phases according to the immersed temperature. The residual stress on the surface of the alloy in these specimens are as shown in Fig. 12, in which a maximum point is also found on the curve. Also in this figure it is shown that the hardness curve and the stress curve have aspects different from each other.

The stress distribution curve of such specimen shows a minimum as shown in Fig. 13, which is quite different from the cases treated in the foregoing sections. It seems to the present writers that the existence of the minimum point may be closely related to the heterogeneity of the structure of the alloy thus treated.

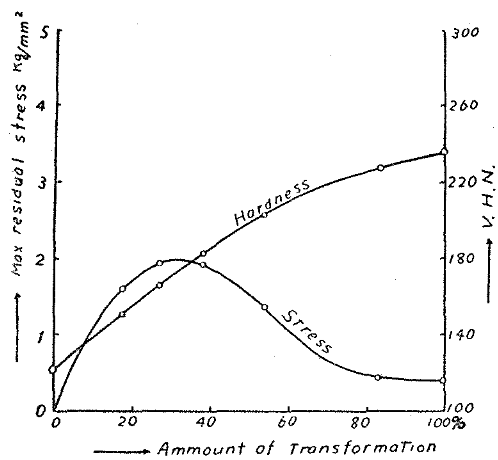

Fig. 12. 29.8 % Ni-Fe.

In the followings the stress restored in the specimen having a non-uniform structure has been measured to see whether the stress distribution curve is simple or not.

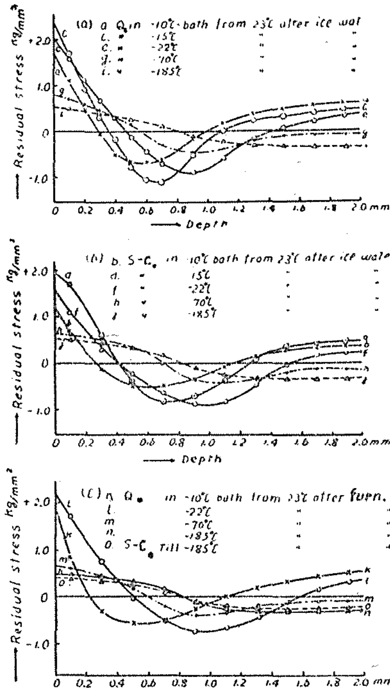

Fig. 13. Stress-depth curves of 29.8 % Ni–Fe.

(3) Residual stress in steels variously heat-treated.

(a) Residual stress in steels quenched or furnace cooled.

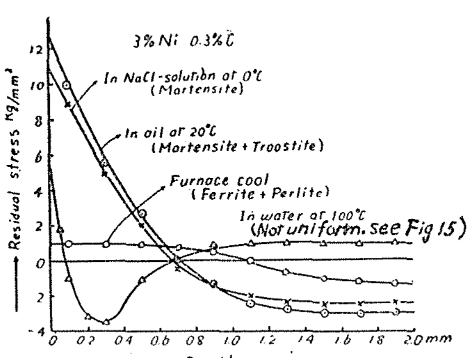

Fig. 14.

The steel containing 3 % Ni and 0.3 % C has been variously heat-treated (quenched from 950° C) and the structure obtained was observed and the restored stress was measured. Fig. 14 shows the results obtained. The structures of the specimens were uniform throughout their section except that of the specimen quenched in water at 100° C, though they are single phased or two phased. While that of the specimen quenched in water at 100° C was not uniform and was troostic in its periphery or external part and sorbitic in the central part, ferrite being mixed in both parts, and the stress distribution curve showed a minimum point only in this specimen. In the other three curves no minimum point is found and the stress is greater as the cooling rate is increased. From the result above explained it seems that the minimum point on the distribution curve is closely related to the non-uniform structures of the specimen which is also proved by the upper curve shown in Fig. 15.

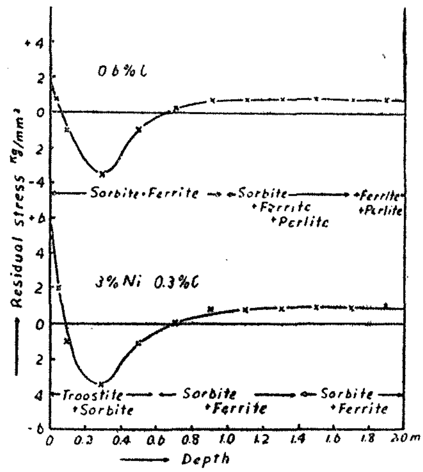

Fig. 15. Residual stress and structure of steel quenched in water from 950° C to 100° C.

The similar minimum point on the distribution curve was also obtained when the structure of the specimen was made non-uniform by the carburization or the decarburization as the following lines show.

(b) In the case of carburized and decarburized steels, furnace cooled or quenched.

The specimens containing 3 % Ni and 0.1 % C were carburized at their two surfaces corresponding to $9 \times 100\,\mathrm{mm}^2$ under various conditions, and the carbon content depth curves obtained being shown in Fig. 16–(a). The carbon content and consequently the structure of such specimen is heterogeneous throughout its section and such heterogeneity of the structure has caused the occurrence of the minimum point on the stress distribution curve of each specimen slowly cooled or quenched, as shown in Figs. 16 and 17. The minimum point is near the point of the eutectoid carbon content in the case of slow cooling and moves to the right in the case of quenching. It is to be noticed that the distribution curves of the quenched specimens are far steeper than those of the slow cooled. In the quenched specimens the structures are all martensitic (thought he carbon content of the specimen is not uniform in its section) and in this meaning their structures are not heterogeneous, but the temperatures at which the Ar″ change takes place are different in every points along the depth of the specimen, and this fact may be taken as the cause of the occurrence of the minimum point.

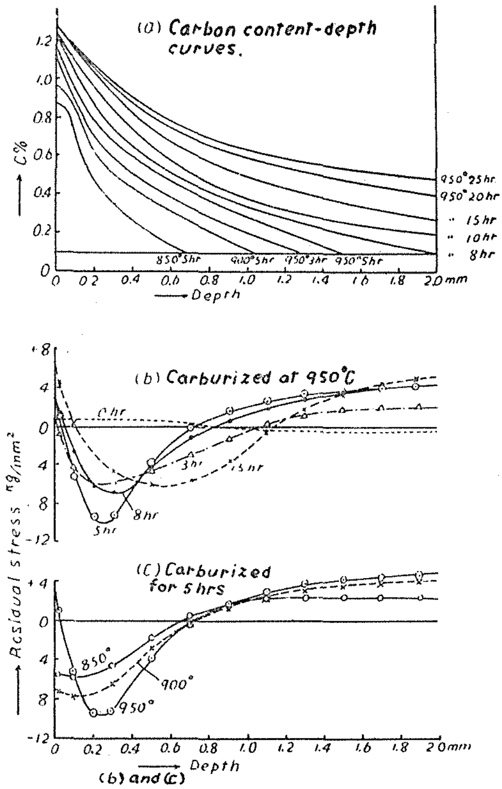

Fig. 16.

Similar sesults have been obtained in the case of the decarburization with the steel containing 1.3 % C as shown in Fig. 18. It is well-known fact that when the steel is decarburized on account of its bad treatment, such steel is apt to bring quenching cracks and this is conceivable from the result shown in Fig. 18.

When a large specimen is heat-treated, the structure will become heterogeneous owing to the mass effect and the occurrence of the minimum point on the distribution curve may move to the left or right by its cooling conditions.

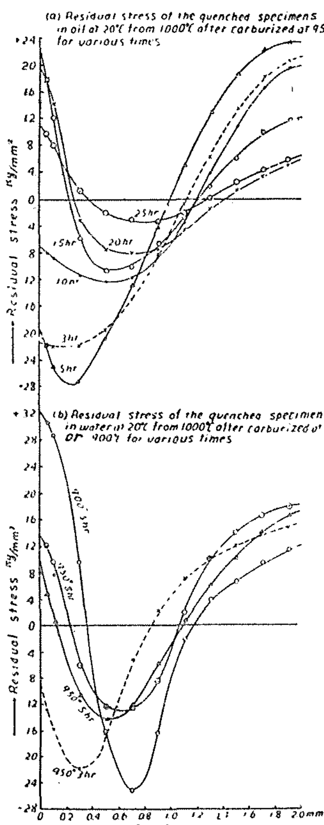

Fig. 17.

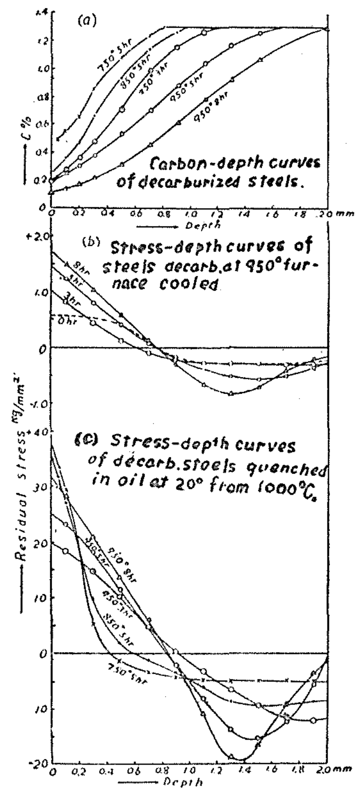

Fig. 18.

(c) In the case of the stepped quenching and of the austempering.

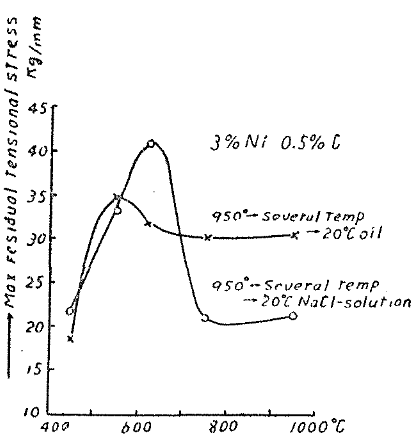

Fig. 19

The steel containing 3 % Ni and 0.5 % C can be quenched in lead bath at the temperature between 450° and 750° C from 950° C with no change of the phase and so this steel was quenched in the lead bath at these temperatures and then quickly in oil or salt solution at 20° C and the stress restored in the specimen was measured with the results as shown in Fig. 19. The curves in this figure show that the stepped quenching should be done in lead bath having the tem-

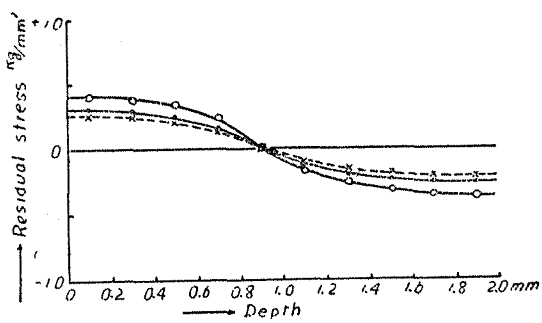

Fig. 20. Residual stress of 0.6 % C after austempering treatment.
× Air cooling after holding 20′ at 700° C Pb-bath (Ferrite+Pearlite)
● Air cooling after holding 12″ at 620° C Pb-bath (Sorbite)
○ Air cooling after holding 7″ at 480° C Pb-bath (Troostite)

perature lower than that corresponding to the maximum point in these curves in order to avoid the greater stress. In the austempering treatment only a very faint stress is restored in the specimen as the result shown in Fig. 20 represents.

(d) In the case of tempering of steel and of the aging of alloys.

By the tempering of the quenched steels the stress is decreased as the tempering temperature is elevated as shown in Fig. 21. In this treatment the separation of the cementite takes place and this may diminish the stress while the heating process may also act similarly. In order to make clear the two effects above explained, ageable silver alloy containing 6.3 % Cu and non-ageable silver alloy containing 6 % Mg have been quenched from 770° C and aged at various temperatures and the stress restored in them were measured. The result obtained is shown in Fig. 22–a, in which the compressive stress on the surface is plotted against the aging time. The precipitation of the copper rich phase has lowered the stress in the alloys.

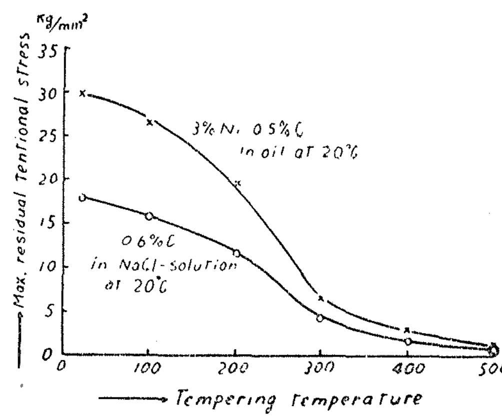

Fig. 21.

The curves in Fig. 22–b represent the results obtained by aging aluminium −4 % copper alloy quenched from 500° C. The aging at 20° C for 24 hrs. does not substancially diminish the stress while that at 150° C increases the stress at first and then decreases just as the hardness of the alloy increases (Fig. 23). The aging at 150° C for 24 hrs., it is said, in the separation of so-called θ'-$CuAl_2$ phase and after this period the θ-$CuAl_2$ phase appears, but this appearance of the θ'-$CuAl_2$ phase has no effect upon the stress curve and the stress decreases steadily from about 2~3 hrs. tempering which is quite different in the case of the change of the hardness. This relation is also seen at 200° C aging. From these results it may be supposed that the stress decreases when the atomic group is formed within the mother solution even if it is not separated from the solution. It is very interesting to compare this with the hardness curves in Fig. 23.

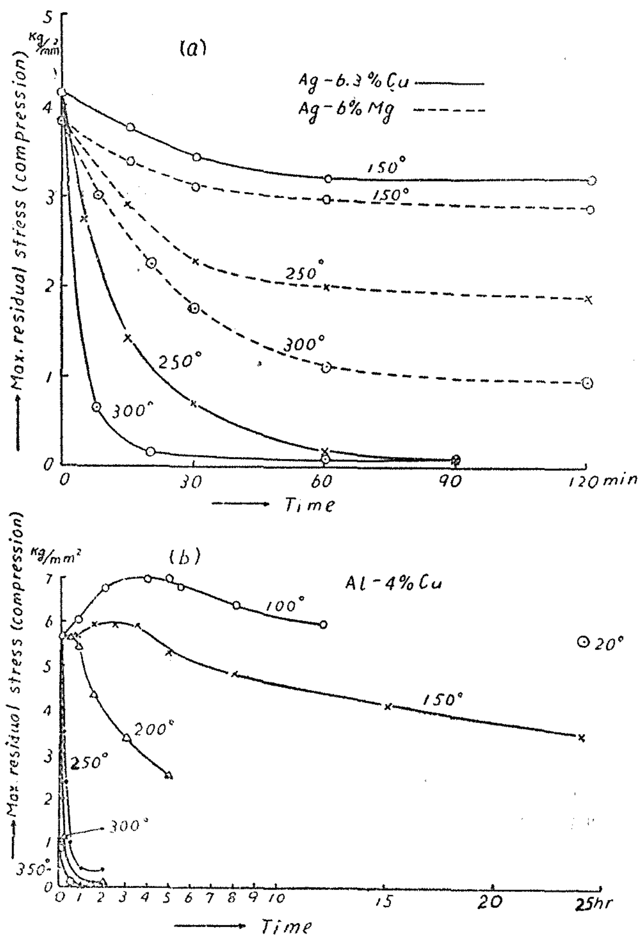

Fig. 22. Relation between maximum residual stress and aging.

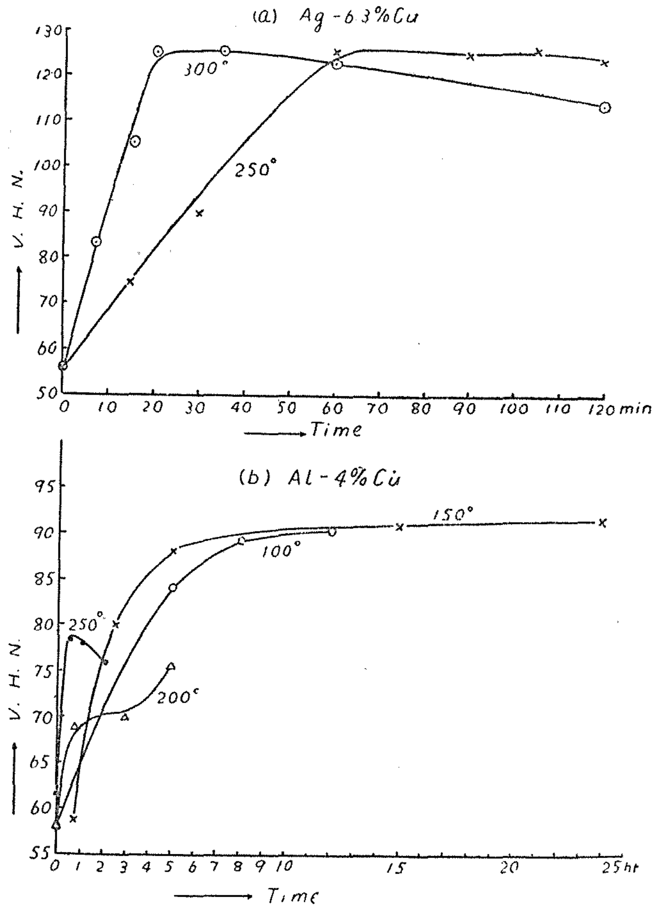

Fig. 23. Age-hardening at various temperatures.

IV. Summary

The cooling as well as transformation stress have been measured with some Fe-Ni and other Fe alloys and some steels using small specimens, and some new facts have been obtained, among which the release with the transformation of the cooling stress, which had been restored in the specimen beforehand, may be applicable in practice.