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Strain Age Hardening of Austenitic Ni-Cr Steel*

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Synopsis

The strain age hardening of austenitic Ni-Cr steels was studied, the chief purpose being to compare the characteristics between the one-phase structure in which austenite is stable and the two-phase one containing martensite.
The results obtained were as follows: (1) The increase in hardness occurred in two stages on annealing below 450°C after severe cold-working, the difference in the characteristics of the one-phase and the two-phase steels not being detected except that in the latter the hardening is very marked from the beginning with the progress of cold working. (2) The hardening was considered to concern mostly a dislocation-dislocation interaction. (3) The characteristics of the hardening are exactly similar to those of carbon steels in the range 150～350°C and copper alloys.

I. Introduction

The strain age hardening treated in the present investigation was confined to a phenomenon observed when a specimen is annealed at relatively high temperatures below the recrystallization temperature after cold working, and it is also called anneal hardening.

In copper alloys such as α Cu-Zn and α Cu-Al alloys,[(1)] the strain age hardening, when annealed in the range 50～280°C, is conspicuous provided the solute concentration is relatively high. In carbon steels,[(2)] the hardening at 150～350°C is more marked when the cementite particles are more numerous and finer within the hypo-eutectoid range. As the one- and the two-phase alloys above mentioned resembled each other in the characteristics of the hardening, it was proposed that both hardenings were of the same cause.[(2)]

Austenitic steels are similar to the copper alloys above stated in respects to the lattice type and high solute concentration. When cold-worked, however, they undergo partially the transformation into martensite, and become a two-phase alloy. The strain age hardening of austenitic steel is more marked when martensite is present than when it is not,[(3)(4)] but the cause is yet unknown. Cementite

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* The 1097th report of the Research Institute for Iron, Steel and Other Metals.

(1) T. Sato. and K. Nishino, J. Japan Inst. Metals, **20** (1956), 115, 704; **22** (1958), 50; **23** (1959), 232, 236.

(2) K. Nishino and K. Takahashi, J. Japan Inst. Metals, **20** (1960), 514, 518.

(3) T. Watanabe, J. Japan Inst. Metals, **21** (1957), 602.

(4) S. Mito, J. Japan Inst. Metals, **24** (1960), 790.

particles in carbon steels or alumina particles in Fe- and Cu-Al_2O_3 alloys affect greatly the hardening.(5)(6) Since the dispersed alumina particles are thermally stable, no chemical change would take place during low temperature annealing. To elucidate the nature of the strain age hardening, the effect of the dispersion of second phase has first to be known. The purpose of the present investigation is first to compare the characteristics of strain age hardening of austenitic Ni-Cr steels with and without martensite with each other, and further to study the correlation of those between carbon steels and copper alloys.

II. Specimens and experimental methods

The alloys used were commercial austenitic stainless and heat resisting steels and the chemical compositions are listed in Table 1. Steel C is a stable austenitic steel in which the martensite formation hardly takes place during cold working, steel B is somewhat unstable and steel A extremely unstable.

Table 1. Compositions and Ms points

Steel \ Element in %	C	Ni	Cr	Si	Mn	P	S	Ms °C
A	0.07	8.57	19.20	0.68	0.92	0.035	0.013	50°
B	0.06	9.28	18.37	0.50	1.30	0.018	0.012	$-70\sim-196°$
C	0.07	20.42	25.10	0.72	1.41	0.024	0.020	$<-196°$

To minimize the carbon content dissolved in austenite phase all specimens were heated in vacuum at 1150°C for 2 hours and then cooled stepwise at every 100°C below 1000°C holding for about 15 hours. The Ms point in Table 1 was estimated dilatometrically and magnetically during and after such a slow cooling. Low temperature annealing after cold working was made by using oil and (Pb-Sn) baths, the heating time being 5 minutes unless otherwise stated. After such treatments, specimens were electro-polished.

Amount of martensite was estimated magnetically by the ballistic method. Magnetic intensities of austenite and martensite were assumed equal to those of steel B in unworked state and 18 Cr-stainless steel, respectively. Effective magnetic field was about 1600 oersteds. Elongation was observed by optical extensometer with tensile specimens 6 mm in diameter and 50 mm in gage length. Measurements of electrical resistivity were made by a precision potentiometer. Dilatation measurements were done with Honda's differential dilatometer, the heating rate being about 1.8°C per minute. Measurements of X-ray diffraction were performed by using a back reflection forcussing camera and Fe-Kα radiation.

III. Results

1. Strain age hardening

In Fig. 1 are shown the strain age hardening curves for various steels rolled to

(5) Y. Imai, and H. Hirotani, Sci. Rep. RITU, **A12** (1960), 168.
(6) K. Nishino and K. Takahashi, Trans. JIM, **3** (1962), 57, 63.

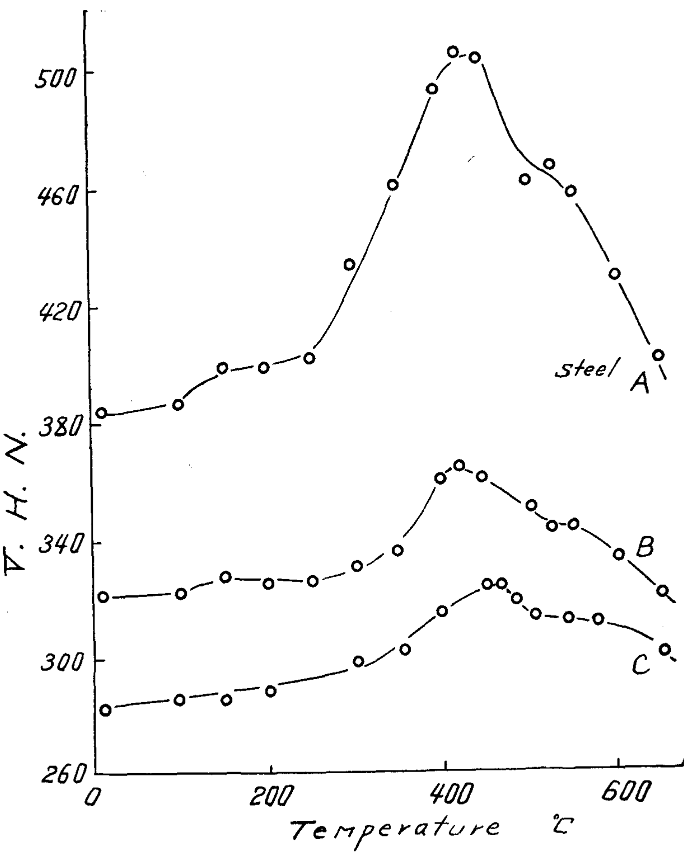


Fig. 1. Strain age hardening curves of various steels 50% rolled.

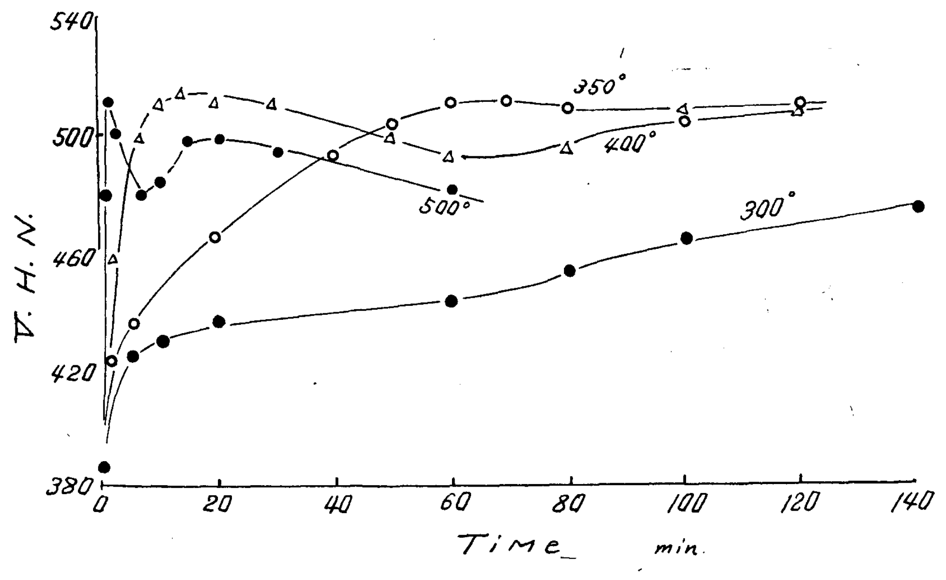


Fig. 2. Strain age hardening curves at various temperatures for steel A 50% rolled.

50 per cent reduction in thickness, herein it is seen that the hardness increases in two stages. The amount of hardening is widely different with different steels, but its trend is conformable to those of copper alloys(1) and carbon steels(2). Fig. 2 shows the annealing time-hardness curves at the illustrated temperatures for

steel A. Since, as seen in Fig. 1, the primary hardening below 200°C is slight, the hardening at an earlier stage in Fig. 2 is considered to correspond to the secondary hardening. The activation energy for the hardening obtained from the results was 26±2 Kcal/mol. On continued annealing, a hardening occurs again. For instance, in the case of 500°C annealing, it is observed in the range of 10∼20 minutes. This hardening probably corresponds to "arrest" at 500∼550°C in Fig. 1, which was also apparent in copper alloys(1) and carbon steels(2).

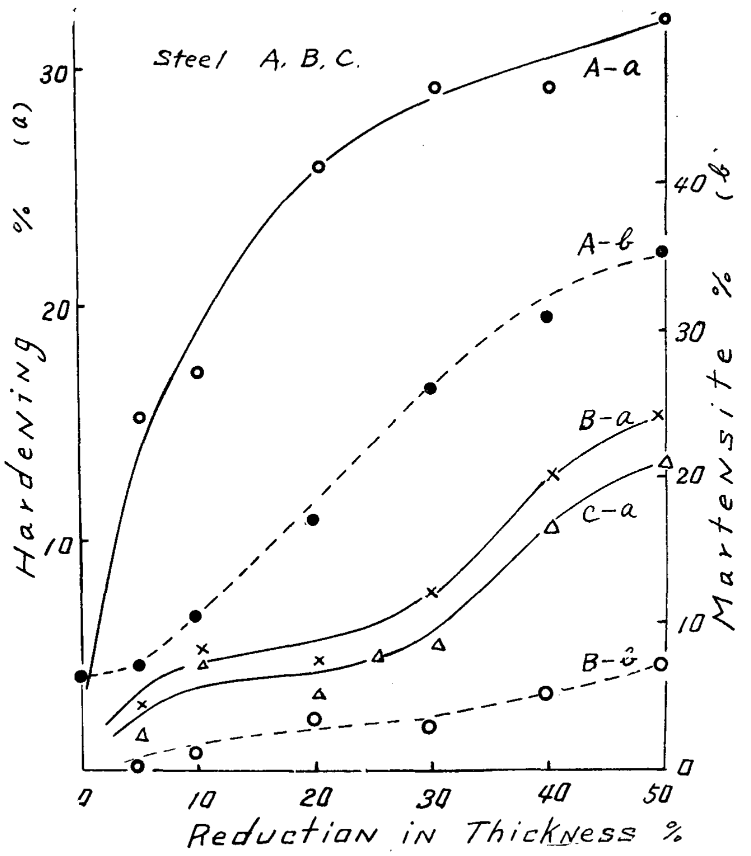


Fig. 3. Rolling degree versus amounts of strain age hardening and martensite.

Fig. 3 shows the relations between the rolling reduction and the amounts of strain age hardening and of martensite. The degree of the hardening in per cent was expressed by $(H_1-H_0)/H_0\times 100$, where H_1 and H_0 are the as-worked hardness and the maximum hardness in the curves in Fig. 1, respectively. In steel A, the martensite is much formed by cold rooling (A-b), and the degree of the hardening increases rapidly from the beginning as the cold rolling proceeds, reaching a high value, 32 per cent at 50 per cent rolling (A-a). In steel C, the martensite is absent even at 50 per cent rolling, and the hardening begins to increase above 20 per cent rolling (C-a). The trends of the increase in the hardening with the progress of cold working for steels A and C exactly correspond to those of carbon steels(2) and copper alloys(1), respectively.

2. Inelastic effect

Figs. 4 and 5 show the relations of tensile properties, rolling reduction and annealing temperature for various steels cold-rolled. In these tests, sheet tensile specimens $12 \times 3\ mm^2$ in section and 50 mm in gage length were used. As the

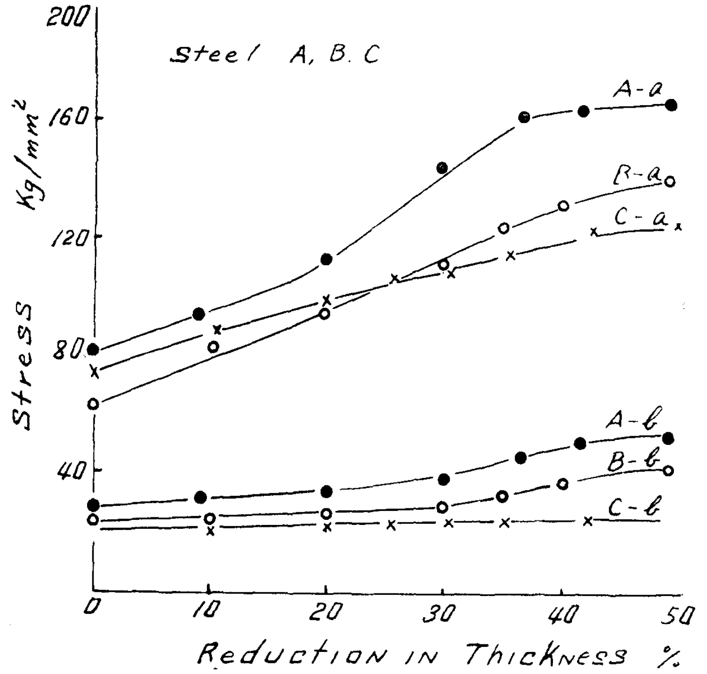


Fig. 4. Tensile properties and rolling degree. a. tensile strength. b. proportional limit.

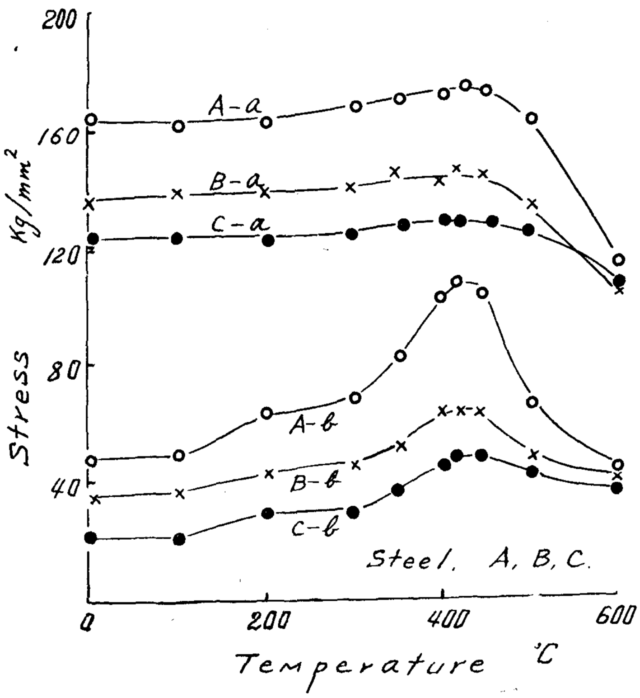


Fig. 5. Tensile properties and annealing temperature of steel A 48% rolled.
a. tensile strength, b. proportional limit

rolling reduction increases, the tensile strength increases remarkably, whereas the proportional limit increases slightly. After low temperature annealing, the amount of the increase in the tensile strength is only 5~8 per cent, while that in the proportional limit is very large, reaching 150 per cent.

In general, strain age hardening is widely different with different directions under a uniaxial stress.(2)(6)(7) To see whether this directionality is present in austenitic steel or not, experiments were made first with the specimen in which the applied stress was the same in type and direction as those of cold working.

Fig. 6 shows the stress-strain curves taken under a tensile stress of the same direction as that of prior extension. Two features are immediately seen. First,

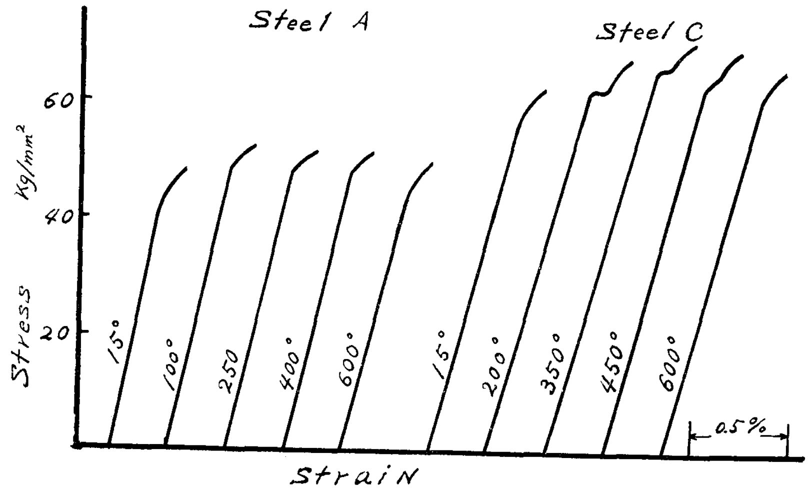


Fig. 6. Stress strain curves of steels A and C 8% extended and annealed at the indicated temperatures.

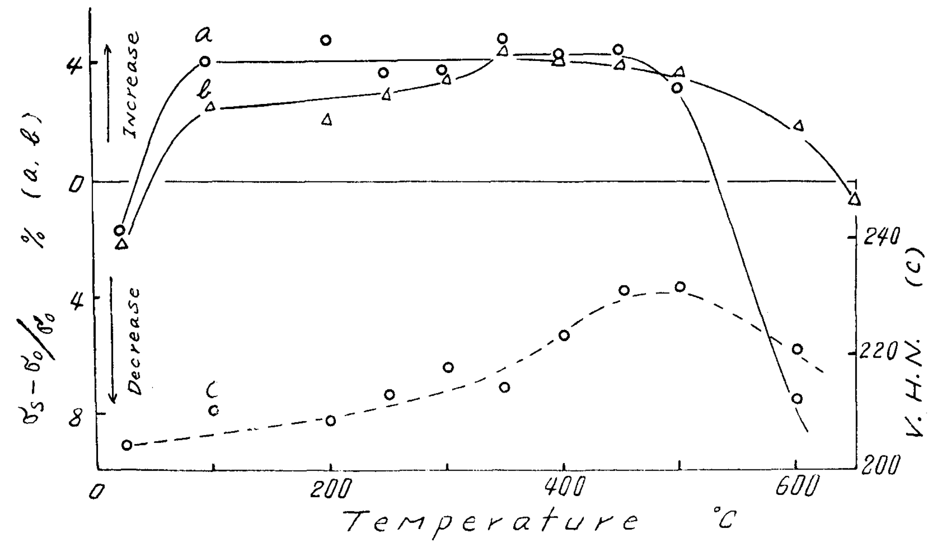


Fig. 7. Changes of yield stress and hardness by annealing of steels A and C 8% extended.
a,c; steel A. δo: flow stress at 8% of extension
b ; steel C. δs: yield stress on reloading

(7) S. Kumakura, Meeting of Japan Soc. Mech. Eng., (1958).

the discontinuous yielding is seen after 100∼200°C annealing but disappears at 350∼450°C, and secondly, the yield stress hardly changes in the range of 200∼430°C.

Curves (a) and (b) in Fig. 7 show the changes in yield stress with annealing temperature obtained from the above stress-strain curves. Although the yield stress rises slightly below 100°C, it hardly varied at 300∼450°C, while the hardness increases as shown in curve (c). As the indentation hardness is relatively insensitive to the directionality of cold-worked specimen, the hardness is considered to relate approximately to an average value of yield stresses in all directions. As the strain age hardening is hardly accompanied by a rise of the yield stress in worked direction, it is guessed that the rise would be remarkable in other directions, especially in the reverse direction. To verify this, an experiment on Bauschinger effect was carried out similarly to the case of copper alloys. But as austenitic steel is hardly machinable, a torsion specimen in the form of thin walled hollow cylinder could not be obtained, the experiment not being performed. The degree of Bauschinger effect, however, can be estimated by other method; that is, as shown at the upper part of Fig. 8, an inelastic deformation begins upon unloading after extension at the point where the stress-strain curve deviates from the straight line, the ratio λ/λ_0 being taken to represent the degree of inelastic effect, i.e. Bauschinger effect. Here, λ and λ_0 are noted in Fig. 8. This ratio is considerably sensitive to conditions of high temperature annealing and to strain rate. So, the annealing was made simultaneously on all the specimens, and the strain rates during extension and subsequent unloading were about 4×10^{-4} sec^{-1} and 4×10^{-5} sec^{-1}, respectively.

Fig. 8 shows the results of various steels. In steel C, in which austenite is very

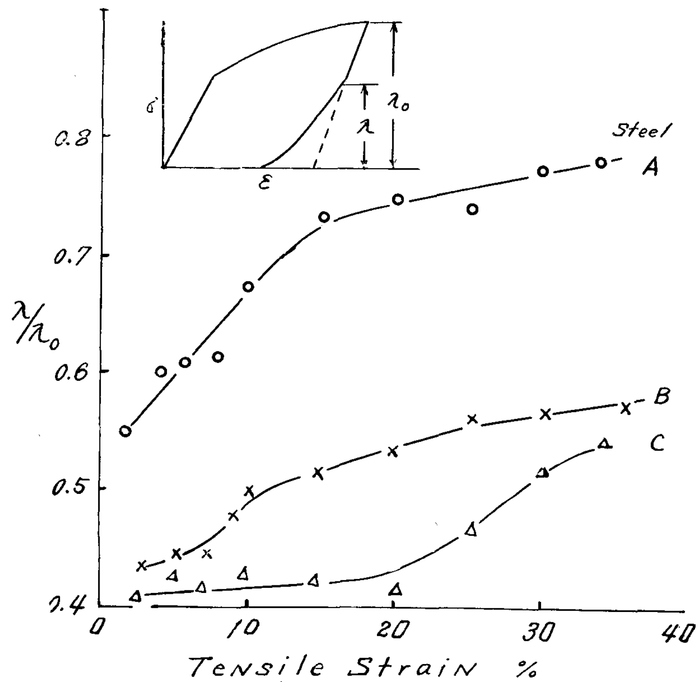
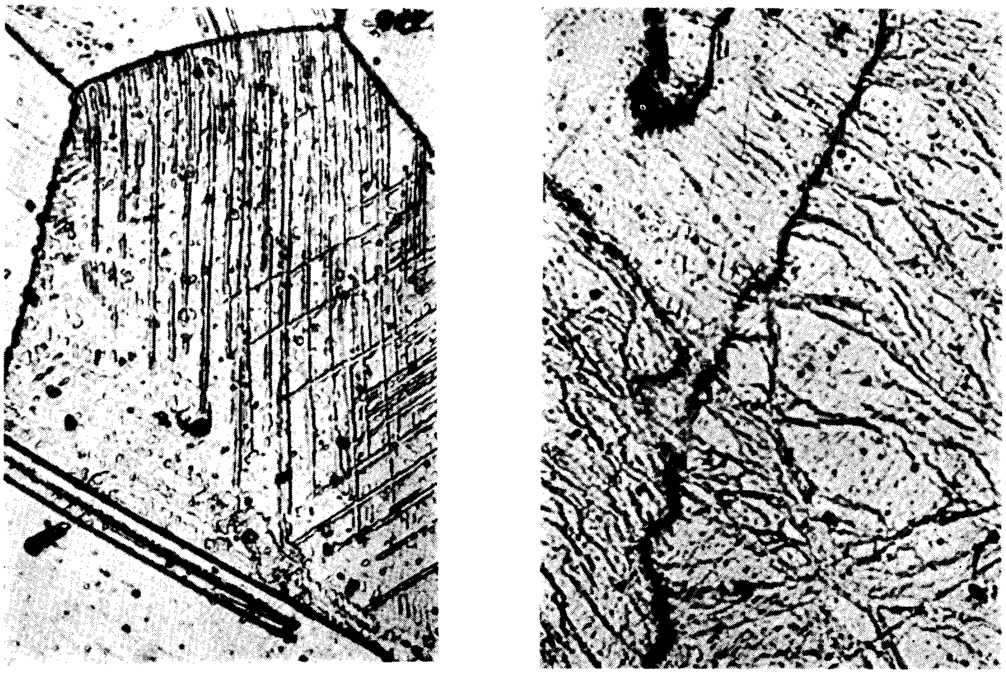


Fig. 8. Inelastic effect and tensile strain.

stable, the ratio λ/λ_0 becomes large above 20 per cent strain with the progress of extension, the trend being exactly similar to that of strain age hardening (Fig. 3), and also to that in copper alloys(1). Above 20 per cent rolling, strain markings were generated very much. Fig. 9 shows microstructures of steel C etched by aqua regia in glycerol. The markings are few at 20 per cent rolling, above which they increase remarkably. With heavy rolling, the structure shown by (b) was observed. A location at which strain markings were detected was likely to offer a high resistance to martensite transformation(8), and thus probably to dislocation movement.



(a) (b)

Fig. 9. Microstructures of rolled steel C. × 600

In steel A containing a large amount of martensite, the ratio λ/λ_0 is very large and increases from an earlier stage with the progress of extension as shown in Fig. 8. It is noteworthy that this tendency exactly corresponds to that of strain age hardening (Fig. 3), and that such a correspondence seems to be a general case regardless of one-phase or two-phase alloys. Although experiment on Bauschinger effect could not be performed as stated above, it might be acceptable that this effect would be eliminated in austenitic steel by low temperature annealing as were the cases of carbon steels(2) and copper alloys(1).

3. Electrical resistivity, thermal dilation and X-ray diffraction

Curve (c) in Fig. 10 shows the change in electrical resistivity with annealing temperature in cold-rolled steel A. The resistivity decreases in the temperature range of strain age hardening, the trend being nearly similar to that of hardness as in the case of copper alloys.(1) Curves (a) and (b) show the changes in thermal

(8) H.M. Otte, Acta Met., **5** (1957), 614.

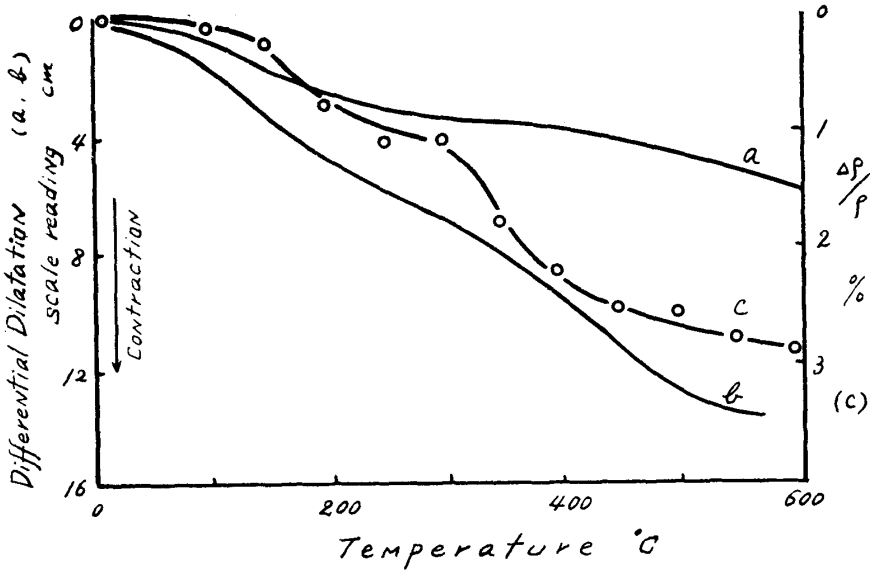


Fig. 10. Dilatation and electrical resistivity curves of steel A
a. 30% extended, b. 50% rolled, c. 50% rolled,
ρ. specific resistance in as-worked stage, 72.14 $\mu\Omega$ cm. $\Delta\rho$. decrease in specific resistance.

dilatation in the worked direction in cold-worked steel A. As the change was measured continuously by the constant rate heating method, it may be unsuitable to compare the results directly with those of other properties measured at room temperature. The trend of the change, however, shows no significant difference from that of the length of specimen measured at room temperature after low temperature annealing, and so, it may be allowed to interpret the results in a similar way. The contraction in the temperature range of strain age hardening is much larger than that in α brass[1], and hardly accompanied by a decrease in volume[9].

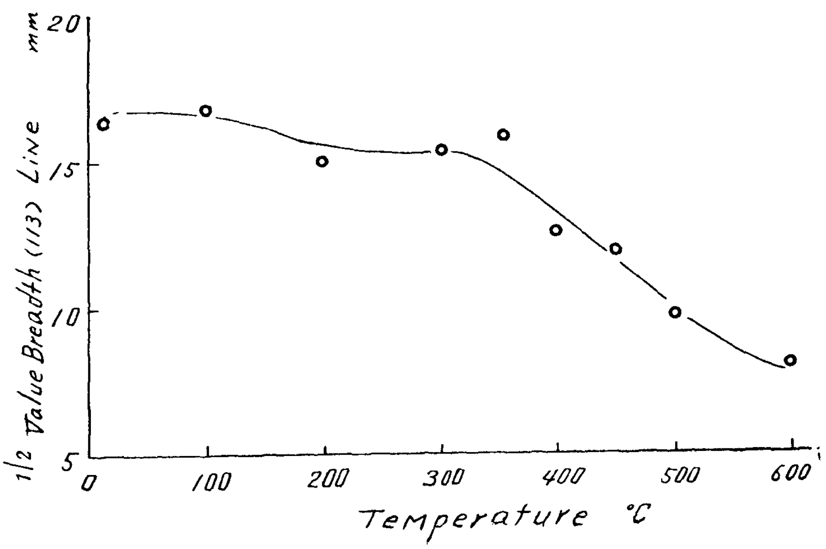


Fig. 11. Change of X-ray line broadening with annealing temperature for steel B 50% rolled.

(9) M. Okada and T. Watanabe, J. Japan Inst. Metals, 22 (1958), 440.

In copper alloys, the contraction becomes large above particular levels of solute concentration and of cold working,[1] the reason for which is probably that the contraction is not merely resulted from a relaxation of Heyn's stress.

In Fig. 11 is given the change in X-ray line broadening with annealing temperature in cold-rolled steel B, which indicates that the broadening is relieved remarkably by annealing in the temperature range of strain age hardening. A similar change was also observed in magnesium alloys[10] and carbon steels[11]. It may generally be said that an alloy showing marked strain age hardening is pronounced in the line broadening and the increase of electrical resistivity after cold working, and also exhibits a marked dilatation change during low temperature annealing.

4. Strain age hardening and martensite

As described previously, the presence of martensite in the matrix affects considerably strain age hardening. To clarify further this effect, the disappearance of martensite by annealing was first investigated. The result is shown in Fig. 12, wherein it is evident that nearly no decrease in the amount of martensite is seen until the annealing temperature rises to 450°C. In accordance, it is considered that there would be no structural change of the martensite in the temperature range of strain age hardening.

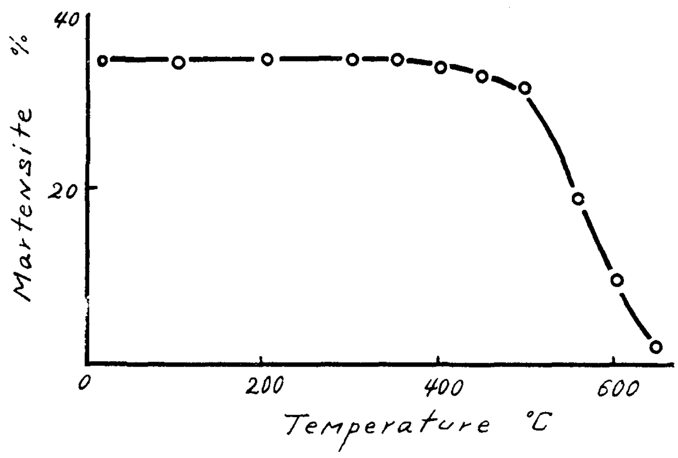


Fig. 12. Disappearance of martensite with annealing temperature in steel A 50% rolled.

To know the effect of the dispersion of martensite, experiments were made with specimens of steels A and B containing various amounts of martensite obtained by cold rolling in the range of $-40 \sim 100$°C. Fig. 13 shows the degree of strain age hardening in the specimens, in which it is evident that the degree increases markedly above a particular amount of martensite. This trend is nearly the same as the case of carbon steel[6], in which the degrees of inelastic effect and strain age

(10) J.C. McDonald, Trans. AIME, **212** (1958), 45.
(11) F.H. Andrew and H. Lee, J. Iron and Steel Inst., **165** (1950), 149 369.

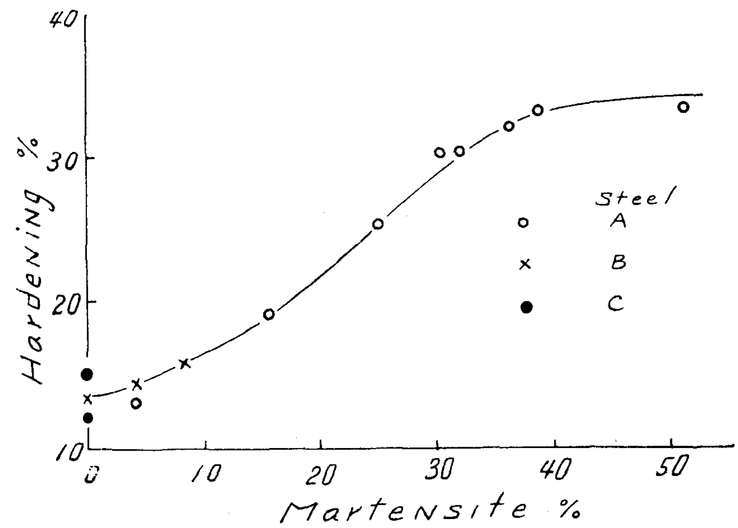


Fig. 13. Degree of strain age hardening and amount of martensite (50% roll)

hardening were approximately proportional to the back stress shown by the equation[12],

$$\sigma = 3\,\mu\,(r/s)^{3/2}\,(Nb/r)$$

or

$$= 3\,\mu f^{3/2}\,(Nb/r)$$

where σ is the back stress from dislocation loops piled-up around particles, μ the rigidity modulus, r the particle size, s the particle spacing, N the number of dislocation loops, b the Burgers vector, f the volume fraction of particles. The above equation concerns the work hardening in the case in which the second phase particles are globular. During or after unloading, dislocations in pile-up groups are easy to move back under the action of the back stress. In regard to the effect of the shape of the dispersed particle, Underwood et al.[13] reported that a platelike particle was more effective as a barrier against dislocation movements than a globular particle, being in good correspondence to the fact that in carbon steels, lamellar pearlitic steel exhibits more marked strain age hardening and inelastic effect than globular one[6]. In spite of the difference according to the shape of particle in such a manner, it may be allowable to say that the above equation can be applied to the case of platelike particle. In two-phase alloys, the back stress depends upon the size, the spacing and the volume fraction of second phase particles, and in addition on the mechanical properties of particle and surrounding matrix.

In carbon steels,[6] the degrees of inelastic effect and strain age hardening were approximately proportional of $f^{3/2}$ provided particle size was nearly constant. With austenitic steels, the degree of strain age hardening seems also, as shown in Fig. 13, to be approximately proportional to $f^{3/2}$ provided f is relatively small, where f is

(12) J.C. Fisher, E.W. Hart and R.H. Pry, Acta Met., **1** (1953), 336.
(13) E.E. Underwood, L.L. Marsh and G.K. Manning, Trans. AIME, **209** (1957), 1182.

the volume fraction of martensite. In higher ranges of the volume fraction, the strain age hardening is appreciably less marked than the expectation from the linear relationship above stated. The reason for this will be as follows: as the volume fraction becomes large, decreases, on the contrary, the volume fraction of matrix, which is considered to contribute substantially to the strain age hardening, whereas the dispersed phase only contributes to the straining of matrix so far as the strain age hardening is concerned, and further, on cold working, the more the second phase is the more the deformation of the second phase and conversely the less the deformation of the matrix.[14]

5. Work softening

In Fig. 14 is shown the work softening of steel A. When a slight working of the same type as that of prior working is again applied after low temperature annealing, the hardness decreases nearly down to the value in the as-primary-worked state. Such a softening is also observable in copper alloys[1] and high carbon

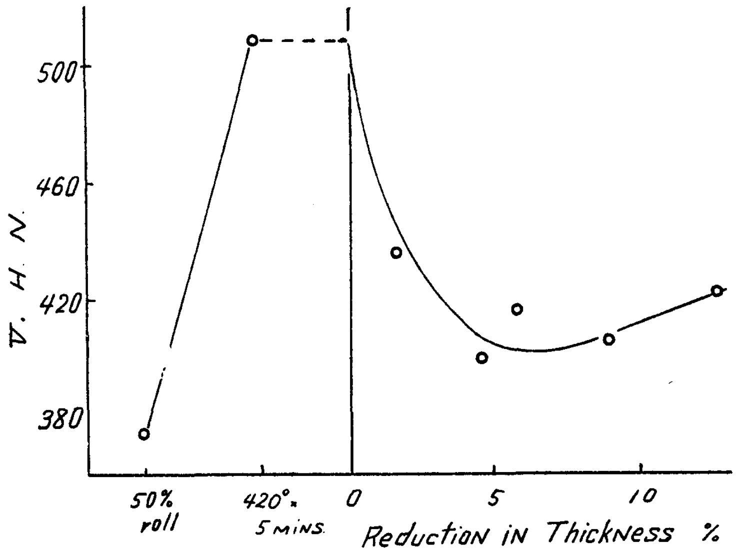


Fig. 14. Work softening of steel A.

steels[6]. In these alloys, the change in hardness such as shown in Fig. 14 was in good correspondence to that in Bauschinger effect. Although the experiment on this effect in austenitic steel could not be made as stated previously, it is presumed that such a correspondence would also be expected. In general, the work softening seems to be present in such alloys as exhibiting a marked strain age hardening.

6. Secondary work hardening

When a secondary working is different in type from the primary one, the hardness change is more complicated than when the both workings are the same in type.

(14) C. Nishimatsu and J. Gurland, Trans. ASM, **52** (1960), 469.

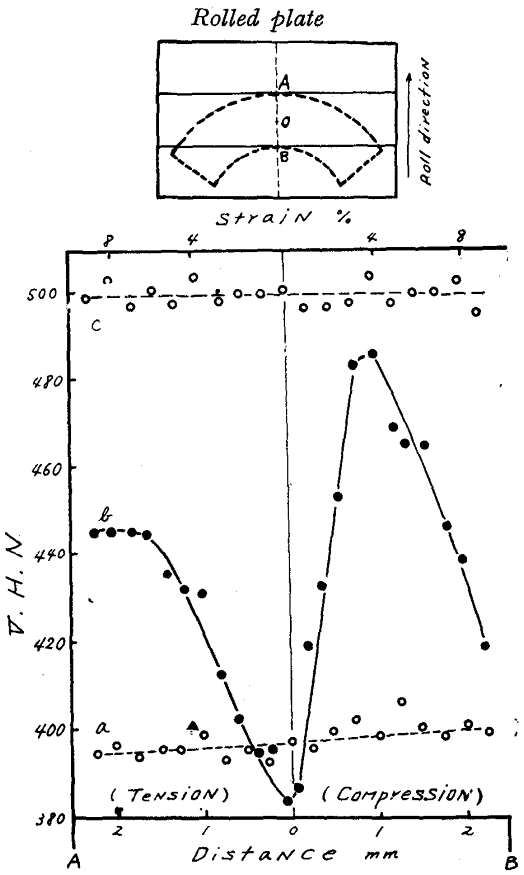


Fig. 15. Hardness changes by bending of steel A 50% rolled.
a. as rolled (50%), b. rolled and bent, c. rolled, bent and annealed at 450°C for 5 mins.

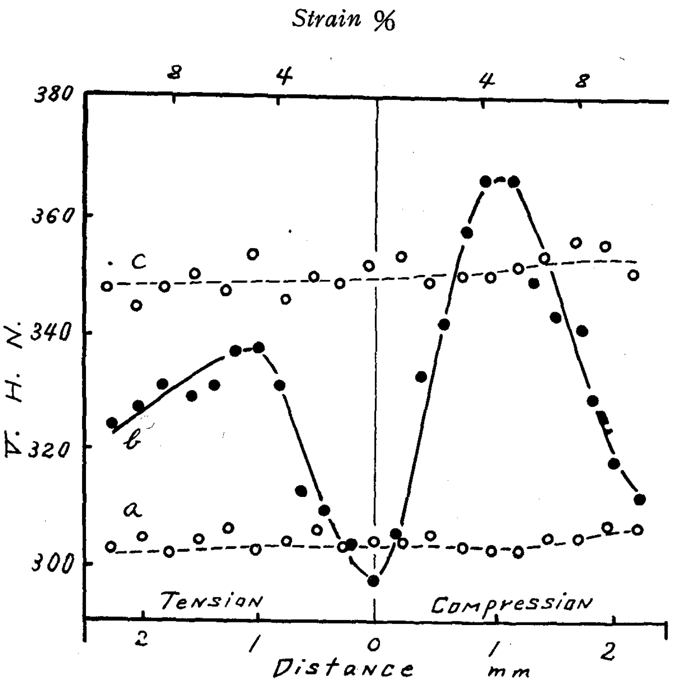


Fig. 16. Hardness changes by bending of steel C 50% rolled.
a. as rolled (50%), b. rolled and bent, c. rolled, bent and annealed at 450°C for 5 mins.

Fig. 15 shows the changes in hardness of cold-rolled steel A by bending. Bending was made on a specimen cut off from cold-rolled sheet of 4 mm in thickness in the way shown at the upper part of Fig. 15. After electropolishing, the hardness was measured along the length AB on the plane parallel to the rolling plane, so that the hardness in the as-primary-rolled state showed nearly a uniform distribution as shown by curve (a). When the bending is small, the distribution changes as shown by curve (b). At the part where the distance indicated on the abscissa is zero, that is, in the neighborhood of the neutral axis of bending, the change in hardness is slight, but as the distance, i.e. the strain increases, the hardness first increases markedly beyond the expectation from the work hardnening, and then decreases. This hardening, say, secondary work hardening, was also observable in copper alloys(1) and carbon steels(6). In addition, it is noteworthy that similar to carbon steels(6), the secondary work hardening is markedly larger at the compression side than at the tension side. The degree of this hardening is much affected by working conditions. In copper alloys,(1) when the primary working was made by drawing and the secondary by rolling, or by combining tensile and compressive deformation, the secondary work hardening nearly equal in the degree to the strain age hardening was brought about. In austenitic steels, however, the results from such combinations considerably scattered. This is probably due to the circumstamces that the specimens easily suffer unfavorable stresses during the handling. As the bending of rolled sheet was rather simple operation, the results could retatively be controlled.

When annealed at 420°C after the secondary working, the hardness increases nearly up to the maximum value in the secondary work-hardened state, and it gains again a uniform distribution throughout the section (curve c in Fig. 15). Fig. 16 shows the results on steel C consisting of very stable austenite, and the hardness distributions are similar to that of steel A containing a large amount of martensite.

On continuing secondary working after the secondary work hardening, a softening occurred as shown by curves (b) in Figs. 15 and 16. In α brass, when the primary working was severe, the secondary work hardening was nearly identical in the degree to the softening(1). Also in austenitic steels it was similar. So it may be said that when the primary or the secondary working is severe, it is in a softened state, and that an intermediate stage, at which the as-primary worked structure transforms into the secondary, is in a hardened state.

Fig. 17 shows the line profiles of (113) reflection in the same specimen as used in the case referring to Fig. 16. (a) is that at the location of the maximum hardness in the tension side, and (b) and (c) are those at the locations of the neutral axis of the bending and of the maximum hardness in the compression side, respectively. The back reflection method was used and the specimen was radiated through a small hole bored into a thin plate of lead with which it was covered. Near the part of neutral axis, that is, in the as-primary-worked state, the peak is indistinct (b), but in the region of secondary work-hardened state, the line becomes

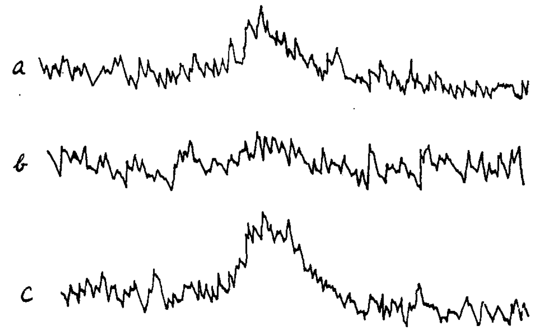


Fig. 17. Recordings of (113) reflections from steel C 50% rolled and then bent using Fe-K_{α} radiation.

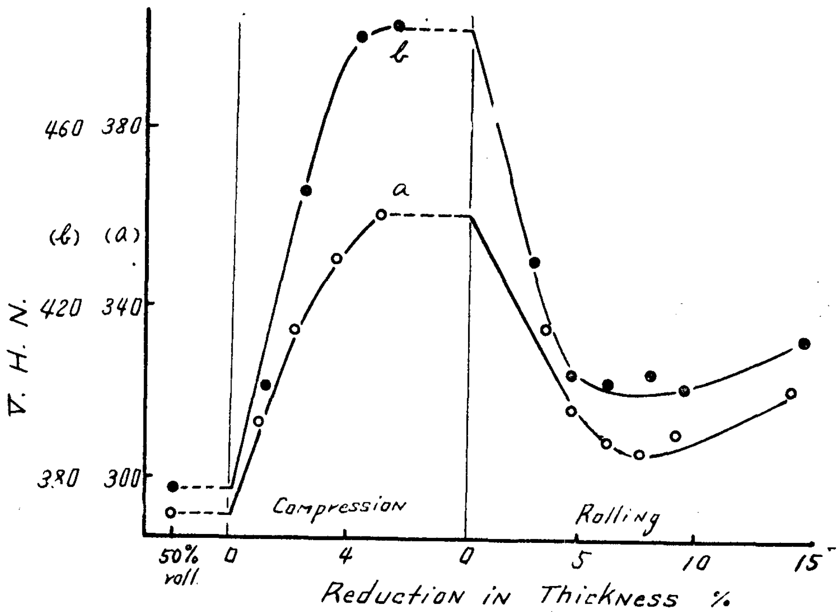


Fig. 18. Hardness changes of steels A (curve b) and C (curve a).

relatively sharp (a and c) similar to the case after low temperature annealing.

When a secondary work hardened specimen was cold-worked thirdly under the same condition as the primary, a softening occurred as shown in Fig. 18. Here, the compressive stress was applied by bending as in the cases referring to Figs. 15 and 16. It is of interest that the softening occurs nearly in the same way as the work softening after low temperature annealing (Fig. 14), in other words, a secondary working of different type gives the same effect as a low temperature annealing on the hardening. Further, it is noteworthy that no essential difference in the hardness changes seems to exist between the austenitic steels with and without martensite, and that these behaviors are quitely similar to those in copper alloys(1) and carbon steels(6).

7. Secondary work hardening and martensite

To see the effect of the presence of martensite on the secondary work hardening specimens containing various amounts of martensite were prepared by varying

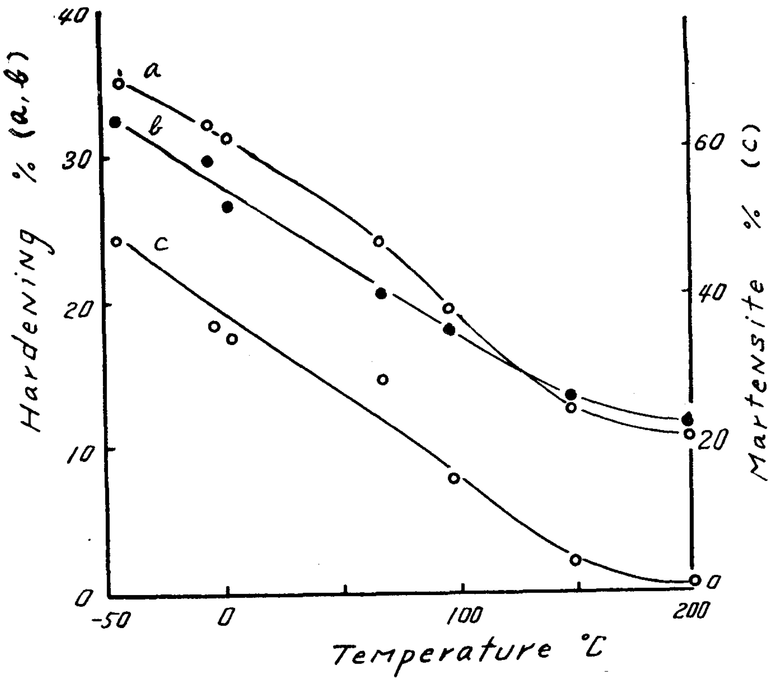


Fig. 19. Working temperature and amounts of hardening and of martensite in steel A. a; strain age hardening, b; secondary work hardening, c; martensite.

rolling temperature in the way similar to the experiments with respect to Fig. 13, and then secondarily bent at room temperature. Fig. 19 shows the relations in 50 per cent rolled specimens between the rolling temperature and the amount of strain age hardening (curve a), secondary work hardening (curve b) and of martensite (curve c). The degree of secondary work hardening in per cent was obtained from $(H_1-H_0)/H_0 \times 100$, where H_1 and H_0 are the maximum hardness in secondary work hardened state and the hardness in as-primary-worked state, respectively. As the rolling temperature rises, the above three decreases in a similar way, probably indicating that the degree of secondary work hardening is, as was the case of strain age hardening, proportional to $f^{3/2}$ in the range where the volume fraction of martensite f is relatively small. The similarity in both hardenings was also observed in α Cu-Al alloys[1], that is, the degrees of both hardenings increased markedly, in any case, above 3 wt per cent Al with the increase of aluminium content.

IV. Discussion

1. The strain age hardening of austenitic Ni-Cr steel was studied chiefly by comparing the cases with and without martensite with each other, and it was ascertained that there was no essential difference between the two cases, and that the behaviors of the hardening are in good similarity to those of carbon steels in the ragne 150$\sim$350°C and copper alloys. The explanation of the strain age hardening of copper alloys has been attempted by many workers on the basis of the

peculiality of face-centred cubic lattice and the diffusion of solute atoms, i.e. the short range ordering[15], the segregation of solute atoms to stacking faults[16], or the formation of GP zones[17].

The strain age hardening of carbon steels at 150~350°C is widely different in the characteristics from that below 150°C, which has been explained by the Cottrell locking. It is the more marked the higher the carbon content within the range up to 0.9 per cent of carbon unlike the hardening below 150°C, and its characteristics resemble those of copper alloys.[1] A similar hardening has also been observed in pure metals[18][19] and two-phase alloys in which alumina particles finely disperse[5][6]. It is thus guessed that the strain age hardening is a general phenomenon in metals and alloys.

2. In two-phase alloys, the inelastic effect is marked from the start of extension, and in carbon steels, it is the more marked the larger the volume fraction of cementite, and further it is inversely proportional to the particle size of cementite provided the volume fraction is constant. Thus, it may be said that the inelastic effect is chiefly induced by the back stress due to dislocation loops piled-up around the particles,[12][21][22] but not to Heyn's stress.[20] The back stress is dependent on the shape, the size and the distribution of dispersed particles, and also on the mechanical properties of particle and the surrounding matrix.

Martensite is surely a strong barrier against the movement of dislocations, inducing a high level of back stresses. Dislocations in piled-up groups are, under the influence of back stress, difficult to move in the direction of external work, but very easy to move back to the reverse direction. A marked inelastic effect and a low proportional limit are considered to be due to the presence of such mobile dislocations. In carbon steels[6], the degree of inelastic effect increased rapidly within the first 10 per cent strain with the progress of extension, but above this it became nearly constant. The final value is probably dependent on the level of back stress inherent in the specimen, although it increases slightly with the progress of cold working probably because of the fragmentation of particles or the strengthening of matrix.

As stated in the previous section, hardening can be brought about not only by low temperature annealing but also by slight deformation. It seems, therefore, that the hardening does not necessarily need annealing, and that it concerns mainly a dislocation-dislocation interaction[23][24] rather than solute-dislocation

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interaction, although the latter cannot quite be ignored. One possible mechanism will be such that dislocations in pile-up groups move back under the influence of the back stress, and encounter other groups with which they can react, so that they lose mobility. Such an interaction of intersecting dislocations[24] may be the easier and the larger the more the groups. Further, it is considered that the activation energy for the reaction W is appreciably high, but it becomes small in the form W-$\alpha\sigma$ because of the presence of back stress σ. Here α is a constant.

According to the observation of thin film by electron microscope, many dislocations are distributed relatively uniformly in as-worked state,[26] but in strain age hardened state, they show a tendency to form cell structure as shown in Fig. 20. The cell formation seems to be a recovery process by thermal migration of dislocations. This will be stated in the next paper.

Fig. 20. Microstructure of thin film of steel B 50% rolled and annealed at 450°C for 5 minutes.

3. In one-phase austenitic steels, the strain age hardening becomes marked above a particular degree of cold working similar to copper alloys[1]. In α Cu-Zn and α Cu-Al alloys of relatively high solute concentrations, the changes in thermoelectric force and density and appearance of strain marking become large abruptly aboe about 15 per cent reduction with the progress of cold drawing. Thus, it seems that above this strain increase abruptly the obstacles against dislocations corresponding to the particles of two phase alloy, and many piled-up dislocations are formed around them. In other words, the obstacles are present prior to cold working in two-phase alloys, while in one-phase alloys, they are generated dur-

ing cold working. As stated, in copper alloys, the smaller the grain size is the more marked the strain age hardening.[25] This is probably due to the circumstances that grain boundaries act as an effective barrier.

When solute concentration is relatively low in copper alloys, the strain age hardening and the secondary work hardening are considerably small, being in good correspondence to the slightness of the changes in electric resistance[18], thermoelectric force and density after cold working. Also it is noticeable that when the solute concentration is low, the cell structure is easily formed after cold working, while in the case of high solute concentration, it is difficult but regular pile-ups of dislocations are formed[26]. It may be said that regular pile-up of dislocations is necessary for the occurrence of the strain age hardening, and further that the effect of alloying on the strain age hardening is to form effective obstacles against the movement of dislocations and to produce a high density of pile-up dislocations with an appreciable resistance to their leakage during cold working, rather than to segregate into stacking faults or to take short range ordering during low temperature annealing. This effect will be discussed in the next paper.

Summary

The strain age hardening of austenitic Ni-Cr steels was studied chiefly by the comparison of the steels with and without martensite, and the results obtained may be summarized as follows:

1. Strain age hardening occurs in two stages on annealing below 450°C after cold working, and no difference in the characteristics of the hardening seems to exist between the one-phase and the two-phase steels.

2. The degree of strain age hardening is approximately proportional to $f^{3/2}$ in the range in which the volume fraction of martensite f is relatively small.

3. A slight application of the secondary working differing in type from the primary causes a marked hardening beyond the expectation from the work hardening. This hardening, say secondary work hardening, seems to be the same in cause as the strain age hardening.

4. The strain age hardening concerns a dislocation-dislocation interaction rather than solute-dislocation interaction. It may be said that the hardening is marked in such alloys as containing large numbers of piled-up dislocations. Martensite is an effective barrier against dislocation movement, and thus, for the generation of piled-up dislocations. In the case without martensite, the obstacles are formed during cold working.

5. The characteristics of the strain age hardening are exactly similar to those of carbon steels in the range 150$\sim$350°C and copper alloys.

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