THE NOTCHED FATIGUE PROPERTIES OF HIGH TENSILE STEELS*

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Abstract—Notched and unnotched fatigue tests have been carried out on several ultra-high strength alloy steels in the as-quenched condition and after tempering at progressively higher temperatures. The effect of residual quenching stress and secondary hardening stresses on the notch sensitivity of the steels has been evaluated. Minimum notch sensitivity was found in the as-quenched condition and after tempering to the peak of secondary hardening. The mechanism of failure is discussed in terms of the effect of residual stresses and metallurgical structure on the relief of applied peak stresses.

1. INTRODUCTION

MANY research programmes are currently being carried out on the basic mechanisms of fatigue. Most of this work has been limited to the pure simple alloy systems of copper and aluminum. These researches are yielding much information on the metallographic changes which occur during fatigue stressing, and provide a foundation upon which a theory of fatigue failure may be developed. However, very little work has been done on the effect of metallurgical structure on the notch sensitivity of various metals, particularly steels. In fact very few of the recently reported researches into the basic mechanisms of fatigue have utilized a study of the variations of notch sensitivity of materials.

A series of steels which showed promise for a study of the effects of metallurgical structure on fatigue notch sensitivity were the secondary hardening steels. These steels show marked resistance to tempering up to 900° F (480° C), and above this tempering temperature show an increase in hardness. Small changes in the tempering temperature result in marked differences in the type and distribution of the carbides produced.

Notched and unnotched fatigue tests were carried out on specimens which were in the as-quenched and the quenched and tempered condition. From these results an attempt has been made to explain the notch

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sensitivity of steel in relation to metallurgical structure and to indicate how these initial experiments may be usefully extended to obtain information from other alloy systems.

2. MATERIALS AND EXPERIMENTAL TECHNIQUES

2.1 *Materials*

The chemical compositions of the steels tested are given in Table 1.

TABLE 1
Chemical Composition of the Steels

Steel	Composition per cent								
	C	Mn	Si	S	P	Cr	Mo	V	Ni
A	0·28	1·21	0·24	0·011	0·004	—	—	1·00	—
B	0·39	0·65	0·44	0·014	0·008	1·34	0·47	0·32	—
C	0·34	0·40	1·17	0·010	0·017	5·08	1·43	0·92	—

Steel A is an experimental steel in which the microstructure can be carefully controlled by the precipitation of vanadium carbide. Steels B and C are examples of commercial low-alloy secondary hardening steels. The latter steel is of current interest for ultra-high tensile strength structural components of aircraft. All of the steels were initially in the as-rolled condition.

2.2 *Experimental Techniques*

Samples of each of the steels were austenitized at 1850 °F (1010 °C) and tempered at tempering temperature up to 1200° F (650° C). Rockwell "C" hardness tests were carried out on each of these specimens and tempering temperature-hardness curves were plotted.

Tensile test specimens 0·375 in. diameter and Krouse fatigue specimens with a test length of 2 in. radius and 0·20 in. diameter were machined from bars of each of the steels. The specimens are shown in Fig. 1. The bars were rough machined prior to heat treatment and then ground to the final dimensions after heat treatment. The fatigue specimens were polished with 000 emery paper until all evidence of transverse scratches was eliminated.

3. EXPERIMENTAL RESULTS

3.1 *The Effect of Tempering Temperature on the Hardness and Microstructures of the Steels*

The hardness tempering-temperature curves for the steels are shown in Fig. 2. The microstructures of these steels in the as-quenched condi-

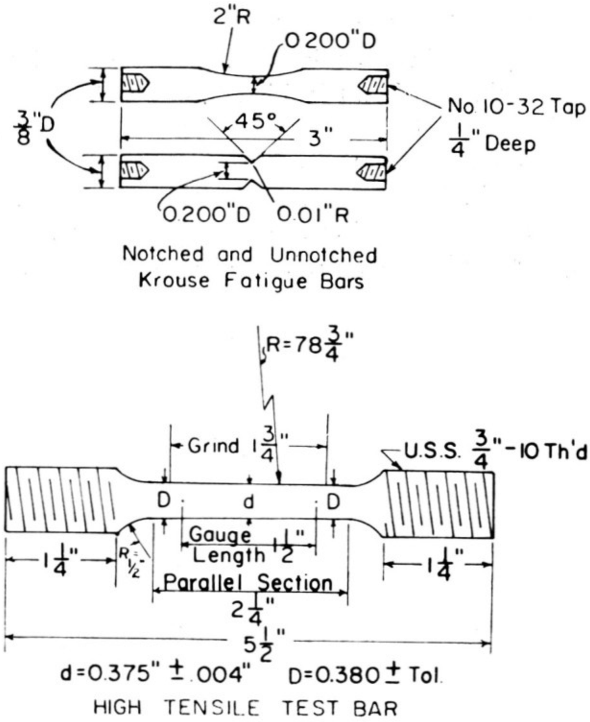

FIG. 1. The dimensions of the tensile and notched and unnotched fatigue specimens.

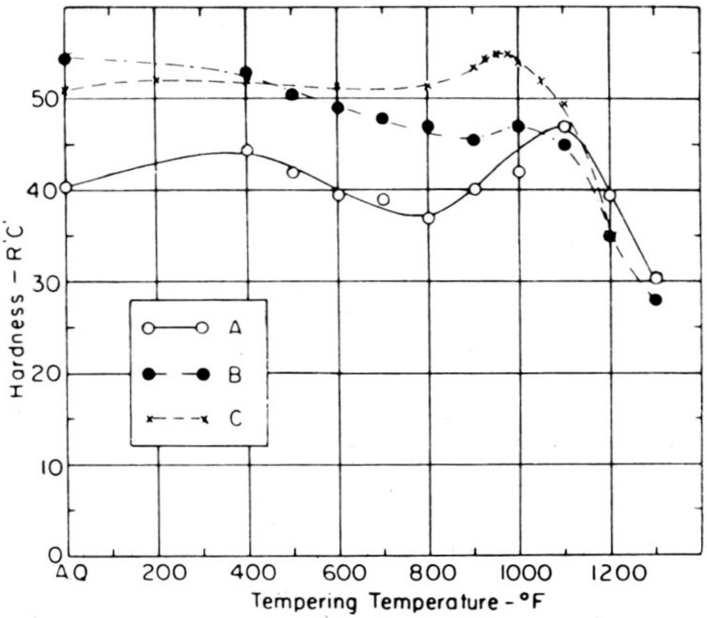

FIG. 2. Relationship between hardness and tempering temperature for the Steels *A*, *B* and *C*.

tion, and after tempering at temperatures corresponding to the minimum, the peak and beyond the peak in the hardness tempering-temperature curves for the steels, are shown in Figs. 3, 4 and 5.

The tempering-temperature hardness curves show three distinct stages:

(a) A small drop in hardness on tempering up to about 800° F (425° C).

(b) An increase in hardness on tempering above 800° F (425° C) to a peak hardness obtained at 1100° F (590° C) for Steel A, 1000° F (540° C) for Steel B and 950° F (510° C) for Steel C.

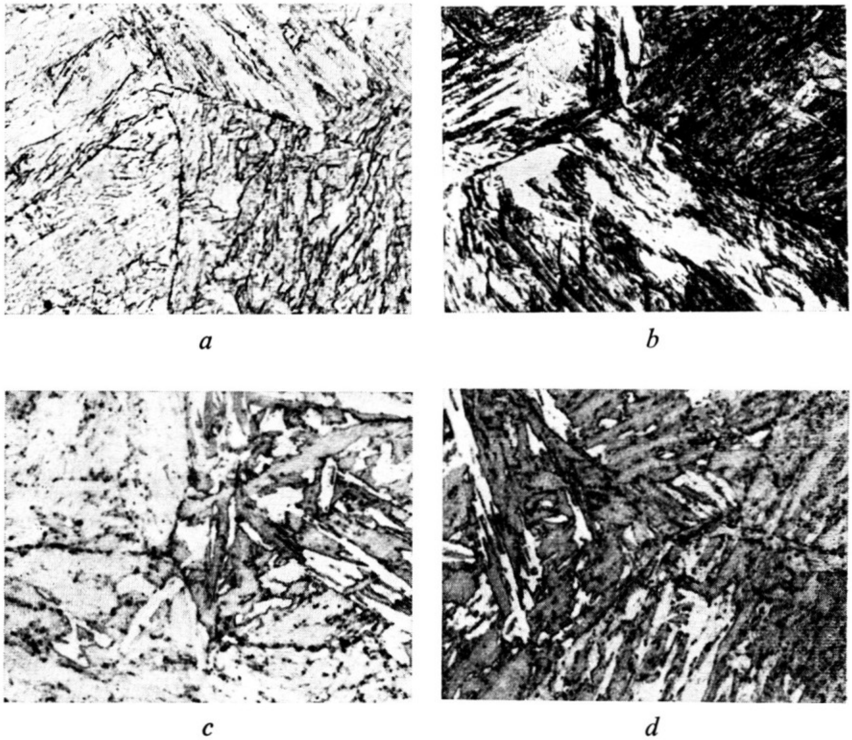

FIG. 3. Microstructures of Steel A. (*a*) as-quenched, (*b*) tempered at 800 °F (425°C), (*c*) tempered at 1100°F (590°C), (*d*) tempered at 1200°F (650°C). All specimens were etched in 2% nital.

(c) Beyond the peak hardness temperatures the hardness decreases rapidly with increasing tempering temperature.

The most marked hardness changes in each stage are shown by Steel A. The microstructures show each steel to be martensitic in the as-quenched condition. On tempering at the temperature corresponding to the decrease of hardness before the peak, the martensite decomposes and the microstructure consists essentially of cementite in a martensitic matrix. At the

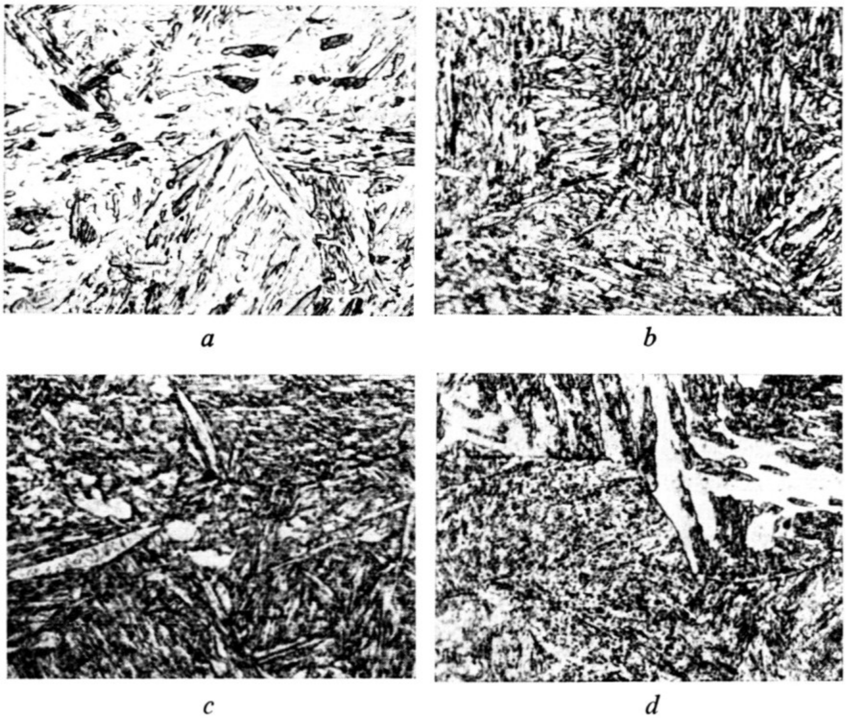

FIG. 4. Microstructures of Steel B. (*a*) as-quenched, (*b*) tempered at 900°F (480°C), (*c*) tempered at 1000°F (560°C), (*d*) tempered at 1150°F (620°C).

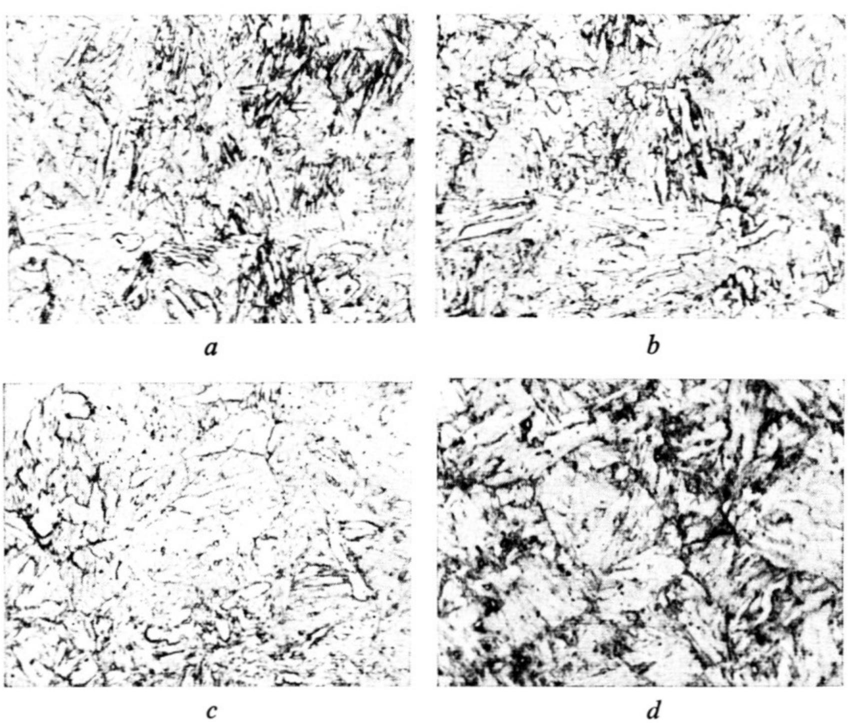

FIG. 5. Microstructures of Steel C. (*a*) as-quenched, (*b*) tempered at 800°F (425°C), (*c*) tempered at 950°F (510°C), (*d*) tempered at 1050°F (565°C).

TABLE 2

Tensile Properties of Steels A, B and C after Various Tempering Treatments

Tempering treatment	U.T.S. kpsi			0·2% P.S. kpsi			Elong. %(4D) in.			R.A. %		
	A	B	C	A	B	C	A	B	C	A	P	C
As-quenched	237·5	296·3	265·7	150·3	186·7	155·0	3·03	1·6	8·3	7·2	9·2	14·6
Minimum*	210·6	223·0	250·7	186·3	190·0	210·5	3·03	11·3	16·0	11·8	31·5	40·7
Peak**	239·4	225·0	280·9	232·4	197·0	211·4	Nil	12·3	12·3	Nil	36·4	36·8
Beyond peak***	224·9	185·1	261·3	165·6	160·0	222·0	2·0	12·6	11·5	6·4	36·9	47·4

* A-800° F (425° C); B-900° F (480° C); C-800° F (425° C).

** A-1100° F (590° C); B-1000° F (540° C); C-950° F (510° C).

*** A-1200°F (650° C); B-1150° F (620°C); C-1050° F (565° C).

peak hardness (Figs. 3*c*, 4*c* and 5*c*) the cementite has been re-dissolved and alloy carbides are beginning to be formed. Beyond the peak (Figs. 3*d*, 4*d* and 5*d*) alloy carbides are clearly visible in a ferritic matrix.

3.2 *The Effect of Tempering Temperature on the Tensile Properties of the Steels*

The tensile properties of the steels tested at room temperature after various tempering treatments are shown in Table 2.

3.3 *The Effect of Tempering Temperature on the Unnotched and Notched Endurance Limits of the Steels*

The endurance limits of the steels are shown in Table 3. All endurance limits were based on no failure after 10^7 cycles.

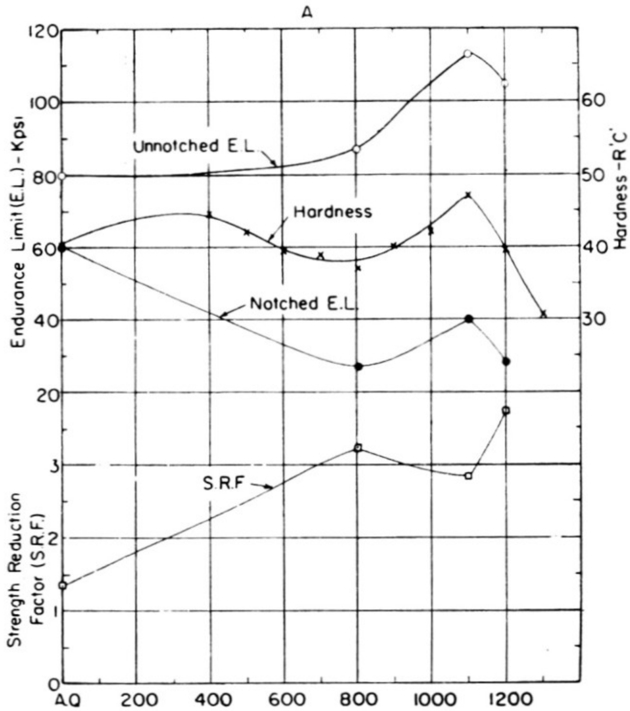

FIG. 6. The relationship between tempering-temperature hardness, notched and unnotched endurance limits and strength reduction factor for Steel A.

The relationships between tempering temperature, hardness, notched and unnotched endurance limits and strength reduction factor is shown in Figs. 6, 7 and 8.

TABLE 3

The Effect of Tempering Temperature on the Fatigue Properties of Steels A, B and C

Tempering treatment	Endurance limit (E.L.) psi at 10 cycles						Strength reduction factor $\frac{\text{Unnotched E.L.}}{\text{Notched E.L}}$		
	Unnotched			Notched					
	A	B	C	A	B	C	A	B	C
As-quenched	80·000	118·000	109·000	60·000	62·000	58·000	1·33	1·90	1·88
Minimum*	87·000	100·000	111·000	27·000	38·000	58·000	3·22	2·63	1·91
Peak**	113·000	107·000	117·000	40·000	56·000	60·000	2·83	1·91	1·95
Beyond peak***	105·000	105·000	119·000	28·000	49·000	60·000	3·75	2·14	1·98

* A-800° F (425° C); B-900° F (480° C); C-800° F (425° C).

** A-1100° F (590° C); B-1000° F (540° C); C-950° F (510° C).

*** A-1200° F (650° C); B-1150° F (620° C); C-1050° F (565° C).

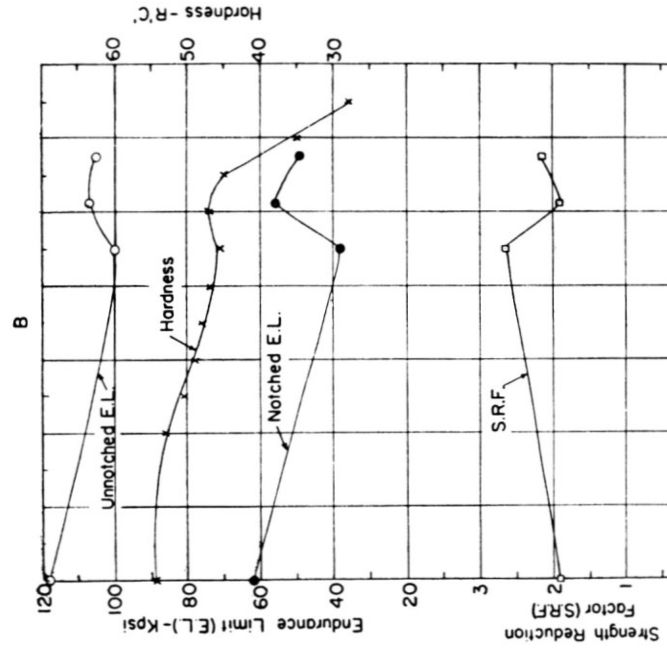

FIG. 7. The relationship between tempering-temperature hardness. Notched and unnotched endurance limits and strength reduction factor for Steel B.

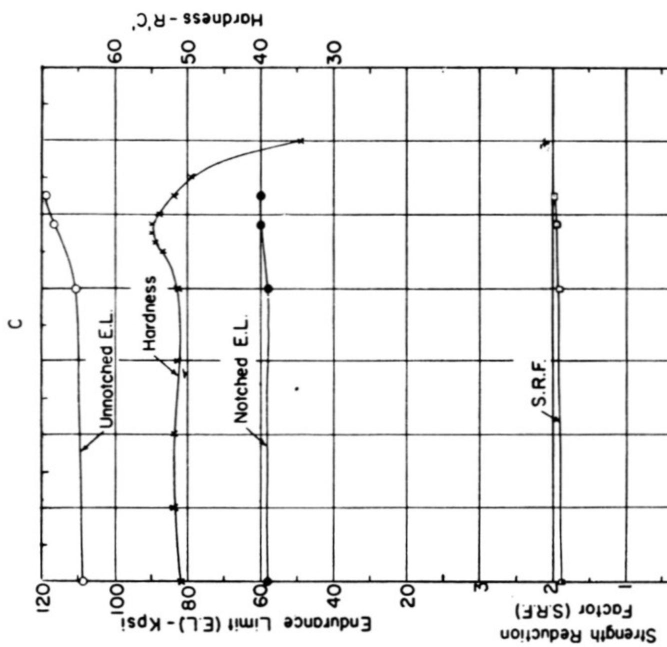

FIG. 8. The relationship between tempering-temperature hardness. Notched and unnotched endurance limits and strength reduction factor for Steel C.

It is clear from these curves that the strength reduction factor for Steels A, B and C is a minimum in the as-quenched condition. It increases on tempering, corresponding to the initial decrease in hardness and then decreases again as the peak hardness is reached. On tempering beyond the peak hardness the strength reduction factor increases, despite a decrease in hardness.

The notched and unnotched endurance limits follow the general form of the hardness tempering-temperature curve.

4. DISCUSSION

The salient point of the results obtained on the steels reported in this paper is that the notch sensitivity or the strength reduction factor does not, as is generally believed, always increase with increasing hardness. The strength reduction factor is related very closely to the metallurgical structure of the steel.

An explanation of this phenomenon may be found by an understanding of the metallurgical properties of the steels in each condition of heat treatment and relating these properties to their effectiveness in relieving peak stresses at the base of a notch. If a peak stress or stress gradient at the base of a notch can be effectively reduced before high levels of stress are developed then the notch sensitivity of the material is reduced. This effect can be understood more clearly if it is discussed in terms of dislocation theory.

During fatigue stressing dislocation movement on active slip planes occurs and according to Stroh[(1)], dislocations may be "piled up" against grain boundaries, substructures, or other barriers in the grain. Theoretical analysis has shown that an active slip plane containing dislocations behaves like a freely-slipping crack, so that it will raise the stress much as a crack will. When the applied shear stress exceeds a critical value the material at the end of the slip line must give way. There are three modes by which it can give way:

(i) It can crack;

(ii) It can slip, the stress being great enough locally to move one plane of atoms over the other without the prior existence of dislocations;

(iii) It can slip, making use of dislocations generated at sources in the vicinity.

It has been shown by Stroh[(1)] that cracking occurs before spontaneous slip so that the relief of a dislocation pile-up is limited to (i) or (iii) above. If (iii) occurs preferentially to (i) then failure will be delayed. It is thus probable that notch sensitivity is related to the ability of the metal to limit the stress concentration at the leading edge of a dislocation pile-up.

Any factors contributing to the generation of dislocations at sources in the vicinity of the leading edge of a pile-up may, therefore, contribute to a lower notch sensitivity or a lower strength reduction factor.

In Steel A the minimum notch sensitivity is found in the as-quenched condition. In this condition, high residual micro-stresses are present, and it is possible for dislocations to pile-up at a martensite plate boundary. In this case, however, it is also possible for the pile-up to be relieved by the easy generation of new dislocations at the martensite plate boundary. Thus, a low strength reduction factor is evident.

When Steel A is tempered at 800° F (425° C) the martensitic structure is decomposed to cementite in a matrix of tempered martensite. This condition is relatively stable and it has been shown[2] that a maximum amount of carbide has been precipitated. In this condition, also, sources for the generation of dislocations in the vicinity of the pile-up (which will probably occur at a grain boundary) are relatively few, and the energy necessary to generate dislocations from the available sources is very much higher than in the as-quenched state. Thus, there is a greater tendency to a high stress concentration at the leading edge of the pile-up and the consequent formation of a crack. The result is a higher strength reduction factor.

On tempering Steel A to the peak of secondary hardness, the cementite re-dissolves and a coherent precipitate of alloy carbide bears a close orientation to the lattice of the matrix and hardening is due to the strain-fields induced in the lattice around the coherent precipitate. In this condition, there is a greater probability of a pile-up being relieved because the application of a stress may be sufficient to cause a change of the coherent precipitate to a non-coherent precipitate by a shearing mechanism with the generation of a new dislocation array. A decrease in strength reduction factor is observed, compared to the 800°F (425°C) tempering treatment, but the strength reduction factor is not as low as that observed in the as-quenched state. This may be due to the fact that the coherent precipitate of vanadium carbide in Steel A is more stable than the other, more complex, alloy steels where the strength reduction factors for the as-quenched and peak-hardness conditions are similar.

Beyond the peak hardness, on tempering at 1200°F (650°C), the coherent precipitate changes to a non-coherent alloy carbide in a matrix of ferrite. In this condition the strain fields around the coherent precipitates are relieved. Thus, when a dislocation pile-up is produced on fatigue stressing it will not be easily relieved, because in the averaged condition the alloy is structurally stable and the generation of dislocations from sources in the vicinity of the pile-up is difficult. Thus, cracking is likely to occur.

Steel B exhibited the same behaviour as Steel A except that the strength reduction factor is a maximum after tempering at 900° F (480° C). Steel C does not show nearly as marked differences in strength reduction factors as Steels A and B but the maximum strength reduction factor is obtained on tempering beyond the peak hardness.

The differences in the strength reduction factors of the three steels for similar tempering points on the tempering-temperature hardness curves can be explained by reference to Fig. 2. Steel A shows a lower notch sensitivity in the as-quenched condition than Steels B and C because the martensitic structure is probably more unstable, thus resulting in the marked change of hardness on initial tempering to 400° F (205° C). Steels B and C have similar as-quenched hardnesses and have similar strength reduction factors in the as-quenched condition.

On tempering to the initial minimum hardness, Steel A shows a greater fall in hardness than Steel B, and Steel B a greater fall than Steel C. The strength reduction factors are also decreased in this order. It is apparent that the greater the drop in hardness the higher is the strength reduction factor. At the same time the greater the fall in hardness, the lower the energy state of a structure, and hence the more limited availability of easily activated dislocation sources.

At the peak hardness of the three alloys, as explained earlier in this discussion, the more stable carbide of vanadium produces a greater strength reduction factor in Steel A than Steels B and C. Also, this carbide loses coherency rapidly beyond the peak and yields a stable matrix containing relatively inert carbides.

Corroborative evidence for the hypothesis outlined above may be found in the results of the work of Dolan and Yen[4] and Russell and Walker[5]. Russell and Walker[5] found that when a high-speed tool steel (0·75% C; 18·5% W; 4·34% Cr; 1·29% V) was tempered at 1340° F (730° C), 1220° F (660° C), and 1130° F (610° C) to produce hardnesses of 40, 55, and 63 Rockwell "C" respectively, the notch sensitivity decreased at the higher hardnesses from 0·73 at 40 Rockwell "C" to 0·42 at 63 Rockwell "C". Although no detailed tempering-temperature hardness curves were given for these steels the peak hardness condition (63 Rockwell "C") corresponds to the secondary hardening peak of this steel. Tempering beyond this peak reduces the hardness and increases the notch sensitivity as shown so clearly by Steels A and B.

Dolan and Yen[4] studied the effect of changes in metallurgical structure on the fatigue strength and notch sensitivity of SAE 1045, SAE 3140 and SAE 2340 steels heat treated to various strength levels. One group of specimens was prepared from oil quenched and tempered stock and another group from austempered stock, showing a bainitic (containing some ferrite and pearlite) and tempered martensitic structure respectively. They found that for equal hardnesses the rapidly quenched samples showed a lower strength reduction factor than the isothermally transformed specimens. This effect, they note, occurs even though it would be expected that steels containing ferrite and pearlite might have more tendency to yield locally, and redistribute the peak stresses over a greater volume, in the region of the notch than steels having finely dispersed carbides. Their results may be explained on the basis of the theory proposed in this present work: the relief of dislocation pile-ups would be easier in a more unstable structure (because of the greater availability and easier activation of dislocation sources) than in a relatively stable isothermally transformed structure consisting of ferrite and pearlite.

The effect of applied static stress on transformation behaviour in steels recently has been studied by Porter and Rosenthal(6), who showed that there is a threshold stress for increased rates of transformation in steels. When the applied stress exceeds a certain value transformation rates are accelerated and when transformation occurs, the yield strength of the steel is suddenly decreased. They explain this on the basis that the threshold stress is the effective stress necessary to cause dislocations to move out from dislocation sources and begin piling up at barriers. The dislocation arrays thus formed produce a large concentration of tensile stress in the vicinity of the leading dislocation. This stress results in increased rates of nucleation and increased rates of transformation. When the nuclei lose coherency with the parent lattice, the dislocation arrays are freed from their barriers. The growing interface acts as a sink, as well as a source for dislocations. Thus dislocation movement is again possible and rapid deformation occurs coincident with transformation. It is interesting to note that, according to their proposed mechanism, either tension or compression should show increased rates of transformation, since the active component of the stress is the shear stress along the slip direction.

The work of Porter and Rosenthal does indicate that transformation processes in steel can lead to the availability of new sources of dislocations which are able to relieve a dislocation pile-up.

The above mechanism of the effect of metallurgical(6) structure is not considered at this time to be accurate or exact in every detail. It is hoped that much more work will be carried out on resistivity measurements and ultrasonic attenuation at various stages of a fatigue test to follow dislocation effects more closely. It is hoped, however, that this discussion does stimulate more thought on the fatigue notch sensitivity of alloy systems and its relationship to metallurgical structure.

The above results and discussion indicates that further work is warranted in studying the notch sensitivity of alloy systems which show the type of behaviour illustrated by the secondary hardening steels. Age-hardenable alloys based on aluminum and magnesium should be tested, and attention should also be paid to the precipitation hardening stainless steels. The present work is to be extended to include the latter alloys, and to study the effect of dislocation pinning by interstitial elements and the grain size of steels on fatigue and fatigue notch sensitivity. It is believed that by studying notch sensitivity it may be possible to contribute to a clearer understanding of the mechanism of fatigue failure.

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