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ACCOMMODATING AND CRACKING MECHANISMS IN LOW-CYCLE FATIGUE

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16. Abstract

The three main stages of fatigue life (accommodation, crack initiation and crack growth) are briefly reviewed. The cyclic behavior of annealed or predeformed face-centered cubic metals is described. Moreover, two types of alloys (Al-4-Cu and WASPALOY) are examined regarding the influence of the interactions between the precipitates and the dislocations on the cyclic behavior. Data on the percent of life to crack initiation (for a microcrack smaller than about 100 μm) are also given. Finally, experimental and theoretical results on crack growth rates in low-cycle fatigue are described.

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ACCOMMODATING AND CRACKING MECHANISMS IN LOW-CYCLE FATIGUE /6*

A. Pineau

ABSTRACT

The three main stages of fatigue life (accommodation, crack initiation and crack growth) are briefly reviewed. The cyclic behavior of annealed or predeformed face centered cubic metals is described. Moreover, two types of alloys (Al-4-Cu and WASPALOY) are examined regarding the influence of the interactions between the precipitates and the dislocations on the cyclic behavior. Data on the percent of life to crack initiation (for a microcrack smaller than about 100 μm) are also given. Finally, experimental and theoretical results on crack growth rates in low-cycle fatigue are described.

INTRODUCTION

The study of plastic fatigue or fatigue early in life ($Nf < 10^4$ cycles) has seen great development in a few years. The interest shown in this type of fatigue is both practical and theoretical. From the practical point of view, calculational methods have been shown necessary for the prediction of lifetime of elements which are greatly stressed a limited number of times (aeronautical and nuclear industries) From a more fundamental point of view, the knowledge acquired by physical metallurgists on the plastic behavior of metals stressed in monotonic fashion has been progressively adapted to the study of the cyclic behavior of metallic materials. Due particularly to observations with the electron microscope, the nature of the metallurgic mechanisms of fatigue damage are beginning to be understood in a more satisfactory manner.

In this article, we examine the physical phenomena involved in plastic fatigue from the structural point of view. Although most often

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the problem of low-cycle fatigue behavior of materials occurs especially when warm (e.g., thermal fatigue), we limit ourselves to the study of plastic fatigue in the neighborhood of ambient temperatures. This important limitation allows us to notably simplify the study of this problem, by eliminating especially what is convenient to call the phenomena of fatigue-creep interaction.

The type of test upon which we will most widely call has now become classic. A thin test sample is submitted to an imposed deformation, most often in traction-compression, the mean deformation ($\pm \Delta\varepsilon p/2$) being maintained constant. This manner of operations avoids the appearance of the phenomenon of cyclic creep which is produced, as a general rule, when the tests are run with an imposed stress. One knows that results concerning the rupture life (Nf) can be represented with the aid of the Manson-Coffin law. These results concerning the rupture include different stages in the lifetime, which we will first recall, and which will then be analyzed. We will examine the accommodation stage in more detailed fashion, calling primarily upon the results obtained on simple alloys in the solid solution state. The behavior of alloys hardened by a second phase will also be called to mind.

I. DIFFERENT STAGES IN LIFE

In performing the type of experiment described above, one first notes that the maximum stress corresponding to the constantly maintained plastic deformation evolves more or less rapidly toward a stable regime, the stress then being essentially constant or changing slowly with the number of cycles. During this stage which we will call that of accommodation and which can represent from 10-50% of the lifetime, the structure of the material changes in order to adapt itself to the imposed cyclic deformation. This accommodation period is followed or overlapped by the stage of the crack initiation, which is manifested by the appearance of microcracks. most often localized at the surface of the sample. One normally notes that at the end of a certain number of cycles, which will be specified later, one of these cracks develops more rapidly. This stage of crack growth, which leads to the complete rupture of the testpiece will also be examined.

II STUDY OF THE ACCOMMODATION STAGE

The accommodation stage has been studied in different metal alloys, notably in copper alloys, to which we will widely refer [1, 2]. An important factor is the initial state of the solid solution which could be either in the annealed state or previously hammer-hardened. With regard to alloys with a second precipitated phase, it seems that one must distinguish the case of alloys containing particles which can be sheared by dislocations from that of alloys containing precipitates which cannot be penetrated by dislocations.

II.1 Accommodation Stage in Solid Solution - Experimental Results

II.1.1 Annealed Initial State-Mechanical Behavior.

The behavior of initially-annealed solid solutions is described schematically in Figure 1. The metal progressively hardens and ends by reaching a stable regime to which one can assign a saturation stress (σ_s). The change of σ_p with the number of cycles (N) or the cumulative /7 deformation ($\Sigma\Delta\varepsilon p$) is schematically-described in Figure 1b where one can note that the stable regime is reached more rapidly as the imposed deformation ($\Delta\varepsilon p$) increases. At saturation, it is convenient to plot what is called the cyclic creep curve which is most often expressed in the form

$$\sigma_s = K' (\Delta\epsilon p/2)^{n'} \tag{1}$$

where n' is the cyclic creep coefficient. This curve can be compared in the curve of monotonic deformation (Figure 1c). In the annealed state one observes a fatigue hardening by comparison to the monotonic deformation.

Several factors can influence the cyclic behavior of metals. By analogy with the monotonic behavior, one expects, especially for metals with cubic centered structures, a notable influence of temperature and speed of deformation. Very few studies, which would nevertheless be very useful for understanding the cyclic behavior, have been done in this area [3, 5]. They could have allowed the decomposition of σ_s into two components $\sigma_3 = \sigma_i + \sigma^*$ where σ_i is the nonthermal component of hardening, related to the long-range internal stresses in the crystal and σ^* is the thermally-activated component, that is, $\sigma^* = f(T, \varepsilon)$ which depends on the nature of the short-distance obstacles which meet the dislocations in their displacement. Another important factor in the case of face-centered cubic metals, is the pile-up defect energy (EFE). In this regard, many studies have been devoted to copper alloys

(Cu-Zn and Cu-Al). One knows, in fact, that the addition of Zn and/or Cu allows one to modify the character of slippage and to avoid or appreciably put off the appearance of deviated slippage. This important point will be recalled later. However, although it is well-established that the EFE of copper decreases with the addition of Zn and Al, one can indicate as of now that the Cu-Zn-Al is undoubtedly not the most

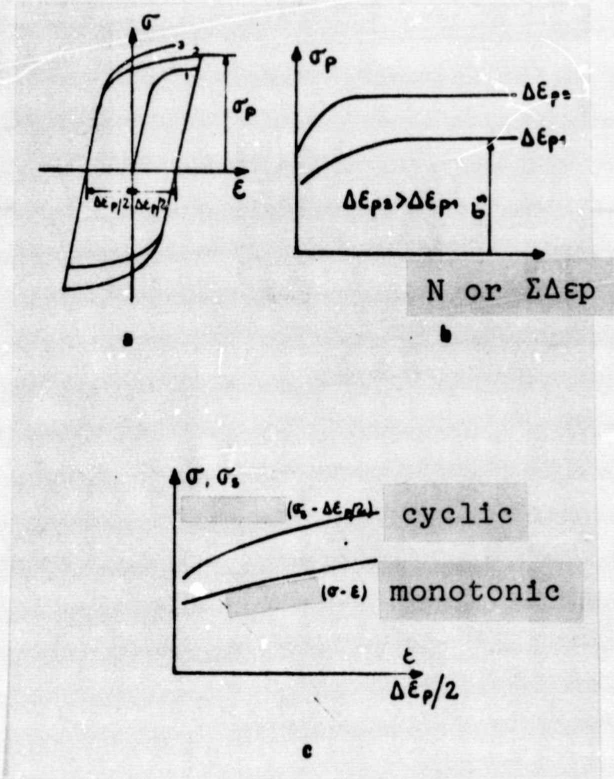

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Figure 1: Cyclic behavior of annealed metals.

judicious choice which can be made to study the influence of the EFE. In fact, in this system the addition of Zn or Al leads to a large solid solution hardening effect, and furthermore, the exact variation of the EFE with composition is not known in this system.

II.1.2 Annealed initial stage - Microstructural changes introduced by the Cyclic Deformation.

The most detailed observations have been carried out on face-centered cubic metals. On copper, during the hardening stage, one observes regroupings of the dislocations into bands following the <112> directions and the formation of numerous dipoles [6, 7; see 6, which is a review]. These microsctructural changes are connected to those observed in stages I and II of the monotonic deformation of copper crystals. As the stress increases, the density of these collections of dislocations increases and one observes the formation of a cellular arrangement of dislocations [1, 2, 6, 7]. The regrouping in cells is better established when the applied deformation is large. On the contrary, in Cu-Zn and Cu-Al alloys the slippage remains very flat and the dislocations preserve the planar configurations. These observations are summarized by Lukas and Klesnel [8].

The typical cells of fatigue at high amplitude have characteristics different from the cells associated with monotonic deformation. The fatigue cells are better formed, their walls are narrower than in simple deformation. In fatigue the walls of cells contain more loops and dipoles and enclose fewer networks of dislocation. One was able to show that the size of the cells becomes smaller as the saturation stress (σ_s) increases [9].

The accommodation stage, likewise, accompanies the formation of point defects. Resistivity measurements indicate that the changes in resistivity recorded in copper are due not only to the increase in density of dislocations during the hardening but equally to the formation of point defects [10].

Another important phenomenon accompanies the saturation stage, that of the formation of bands of intense slippage which was studied in a detailed manner in the range of intermediate and long lifetimes. It is manifested by a heterogeneous deformation localized in these bands as has been recently shown by Finney and Laird [11]. In these bands, one observes that the substructure of the dislocations is different from that of the matrix. In the domain of small deformations, to which correspond arrangements of dislocations in bands in the matrix, these are replaced by cells in the bands of intense slippage, thus showing the localization of the deformation.

II.1.3 Predeformed Initial State - Mechanical Behavior.

In previously deformed states, one observes, as a general rule, a softening once the magnitude of the predeformation exceeds the imposed cyclic deformation. The corresponding behavior is schematically indicated in Figure 2, where one can note that the softening is rapid, especially at the beginning of the cycling and that, in a general fashion, one reaches a stationary regime (σ_s). The cyclic hammerhardening curve is thus located below the curve for monotonic deformation (Figure 2c). This softening phenomenon is also produced in structures which are naturally hardened and results from a phase change of martenistic type, like the martensitism in laths of maraged steel [12]. /8

One can ask under these conditions to what extent a given metal does not possess an applied cyclic deformation curve ($\Delta\varepsilon p = 10^{-2}$ to 10^{-3}). When the cyclic deformations are smaller ($\Delta\varepsilon p = 10^{-4}$ to 10^{-6}), Lukas and Klesnil have shown that there, nevertheless, exists a difference in the deformation curve independent of the former history undergone by the material [2]. This is all the better verified because the applied cyclic deformations are larger (CBp x 10^{-2} to 10^{-3}), when the cyclic deformations are smaller (CBp x 10^{-4} to 10^{-6}). Lukas and Klesnil have shown that there, nevertheless, exists a difference between the cyclic deformation curve obtained starting from the annealed state and that corresponding to a previously hammer-hardened state [8]. In a recent detailed study, Laird et. al. have shown that for copper and nickel, there exists a predeformation below which the cyclic behavior at saturation is independent of the initial state, and above which a residual effect of the predeformation persists [13]. However, in these metals, the difference between the annealed state and the pre-hammer-hardened state remains small, as a general rule.

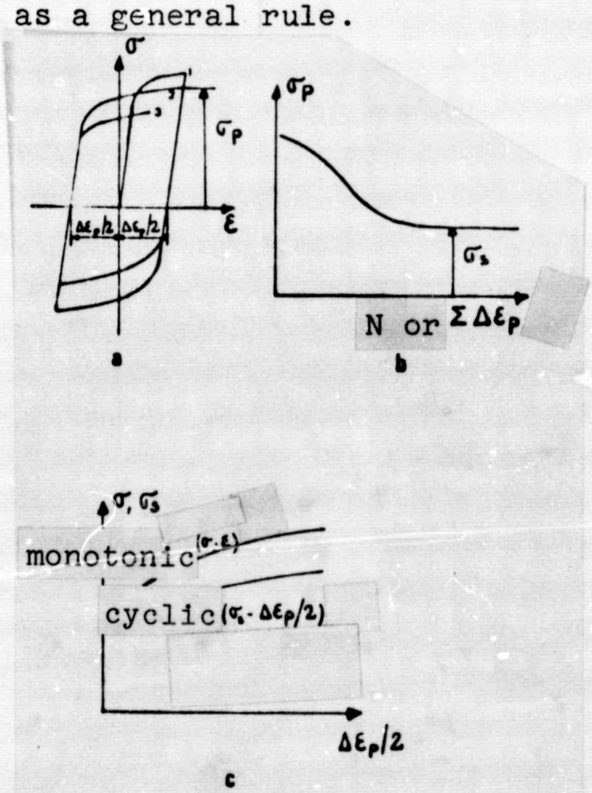

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Figure 2: Cyclic Behavior of predeformed structures.

On the contrary, in Cu-Zn and Cu-Al, it is admitted that the stress at saturation (σ_s) of the predeformed state always remains clearly above that corresponding to the annealed state. It thus seems that the CFC metals with small EFE give rise to a large reversibility contrary to the case of CFC materials with a large EFE.

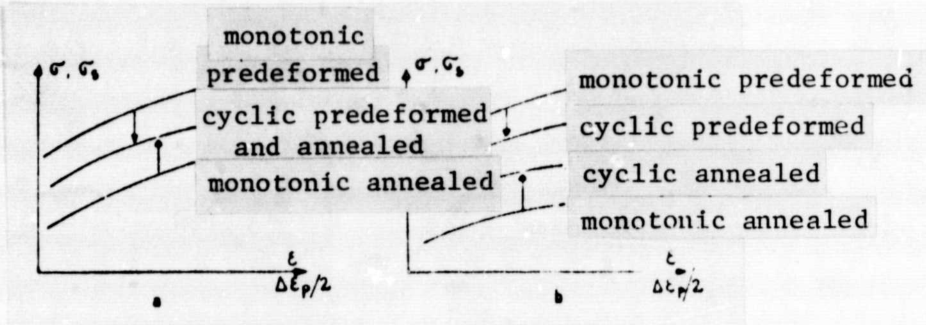

Figure 3: Curves of cyclic and monotonic hardening in the annealed and predeformed state. a) Metals with a reversible behavior (Al, Cu, Ni) b) metals having an irreversible behavior (Cu-Zn, Cu-Al, etc.)

II.1.4 Predeformed initial State - Structural Modifications Introduced by a Cyclic Deformation.

Alloys such as Cu-Zn and Cu-Al have been studied less than the pure metals. In metals like copper and nickel, the dislocation cells introduced by the monotonic predeformation can reorganize themselves in their size and their degree of perfrection during the cyclic deformation. The observations made especially by Feltner and Laird have shown that, when the cyclic deformation is large, the cells corresponding to the monotonic deformation are progressively erased by the cycling of the deformation and one attains, at saturation, sizes of cells which correspond very closely to those obtained by cycling an annealed state to the same deformation [2, 13]. There is thus, reversibility of the microstructure and one realizes that under these conditions there would not be an effect due to the history of the stress at saturation (σ_s). When, on the contrary, the cyclic deformation is small ($\Delta\varepsilon p = 10^{-4}$ to 10^{-6}), the cellular structure of monotonic deformation is not erased. Thus, in this case, while one obtains values for σ_s which are quite close starting from either an annealed or predeformed state, the dislocated substructures in the saturation regime are notably different.

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REMARK:

Supplementary Observations on the reversibility of the microstructure introduced during a cyclic deformation.

One can ask to what extent the absence of influence of the predeformation on the stabilized regime can be applied to the case where the initial state is predeformed in a cyclic and not in a monotonic manner.

Tests run on Cu, Al, and Fe have shown that there does not exist any effect of the cyclic predeformation $\Delta\varepsilon p_1$ on the saturation stress σ_s corresponding to $\Delta\varepsilon p_2$ [2, 14]. Microstructural observations have shown, however, that the situation is more complex [12]. When the cyclic deformations $\Delta\varepsilon p_1$ and $\Delta\varepsilon p_2$ are large in both cases ($\sim 10^{-2}$), one notes a reversibility in the size of the cells. On the contrary, when the cyclic deformations are very different of such a type that the application of $\Delta\varepsilon p_1$ leads to the formation of cells and that $\Delta\varepsilon\, p_2$ gives rise, starting from an annealed state, to an arrangement of debris and bands of dislocations, the deformation $\Delta\varepsilon\, p_2$ taking place with the cycling of $\Delta\varepsilon p_1$, does not permit the erasure of the cellular arrangement of dislocations [8]. Nevertheless, in this case, as in the case of a monotonic predeformation, the stabilized constraint is essentially independent of the history of the material.

II.2 Accommodation Stage in Solid Solutions - Discussion

We limit ourselves to the discussion of the phenomenon of rapid hardening and saturation observed in metals with large EFE, like Cu, Ni, and Al. In fact, the study of the softening of predeformed states is less deep. Furthermore, with regard to the alloys of the Cu-Zn and Cu-Al type, the situation seems less clear to us. It seems curious that in these alloys which lead easily to twinning by deformation, no author has reported the influence of this important and no doubt not easily-reversible mode of deformation on the evaluation of cyclic deformation structures.

With this limitation, at least two important questions remain. 1) By what mechanisms does the crystal progressively harden? 2) In the stabilized regime, how does the deformation operate, or further, what mechanisms control the stationary regime?

II.2.1 Rapid hardening

All observations suggest that this stage is controlled by a phenomenon of defect multiplication. The repetition of the monotonic deformation induces the dislocations to accumulate in bands under the effect of local blocking mechanisms, as in a monotonic deformation. The mobile dislocations are blocked in these bands, and new sources first belonging to the primary slippage system must be activated; this has as a consequence an increase in the density of dislocations and a reduction in the slippage distance. As the stress increases, new slippage systems initially oriented in a less-favorable manner can be activated and it is thus possible to progressively obtain a three-dimensional arrangement of cells. This phenomenon will be produced more easily as the applied deformation increases.

In this regime of rapid hardening, the consolidation mechanisms are those which exist in monotonic deformation; deformation in traction, like that in compression, produces defects. One can think that, as in monotonic deformation, the increase of the stress necessary to pursue the accumulation of the deformation is related to the dislocation density (ρ) and is written $\sigma = A\sqrt{\rho}$ [2]. It must be noted that one part of the observed hardening can equally be due to the creation of point defects.

II.2.2 Stabilized Regime /9

Upon approach to the saturation stress, the accommodation mechanisms change and become characteristic of fatigue. The observations which have shown that the substructure of dislocations is not modified, although large deformations may be accumulated, has led several authors to suggest that, in the domain of high-amplitude fatigue, the imposed deformation is accommodated by the back-and-forth motion of dislocations inside the cells. Grosskreutz [6] has recently discussed this interpretation and shown that in order for this mechanism to be able to operate without the defect density increasing, it is necessary that these be practically the same defects which permit the back-and-forth movement.

The origin of the stress at saturation (σ_s) is discussed further. It is not very reasonable that it is related directly to the size of the dislocation cells, as is nonetheless suggested by the correlation established by Pratt [9]. In effect, the experiments of Lucas and Klesnil have shown that one can obtain the same values of σ_s with very different arrangements of dislocations [8]. Likewise, Chopra and Gowda [14] have shown in iron that if during cycling, one goes from a high value of $\Delta\varepsilon p$ to a smaller one, there is no important modification in the arrangement of cells during the major part of the softening.

II.3 Accommodation Stage in Alloys Containing a Second Phase - Experimental Results and Discussion.

The important class of alloys hardened by the presence of a second phase has been studied in less detailed fashion. One can refer, however, to the study made by Calabrese and Laird on the Al-4-Cu alloy [15]. These authors have studied the cyclic behavior, on one hand, of the Gp and θ" structures which give rise to the shearing of particles by the dislocations and, on the other hand, the case of the θ' precipitates which cannot be penetrated by dislocations. One study made at CM (Centre des Materiaux) and inspired by the same idea was conducted on a nickel-based alloy (WASPALOY) in which, by adjusting the thermal treatment, we have studied the cyclic behavior of this alloy treated to contain the different sizes of the γ' hardening particles (Ni_3TiAl) [16]. We have examined, on one hand, the case of small γ' particles ($\Phi \simeq 80$ Å) which could be sheared by the dislocations and, on the other hand, the case of large γ' particles ($\Phi \simeq 900$ Å) which are, on the contrary, primarily bypassed by the dislocations. These two studies suggest that the two extreme cases corresponding to shearing and bypassing must be distinguished.

II.3.1 Cyclic Behavior in the Presence of Shearing of Precipitated Particles.

In Figure 4a, we have repeated the curves obtained by Calabrese and Laird on Al-Cu-θ", giving the evolution of the maximum stress with the number of cycles for different imposed deformations. The observed behavior is different from that obtained on solid solutions. After a stage of rapid hardening which lasts less long as the deformation increases, one notes a softening. Examples of curves obtained for WASPALOY treated to contain small γ' particles ($\Phi \simeq 80$ Å) are also given in Figure 4b. In this case, one also observes a hardening stage followed by a deconsolidation. However, one can note that contrary to Al-4-Cu-θ' the number of cycles corresponding to the beginnings of the

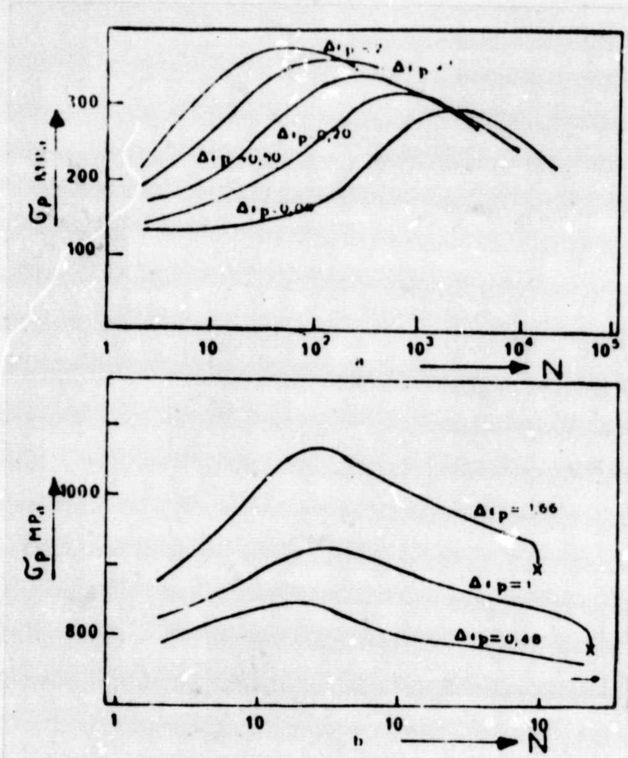

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Figure 4: Cyclic behavior of alloys containing particles which can be sheared by dislocation. a) Alloy: Al-4-Cu, precipitates: θ" [15]. b) WASPALOY, θγ' = 80 Å [16]. The deformations are indicated in %.

softening is independent of the imposed deformation.

Thus for these two alloys, the softening phenomenon is the most characteristic. It has, furthermore, been demonstrated in other alloys [17, 18]. The explanation of the softening is controversial and different explanations have been advanced. One has invoked a return to solution of the precipitates in the slippage bands [10]. This hypothesis which has not been able to be clearly demonstrated is not very reasonable in systems where the diffusion is slow at the temperature of the test. Another hypothesis consists of thinking that a part of the hardening of chemical origin associated with the shearing of the particles is progressively lost during the cycling of the deformation [15]. In fact, for example, in the case of WASPALOY containing small γ' precipitates, electron microscope observations have shown that the /10
deformation remains localized in the bands [16]. In these bands, the travelling back-and-forth of the dislocations which contain flaws is going to progressively diminish the mean area occupied by the particles

in a slippage plane and, consequently, to lead to a decrease in the stress necessary to cross the particles. In schematic fashion, the global cyclic behavior observed results from two antagonistic phenmena (Figure 5). As in the solid solutions, a hardening component is due to the increase in the dislocation density. Upon this is superimposed the softening effect due to the progressive decrease of the efficiency of the hardening of chemical origin associated with the second phase.

II.3.2 Cyclic Behavior in the Presence of Bypassing of Particles by Dislocations.

The curves giving the evolution of the stress with the number of cycles in the case of the Al-4-Cu aged to contain platelets of the phase θ' and of WASPALOY containing large γ' precipitates (Φ $\simeq$ 900 Å), which are bypassed by the dislocations during a major part of the lifetime, are indicated in Figure 6. In the two cases, one observes after

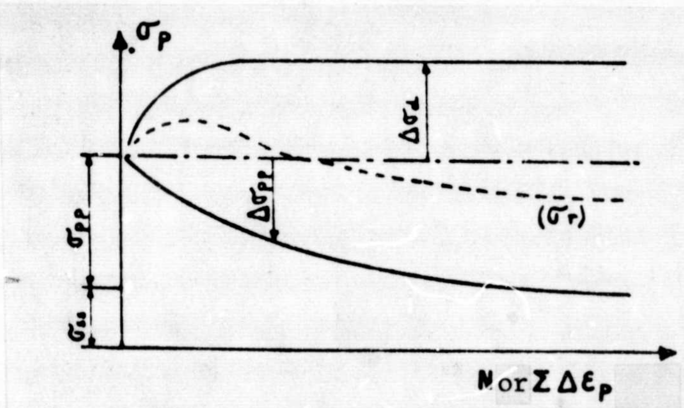

Figure 5: Schematic evaluation of the stress (σ_r) with the cumulative deformation in the case where a part of the precipitation hardening ($\Delta\sigma_{pp}$) is lost. $\Delta\sigma_d$ represents the part of the hardening due to dislocations in a solid solution. σ_{ss} and σ_{pp} designate the hardening of the solid solution and that due to precipitation in the initial state, respectively.

a stage of rapid hardening, a saturation regime. This behavior seems to be characteristic of systems which contain non-sheared particles [20]. Calabrese and Laird [15] have analyzed the structures introduced by the cyclic deformation for two distances (λ) between θ' particles. They chose on the one hand a relatively dense precipitation (λ = 0.18 μm) and on the other hand, a rather large repartition (λ = 0.58 μm). Their observations have shown that, in the two cases, one observes dense skeins of dislocations around the particles, which are those demonstrated after monotonic deformations. In the case of the large repartition and for large cyclic deformations, they further noted of dislocation cells which partly press upon the particles but which are equally situated between the precipitates. This interesting observation underlines the possible interaction between the size of the cells, when these exists, and the distance between particles. In particlar, it showed that in this state (λ = 0.58 µm) there exists, contrary to the other states, a strong influence of the mechanical history undergone by the metal on its cyclic behavior. Thus, a cyclic deformation of $\pm$ 1% which corresponds to the skeins around the particles and to cells, is not completely erased by a second deformation below $\pm$ 0.1%, which, carried out starting with the annealed state, normally corresponds to arrangements of dislocations around precipitates. It is thus not possible under these conditions to define a unique cyclic hardening curve.

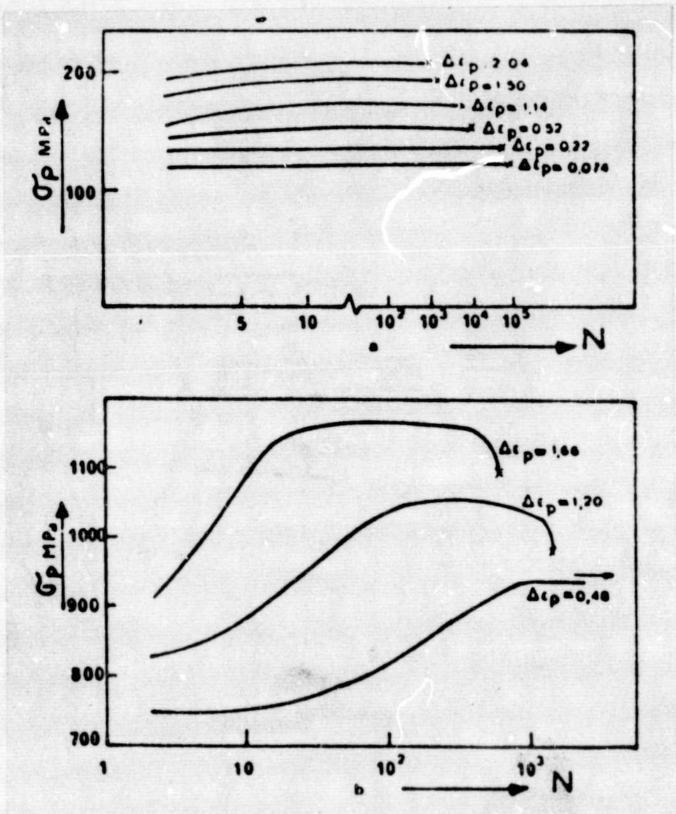

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Figure 6: Cyclic behavior of alloys containing particles which are not sheared. a) alloy Al-4-Cu. Gross repartition of θ' precipitates ($\lambda \sim$ 0.58 µm) [15]. b) Alloy WASPALOY, Large γ' precipitates ($\bar{\lambda} \simeq$ 900 Å) [16]. The deformations indicated are in %.

III. INITIATION OF FATIGUE CRACKS

III.1 Number of Cycles at Initiation

The detailed examination of the initiation of cracking in low-cycle fatigue, as in the domain of vibrating fatigue, strikes at the very definition of a microcrack which depends directly on the resolution of the means of observation used. Further, in the domain of short lifetimes, the definition of initiation also strikes at the fact that, on a test piece, one habitually observes several starting sites.

Also, in the absence of a general agreement on the definition of initiation, it is no doubt preferable to relate the data concerning the number of cycles at initiation (Ni) for a given length of defect (a_o). This length could be of the order of 100 μm, for example. In fact, this dimension could be detected rather easily. It corresponds, furthermore, to a defect size comparable to the size of a grain. Finally, experience teaches that in a good number of cases, as soon as a microcrack reaches a depth near this dimension it propagates regularly across the section.

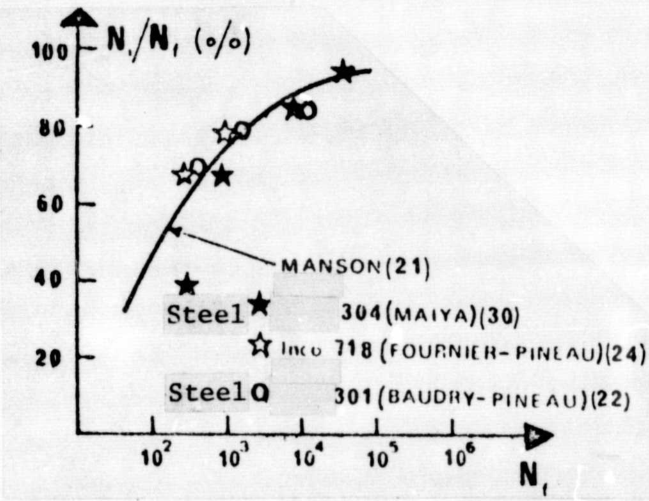

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Figure 7: Evolution of the ratio (Ni/Nf) with the lifetime Nf. The principle curve is taken from S. S. Manson, to which are added recent results.

The variation of the ratio Nf/Ni with the lifetime at rupture (Nf) for a_o = 500-1000 μm is shown in Figure 7, which includes data concerning different types of materials. These results indicate that in the domain of lifetimes on the order of 10^3 cycles, the number of cycles passed at the initiation, such as is defined, represents a non-negligible proportion of the lifetime.

III.2 Nature of Initiation Sites

While the study of the initiation of plastic fatigue is less developed than that relating to the domain of vibratory fatigue, it seems that the nature of the sites of the start of cracking are the same in the two cases.

It is thus that one observes, as a general rule, the initiation is produced at the surface of the test sample or also at interfaces (matrix-inclusion interfaces, grain junctures). Even in the absence of microstructural embrittlement of joints or emrbittlement of joints related to the envirionment (hot fatigue oxidation) the junctures of particles can constitute starting sites which are more privileged as the deformation increases [23]. At large amplitudes, the initiation mechanism could in fact be purely geometric. Under the influence of strong imposed deformations, the slippage bands completely invade the particles and the free surface of the test piece takes on a tormented aspect, with notable changes in form at the level of junctures which cut the free surface. One thus obtains a natural groove effect and deep intrusions develop at the joints which can give birth to micro-cracks in materials such as Cu and Al. This effect will become more important as the particle size increases, as is seen in the results published by Boettner, et al. [23].

When the initiation is intragranular, the bands of intense slippage are the natural starting sites for fatigue cracks. The mechanisms which control initiation in the slippage bands are not yet perfectly elucidated. In the case of alloys which include a second phase, the nature of the interaction between the dislocations and the particles is important.

If the precipitated particles can be sheared in the slippage bands and this mode of crossing leads to an easier accumulation of the deformation in the slippage bands, one can think that this will lead to a relative diminuition of lifetime at initiation. An example of this type of initiation is given in Figure 8, relatively to the alloy INCO 718. In this type of alloy, we have shown that the hardening particles (phase γ'', type Ni_3Nb) can be sheared and that the deformation concentrates in bands containing mechanical twins [24]. The concentrations of the deformation in these bands gives birth to intrusion-extrusions and to the beginning of cracking in the bands or at the interfaces between the bands and the matrix.

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Figure 8: INCO 718 alloys, $\theta = 25°C$, $\Delta\varepsilon p/2 = \pm 0.62\%$. Surface of the test sample observed with a scanning microscope - Note the extrusions across the deformation bands.

It thus seems that every structural modification which will lead to a more homogeneous repartition of the deformation across the grain will be reasonably favorable for the behavior at initiation. In particular, it can be preferable to use overaged states for which the particles are bypassed, rather than underaged states most often associated with the shearing of the precipitates.

IV. PROPAGATION OF FISSURES IN PLASTIC FATIGUE

The propagation state represents an important aspect of the lifetime, because it is small. The usual classification of the geometric factor of propagation in stage I and stage II also applies to /12
plastic fatigue tests. The extent of stage I will be smaller the greater tha applied deformation, in such a way that the traction propagation corresponding to stage II will become larger.

IV.1 Crack Propagation Mechanisms

The appearance of a fracture surface of a stainless steel tested in the low-cycle fatigue domain is shown in Figure 9. For relatively short lengths of crack, and thus rather slow speeds of propagation per cycle, (da/dN), one observes the classic, well-defined stria. For larger speeds, undulations remain present. The distance between these undulations corresponds to the advance of the crack per cycle. Further, between these undulations, one notes the presence of fine lines essentially parallel to the crest of the undulations. These observations correspond to what is commonly seen in fracture surfaces of test samples in low-cycle fatigue [25]. One also often notes the presence of inclusions, of ductile tearing in the neighborhood of those, and secondary cracks.

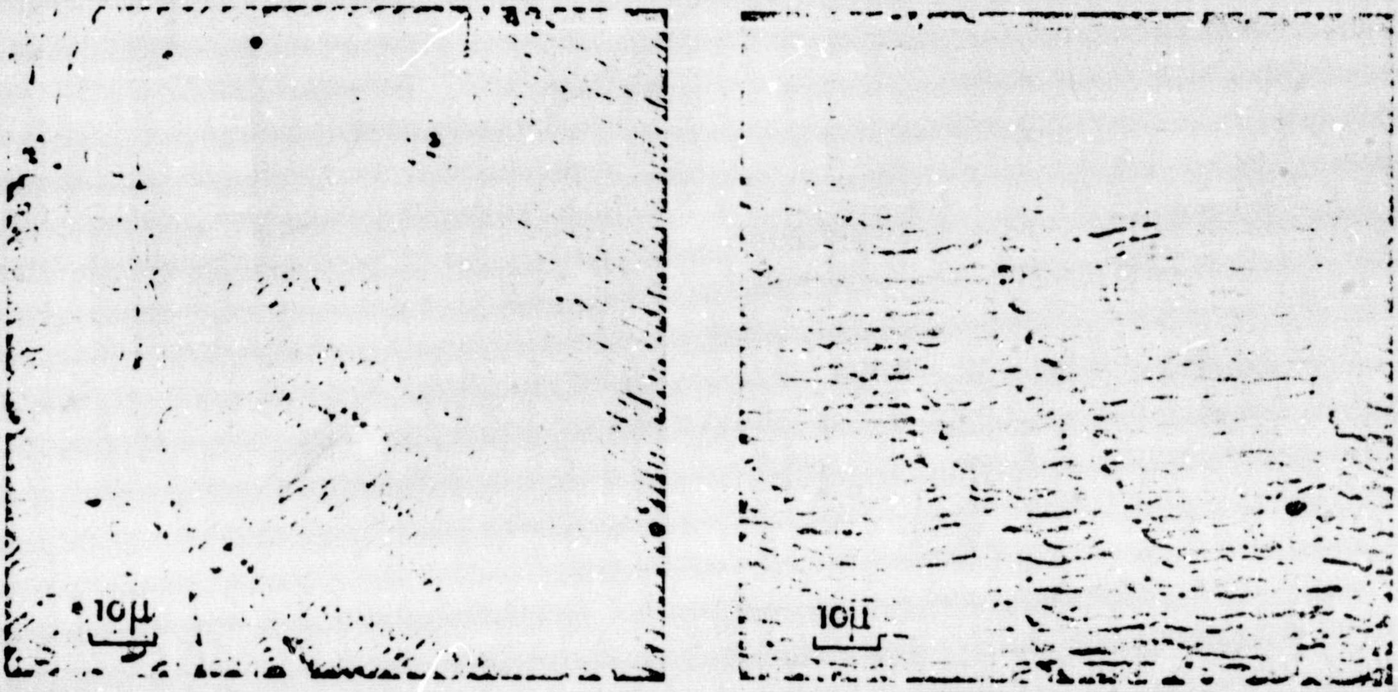

Figure 9: 301 Steel Plastic fatigue, $\theta = 200°C$, $\Delta\varepsilon p/2 = \pm 0.4\%$. Appearance of fracture surfaces at two crack depths.

a) a = 0.70 mm, da/dN = 1.8 μm/cycle.
b) a = 1.50 mm, da/dN = 4 μm/cycle.

The mechanism of formation of stria proposed by Laird [26] has been taken up and studied by Tomkins and Biggs [25] for the case of the propagation of cracks in the short lifetime domain. It allows the explanation of the appearance of undulations.

IV.2 Speed of Propagation of Cracks in the Plastic Fatigue Domain.

Few studies have been devoted to the measurement of the speed of propagation of cracks (da/dN) in the domain of the short lifetimes. The existing results most often concern indirect measurements, that is to say, the determination of the distance between stria or undulations as a function of the depth (a) of the principal crack. Direct measurements have also been made [27, 28].

The results obtained to present seem to indicate that the speed of propagation is regulated by purely mechanical phenomena, to the extent which it is possible to correlate da/dN with a, $\Delta\varepsilon p$ or the stress σ_s corresponding to the stabilized regime.

Such a correlation can be written in the form:

$$da/dN = A.a\,(\Delta\epsilon p)^{\alpha} \tag{3}$$

where A is a constant for a given material. Examples are given in Figure 10, which summarize the tests conducted by H. Solomon on a carbon steel [27]. We have also included results we ourselves obtained on an unstable stainless steel of 301 type, tested at 25°C and 200°C [22]. At 200°C, this steel remains austenetic throughout its lifetime. On the contrary, at 25°C, it can be transformed into martensite ε and α', which has the effect of considerably modifying the form of the cyclic hardening curves. In the interval of $\Delta\varepsilon p$ where the velocity measurements were done, one finds a value of n' at 25°C clearly higher than at 200°C, since at 25°C n' = 0.80, while at 200°C, n $\approx$ 0.45.

These few results suggest that the slope depends directly on the cyclic behavior, and in particular, on the value of the coefficient n'. Such a dependence appears directly in the model proposed by Tomkins for the propagation of cracks in the plastic domain [20]. This author has proposed that, for ductile materials, the progression of the crack per cycle (da/dN) is directly proportional to the crack base /13 opening, and he has established the following relation:

$$da\,dN \alpha\, a\left(\frac{\sigma_s{}^2}{T}\right)(\Delta\epsilon p) \tag{4}$$

where T is the "resistance to cyclic traction" of the material. While the manner in which this relationship was established could be criticized, equation [4] has the advantage of emphasizing the importance of

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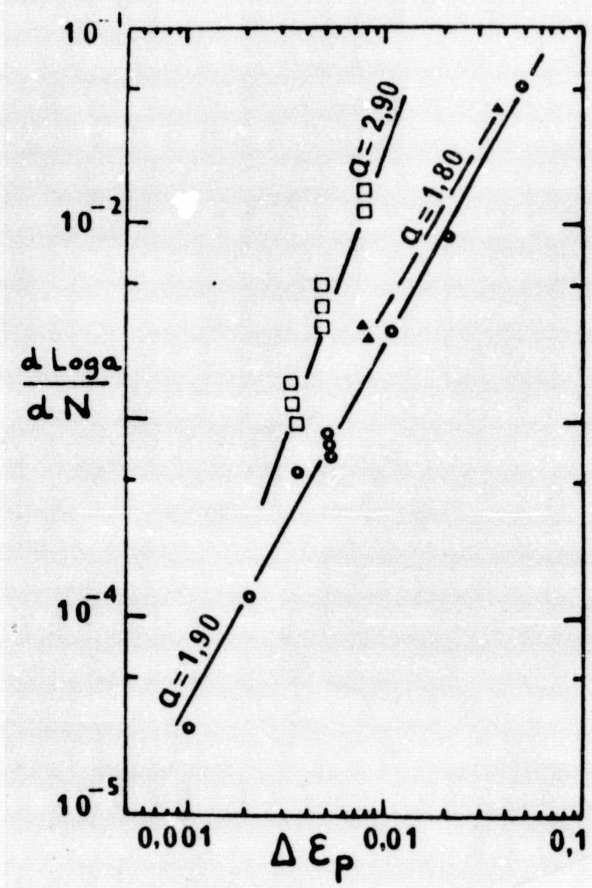

Figure 10: Variation of the speed of propagation in low-cycle fatigue with p for two types of steel.

O——O Carbon steel (n' = 0.26), from H. Solomon [27].

Δ——Δ 301 Steel, θ = 200°C, (n' ≈ 0.45) } [22]

□——□ 301 steel, θ = 25°C, (n' ≈ 0.80)

stress as the results obtained for 301 steel suggest, using the expression of the cyclic hardening low to relate α and n', since $\alpha = 1 + 2n'$.

One can also note that the integration of relation [4] permits the prediction that the propagation lifetime is regulated by a law of the Manson-Coffin type. This law is most often established for rupture lifetimes. As the ratio Ni/Nf for a_o = 50-100 μm is on the order of 0.50 in the range $10^{\pm 2} < Nf < 10^4$, one can remark nevertheless that to within a factor of about 2, the rupture lifetime can be predicated by taking into consideration only the propagation stage. Further, if one thinks of applying it in the case of biphase alloys which had been previously indicated concerning the influence of the mode of crossing the precipitate, one can expect that in a good number of cases, the rupture data would be relatively independent of the state of precipitation. The influence of shearing and bypassing on the initiation and on the propagation can counterbalance each other. The shearing can lead to a premature initiation, but the softening which is associated with it as a general rule will permit a slowing of the progression of the cracks.

CONCLUSIONS

1. Studies concerning the phenomena of cyclic hardening and softening have permitted the extraction of precise information for the metallurgist and mechanical engineer. However, even in the case of solid solutions, there remains much to do to attempt to relate the stabilized stress (σ_s) to the substructure introduced by the cyclic deformation. In this regard, studies aiming to decompose σ_s into its thermal and nonthermal components would be seen. The results concerning the reversibility of the cyclic behavior are useful to the mechanical engineer interested, in particular, by these aspects in programmatic fatigue.

2. The number of cycles necessary to initiate and propagate microcracks of dimension $a_o \simeq 100$ µm represents a non-negligible part of the lifetime.

3. The cyclic behavior of materials and their behavior relative to the initiation depends narrowly on this mode of deformation. In these stages, the metallurgist could think of developing microstructures favorable to bearing in low-cycle fatigue. In particular, in the case of alloys with a second phase, one can think that the dispersion of the slippage associated with the bypassing of the particles by the dislocations will be beneficial to the bearing at initiation.

4. The experimental and theoretical studies relating to the propagation phase show that the speed of propagation of cracks in plastic fatigue is regulated by continuous phenomena. It is possible to correlate the speed of propagation to the amplitude of the imposed deformation and the corresponding stabilized stress.

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