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I have reviewed this completed senior honors thesis with this student and certify that it is a project commensurate with honors level undergraduate research in this field.

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ADVANCES IN ULTRA-HIGH CYCLE FATIGUE

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ABSTRACT

The fatigue of metals has been extensively studied. However, most published research does not extend past around 10^7 cycles. Because plots of the stress versus number of cycles to failure (S-N curves) of ferrous alloys and some other metals apparently reach a horizontal asymptote, it was assumed that specimens tested at stresses below the asymptote, called the fatigue limit, would have infinite lives. However, recent research has discovered fatigue failures at stresses below the fatigue limit and lives above 10^7 cycles, termed ultra-high cycle fatigue (UHCF). This paper reviews published research in this area. The S-N curves found in this research are presented, illustrating the location of crack initiation in the UHCF region, the existence of primary and secondary plateaus, and the effects of test frequency, temperature, and environment on fatigue behavior in the UHCF region. This paper also reviews several mechanisms and models for UHCF, including the $\sqrt{area}$ parameter model, the slip mechanism, hydrogen-embrittlement mechanism, and fatigue crack initiation at porosities and inhomogeneities.

INTRODUCTION

Traditional fatigue analysis identifies in some metals a fatigue limit, which is the stress limit below which the metal will not fail after an infinite number of cycles [1]. In these materials, primarily steels, plots of the stress versus number of cycles to failure (S-N curves) exhibit an apparent horizontal asymptote around 10^5 to 10^6 cycles. Failure at lives below about 10^5 cycles is termed low-cycle fatigue (LCF). Failure at lives above about 10^5 cycles has been termed high-cycle fatigue (HCF). However, the calculation of traditional fatigue limits is based upon measurements to a maximum of about 10^7 cycles. This approach was safe in the past because the fatigue lives of machinery were well below 10^6 cycles. However, many modern applications, for example in the transportation [2] and electronic [3, 4] industries, can require fatigue lives of over 10^8 cycles. In recent years, research has begun to extend S-N curves above 10^7 cycles, termed ultra-high cycle fatigue (UHCF). In the mid 1980's, failure above 10^7 cycles was first reported. However not until recently, in the late 1990's, did UHCF research begin in earnest. This area of research is just beginning to be fully explored. The amount of published results of UHCF is still relatively small. Failure at stresses significantly below traditional fatigue limits has been discovered. In order to ensure safe design, the fatigue behavior of materials must be examined above 10^7 cycles. This paper reviews the recent research conducted on the UHCF of metals.

ULTRA-HIGH CYCLE FATIGUE (UHCF)

Several studies have been completed, exploring the gigacycle fatigue behavior of metals. Some of their results are summarized in the following sections. The researchers found fatigue failures in the UHCF region at stresses below the conventional fatigue limit. A typical example is seen in the S-N curve for a 1Cr-Mo steel with $R = -1$ (Figure 1, [5]), where R is the ratio of the applied minimum to maximum stresses. A plateau corresponding to the conventional fatigue limit can be observed from approximately 7×10^5 to 2×10^7 cycles, followed by failures in the UHCF region, up to about 6×10^7 cycles.

The fatigue behavior of non-ferrous alloys has also been examined [1, 6, 7]. UHCF behavior similar to that for steels has been found for a 2024/T3 aluminum-base alloy and an ULTIMET cobalt-base superalloy (Figures 2 and 3, respectively, [1, 6, 8]).

Location of Fatigue Crack Initiation

When the researchers examined the location of the fatigue crack initiation, all found that the fatigue cracks of specimens that failed in the UHCF region originated in the interior of the specimen, usually at non-metallic inclusions. Conversely, specimens that failed earlier did so by cracks that originated on the surface. The boundary between surface initiation and internal initiation is generally at the HCF plateau and very distinct. This observation holds true for both ferrous (Figures 4 [9] and 5 [10]) and non-ferrous alloys, specifically the ULTIMET superalloy (Figure 6, [6, 8]). The mechanism behind this phenomenon will be examined later in this paper.

In order to further investigate the influence of inclusions on UHCF, Bathias [7] conducted fatigue tests on both standard specimens of a N18 nickel-base alloy and specimens of the same alloy seeded with inclusions. The fatigue strengths of the seeded specimens were found to be lower than those of the standard specimens, particularly with an R ratio of 0.8 (Figure 7, [7]). The effect was less pronounced with an R ratio of zero. These results confirm the role of inclusions and suggest that their importance is greater at larger R ratios. In Figure 6, smaller R ratios result in larger stress amplitudes, and therefore, more damage, which masks the inclusion effect on fatigue life. Murakami, Toriyama, Tsubota, and Furumura [11] went the other way;

instead of increasing inclusions, they decreased them. Super-clean specimens of SAE52100 bearing steel were prepared by an electron beam remelting process. It was observed that some fatal cracks initiated at bainite inhomogeneities rather than inclusions. In the super-clean specimens, the size of the inclusions was so small that the effect of inhomogeneities became prevalent.

Existence of Primary and Secondary Plateaus

Some materials were observed to fail in the UHCF region, but did not clearly show the HCF plateau seen in other metals (as in Figures 1 [5] and 2 [1]). For example, no clear HCF plateau was observed in S-N curves of SCM 435 Cr-Mo steel with a carburized/nitrided surface (Figure 8 [10]). Nishijima and Kanazawa [5] suggest a possible explanation for the absence of a HCF plateau. They argue that due to the lack of plasticity, the internal crack growth rate for harder materials is faster than for less hard materials. Therefore, the UHCF region of the S-N curve, attributed to internal cracks, will move to shorter lives, shrinking the HCF plateau transition region, eventually to the point where it is no longer observed. The disappearance of the plateau in a hardened steel (Figure 8 [10]) supports this assertion.

The S-N curves of some materials suggest that a second plateau exists at extremely long lives. For example, a secondary, UHCF, plateau is seen for N18 nickel alloy in Figure 6 tested at 450°C. The S-N curve of ULTIMET superalloy (Figure 3 [6, 8]) shows a possible secondary plateau starting near 5×10^6 . The S-N curve for a medium carbon steel (Figure 5 [10]) shows a possible secondary plateau starting near 5×10^8 . Possible explanations for the presence of the secondary plateau will be discussed later in this paper.

Environmental Effect

The effect of environment on UHCF has also been studied. Fatigue test results from an ULTIMET superalloy (Figure 9 [6, 8]) show that conducting the test in a vacuum, instead of air, shifts the S-N curve to longer lives [6, 8]. It is also observed that in the UHCF region, the environmental effect is less. The longer fatigue life in vacuum than in air is explained by the absence of environmental degradation of the surface when the test is performed in vacuum. In the UHCF region, crack initiation is internal. Therefore, the environmental effect would be

expected to be less in this region than in the LCF and HCF regions. This is verified by the experimental data above.

Frequency Effect

One important topic to consider when discussing UHCF is test frequency. In order to conduct long life tests in practical amounts of time, they must be run at high frequencies (in the range of kHz). For example, a test up to 10^{10} cycles would require almost 16 years at 20 Hz, but only 6 days at 20 kHz. Ultrasonic machines have been constructed, which allow testing at high frequencies in the kHz range [12, 13]. Another type of high frequency machine is electrohydraulic. For example, the MTS 1,000 Hz 810 electrohydraulic machine is capable of testing at frequencies up to 1,000 Hz [14, 15]. Muhlstein, Brown, and Ritchie [3] have developed a method to test the fatigue behavior of thin films of silicon at very high frequencies, namely 40 and 50 kHz. The micron-sized structure tests a notched, cantilever beam, which is attached to a resonant mass and vibrated at a resonant frequency.

Some results seem to indicate that the frequency has a significant effect on the results of fatigue testing [14, 16]. Examining the HCF plateau and beginning of the UHCF region of Figure 6 [6, 8] shows that shorter lives are seen at 1000 Hz than at 20 Hz. Fatigue tests run on 316 LN stainless steel (Figure 10 [16]) show that the fatigue lives in air at 10 Hz are longer than those at 700 Hz. It was observed that the specimen temperature reached a maximum of 270°C during the 700 Hz test. The elevated temperature decreased the strength of the material, thereby reducing the fatigue life. When the test at 700 Hz was carried out with nitrogen cooling to 20 to 70°C, the frequency effect was smaller. These results support the conclusion that the majority of the frequency effect is due to specimen heating at high frequencies. Other researchers [17] have also found that high frequency data corresponds well with low frequency data when the specimen is cooled. Other frequency effects are a result of time-dependent processes, such as environmental and diffusional processes [18]. Since a high frequency test takes a shorter amount of time for the same number of cycles, there are fewer opportunities for environmental damage and diffusion, and therefore, the life would be expected to be longer than for a low frequency test. Test frequency can also affect fatigue performance if the frequency is greater than the dislocation motion, in which case, the motion of dislocations will be impeded, increasing fatigue life [19].

Temperature Effect

Elevated temperatures, such as those associated with high frequency testing, can adversely impact fatigue performance. Lowered temperatures, on the other hand, can increase fatigue performance as shown in Figure 11 [7], which plots the S-N curves of a Ti6246 titanium alloy at 24°C and -196°C. The fatigue strength in the UHCF region is much higher at the lower temperature. The UHCF plateaus are near 375 and 625 MPa for the high and low temperatures, respectively.

UHCF of Silicon

Using the micron-scale structure previously discussed, research has begun to explore UHCF of silicon [3,4]. The high frequencies of the tests (40 kHz and 50 kHz) have allowed tests to be run up to 10^{11} cycles. The S-N curves of both monocrystalline and polycrystalline thin film silicon are presented in Figure 12 [3, 4]. Both types of silicon exhibited fatigue failures up to 10^{11} cycles at stresses as low as half of the fracture strength. Both also exhibited asymptotic behavior, suggesting that a true fatigue limit might exist. A second plateau, as seen in metals, was not observed. The fatigue strength of the monocrystalline silicon was observed to be greater than that of the polycrystalline silicon, as expected, because a single crystal of silicon has a higher fracture strength than polycrystalline silicon. UHCF of thin films of silicon has direct applications in microelectromechanical systems (MEMS). MEMS are generally constructed of silicon and often used in applications such as computing, which subject the MEMS to many cycles at low loads.

MECHANISMS

In the previous section, results from UHCF tests have been presented and the effect of testing parameters examined. In this section, the micromechanisms behind UHCF will be studied.

Internal Crack Initiation

Fatigue cracks in the LCF and HCF ranges are generally initiated on the surface. However, it is accepted that fatigue crack initiation in the UHCF range moves to the interior [5,6,7,9,10,20,21]. Though internal cracks can also initiate at porosities and inhomogeneities, they most often initiate at inclusions, giving rise to fisheye-type microstructures. A fisheye is a radial microstructure with a small bright spot, the inclusion, at the center. Cracks initiate at inclusions by three possible mechanisms: cracking of the particle, debonding of the particle from the matrix, and formation of slip bands near the particle [21].

$\sqrt{area}$ Parameter Model

Several researchers have investigated the effect of inclusion size on fatigue behavior [11,19]. Murakami, Toriyama, Tsubota, and Furumura [11] have developed a model, below, to predict the fatigue limit (σ_w) of metals with non-metallic inclusions, based upon the size of the inclusion. This model is called the $\sqrt{area}$ parameter model.

$$\sigma_w = \frac{C(HV + 120)}{(\sqrt{area})^{1/6}} \cdot \left(\frac{1 - R}{2} \right)^\alpha \quad (1)$$

The constant, C , depends on the location of the inclusion ($C = 1.43, 1.56$, and 1.41 for an inclusion at the surface, in the interior, or just below the surface, respectively), $\sqrt{area}$ is the square root of the area of the inclusion, and HV is the Vickers hardness. The coefficient, α , depends on the hardness as follows, $\alpha = 0.226 + HV \times 10^{-4}$. The effect of inclusion size on fatigue behavior can be normalized by plotting the relative stress versus number of cycles to failure, where the relative stress is the ratio of the applied stress to the predicted fatigue limit calculated using Equation 1. Figure 13 [19] shows both the normal and modified S-N curves of

the same data for SNCM439 steel. In the unmodified plot (Figure 13a), the scatter in the data is quite large, even among tests run at the same frequency. A plateau seems to emerge near 10^9 cycles. Normalizing the data for the inclusion size, using fatigue limits calculated with Equation 1, lowers the scatter of the data, allowing a clear trend line to be drawn in the modified curve (Figure 13b). It may also be seen that the plateau near 10^9 cycles closely corresponds to the fatigue limit predicted using Equation 1. Wang, Berard, Dubarre, Baudry, Rathery, and Bathias [17] have developed an empirical modification of the $\sqrt{area}$ parameter model, which predicts the fatigue life at a specified number of cycles, rather than the fatigue limit. In this modified model, the constant, C , is replaced by a parameter, β , which equals $3.09-0.12 \log N_f$ for inclusions in the interior and $2.79-0.108 \log N_f$ for inclusions on the surface.

Slip Mechanism

Several papers [5,18,21] propose a fatigue mechanism based upon slip. Mughrabi [21] classifies materials into Type I, which are single-phase materials with no internal defects, and Type II, which contain internal defects. According to this model, fracture in the LCF and HCF regions for both Type I and Type II materials occurs due to crack initiation on the surface, resulting from the formation of surface roughness by persistent slip bands (PSBs). The HCF limit, the first plateau in the S-N curve, corresponds to a threshold value below which PSBs will not form.

In Type II materials, once the stress is sufficiently low to remove the possibility of formation of surface roughness by PSBs at the surface, cracks initiated at internal inclusions become dominant. Cracks at internal inclusions dominate the UHCF region, because the probability of finding an inclusion in the interior of the specimen is greater than on the surface. This was quantified by Mughrabi [21] for the case of a cylindrical specimen in Equation 2 below, where N_s and N_v are the number of inclusions in the surface layer and in the entire volume, respectively, and d_i and d are the diameters of the inclusion and the specimen, respectively.

$$\frac{N_s}{N_v} = 4 \frac{d_i}{d} \quad (2)$$

The volume density required to find at least one inclusion on the surface may be calculated using Equation 3 below, also developed by Mughrabi [21], where l is the length of the specimen.

$$n_{crit} \geq \frac{1}{\pi \cdot d \cdot d_i \cdot l} \quad (3)$$

Though the probability of finding an inclusion in the interior is greater than on the surface, cracks initiated at internal inclusions will grow more slowly than cracks initiated on the surface, due to a lower fraction of irreversible slip. The lower fraction of irreversible slip could be due to the isolation of internal cracks from the environment. In LCF and HCF, oxidation speeds crack growth on the surface by preventing full crack closure, thereby decreasing the life. However, in UHCF, interior cracks are isolated from the effects of the environment, causing the crack growth rates to be slower and resulting in longer fatigue lives. Nishijima and Kanazawa [5] also point out that the stress intensity factor for internal cracks is less than for external cracks, contributing to the slower crack growth in the interior as compared to the surface. Therefore, failure by internal cracks will be seen at longer lives, in the UHCF region, than failure by surface cracks, in the LCF and HCF regions.

In Type I materials, there are no internal defects for cracks initiation. However, surface roughness will still form, below the PSB threshold, due to irreversible slip, allowing surface crack initiation. This process is slower by slip than by PSBs, leading to longer lives. Surface roughness also forms in the UHCF region in Type II materials, but its effect is hidden by internal initiation at defects. The slip mechanism proposes a second plateau, which corresponds to the stress value below which the fraction of irreversible slip becomes negligible.

Hydrogen Embrittlement Mechanism

Murakami, Nomoto, and Ueda [10] propose a different mechanism for UHCF. This mechanism is still based upon internal initiation at inclusions, but it proposes the underlining cause to be hydrogen embrittlement combined with fatigue. Hydrogen tends to gather at inclusions, especially over the long times considered in UHCF. The presence of hydrogen assists the formation and growth of cracks at inclusions until they reach a critical size above which they grow on their own in the typical manner. This assertion is supported by the observance of elevated amounts of hydrogen in optically dark areas (ODAs) surrounding fisheye fractures. Because this mechanism depends on a time-dependent process, diffusion, it will be manifested at

longer times and lives. In addition, because it is time-dependent rather than cycle-dependent, test frequency would be expected to affect UHCF behavior. However, further research by Furuya, Matsuoka, Abe, and Yamaguchi [19] indicates that the size of ODAs is independent of the test frequency, and therefore, the length of time, of the test. These researchers suggest that the formation of ODAs is influenced by both time-dependent hydrogen damage and cycle-dependent fatigue damage.

Initiation at Porosities and Inhomogeneities

In the absence of inclusions, internal cracks from UHCF can initiate at porosities and inhomogeneities. As found by Bathias [7], crack initiation at porosities can be significant when the R ratio is low, even when inclusions are present, particularly in nickel alloys. Investigation of seeded and standard N18 nickel alloy (Figure 7 [7]) shows that at $R = 0$, the effect of inclusions is less than at $R = 0.8$, as expected because initiation at porosities would mask the effect of the inclusions. Murakami [11] investigated the effect of inhomogeneities. Murakami found that when the size of inclusion is very small, as in super-clean steels prepared by electron beam remelting, initiation at inhomogeneities, such as bainite, can become dominate, because the size of the inhomogeneities is larger than the size of the inclusions.

UHCF in Silicon

Silicon does not exhibit the stepwise S-N curve seen in metals, because fatigue in silicon operates by a fundamentally different mechanism. Because silicon does not experience room-temperature plasticity, extrinsic toughening, or susceptibility to stress-corrosion cracking, it is not expected to fail by fatigue in air at room temperatures. However, in both mono- and polycrystalline silicon thin films, fatigue failures have been observed (Figure 12) [3,4]. The fatigue behavior appears to be continuous across a range of lifetimes from approximately 10^5 to 10^{11} cycles, suggesting that a single fatigue mechanism acts in all regions of the S-N curve. Muhlstein, Stach, and Ritchie [4] suggest that fatigue of thin films of silicon occurs by an environmental mechanism. Reaction with the atmosphere causes the formation of a layer of SiO_2 on the surface of the silicon. Crack initiation occurs in the oxide layer. Then the freshly exposed silicon oxidizes. This process repeats until a critical crack size is reached, at which point the silicon fractures.

CONCLUDING REMARKS

The studies reviewed in this paper covered a wide range of materials, including plain-carbon and alloy steels, cobalt-, nickel- and aluminum-base alloys, and silicon, tested under temperatures from -196°C to 450°C , frequencies from 20 Hz to 20 kHz, and different environments, including air and vacuum. Despite this variety of data, several conclusions are drawn and summarized in Figure 14:

- 1 Most importantly, fatigue failure can and does occur at lives above 10^7 cycles and stresses below traditional fatigue limits. This was determined for both ferrous and non-ferrous alloys.
- 2 Most of the materials studied exhibited a stepwise S-N curve as in Figure 14. An initial downward sloping region is seen, followed by a HCF plateau. The HCF plateau is related to the threshold value for persistent slip bands (PSBs), which lead to crack initiation at surface roughness. At lives longer than the HCF plateau, another downward sloping curve is seen in the UHCF region. Several materials then showed a second, UHCF, plateau, which is related to the threshold value for irreversible slip. In silicon, only one plateau is seen, because the mechanism governing fatigue in silicon is fundamentally different.
- 3 The location of crack initiation moves from the surface for LCF and HCF to the interior for UHCF. Cracks that initiate in the interior usually do so at inclusions, resulting in a fish-eye microstructure. However, internal cracks can also initiate at porosities and inhomogeneities.
- 4 UHCF failures by internal cracks can occur when the stress is sufficiently low to prevent the formation of surface cracks. Several mechanisms are proposed to explain why internal cracks cause failure at longer lives. Internally initiated cracks, being isolated from the environment, will exhibit more complete crack closure and lower fractions of irreversible slip, which leads to slower crack growth rates, and therefore, longer lives.

Internal cracks will also have a smaller stress intensity factor than surface cracks. In addition, the long lifetimes associated with UHCF allow hydrogen embrittlement due to the diffusion of hydrogen to internal inclusions to become significant, leading to fatigue crack growth by hydrogen embrittlement, and, ultimately, failure.

- 5 High test frequencies can be detrimental to fatigue limits due to specimen heating. Cooling the sample can mitigate the frequency effect. Other testing parameters that can affect the fatigue results include temperature and environment. Lower temperatures generally lead to higher strengths and longer lives. A corrosive environment can decrease the life in LCF and HCF due to oxidation of the surface. However, the environment does not have a large effect in the UHCF region, because the internally initiated cracks dominate in this region are isolated from the environment.

Research into UHCF is still at its beginning stages. Several areas for future research present themselves. Future research is necessary to determine the UHCF behavior of other materials not covered in this paper. In addition, future research is needed to confirm the existence of the second, UHCF, plateau, which is strongly suggested in some studies presented here. Research is also needed to further examine the effect of temperature evolution during high-frequency tests. Once the temperature evolution is fully understood, it may be applied to high-frequency applications, both cooled and not cooled, according to the environment in which they are employed. However, unless specimen heating due to high frequency is specifically being addressed, it would be advisable, for comparison purposes between laboratories, to run tests of different frequencies at the same temperature, by cooling high-frequency test specimens.

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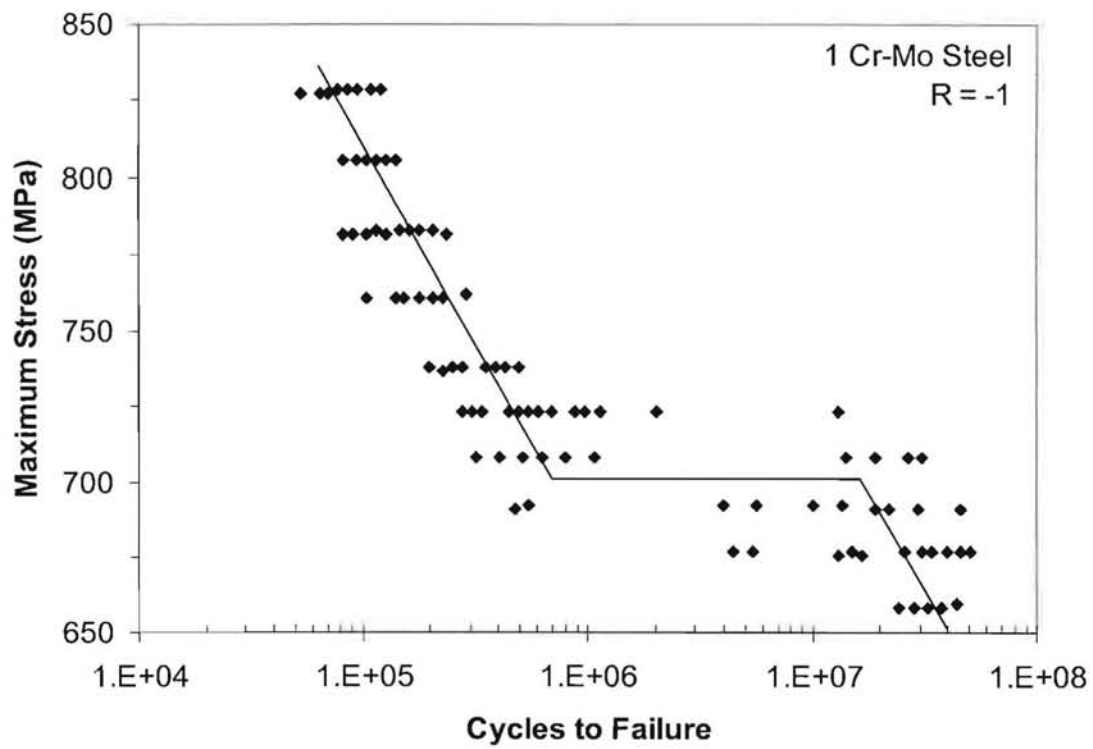


Figure 1: S-N curve of 1Cr-Mo steel with $R = -1$ in air at room temperature [5].

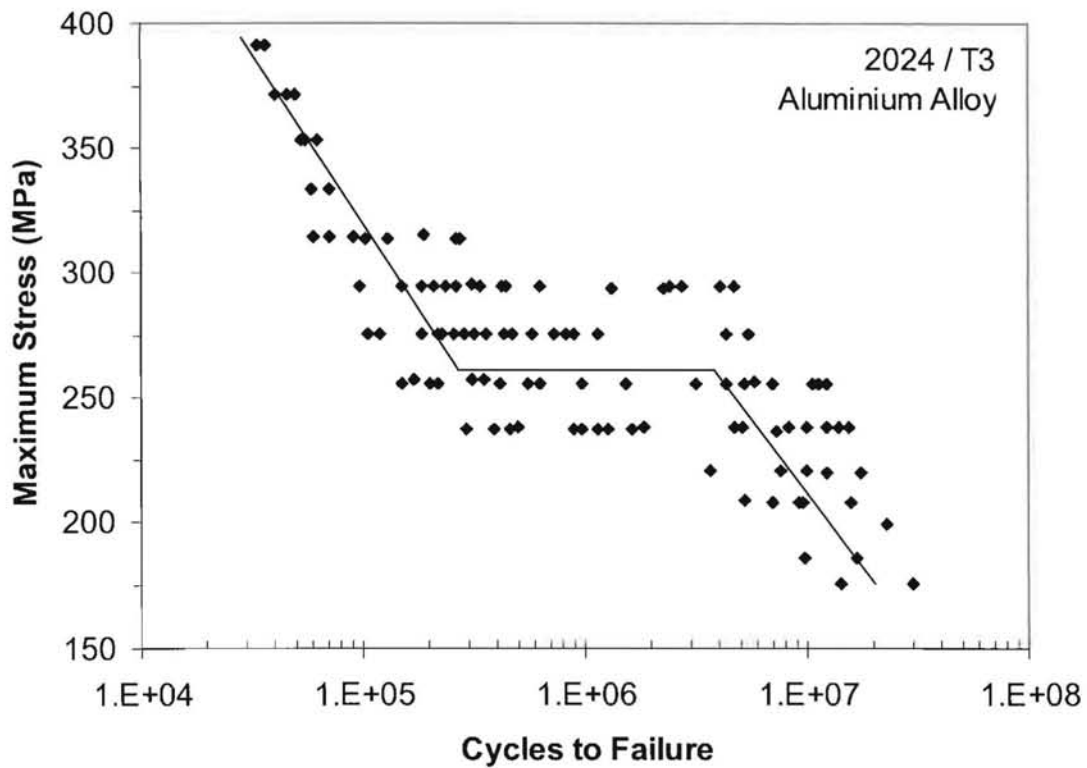


Figure 2: S-N curve of 2024/T3 aluminum alloy in air at room temperature [1].

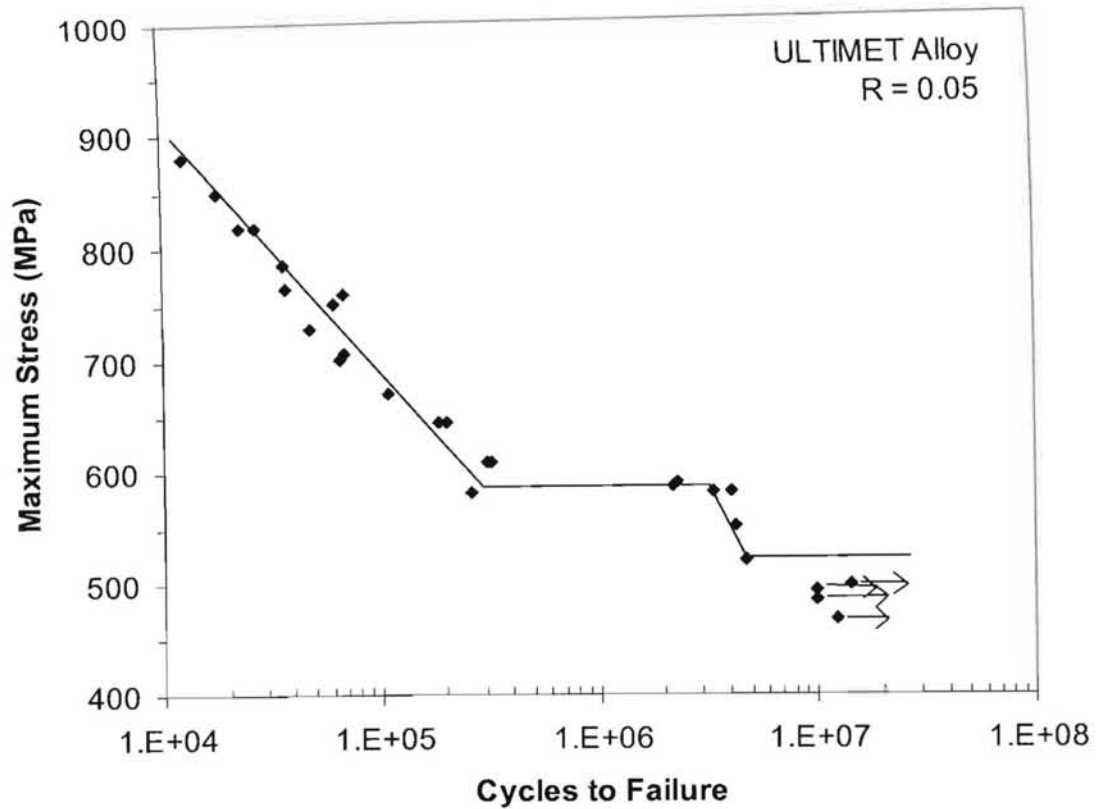


Figure 3: S-N curve of ULTIMET superalloy with R = 0.05 in air at room temperature [6, 8].

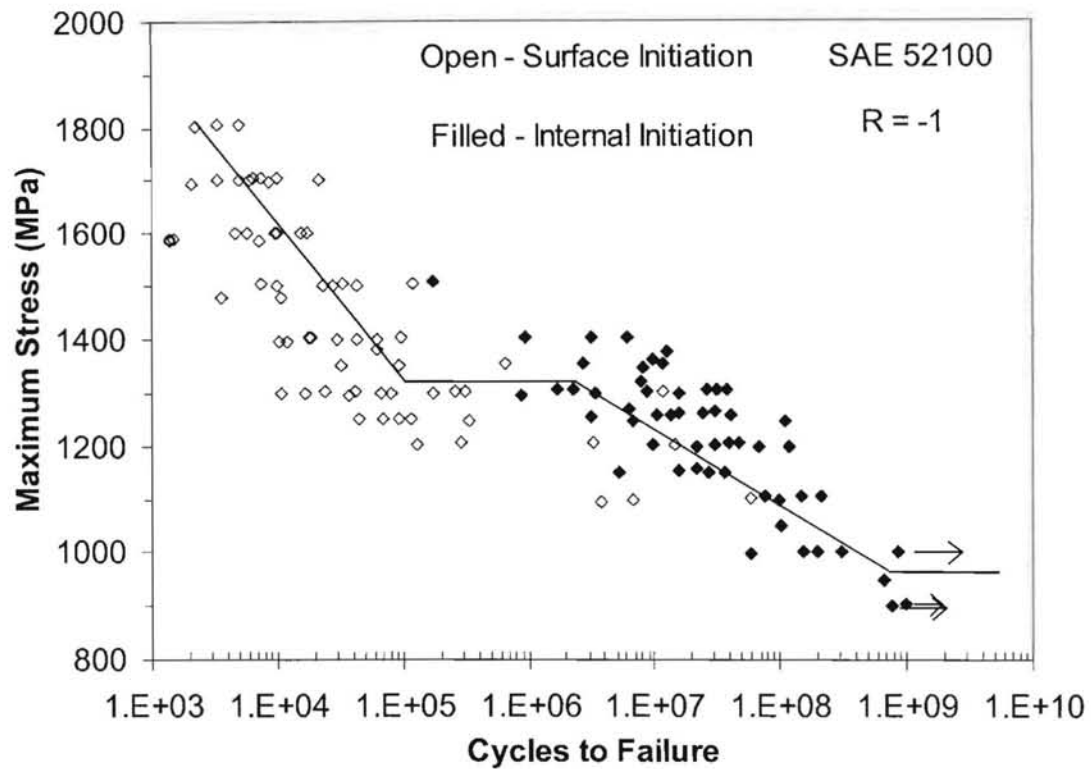


Figure 4: S-N curves of SAE 52100 bearing steel showing initiation location with R = -1 in air at room temperature [9].

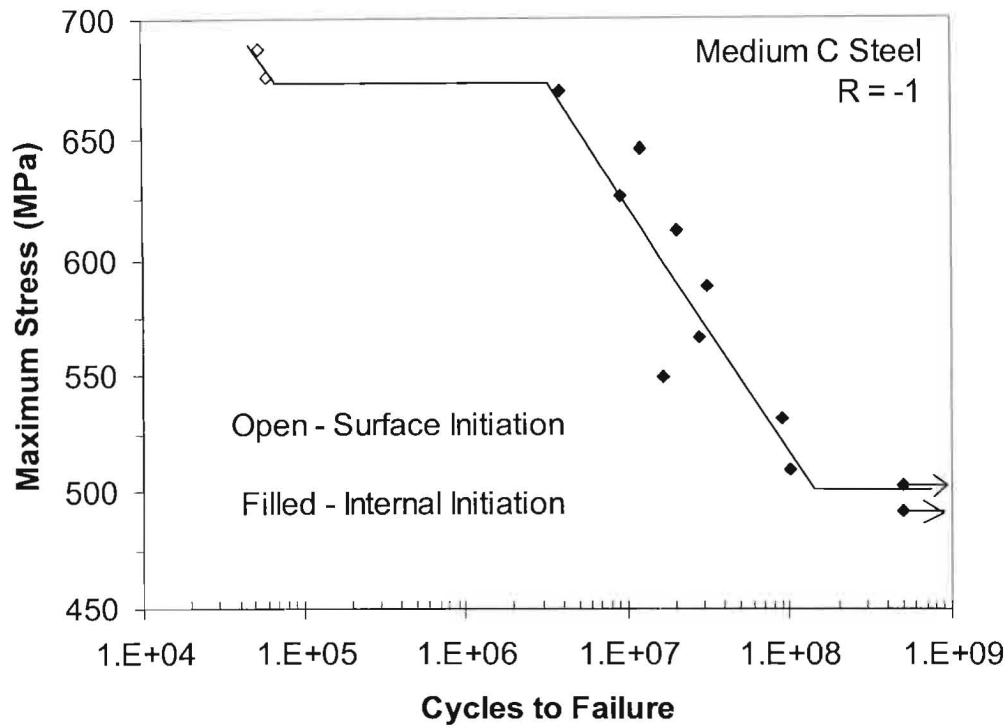


Figure 5: S-N curve of medium carbon steel showing initiation location with $R = -1$ in air at room temperature [10].

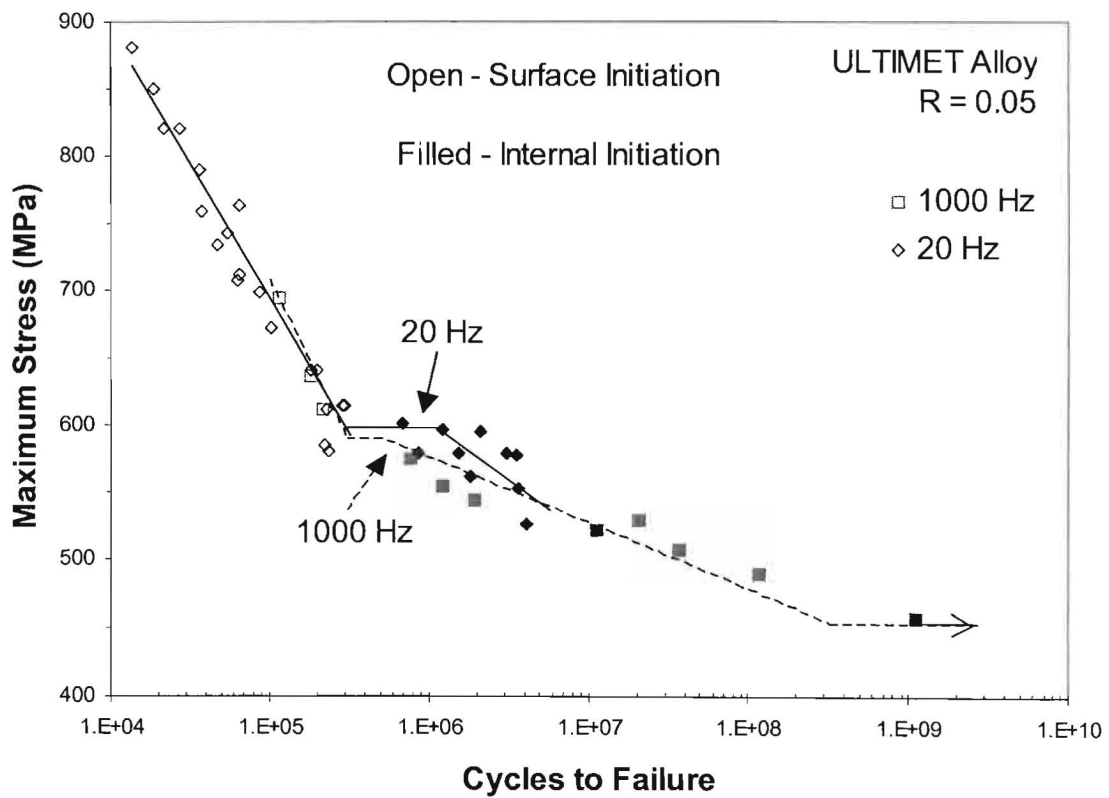


Figure 6: S-N curve of ULTIMET alloy with $R = 0.05$ in air at room temperature [6, 8].

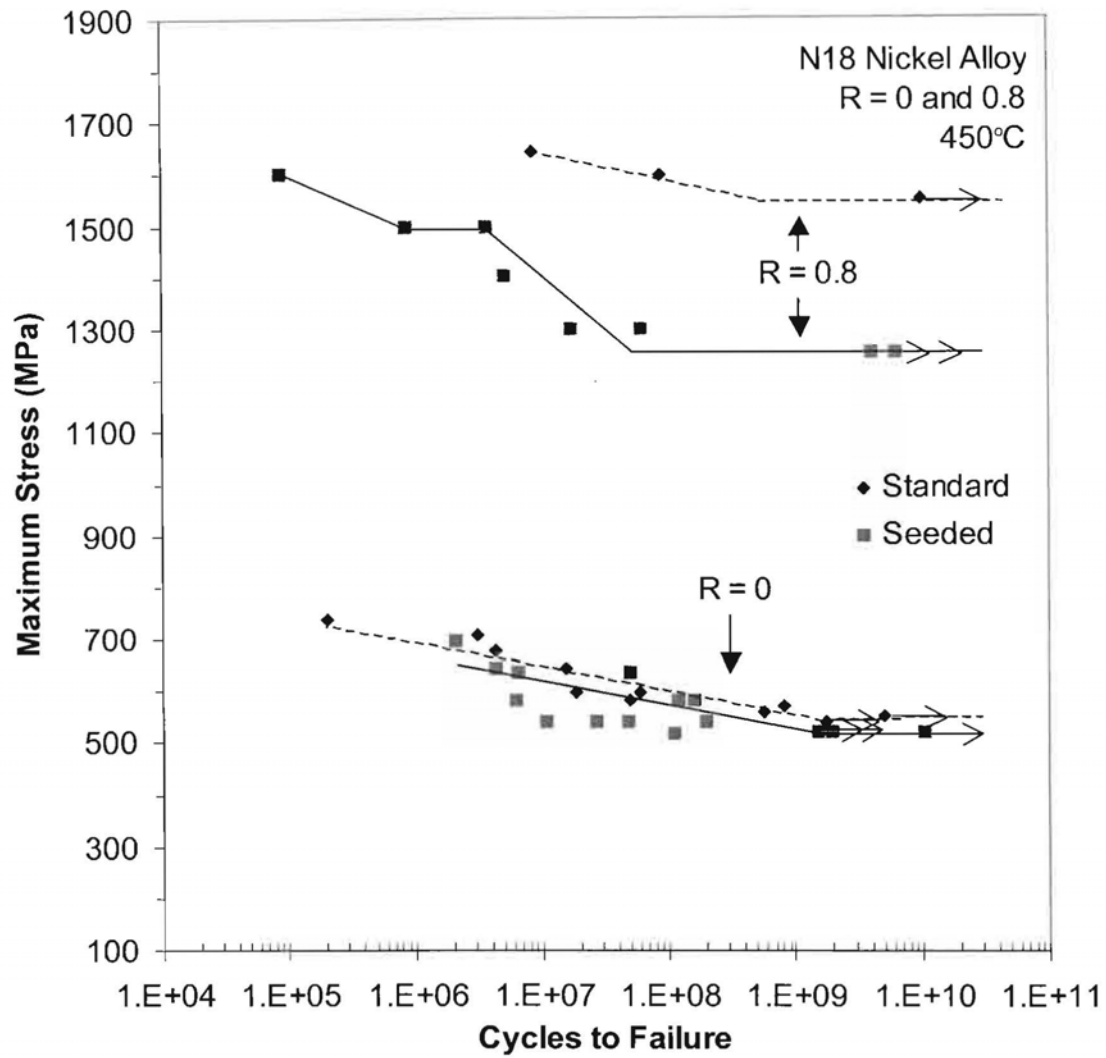


Figure 7: S-N curves of seeded and standard N18 nickel alloy with $R = 0$ and 0.8 in air at 450°C [7].

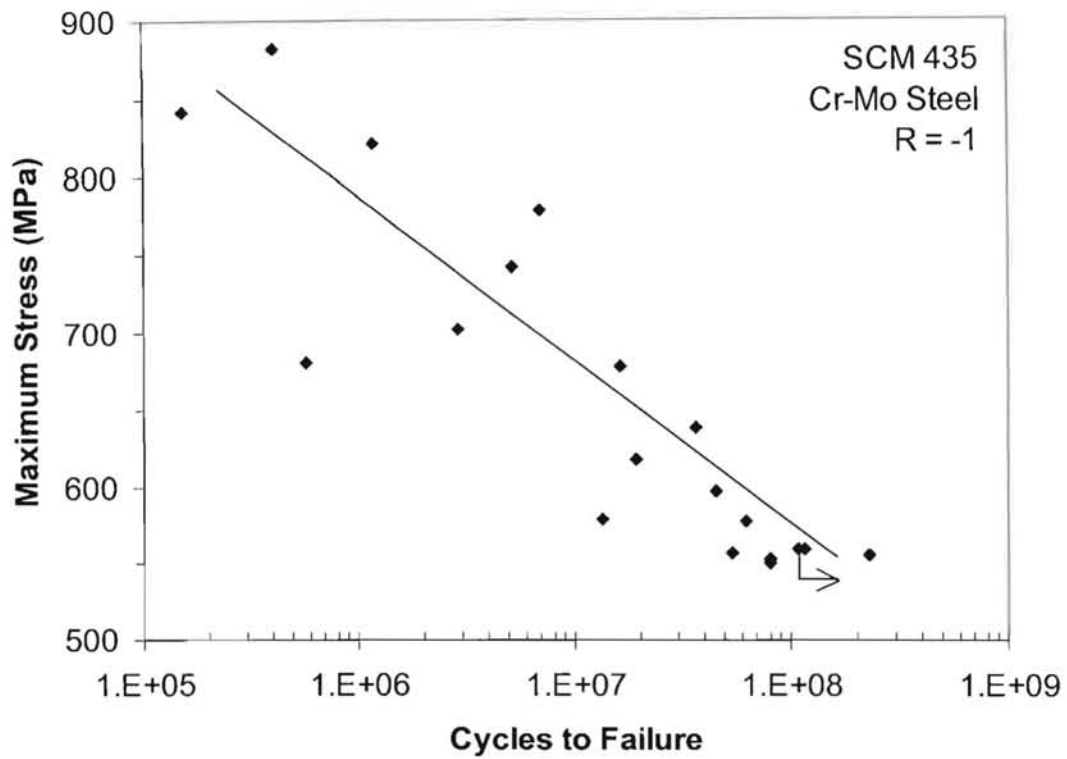


Figure 8: S-N curve of carburized/nitrided SCM 435 Cr-Mo steel with R = -1 in air at room temperature [10].

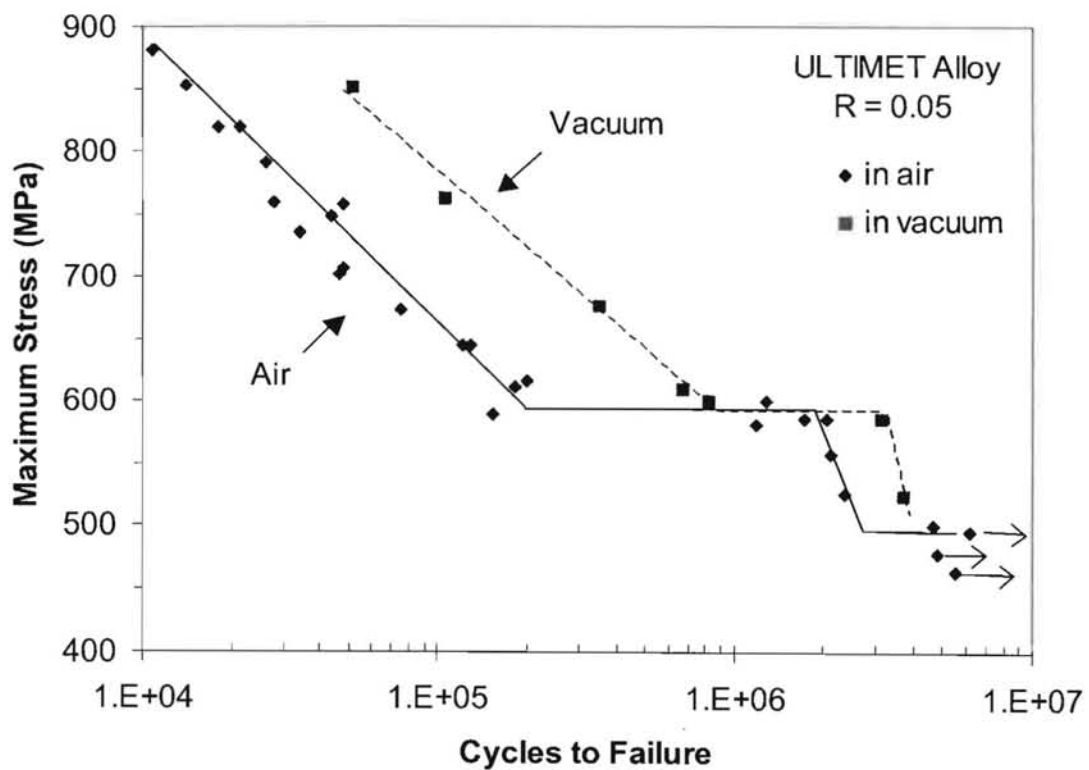


Figure 9: Effect of environment on S-N curve of ULTIMET alloy with R = 0.05 in air at room temperature [6, 8].

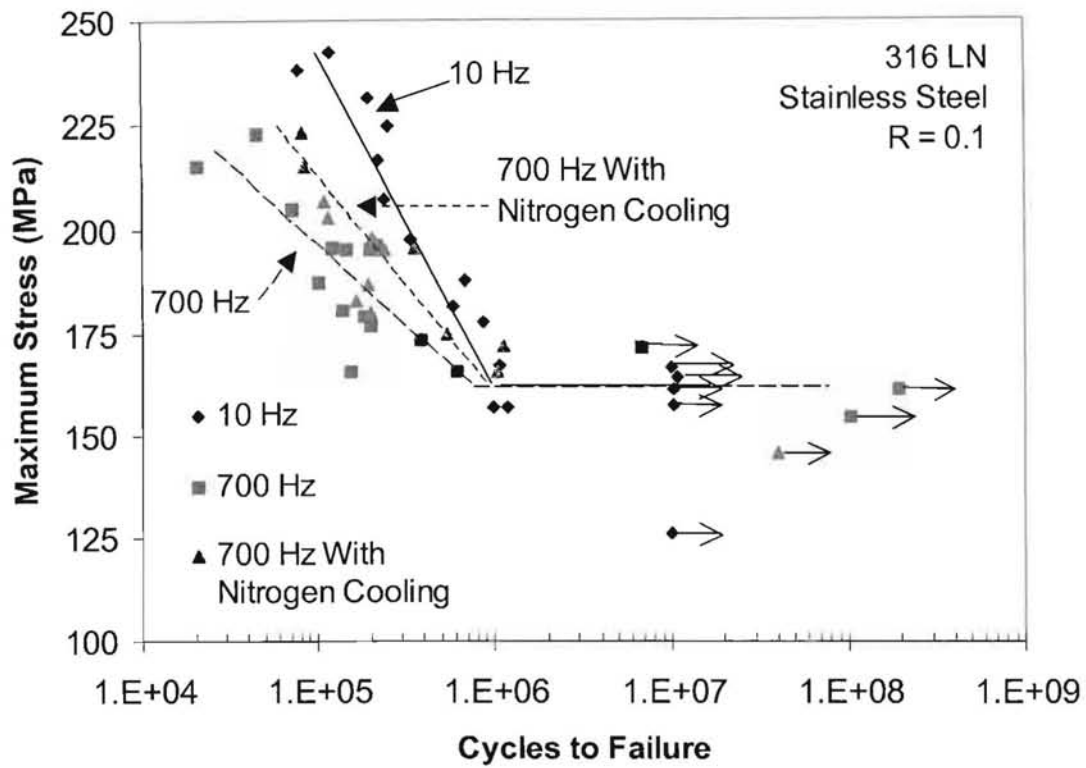


Figure 10: Effect of frequency on S-N curve of 316LN stainless steel with $R = 0.1$ in air at room temperature [16].

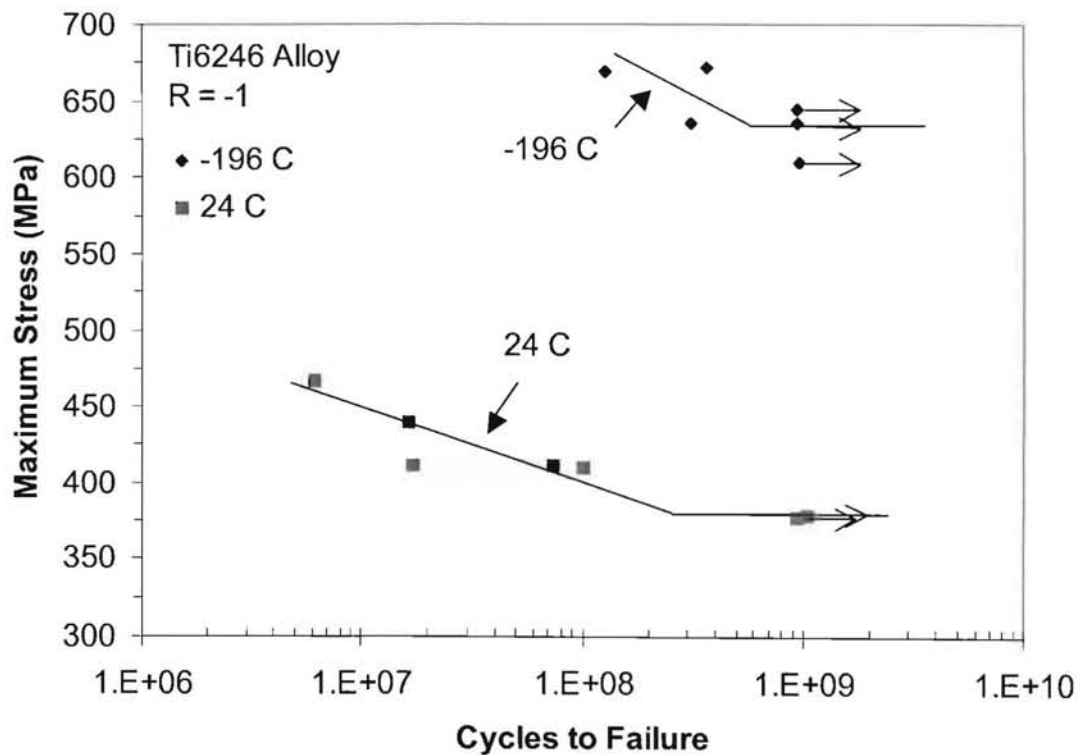


Figure 11: Effect of temperature on S-N curve of Ti6246 titanium-based alloy with $R = -1$ in air at -196°C and 24°C [7].

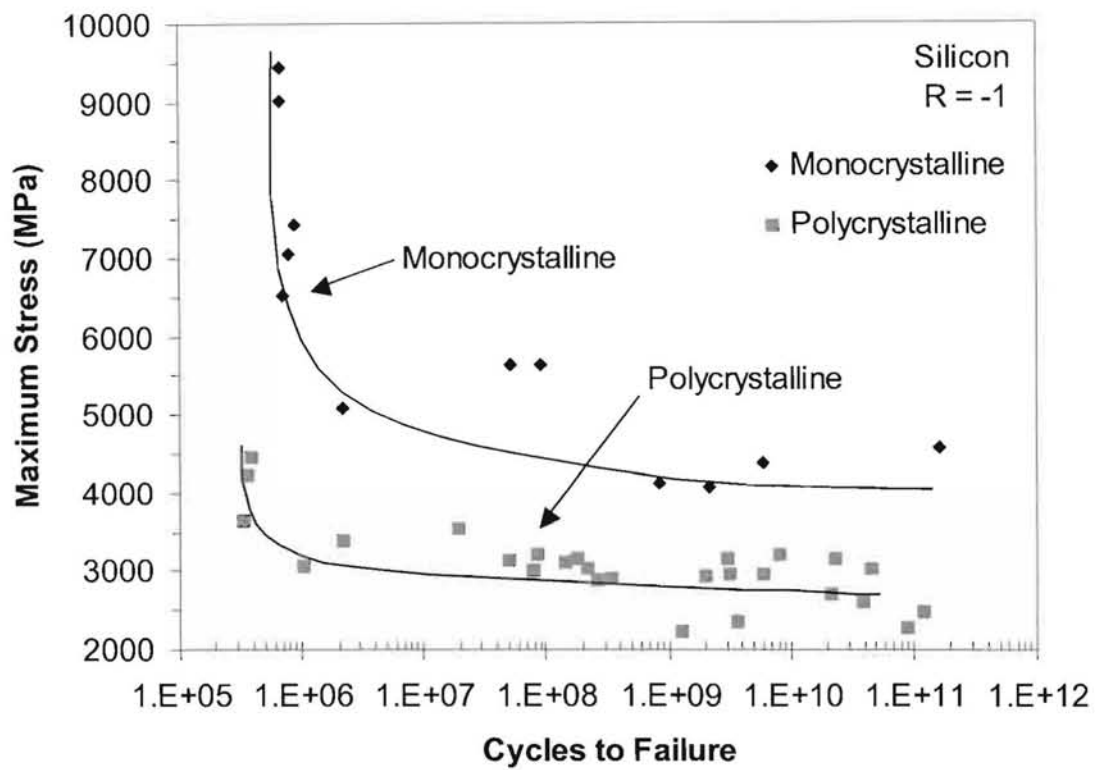
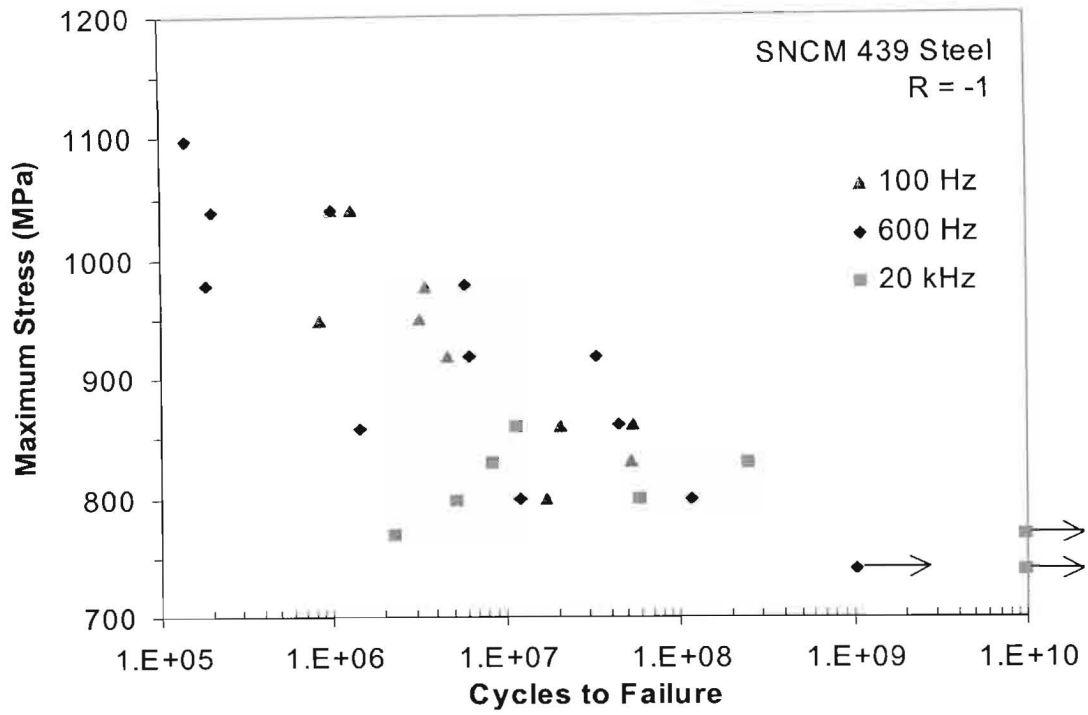
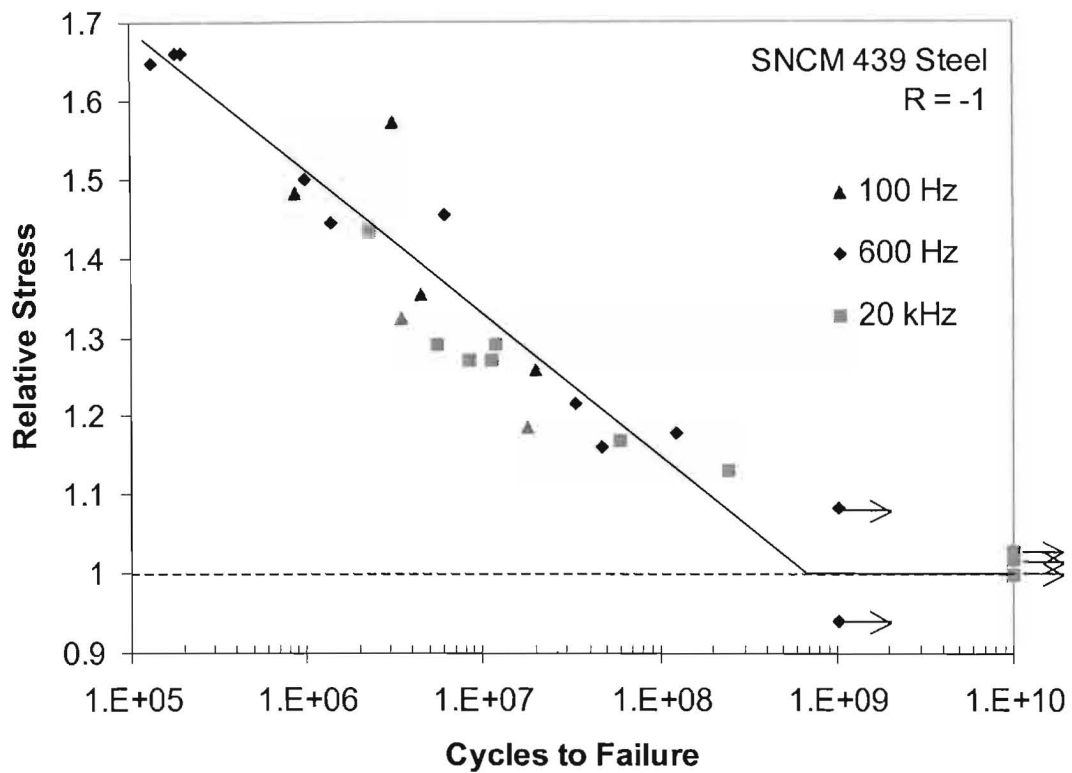


Figure 12: S-N curves of mono- and poly-crystalline thin-film silicon with $R = -1$ in air at room temperature [3, 4].



(a)



(b)

Figure 13: Effect of frequency on S-N curve of SNCM439 steel with R = -1 in air at room temperature, showing (a) normal and (b) modified plots [19].

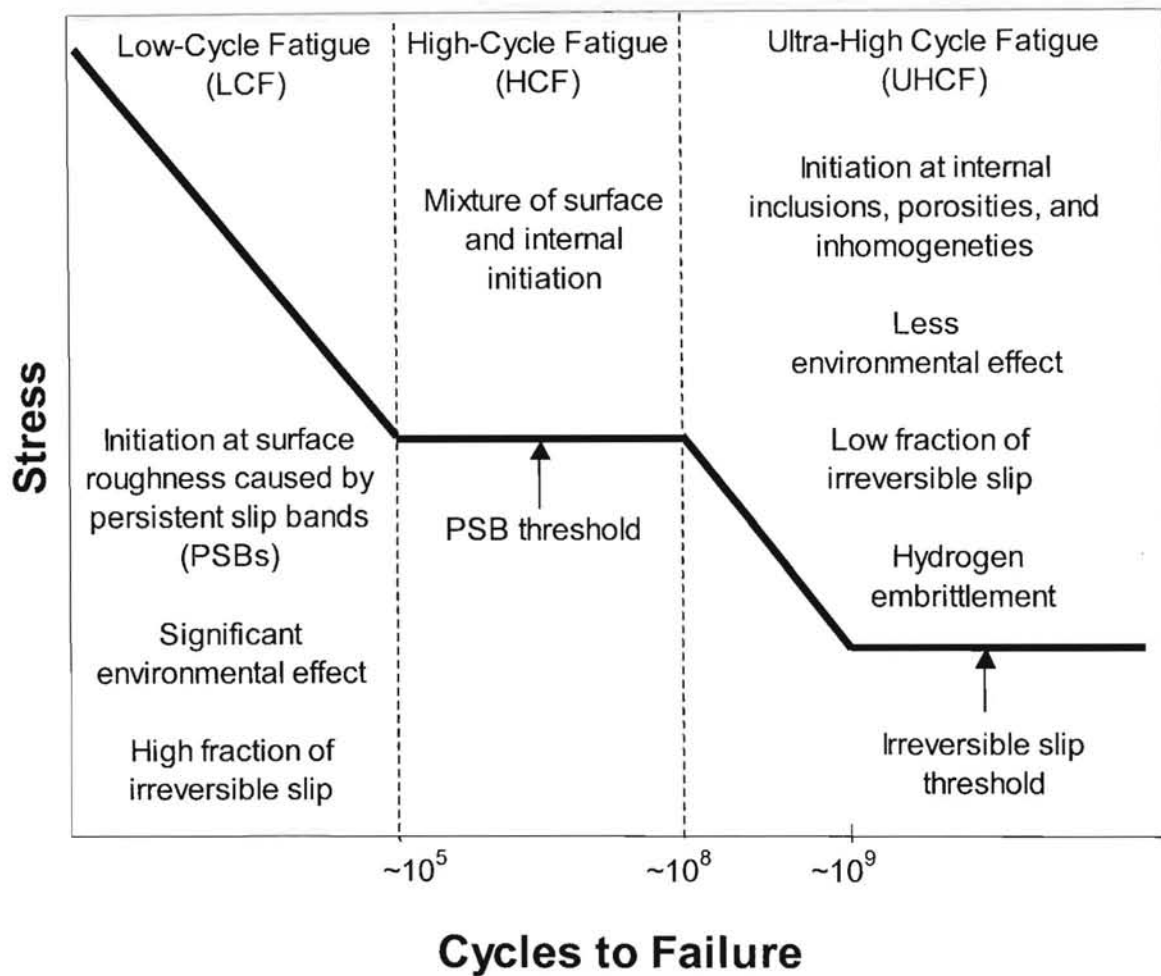


Figure 14: Schematic of a typical S-N curve with a HCF plateau around 10^6 to 10^8 cycles and a UHCF plateau above about 10^9 cycles.